THESIS FOR THE DEGREE OF

DOCTOR OF PHILOSOPHY

ON

EFFECTS OF ALLOYING ELEMENTS ON THE PROPERTIES OF ALUMINIUM AND ITS ALLOYS

By

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CERTIFICATE

Certified that the thesis entitled "EFFECTS OF ALLOYING ELEMENTS ON THE PROPERTIES OF ALUMINIUM AND ITS ALLOYS" which is being submitted by Sri Rajendra Sharan in fulfilment for the award of the Degree of Doctor of Philosophy in Metallurgical Engineering of University of Roorkee is a record of his own work carried out by him under my supervision and guidance. The matter embodied in this thesis has not been submitted for the award of any other Degree.

This is further to certify that he has worked for a period of 36 months for preparing his thesis for Ph.D. Degree at the University.

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ABSTRACT

The present investigation forms a part of a series of investigations that have been undertaken in this Department on a study of aluminium and its alloys. The various physical and mechanical properties studied include metallographic structure, lattice parameter, density, coefficient of thermal expansion, hardness, tensile strength, percentage elongation and machinability.

Most of the earlier investigations of the lattice parameter were carried out at ordinary temperature and very little work of the study of lattice parameter at elevated temperatures has been reported. In the present investigation, an attempt has been made to determine as accurately as possible the changes in lattice spacing with temperature and composition under the equilibrium conditions of aluminium, aluminium-copper and aluminium-silver alloy systems by using high temperature Unicam X-ray diffraction camera. The knife edge of the camera was calibrated by taking silver standards. The camera was also calibrated for mean specimen temperature corresponding to thermo-couple temperature by taking photographs of silver specimens at different temperatures. Nelson Riley's extrapolation method was adopted for the precise determination of lattice parameter. The equations representing the relation between lattice parameter and temperature

have been determined by using a digital computer. The values of lattice parameter as determined from the equations are in close agreement with the experimentally determined values. The density and coefficient of thermal expansion were found out in single phase regions. In the case of pure aluminium, the coefficient of thermal expansion as determined with reference to lattice parameter at 0° C decreases slightly with the increase in temperature. In the case of aluminium-copper system, the values at 525° and 548° C as determined with reference to lattice parameter at the same temperature, were found to remain almost unchanged upto 4% copper addition. Beyond this the values of coefficient of thermal expansion decrease at an appreciable rate as the copper concentration increases. The coefficient of thermal expansion for the alloy containing 7.5% silver is lower and for the alloy containing 19.9% silver is higher than that of pure aluminium at a given temperature and determined in the single phase region. The value of lattice parameter for alloys containing 7.5% and 19.9% silver decreases and increases respectively with rise in temperature. The solid solubility line as determined by high temperature X-ray diffraction method between 475° and 548° C was found in agreement with that reported by earlier workers based on other methods. The maximum solid solubility at the eutectic temperature was found out to be 5.7 weight percent confirming the results obtained by Raymor. However, the value does not agree with that reported by Ellwood and Silcock which is 5.85 weight percent.

Survey of the literature has indicated that although the effects of common alloying elements on the properties of aluminium have been thoroughly investigated, very little work on the effects of tantalum and rare earths additions has been reported. The results so far on aluminium and aluminium alloys have been rather conflicting. Practically no work has been reported in the case of aluminium-silicon and aluminium-magnesium alloys. Most of the work reported earlier, has been carried out by using mixtures of rare earth metals and not individual metals. In the present investigation the effects of tantalum and rare earths on the microstructure, hardness, machinability, tensile strength and percentage elongation both at ordinary and elevated temperatures of aluminium and its alloys have been studied. Power measured by drill calorimeter and chip formation study have been taken as indices for the evaluation of machinability. The metallographic study has been made in all the cases to explain the changes in properties on alloying additions. The rare earths have been added in the form of misch metal, lancer-amp[2], rare earth fluorides and individual elements viz., cerium, lanthanum and neodymium. The work has been carried out on aluminium, aluminium-magnesium alloy 218, aluminium copper alloys 195 and 122 and aluminium -silicon alloys. Systematic and extensive investigation has been undertaken in the case of aluminium-silicon alloys with a view to explore the possibility of tantalum and rare earth metals primarily as modifier and secondarily as beneficial alloying additions.

Following beneficial effects of additions of tantalum and rare earths to aluminium and its alloys at ordinary and elevated temperatures have been established:-

Aluminium :

i) Large refining effect with optimum additions of 0.01% Ta, 1.5% Ce, 0.1% La, 2% Misch metal and 2% Rare earth fluorides.

ii) Improvement in percentage elongation with optimum additions of 0.3% La, 0.2% Nd, 1.5% Rare earth fluorides, 0.5% Misch metal and 0.5% Lancer-amp^2. Increase in percentage elongation at 400° C with optimum addition of 0.75% Misch Metal. Large increase in hardness with optimum additions of 3% Misch metal, 1.5% Rare earth fluorides and 2% Lancer-amp^2 .

iii) Large improvement in machinability with optimum additions of 0.5% Ce, 0.05% La, 0.1% Nd and 1% Misch metal.

Alloy 218 :

i) Large refining effect with optimum additions of 0.15% Ta, 0.2% La.

ii) Large improvement in tensile strength at 200° and 400° C with optimum addition of 4% misch metal. Increase in percentage elongation with additions upto 0.15% Ta, 0.5% Ce, 0.3% La and 0.1% Nd. Increase in hardness with optimum additions of 1.5% Ce and 1% Misch metal.

iii) Large improvement in machinability with optimum additions of 0.075% Ta, 0.5% Ce, 0.1% La, 0.1% Nd and 4% Misch metal.

Alloy 195 :

i) Large refinement with optimum additions of 0.75% Ce, 0.2% Nd and 2% Misch metal.

ii) Increase in percentage elongation with additions upto 0.1% Ta, 1.5% Ce, 0.3% La and 0.1% Nd. Large increase in hardness with additions of 0.3% La, 0.1% Nd and 1% Misch metal.

iii) Large improvement in machinability with optimum additions of 0.1% Ta, 0.1% La and 3% Misch metal.

Alloy 122 :

i) Pronounced refining effect with additions of 0.5% Ta, 0.75% Ce, 0.2% La and 1% Misch metal.

ii) Large improvement in tensile strength with optimum additions of 0.05% Ta, 0.5% Ce, 0.1% La, 0.3% Nd and 3% Misch metal. Large increase in hardness and percentage elongation with the addition of 3% Misch metal.

iii) Large improvement in machinability with optimum additions of 1.5% Ce, 1% Misch metal and only slight effect with 0.1% Ta addition.

Aluminium-silicon alloy (Hyper)

i) Partial modification effect produced in microstructure with optimum additions of 0.015% Ta, 0.75% Ce and 0.05% La. Complete modification effect produced with optimum additions of 1% Misch metal, 1.5% Lanceramp^2 and 0.2% Rare earth fluorides.

ii) Large improvement in tensile strength at ordinary temperature with optimum additions of 0.025% Ta, 0.75% Ce, 0.05% La, 0.3% Nd, 1% Misch metal, 3.5% Lancer-amp^2 and 0.2% Rare earth fluorides. Large improvement in tensile strength with the addition of 1% Misch metal. Large increase in percentage elongation at ordinary temperature with optimum additions of 0.05 % La, 1% Misch metal, 2% Lancer-amp^2 and 0.2% Rare earth fluorides. Large beneficial effect of the addition of 1% Misch metal on percentage elongation at 200° and 400° C.

iii) Large improvement in machinability with optimum additions of 0.1% Ce, and 0.5% Misch metal.

<u>Aluminium-silicon alloy (hypo):</u>

i) Large refinement with addition upto 0.2% Rare earth fluorides. Increase in the size and amount of primary alpha in the microstructure with lancer-amp^2 and misch metal additions.

ii) Improvement in tensile strength and percentage elongation with optimum addition of 0.2% Rare earth fluorides. Large beneficial effects on hardness with addition of 0.2% Rare earth fluoride and only slight beneficial effect with the additions of both misch metal and lanceramp[2] between 1.5 and 3%.

It is hoped that these investigations will prove useful in the development of new aluminium alloys with better machinability and mechanical properties for use at ordinary and high temperatures.

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CHAPTER I

INTRODUCTION

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INTRODUCTION

1.1 OBJECT AND SCOPE OF THE PRESENT INVESTIGATION.

1.1.1 General introduction :

A striking development of modern industrial societies is the way in which traditional materials are progressively being replaced by new ones. One such material is aluminium, which is actually preferred for several applications which were formerly in the province of steel, copper, lead, zinc and other long established materials. The chief reasons for this are availability, lower price and technical suitability. The properties of aluminium that make this metal, and its alloys, the most economical for a wide variety of uses are appearance, lightness, fabricability, corrosion resistance, physical and mechanical properties. Because of its low neutron capture cross-section and corrosion resistance, aluminium warrants favourable consideration as a structural material for nuclear reactors and for canning and cladding of nuclear fuel elements. It has also found extensive use in electrical conductors, and air-craft and other light-weight structures. However, its low melting point, decreased high-temperature strength, lowered corrosion strength of most of its high strength alloys and the high coefficient of thermal expansion, limit the use of aluminium and its alloys mainly to low temperature applications.

Extensive research of fundamental nature is being carried out to understand the mechanism of alloy-formation and correlate the various changes in its mechanical properties due to alloying and heat-treatment, with change in crystal and metallographic structure. Attempts are also being made to develop new alloys of aluminium with new elements such as rare-earths etc. for specific purposes.

The present investigation forms a part of a series of investigations that have been undertaken in this Department on a study of aluminium and its alloys. The various properties studied include lattice parameter, density and coefficient of thermal expansion at elevated temperatures, metallographic structure, machinability and mechanical properties at room and elevated temperatures. The whole investigation can be broadly classified as follows :-

i) High temperature X-ray diffraction study of aluminium, aluminium- copper and aluminium-silver alloys.

ii) Effects of tantalum rare-earth elements and mixtures of rare-earth elements on the micro-structure , machinability and mechanical properties at ordinary temperature of aluminium, aluminium-copper, aluminium-magnesium and aluminium-silicon alloys.

iii) Effects of misch-metal additions on the high temperature mechanical properties of aluminium, aluminium-copper, aluminium-magnesium and aluminium silicon alloys.

1.1.2 High temperature X-ray diffraction study of aluminium, aluminium-copper and aluminium-silver alloys :

Since the formation of a solid solution is accompanied by variation in mechanical properties and lattice parameters, a study of variation in lattice parameters is useful to understand the changes in the physical and mechanical properties. The densities, thermal expansion as well as vacancies on which mechanical properties depend, can also be studied by lattice spacing measurement. In fact, much work has been carried out by other investigators to study the effects of various common alloying elements on the metallographic structure and mechanical properties of aluminium (1)*. A systematic study of the quantitative evaluation of machinability of some commercial aluminium alloys has been made by the author in his earlier investigation (2). The effects of alloying elements viz., silicon, copper, zinc and magnesium (3,4,5,6) on the liquid densities, and total shrinkage during solidification have also been reported in earlier publications.

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* The numbers in paranthesis indicate the Reference Numbers given at the end.

Influence of mould materials on apparent shrinkage and mould wall movement of aluminium alloys has also been thoroughly investigated (7). Although the engineering properties of various aluminium alloys have thus been extensively investigated, we find that most of the earlier investigations of the lattice parameter were carried out at ordinary temperature where the solid solubility in most cases is very small. In such cases, in order to determine the dependance of lattice parameter on the solute concentrations, supersaturated solid solutions were obtained by quenching the alloy from the high temperature at which the solid solubility was greater. It should be emphasized here that the state of the supersaturated solid solution should be different from equilibrium solid solution. Survey of the literature has indicated that very little work on the study of lattice parameter of aluminium alloys, at high temperature has been reported. **Ellwood** & **Silcock** (8) were the first to examine lattice spacings of aluminium-copper alloys in a temperature range of 400° - 600° C. The lattice spacings at lower temperature in equilibrium solid solution have not been reported by them. They have , however, reported low accuracy in measurement. There also appears to be some discrepancy in calibration with regard to temperature measurement. The variation of lattice parameter of alloy in equilibrium condition with composition at different temperatures was not studied. Hume-Rothery and Boultbee (9) made the measurement on the alloy of Al-Cu containing 1.62 atomic percentage Cu, Al - Mg solid solution and alloys of Al-Si in the temperature range of -50° to 200° C. Lattice spacings at high temperatures have also not been reported by them. Practically no work has been reported in case of aluminium silver alloys at elevated temperatures. Thus there is lack of systematic data on the lattice spacing measurements of aluminium alloys in equilibrium at elevated temperatures and there seems to be a need of thorough investigation in this direction.

In the present investigation lattice spacing measurements at elevated temperatures have been made by using high temperature Unicam X-ray diffraction camera. The camera was calibrated for mean specimen temperature, corresponding to thermocouple temperature. Nelson-Riley's extrapolation correction was applied for precision determination of lattice parameter. The knife edge of the camera was also calibrated by taking silver standards. The work has, however, been limited to study the effect of copper and silver on the lattice spacing of aluminium at elevated temperature. The work has not been undertaken in case of aluminium alloys with tantalum and rare earth additions although very elaborate work has been carried out on machinability and mechanical properties because of negligibly small solid solubilities of these elements in aluminium. The equations were determined for lattice parameter vs. temperature for each composition with the help of computer. In the case of pure aluminium the lattice parameter, thermal expansion and densities were calculated for different temperatures. In case of aluminium -copper system the solubility line was redetermined by high temperature X-ray method, between 475° and 548° C. The values of coefficient of thermal expansion and densities at elevated temperatures have been calculated for both aluminium copper and aluminium silver alloys.

1.1.3 Effects of tantalum, rare earth elements and mixtures of rare earth elements on the microstructure, machinability and mechanical properties at ordinary temperature of aluminium, aluminium copper, aluminium magnesium and aluminium silicon alloys :

We are rapidly moving in the world of supersonic jet flights, missiles etc. Resources of metals in general are getting depleted at an enormous rate, for example, proved world reserves of copper will barely last 30 years and even if its production continues to multiply three or four fold over the next two to three decades, the supplies would still not be adequate. We shall then have to look around for metals of the rare earth group, which today have relatively much smaller industrial scale uses. It has been clearly shown that metals of the rare earth group are valuable alloying elements both in micro and macro additions (10).

Metallurgical applications of rare earths may be classified in two broad headings :-

i) Applications in which rare earths form an alloy with some other metals e.g. with aluminium and magnesium.

ii) Applications where rare earths play a weak role as alloying elements but function fundamentally as purifying elements.

In each case rare earths are added in the form of mixture, and not in their pure state . Two types of mixtures Misch Metal and Lancer-Amp. are widely used. Mixture of fluorides and oxides have been also used in many cases of ferrous and non-ferrous alloys. The applications of rare earths as mixed metals or their oxides have been extensively explored in the past decade and found capable of providing a variety of alloys with markedly improved properties.

From the stand point of the amounts utilized in the industry misch metal is the most important of the rare earth metals. Approximately one quarter of the rare earth compounds produced in the United States goes into the manufacture of misch metal and cerium. Misch metal is a mixture of cerium group rare earth metals and is prepared either from monazite or bastmasite. Some examples of the current uses of misch metal are given in the following list (11).

i) Lighter flints contain 70 - 80 percent misch metal with iron as remainder (12).

ii) Important component in effecting nodularization of cast iron in the ductile form (13) in improving the machinability and mechanical properties of grey (14) malleable (15) and alloyed cast iron (16).

iii) Used as an additive to low alloy steel castings, misch metal improves their ductility, impact resistance and other physical properties (17).

iv) Misch metal appears to improve hot working properties of stainless steel (18).

v) Creep resistance (19) and tensile strength of some high temperature Zr-Mg alloys for jet engines have been improved by adding as little as 3 percent misch metal.

vi) Good deoxidizer and desulphiser in steel (12)

vii) An alloy of misch metal, thorium and aluminium has been shown to be an efficient gating agent for use in electron tubes (20).

The use of rare earth metals as beneficial additions to light alloys has aroused considerable interest in recent years (10). The results so far on aluminium and aluminium alloys have been rather conflicting, although claims have been made for grain refinement and improvement of mechanical properties in quite a few cases. (21-29). The available data is poor and justifiable conclusions cannot be drawn. Practically no work has been reported in the case of aluminium-silicon and aluminium-magnesium alloys. Most of the work reported earlier **have** been carried out by using mixtures of rare-earth metals and not individual metals. Effects of tantalum and **individual rare** earth metals viz., cerium, lanthanum, and neodymium have not at all been investigated.

The aim of the present investigation has been to study systematically the effects of tantalum and rare-earth additions on the micro structure, hardness, ductility, **ordinary** temperature tensile strength and machinability of aluminium and its alloys. A systematic study has been initiated with the object of **developing** aluminium alloys with the rare earth metals as the alloying elements ,in view of the facts that India has plenty of rare earth metals available from monazite residues left after the extraction of atomic energy metals **viz.**, uranium, thorium, etc. The rare-earths have been added in the form of misch metal, lancer-amp-2, rare-earth fluorides and individual elements viz., cerium, lanthanum and neodymium. The work has been carried out on aluminium, aluminium-copper, alloys 195 and 122, aluminium-magnesium **alloy 218,** and the possibility of uses of the above mentioned metals as beneficial additions from strength and machinability stand point, have been explored. The systematic and extensive investigations have been undertaken in the case of aluminium - silicon alloys with a view to explore the possibility of the modification of aluminium alloys by rare earth additions. The microstructure, mechanical properties, as well as machinability of hyper and hypo-alloys treated with different amounts of tantalum and rare earth metals and cast in permanent as well as in sand moulds, have also been studied.

The chip formation study and measurement of power consumption by calorimetric method have been taken as indices for evaluating machinability of different alloys. The chip formation study includes the measurement of chip thickness ratio and shear plane angle ϕ . The metallographic structure has been taken in all the cases to correlate the changes in the properties with changes in microstructure of the alloys on additions.

1.1.4 Effects of misch metal addition on the high temperature mechanical properties of aluminium, aluminium copper, aluminium magnesium, and aluminium silicon alloys :

Literature survey has indicated that although little but conflicting reports are available on the effects of misch metal on the mechanical properties of aluminium alloys, practically no work has been reported on the effects of additions on the high temperature mechanical properties ;

although claims have been made for improving the mechanical properties in quite a few cases of magnesium base alloys(30) plain carbon and alloy steels (31). In the present investigation the effects of misch metal additions on the tensile strength and percent elongation of aluminium, aluminium - copper alloy 122, aluminium magnesium alloy 218 and aluminium-silicon alloys at elevated temperatures have been investigated, with a view to develop alloys for high temperature service conditions. Misch metal has been preferred over the individual elements because of its being cheaply available in this country. Avery tensile testing machine was used for the determination of tensile strength at elevated temperatures. Provision was made for three thermocouples that could be introduced through a series of radial holes with the help of which the temperature could be measured at the two ends and in the middle of the specimen. Duplicate and triplicate tests have been made in all the cases. As no systematic work seems to have been carried out, on these lines it is hoped that these investigations will prove useful in the development of new lighter alloys with better machinability, structural and mechanical properties for ordinary and high temperature service conditions.

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CHAPTER II

MEASUREMENT OF LATTICE SPACINGS IN ALUMINIUM AND ITS ALLOYS.

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MEASUREMENT OF LATTICE SPACINGS IN ALUMINIUM AND ITS ALLOYS

2.1 GENERAL INTRODUCTION.

According to Vegard's law the lattice spacing vs. composition curves for solid solution of salts of the same crystal structure are straight line joining the values for the two pure salts. This law was originally (32) advanced for salts in which oppositely charged ions are drawn togethef by coulomb forces until their electron clouds overlap and produce intense repulsive forces which balance the attraction. It was later extended to metals. Detailed examination of lattice spacing vs. composition relationships by Hume Rothery (33) and Owen (34) has shown that deviations from the Vegard's law occured in nearly all metallic solid solutions.

Vegard's law can be extended to cases where solute and solvent atoms possess different crystal structures, by converting the lattice spacing into d, the closest distance of approach of atoms (35). In the face centred cubic structures d = a/2 and corresponds to the spacing of atoms in the close packed (111) planes, while in the closed packed hexagonal structure d = a, and corresponds to the spacing of atoms in the basal plane (0001). When there are several different distances of approach in the structure of an element e.g in gallium, the closest distance of approach does not adequately express the size of the gallium atoms in a solid solution. In such a case a measure of the effective size may be obtained by extrapolating to 100% solute the lattice spacing/composition curve of gallium dissolved in a simple solvent e.g. copper. From the apparent atomic spacing a of the solute thus obtained; the corresponding apparent atomic diameter (A.A.D.) can be calculated by dividing(a) by $\sqrt{2}$. For gallium in copper the A.A.D. = 2.75 A°.

The apparent atomic diameter method is however, strongly dependent on the nature of the solvent element, bacause the extrapolated lattice spacing/composition curve contains both the atomic size contributions due to the solute and the valency and other electronic differences existing between the solute and the solvent. Thus the A.A.D. of gallium in gold is 2.83 A°

2.2 PRECISION DETERMINATION OF LATTICE PARAMETERS:

2.2.1 Instrumental errors and their corrections.

Precision work with Debye cameras is hampered by many factors:

i) Film shrinkage.

ii) Errors in measuring camera radius.

iii) Displacement of the specimen from the centre of the camera by improper adjustment.

iv) Displacement of the effective centre of the specimen from the centre of the camera by absorption of the rays in the sample.

v) X-ray beam divergence.

vi) Finite height of slits.

The film shrinkage alters the approximate circumference of the camera. A remedy for it is to print a scale on the film at the time the photograph is made. Subsequent measurement of this scale then reveals amount of shrinkage that has taken place. The same percentage correction can be applied to all measured distances on the film.

Straumanis (36) originated one method of correcting for film shrinkage. The film is wrapped around the camera in such a way that the ends come together on one side midway between entrance and exit holes. The position $\theta = 0$ is then determined as midway between right and left hand arc of small angle Debye rings, and the position $\theta = 90°$ is similarly determined from the large angle rings. Other θ values are then made by interpolation.

Another common method is to make a standard distance on the film. Two knife edges are built into the camera, so that they will form sharp limits to the exposed portion of the film and θ is calculated from the formula:

$$\frac{\theta°}{\theta°_1} = \frac{S}{S_1} \qquad \ldots (2.01)$$

where θ_1 is the diffraction angle exactly at the knife edges and S_1 is the circumferential distance between them.

The principle of calibration can be applied to individual films if the powders of the unknown sample and the standard substance are mixed and the super-imposed spectra are obtained. The angles of reflection from the standard powder which are known accurately are plotted against the measurement of the corresponding lines to give a calibration curve for the film. Then the corrected angles for the other lines of unknown substance are read from the curve.

2.2.2 KETTMANN 'S EXTRAPOLATION: (37)

To compute the error in lattice spacing d that results from an error in the measurement of θ, we may put Bragg's Law in the form.

$$d \sin \theta = n \lambda/2 \quad .. (2.02)$$

which by differentiation yields:

$$d \cos \theta \Delta \theta + \sin \theta \cdot \Delta d = 0 \quad .. (2.03)$$

from which $\frac{\Delta d}{d} = - \cot \theta \Delta \theta$ .. (2.04)

Thus the percentage error in spacing measurement caused by a given error in angle measurement approaches 0 as cot θ approaches 0, i.e. θ approaches 90°. A smooth curve drawn through the plotted points and extrapolated to $\theta = 90^\circ$ gave the corrected value of lattice constant.

Such plots have the disadvantage of non-linearity, making it difficult to extend the curve accurately to $\theta = 90^\circ$.

2.2.3 Bradley and Jay's extrapolation (38).

Bradley and Jay found that most errors depended on various functions of θ which can be combined into a single expression which is linear with $\cos^2\theta$ for $\theta > 60^\circ$. The advantage of the use of $\cos^2\theta$ is that it is easy to extrapolate along a straight line. It should be noted that since $\sin^2\theta = 1 - \cos^2\theta$, the errors also vary linearly with $\sin^2\theta$. If $\cos^2\theta$ is used, the plotted d values are extraploted to $\cos^2\theta = 0$ and if $\sin^2\theta$ is used, d values are extrapolated to $\sin^2\theta = 1$. If the radius error is eliminated by the asymmetric film-mounting arrangement and the beam divergence error is ignored, then only two principal errors remain namely the absorption error and the specimen eccentricity error. Both errors depend primarily on $\cos^2\theta$ for large values of θ, since $\sin\theta$ tends to unity as θ approaches 90°. The plot against $\cos^2\theta$ is linear for θ values greater than 60°.

2.2.4 Taylor & Sinclair, Nelson & Riley's extrapolation:(39,40)

Recently Taylor & Sinclair and independently Nelson & Riley have shown that even better extrapolation function to use is $1/2 \left(\frac{\cos^2\theta}{\sin\theta} + \frac{\cos^2\theta}{\theta} \right)$. This represents the mean between the different functions which are proportional to the error caused by absorption and the error caused by beam divergence. Since modern powder cameras eliminate the radius errors and minimise the eccentricity errors, absorption and the divergent radiations remain the chief sources of error. The plot against $\cos^2\theta$ is linear for θ values greater than 60°. The plots against either $\frac{\cos^2\theta}{\sin\theta}$ or $\frac{\cos^2\theta}{\theta}$ are linear for much smaller values of θ. It is for this reason that Nelson and Riley recommended the use of arithmetic mean of these two functions whenever absorption is the chief source of error.

2.2.5 Cohen's method of least squares: (41)

Cohen has shown that the extrapolation to eliminate systematic errors can be performed with increased accuracy by an analytical method.

Bragg's Law, by squaring and taking logarithms can be put in the form

$$\log \sin^2\theta = 2 \log (\lambda/2) - 2 \log d \qquad .. \; 2.05$$

and differentiating gives

$$\frac{\Delta \sin^2\theta}{\sin^2\theta} = - \frac{2 \Delta d}{d} + 2 \frac{\Delta \lambda}{\lambda} \qquad 2.06$$

since λ is known accurately, $\Delta \lambda = 0$

Therefore we get $\frac{\Delta \sin^2\theta}{\sin^2\theta} = - 2 \frac{\Delta d}{d}$

The errors in measured d values caused by absorption and specimen eccentricity can be set proportional to $\cos^2\theta$ that is,

$$\frac{\Delta d}{d} \propto \cos^2 \theta \qquad \ldots\ldots\ldots 2.07$$

$$\text{Thus } \Delta \frac{\sin^2\theta}{\sin^2\theta} \propto \cos^2 \theta \qquad \ldots\ldots\ldots 2.08$$

$$\text{or } \Delta \sin^2 \theta \propto \sin^2 \theta \cdot \cos^2 \theta \qquad \ldots\ldots\ldots 2.09$$

$$= D \sin^2 2\theta \qquad \ldots\ldots\ldots 2.10$$

where D is proportionately constant having the same value for all angles θ of the film but having different values for different films.

The relation determining the values of $\sin^2 \theta$ for isometric crystal can be written in the form :-

$$\sin^2 \theta = \lambda^2 / 4 d^2 = \frac{\lambda^2 (h^2 + k^2 + l^2)}{4 a^2} \qquad \ldots\ldots\ldots 2.11$$

$$= A_0 (h^2 + k^2 + l^2) \qquad \ldots\ldots\ldots 2.12$$

where $A_0 = \lambda^2/4a^2$

Adding the corrective term we get

$$\frac{\lambda^2}{4 d^2} = \sin^2 \theta - D \sin^2 2\theta \qquad \ldots\ldots\ldots 2.13$$

An equation of the type is written for each line of the pattern expressing d in terms of known indices and the unknown lattice constants.

For cubic crystals Cohen rewites equation in the form

$$A_o\ (h^2 + k^2 + l^2) + D \sin^2 2\theta = \sin^2\theta \quad \ldots\ldots.2.14$$

Let $\alpha = h^2 + k^2 + l^2$,

$\delta = 10 \sin^2 2\theta$, so equation 2.14 can be written as

$$\alpha A_o + \delta D = \sin^2\theta \quad \ldots\ldots.2.15$$

Equation relates known value of to the measured values of θ provided only systematic errors occur. As random observational errors also occur, the equation should be put in the form ;

$$\alpha_1 A_o\ \delta_1\ D - \mathrm{Sin}^2\theta_I = \epsilon_1 \quad \ldots\ldots.2.16$$

Squaring each term on both sides, equation 2.16 becomes

$$\sum_i (d_i A_o + \delta_i D - \mathrm{Sin}^2\theta_i) = \sum_i \epsilon_i \quad \ldots\ldots.2.17$$

Squaring each term on both sides, equation 2.17 becomes

$$_i(\alpha_i A_o + \delta_i D - \mathrm{Sin}^2\theta_i)^2 = {}_i\ \epsilon_i^2 \quad \ldots\ldots.2.18$$

In order that 2.18 be minimum, the first derivatives of $\sum_i \epsilon_i^2$ with respect to A_o and D must vanish

$$\therefore \frac{\partial}{\partial A_o} \sum_i \epsilon_i^2 = 0 = \sum d_i(d_i A_o + \delta_i D - \sin^2\theta_i) \quad \ldots.2.19$$

and $$\frac{\partial}{\partial D} \sum_i \epsilon_i^2 = 0 = \sum_i^i \delta_i\ (d_i A_o + \delta_i^i D - \mathrm{Sin}^2\theta_i) \ldots\ldots.2.20$$

Multiplying through by α_i in 2.19 and by δ_i in 2.20 and rearranging the terms, the so-called normal equations are obtained ;

$$(\sum_i \alpha_i^2) A_o + (\sum_i \alpha_i \delta_i) D = \sum_i \alpha_i \sin^2 \theta_i \qquad \ldots\ldots 2.21$$

$$(\sum_i \delta_i \alpha_i) A_o + (\sum_i \delta_i^2) D = \sum_i \delta_i \sin^2 \theta_i \qquad \ldots\ldots 2.22$$

These two simultaneous equations can be solved to give the values of A_o and D.

From the value of A_o, a is calculated by equation $a = \lambda^2/4 A_o$.

If a real choice between graphical method and Cohen's method exists, it must be resolved on the basis of whether the random errors are large or are negligible. In the graphical method, they manifest themselves in the scatter of observational points about the straight line that must be drawn to determine the extrapolated value. When such scatter is small, the line can be drawn and the graphical method is very accurate. On the other hand, when the scatter is great, least square method should give better results, since it is exclusively designed to eliminate random errors.

2.3 CONSIDERATION IN HIGH TEMPERATURE LATTICE SPACING MEASUREMENTS (42).

X-ray measurements can be used effectively to determine the thermal expansion of metals and alloys in the range - 196° to 1000 °C and with rather less convenience and accuracy at low and high temperatures at least upto 2000°C, and probably as high as 2800 °C . X-ray measurements provide the only method of obtaining the thermal expansion coefficients in cases where only a little material is available or in cases of crystals of low symmetry along the crystal axes, for the preparation and cutting of large crystals for macroscopic measurements would be tedious.

Expansion coefficients are normally obtained from the change in accurately determined lattice spacings with temperature. Sometimes it is not possible,for reasons such as high absorption, temperature factor etc., to obtain suitable high angle back reflection lines in powder photographs. Still accurate expansion coefficient may be obtained by determining the change of interplaner spacing of suitably reflecting planes. In this case position and diameter of the specimen must not be altered during a series of exposure at different temperatures, so that the absorption symmetry errors similarly affect the reflections from any particular plane at various temperatures.

The accurate expansion coefficient is then calculated from the relative change of interplaner spacing with temperature even though the true values of interplaner spacing are known.

There are four main points of importance in the determination of the expansion coefficients.

i) The first is the exact determination of the specimen temperature which can be made by taking photograph of pure silver, at different temperature the lattice spacings of which are accurately known at different temperatures; and then calculating the exact mean specimen temperature from the measured values of lattice spacing at different observed temperatures.

ii) There may be possibility of specimen contamination during exposure at high temperature, unless the contamination is large it will in general have a second order effect on the expansion coefficient. But a contamination leading to an expansion of the lattice of say 0.0010 KX at room temperature will also increase the lattice spacing measured at high temperature by approximately the same amount. It is, therefore, necessary to apply the appropriate, correction to the observed high temperature lattice spacing, if it is found that the lattice spacing at room temperature has changed following the high temperature exposure.

iii) The third point concerns the examination of alloys. If one component of the system is very susceptible to oxidation at high temperature or is easily lost by evaporation, or by reaction with the enclosing silica capillary, then the composition of the specimen may change before and during exposure at high temperature, thus affecting the apparent expansion coefficient.

iv) All specimens used in accurate determinations of expansion coefficient should previously be annealed at a temperature at least as high as that of the highest exposure for a time relatively long compared with the exposure time. This treatment will complete all recovery recrystallization and grain growth processes that can take place upto temperatures of highest exposure. The specimens should not then change their properties during exposures subsequently made at lower temperatures. The treatment is important because the processes of recovery recrystallization and grain growth may be accompanied by a change of lattice parameter.

2.4 GRUENEISEN'S RULE AND THERMAL EXPANSION OF METALS.

Grueneisen (43) has put forward an empirical rule connecting the volume expansion of solids due to rise in temperature with the atomic heat of the solid and its compressibility at absolute zero. This rule has been stated in various forms. Hume Rothery (1945) has recommended the form due to Simon and Vohson (1928) namely ;

$$\frac{V_T - V_o}{V_o} = \frac{E_t}{Q - KE_T} \qquad \ldots\ldots.2.23$$

where V_T and V_o are the volumes of the Solid at T °K and 0 °K respectively. E_t is the total heat content of the solid at T °K and Q and K are constants. Thus

$$\frac{V_o}{V_T - V_o} = \frac{Q}{E_t} - K \qquad \ldots\ldots2.24$$

$\frac{V_T - V_o}{V_o}$ may be substituted by $3\,\frac{a_T - a_o}{a_o}$, where a_T and a_o are lattice parameters of a cubic crystal at T °K and 0 °K respectively. Similarly we may write for E_T the Debye expression $9\,R_T D(\theta/T)$. Substituting these values in equation, it becomes evident that the curve of $\frac{a_o}{a_T - a_o}$ vs. $\frac{1}{TD\,(\theta/T)}$ will become a straight line. These curves have been drawn for ionic crystals such as sodium and potassium chlorides, potassium and Caesium bromides, lithium fluorides, potassium and caesium iodides and in each of these cases a straight line has been obtained.

Mitra (43) undertook an investigation to find out if the same relation was obeyed for pure metals. For this purpose he studied the cell edges of pure copper, aluminium, silver and platinum at various temperatures by the X-ray diffraction method. Measurements were carried out with the help of a 19 cm camera and filtered X-rays were used. The cell edges were determined by the extrapolation technique due to Sinclair and Taylor (39). Extrapolation to absolute zero was carried out by the technique described by Fiochmeister (43) They concluded that unlike the curves for ionic crystals, for metallic crystals, the curves are non linear.

2.5 LATTICE SPACING RELATIONSHIP IN ALUMINIUM ALLOYS .

2.5.1 Alloying elements contracting the lattice :

The relation between lattice spacings and composition was obtained by Axon and Hume Rothery (33) for Li, Si, Cu, Ag and Zn where these elements indicate lattice contractions. In the system aluminium-zinc, the lattice spacing/composition relation is not linear but is a curve such that the lattice contraction per atomic percent of solute increases with the concentration. In the system aluminium-copper the lattice spacing/composition relation is linear. In the system, aluminium-silicon, a straight line can be drawn through all points within limits of experimental accuracy. In this system the solid solution is restricted and hence Axon and Hume Rothery employed different rates of quenching in order to ensure that partial decomposition had not occurred. In the system aluminium-lithium, a contraction of lattice spacing takes place. The lattice spacing/composition line is slightly curved.

Similar measurements were made by Ellwood and Silcock (8) on alloys of aluminium and copper. Lattice spacings accurate to better than ± 0.0003 A° were given in A° and not KX units as stated in the paper. Gulijaev & Trusova (44) also measured the lattice spacings of several alloys upto 0.87 atomic percent copper. Dorn and others (45) measured the lattice spacings of very dilute solutions of aluminium-copper. The results were in agreement with that of Hume Rothery (33).

Ellwood (46) in his investigation on lattice spacing in aluminium zinc alloys upto 35 atomic percent Zn has found excellent agreement with the values of Axon and Hume Rothery in the range 0.6 percent Zn. The lattice spacings curve as reported by him at 250° C is unusual in that it can be divided into 4 parts. In the ranges 0 - 4.0 and 18.5 - 25.0 percent zinc, a non-linear relationship between lattice spacings and composition is found whereas in the range 4.0 - 18.5 and 25.0 - 35.0 percent zinc, a linear relationship exists. The lattice spacings at 360° C show similar anomalies except that the curved portions extend over a greater range of composition, the approximate limits being 0 14.0 and 17.0 - 30.0 percent zinc.

Lattice parameters of the alloys of aluminium - zinc and aluminium-silicon were also measured by Gulijaev and Trusova (44). Lattice spacings of aluminium - zinc alloys were found higher than that of the values determined by Axon and Hume Rothery (33) and Ellwood (46), the later both being in agreement are preferred. The results of lattice spacing measurements of aluminium-silicon alloys as determined by

Gulijaev and Trusova (44) were also found higher than the values determined by Axon and Hume Rothery (33).

In the case of aluminium - lithium solid solution there are conflicting reports about the effects of lithium on the lattice spacing of aluminium in solid solution. Earlier measurements by Vosskuhler (47) indicated a slight contraction of aluminium lattice with lithium in solid solution and were found in agreement with the results determined by Hume Rothery (33). But Zuitl and Woltersdrof (48) had previously found the expansion of lattice in alloys melted in LiF lined zirconia crucibles. Komovsky and Maximova (49) found no change in aluminium lattice spacing in the aluminium-lithium solid solutions.

The aluminium - chromium solid solution lattice spacings have been measured by Hofmann & Herzer (50) and Hofman and Wiehr (51) using back reflection methods. Similar measurements have been discussed and shown in the diagrams by Falkanhagen and Hofmann (52) and also by Knappwost and Nowotny (53). All report contractions in lattice spacings in aluminium-chromium solid solutions. Knappwost and Nowotny found a considerable larger lattice spacing in the one homogeneous alloy examined.

In case of aluminium-cadmium there is very little terminal solid solubility. Dorn (45) measured the lattice spacings of very dilute solid solutions upto 0.065 atomic percentage of cadmium. A contraction of 0.0002 KX was reported with the addition of 0.065 atomic percentage cadmium.

Slight contraction in the lattice spacing has also been reported in the case of aluminium-beryllium dilute solid solution by Makaror and Tarschisch (54) and by Mikheeva (55). They measured the lattice spacing by Sachs method but the temperature to which the measurements refer is not stated. Aluminium-manganese solid solution lattice spacings have been measured by Hofmann (56) and by Falkenhagen and Hofmann (52) by back reflection photographs using gold for calibration. Although the solubility of the manganese in aluminium at the eutectic temperature is only 1.4 wt. percentage, it is possible to obtain upto 9.5 wt. percent Mn in metastable solid solution by quenching liquid alloys with appropriately cooled copper moulds. The contraction in the lattice has been reported.

Falkenhagen and Hofmann (52) also carried out the lattice spacing-measurements in the case of aluminium-titanium solid solution and report contraction in lattice spacing. They used rapid quenching from the liquid state to obtain supersaturated solid solution.

Brauer (57), however, reported an increase in the lattice spacing from a = 4.040 KX for pure aluminium to a = 4.044 KX, for aluminium saturated with titanium in·

powders, annealed between 320° C and 460° C.

2.5.2 Alloying elements expanding the lattice :

Hume Rothery determined the lattice spacings of aluminium rich solid solutions of the system aluminium-magnesium, aluminium-germanium and aluminium-silver (33). The lattice spacing of aluminium is expanded considerably by the solution of magnesium and germanium and is unaltered by silver. In the system aluminium-magnesium the limits of aluminium rich solid solution are approximately 16 atomic percent magnesium at 450° C. and 3 atomic percent at 200° C, and hence the decomposition during quenching was prevented by cooling in a blast of air. The lattice spacing composition curve beyond 2 atomic percent magnesium is practically a straight line, but shows a definite curvature at the lower percentage. It is seen that addition of less than 1 atomic percent of magnesium produces a slightly smaller expansion in the lattice as compared with the effect of larger percentages.

Poole and Axon (58) determined the lattice spacing of aluminium rich solid solutions of the system, aluminium-magnesium. The curves show that small amount of impurities in the magnesium have a pronounced effect on the lattice spacing relationship. They have shown that the change of slope in the curves which was found generally in the first 0.5 atomic percentage solute addition is the result of impurities into the alloys with magnesium.

Ellwood (46) has also observed a slight curvature in the lattice spacing composition relation for aluminium-magnesium solid solutions alloys at low magnesium contents.

Aluminium solid solution lattice spacings have been also measured by Dorn (45), Gulijaev and Trusova (44), Hume Rothery and Boultbee (9), Kustner (59), Kuznetsor and Guseva (6), Siebel and Vosskuhler (61) and Trillat and Paic (62).

There is good agreement between the measurements of Ellwood (46) and of Poole and Axon (58). Axon and Hume Rothery's (33) measurements tend to be rather lower and Kustner's slightly higher particularly at larger solute concentrations.

In case of aluminium-germanium Stohr and Wklemn (63) have determined the solid solubility of germanium in aluminium as 2.8 at. percent, Lattice spacing of the alloys upto 1.98 atomic percent germanium was measured. In the system the lattice spacing/composition relation is very slightly curved. Dorn (45) also measured the lattice spacing of very dilute solutions of aluminium germanium upto 0.145 atomic percentage and have shown an expansion of only 0.0001 A^{o}.

There are conflicting reports about the effects of silver, on the lattice parameter of aluminium in solid solution. Hume Rothery's work (33) upto 6 atomic percent Ag has shown that lattice parameter remains unchanged. Dorn (45) carried out the lattice spacing measurement of very dilute solid solution of aluminium-silver richest alloy containing maximum 0.194 atomic percentage silver, and has indicated only a slight increase of 0.0001 A° with the addition upto 0.194 atomic percentage silver.

Gulijaev and Trusova (44) measured the solid solution lattice spacings from powder photographs. The lattice spacings increased relatively rapidly with concentration and did not agtee with the results of Axon and Hume Rothery (33). Aluminium-silver solid solution lattice spacings were also measured by Ellwood (46). The lattice spacing/composition relation is given in Fig. 6. It is seen that the lattice spacing of aluminium is practically unchanged by addition upto 6 atomic percentage silver, increases slightly between 6 and 14 atomic percentage silver and remains fairly constant in the range 14 - 27 percent silver. His results upto 6 atomic percentage silver are in agreement with Hume Rothery's work, but does not agree with the results of Gulijaev and Trusova. He has shown that the increase in the lattice spacing between 6 and 14 atomic percentage silver is associated with the filling of a Brillioun Zone. The zone is filled as the electron atomic ratio decreases. The zone appears to be filled when 14 percent Ag is present corresponding with an electron atomic ratio of 2.72.

Hull and Axon have examined (64) the aluminium solid solution lattice spacings in alloys containing upto 20 atomic percentage silver. They emphasize the importance of quenching sufficiently rapidly from hot to a low temperature to prevent separation of a second phase in alloys containing more than 10 atomic percentage silver. Their measurements show no lattice change or a slight decrease in the lattice spacing upto 14 atomic percentage silver after which the lattice parameter rises. Between 10 and 20 atomic percentage silver the measurements of Hull and Axon lie some 0.0001 KX lower than that of Ellwood.

In the system aluminium gallium practically no work has been carried out up till now although aluminium dissolves 6 atomic percentage gallium at room temperature. Owen liu and Morris (65) gave the lattice spacing of an alloy containing 0.53 atomic percentage gallium as a - = 4.04115 KX at 18 ^{o}C. After that no data is available in this system.

In case of aluminium-calcium system terminal solid solutions are of limited extent. Nowotny (66) examined alloys prepared from 99.9985 percent aluminium and 98.0 percent calcium and determined the lattice spacings by the

Straumanis method. Lattice spacing was found to increase considerably with the addition of calcium in solid solution.

In aluminium-thorium system the aluminium solid solution of thorium, is very restricted only to 0.24 atomic percent thorium at 610 °C. Grube and Botzenhardt (67) examined alloys prepared from 99.73 percent aluminium and 98.7 percent thorium, by melting them together at 1000 °C. Filings were prepared and annealed at 450 °C to 600 °C. for 200 hours and quenched. Lattice has been found to expand considerably with the addition of thorium in gold solution.

2.6 APPARENT ATOMIC DIAMETER IN ALUMINIUM ALLOYS .

Aluminium is a trivalent solvent with face centered cubic structure. The first Brillioin Zone can hold only two electrons per atom and must therefore, be overlapped. From the results of Axon and Hume Rothery (33) on the lattice spacings of solid solutions of Li, Mg, Si, Cu, Zn, Ge and Ag in aluminium it is seen that in the alloys with silicon, copper and silver the lattice spacing composition curves are straight lines so that the A.A.D. of these solutes are easily obtained by extrapolation to 100 atomic percent solute. In the alloys with Li and Zn a similar procedure can be applied to the initial portions of the curves but magnesium produces a considerable curvature and hence the A.A.D. in the alloys with magnesium as the solute, is uncertain (68).

When the A.A.D. of various solute elements in Aluminium are compared with similar values in other solvents (33) it is seen that quadrivalent elements (Si, Ge,) give larger A.A.D. values in aluminium than in univalent solvents. On the other hand univalent elements (Li, Ag and Cu) produce a reverse trend. Divalent elements (Mg, Zn) show relatively little difference whether dissolved in aluminium or in univalent solvents, but the A.A.D. of magnesium in aluminium and zinc in aluminium, are smaller than distances of approach in the pure elements in the direction of main Brillioüin Zone overlaps.

Hume Rothery has shown that the differences between the A.A.D. values of any given element when dissolved in various solvents are controlled by the following four factors :-

i) The relative volume per valency electron in crystals of solute and solvent.
ii) The relative radii of the ions of solute and solvent.
iii) Brillioüin Zone effects and
iv) The differences between solvent and solute in the electro-chemical series.

The influence of the first factor is such that if a solute with a high mean volume per valency electron is dissolved in a solvent of a relatively low mean volume per electron, the lattice spacing is contracted.

When Lithium is dissolved in aluminium, relative volume per valency electron Ve, for Li (21.7 KX^3) is greater than that for aluminium (5.51 KX^3) such that A.A.D. of Lithium is observed to be small and the mean lattice spacing is contracted. In the case of germanium (Ve = 3.6) in aluminium, the reverse is true and A.A.D. is large.

2.7 HIGH TEMPERATURE X-RAY DIFFRACTION STUDY OF ALUMINIUM AND ITS ALLOYS.

Lattice spacings of aluminium at elevated temperatures were measured by several workers viz. Straumanis and Ievins (69) Ener (42) in 1958, Wilson (71,140) and Ellwood and Silcock. In the high temperature range, the measurements of Ellwood and Silcock, Wilson and Straumanis and Ievins are in good agreement.

The results of Ener (42), are in good agreement with the measurement of Wilson below 200° C., but at higher temperatures greater expansion was found.

The expansion of annealed aluminium below room temperature has been examined by Hume Rothery and Straubridge (72) by Hume Rothery and Boultbee (9) by Pearson (73) and by Figgings (74).

Ellwood and Silcock (8) examined the lattice spacings of aluminium-copper alloys as a function of temperature. They showed that the effect of equivalent atomic percentage of copper on the aluminium lattice, increases with increasing temperature, owing to their relatively greater differences in the atomic radii caused by the difference in the thermal expansion. The coefficient of thermal expansion at 548° C. was measured for each alloy. The decrease in coefficient with increase in copper content was reported. Authors appeared to have confused KX units, crystal Aungstron units and absolute Aungstron units, and in correspondence on paper they reported the correct values of lattice spacing in KX units.

As regards temperature measurement there appeared to be some discrepancy in calibration and accuracy in measurement was reported low.

Hume Rothery and Boultbee (9) made the measurements on alloys of aluminium-coppper containing 1.62 atomic percent copper, on the aluminium magnesium solid solutions and on the alloys of aluminium-silicon between - 50 to 200° C. The work was not carried out at higher temperatures. The results of the measurements are given in Table No. 2.01, 2.02, and 2.03.

—— Contd.

TABLE No. 2.01

ALUMINIUM-COPPER ALLOY (1.62 ATOMIC % COPPER)

Temperature °C	a in KX	$\alpha \times 10^6$
- 50	4.0267	25 - 50 = 22.5
+ 25	4.0335	
+ 200	4.0511	200-25 = 24.8

TABLE No. 202

ALUMINIUM-MAGNESIUM ALLOY

Atomic % magnesium	Temperatures		
	-50°C	25° C	200° C
0	4.0347	4.0414	-
2.36	4.0427	4.0496	-
6.36	4.0614	4.0685	4.0858
10.56	4.0782	4.0861	4.1037

TABLE 2.03

ALUMINIUM-SILICON ALLOY CONTAINING 0.92 ATOMIC % SILICON

Temperature	a in KX	α
-50	4.0329	- 50 to 25 °C
25	4.0396	22.4×10^{-6}

There results have indicated that in case of aluminium-copper alloys thermal expansion increases with increase of temperature upto 200° C. The work was carried for : one alloy and the effect of copper on thermal expansion at different temperatures could not be studied. In case of aluminium-magnesium alloy their results indicate that both the composition and the temperature effect the values of thermal expansion. Both, increase in temperature, as well as magnesium concentration, increase the values of thermal expansion of aluminium.

The lattice spacings of the aluminium-zinc solid solution at high temperatures have been measured most recently and extensively by Ellwood (46). They lead to boundaries which are generally in good agreement with the equilibrium diagram given by Rayanor. Petrov & Vadera (~~1947~~) have also made high temperature photographs of a series of alloys containing from 20 to 96 percent Zn at 370° C. and 300° C. Among earlier investigations (42) of aluminium-zinc solid solutions by high temperature X-ray photographs, the following may be mentioned; Ellwood, Owen and Pick up (~~1935~~), Kissolaper and Trapesnikor (~~1935~~), Owen and Iball (~~1934~~); Schmid and Wasserman (~~1932~~, ~~1934~~), Fink & Van-Horn(~~1932~~).

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CHAPTER III

EFFECTS OF RARE EARTH ADDITIONS ON THE PROPERTIES OF METALS AND ALLOYS.

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EFFECTS OF RARE EARTH ADDITIONS ON PROPERTIES OF METALS AND ALLOYS

3.1 GENERAL INTRODUCTION:

Rare earth elements which have lately assumed importance have numerous unique properties and now are available on a commercial scale. They are the elements of stomic numbers 57 - 71. Some of their properties to research metallurgists are as follows:

1. Large atomic volume.
2. An unusually strong chemical affinity for non-metallic elements normally associated with steel (S,O,N & C) in many cases forming refractory compounds which considerably modify the crystal structure and thus their properties.
3. Their great capacity to occlude hydrogen exothermally. . They can dissolve considerable quantities of hydrogen at low temperatures as compared to many other metals including iron.

The rare earths are widely scattered in nature at low concentrations and they occur in a number of concentrated minerals such as monazite, Xenotime, bastnasite, gadolinite, samarskite, fergusonite and euxenite to mention a few. Rare earths are about 10 times more abundant in monazite than thorium . Since monazite is the principal source of thorium, it is clear that as thorium becomes important as a fuel for nuclear power, large quantities of rare earths will be available as a by-product. In addition, rare earths are frequently found associated with tantalum and columbium ores, these high temperature materials increase in importance, again considerable quantities of rare earths will be available as by-products.

It has now been clearly shown that metals of the rare earth group are valuable alloying elements both in micro-and macro-additions in ferrous and non-ferrous alloys. This chapter presents briefly the uses of rare earths as alloying elements in metals and alloys. Current developments in rare earths applications have also been mentioned briefly.

The applications for rare earths as mixed metals or their oxides have been extensively explored in the past decade, and found capable of providing a variety of alloys with markedly improved properties. This phase of rare earths development is fairly well established.

As regards individual rare earth elements and their effect on ferrous and non-ferrous alloys, the information is still of qualitative nature. The number of organisations that are producing rare earth metals have increased rapidly in the past few years. The work reviewed here has served to define a few applications which have a real potential. However, the extent to which producers will find a market for the rare earth metals, depends on the ability of industry to translate laboratory results to useful high volume product application.

In the words of Felix Trombe " The rare earths are no longer of only academic interest. Having penetrated into the industrial field, they occupy a place which only remains to be developed."

3.2 PRODUCTION AND PROPERTIES OF RARE EARTH METALS

3.2.1 <u>History of rare earths</u>:

Rare earth history began in 1794 when Gadolin (31) separated a new earths called yesterly [illegible]. The name Yttria was given to the oxide and gadolinite to the mineral. Although Yttria was at that time considered to be the oxides of a single metal it was in reality a mixture of at least 15 or 16 different metals.

J.J.Berzelius and W.Hisinger in 1804 (71) discovered the mineral known as Cerite. M.H.Klaproth made a similar discovery independently. After this work, it was soon found that the original oxides were complex mixtures. The rare earths were gradually isolated from that time on.

During the next hundred years, 13 of the 14 rare earths were identified and isolated. Centres of the rare earths research developed at numerous locations and great chemists such as Mosander, Marignac, Crookes, Nilson, Clive and Demarcay, to mention only a few, were associated with these researches. In the later part of the nineteenth century and the first part of the twentieth, five centres should be particularly cited for their contributions e.g. the laboratories of Urbain in France, Auervon Welsbach in Austria, James in New Hampshire, Hopkins in Illinois and McCoy in Chicago. It is interesting to note that at the turn of the century, Auervon Welsbach had achieved fame and amassed quite a fortune as a result of his researches in rare earths. In this century a very respectable industry was developed at Lindsay with which McCoy was associated and at Maywood dealing with cerium, lanthanum and the mixed rare earths. During the period 1796 to 1940 all the naturally occurring rare earths were isolated and a great deal about their properties was described, thus providing the basic information from which the modern developments steamed.

The original names given to rare earths are as interesting as their properties. Cerium was named by Berzelius in 1803 from ostroid ceres; dysprosium, Greek "hard to get at", Gadolinium after Johann Gadolin; holnium from Stockholm; lanthanum, Greek, " to be concealed;" lutetum from lutetia, a town in Gaul, now Paris; neodymium, Greek new twin; "samarium" after Col. Samarsky, a Russian mine official; thulium after Thule, a distinct mysterious region. Erbium, terbium, ytterbium, yttriumm are derived from Yeterly, Sweden and europium, scandium are easily recognisable geographical names.

Information received from various rare earths producers in the United States indicates that rare earth metals and alloys prapared commercially can be divided into the following groups: misch metal, cerium and lanthanum metals and didymium metal.

The traditional source of rare earth elements is monazite sand. In the 1890 thorium was extracted from monazite for use as thorium nitrate in Welsbasch gas mantles. The residue containing cerium, lanthanum, neodymium, preseodymium and other rare earth compounds in lesser amounts was discarded. Welsbasch found that a pyrophoric metal (misch metal) could be made from this residue. In 1908 in Austria he turned these pyrophoric properties to use in the manufacture of lighter flints and founded the Treibash Chemical Works, thus becoming one of the first commercial producers of misch metal.

With the decline in the use of gas lamps, the demand for thorium decreased and monazite sand was used principally as a source for misch metal. In recent years nuclear requirements have shifted the balance back to thorium as the main product from this raw material. In 1949 the discovery in California of the Mountain Panbastnasite deposit with an estimated reserve of 5 million tons of rare earth oxides (48 % CeO_2 , 34 % La_2O_3) providing an additional large source for misch metal and related alloys. At present misch metal is made from both monazite and bestnasite.

3.2.2. Preparation of rare earth metals:

Rare earths are divided into cerium, terbium & yttrium earths, on the basis of complex salt solubility in a saturated K_2SO_4 solution. The cerium family is insoluble, the terbium family is moderately so and the yttrium family is readily soluble.

The ores are opened by H_2SO_4 and the lanthanum sulphate is dissolved by water and precipitated as oxalates or sodium double sulphates. Cerium is separated by oxidation to the tetravalent state. Members of the individual group are separated on a large scale by fractional crystallization, solvent extraction and ion exchange processes. Metal is produced either by fused salt electrolysis or by thermal reduction with calcium, barium, magnesium, sodium amalgam, aluminium or zirconium (31).

Large scale commercial production of misch metal is presently limited to the electrolysis of fused anhydrous chlorides (75). The requisite dehydration of rare earth chloride before electro-winning is normally accomplished by one of two methods. In the first method the chloride is melted in cast iron, steel or cerium vessels from which air is largely excluded. Heating is continued until a porous, solid, nearly anhydrous product is obtained. The product contains upto 10 % water insoluble basic chlorides. The second method for drying these chlorides uses a vacuum chamber (76). The anhydrous chloride was reported to contain 1.5 % oxychlorides.

Most cells for electrowinning misch metal have an iron, carbon, graphite or refractory lined steel vessel to contain the molten bath. The bath container or an iron or carbon block at the bottom of the container serves as the cathode. One or more carbon or graphite rods extend vertically into the cell through its top and serve as anodes.

Just before electro-winning a mixture of anhydrous rare earth chloride and NaCl, KCl, or $CaCl_2$ is charged into the pot. In one procedure the charge is melted by external burners. In another procedure anodes are lowered until they contact broken pieces of cerium metal placed at the bottom of the pot. The direct current is turned on and when sufficient electrolyte has been melted, the anodes are raised and electrolysis commences.

The misch metal product can be collected in various ways. (11) One procedure is to remove the liquid metal by scooping it out with a ladle. Other manufacturers prefer to pour the entire contents of the cell, bath included into moulds, in which it is kept molten. Once the misch metal has settled to the bottom of the mould, the liquid bath is returned to the electro-winning cell and electrolysis is begun again. Electrolysis is carried out at 1560°F. Cerium, lanthanum and didymium metals are electro-won commercially from their fused anhydrous chlorides by essentially the same procedure as that previously described for misch metal.

3.2.3 Properties of rare-earth metals;

The metallurgical importance of rare earths lies in their unusually strong chemical affinity for non-metallic elements, normally associated with ferrous and non-ferrous metals, sulphur, oxygen, hydrogen, nitrogen and carbon. Though the melting point of rare earths is not very high, they form stable refractory compounds, which can have profound influence on the metal structure even when present in small quantities. Tables 3.01, 3.02 & 3.03 (76) give some of the physical, chemical and thermodynamic constants of the rare earth metals.

Table 3.01 Physical and Chemical Constants of the Rare Earth Metals.

Element	Thermal-neutron-absorption cross section, barns/atom.	Density* gm/cm^3	Melting point $^{\circ}$C	Boiling point $^{\circ}$C	Heat of fusion Kcal/mole.	Heat of vaporization, Kcal/mole.
Scandium	13 ± 2	2.5	1400	-	4	93
Yttrium	1.38 ± 0.14	5.51	1475	-	4	103
Lanthanum	8.9± 0.3	α 6.194 β 6.18	866±10	4340 ± 260	2.3	79.5±1.7
Cerium	0.70±0.08	α 6.78 β 6.81	780±10	2420 ± 70	5.5	107.7± 2
Praseodymium	11.2 ± 0.6	α 6.776 β 6.805	950± 10	3020 ± 90	3.1	79.5 ± 1.1
Neodymium	44 ± 2.0	6.998	820± 10	-	-	86
Promethium	-	-	-	-	-	-
Samarium	6500± 1000	6.93	>1300	-	-	-
Europium	4500 ± 500	5.224	-	-	-	-
Gadolinium	44000±2000	7.948	~1200	-	-	-
Terbium	44 ± 4	8.332	>1400	-	-	-
Dysrosium	1100± 150	8.562	1400	-	-	-
Holmium	64± 3	8.764	~1200	-	-	-
Erbium	166 ± 16	9.164	>1140	-	-	-
Thulium	118 ± 6	9.346	>1400	-	-	-
Ytterbium	36 ± 4	7.010	~1800	-	-	-
Lutetium	108 ± 5	9.740	-	-	-	-

* Gm/cm^3 x 62.43 = lb/cu.ft.

Table 3.02 Thermodynamic Constants of the Rare Earths

Specific heat, cal /(mole) (^{o}K)

Lanthanum 298° to 800°K	$C_p = 6,17 - 1.60 \times 10^{-3}T$
Cerium 298° to 800°K	$C_p = 4,40 - 6.00 \times 10^{-3}T$
Praseodymium 293° to 373°K	$C_p = 6.85$
Neodymium* 0° to 400°C	$C_{p_o}^{t} = 6.479 + 7.065\times10^{-3}t$ $+4.463 \times 10^{-6}t^2$

Enthalpy, or heat contant, cal/mole

Lanthanum ** 298° to 800°K	$H_T - H_{298.16} = 6.17T + 0.80\times10^{-3}T^2 - 1911$
Cerium@ 298° to 373°K	$H_T - H_{298.16} = 4.40T + 3.00\times10^{-3}T^2 - 1579$
Praseodymium 293° to 373°K	$H_T - H_{298.16} = 6.85T - 2042$
Neodymium* 0° to 400°K	$H_t - H_o = 6.479t\times3.533\times10^{-3}t^2$ $+ 1.488 \times 10^{-6}t^2$

Entropy at 25°C., cal/(mole) (°C)

	Cerium	Lanthanum	Letutium
Solid	13.8 ± 0.8	13.7 ± 0.8	14.5
Gas	48.09± 0.05	43.57± 0.01	

Entropy ($S_T - S_{25°C}$)
cal/(more)(°C)

Entropy $(S_T - S_{25^\circ C})$
CAL/ (mole)(°C)

Temp. °C	Cerium	Lanthanum	Neodymium.
127	1.93	1.96	2.18
227	3.51	3.52	3.97
327	4.90	4.80	5.54
427	6.18	5.92	6.94
527	7.36	6.88	8.22
627	-	-	9.41

Vapor pressure, mm Hg

Lanthanum

$$\log P = \frac{-17{,}185 \pm 372}{T} + 6.605 \pm 0.201$$

Cerium

$$\log P = \frac{-23{,}400 \pm 440}{T} + 11.58 \pm 0.27$$

Preseodymium

$$\log P = \frac{-17{,}188 \pm 243}{T} + 8.098 \pm 0.156$$

Samarium — Reported to have higher vapor pressure than gadolinium in range 1600° to 1700°C

Electrical resistivity

(room temperature) µohm-cm

Lanthanum	65-75
Cerium	75-90
Praseodymium	75
Neodymium	70

* t = °C. ; ** Formula accurate within 2 percent.

@ Formula accurate within 3 percent.

Table 3.03 Crystal Form and Lattice Structure of the Rare Earths.

Element		Crystal form*	Lattice Constants, A	
			a_o	c_o
Scandium	(α)	F.c.c.	4.541	-
Scandium	(β)	H.c.p.	3.300	5.256
Yttrium		H.c.p.	3.637	5.762
Lanthanum	(α)	F.c.p.	5.305	-
Lanthanum	(β)	H.c.p.	3.762	6.075
Cerium	(α)	F.c.c.	5.150	-
Cerium	(β)	H.c.p.	3.657	5.972
Praseodymium	(α)	H.c.p.	3.669	5.920
Praseodymium	(β)	B.c.c.	5.161	-
Neodymium		H.c.p.	3.657	5.902
Promethium		F.c.c.	-	-
Samarium		H.c.p.	-	-
Europium		B.c.c	4.582	-
Gadolinium		H.c.p.	3.629	5.760
Terbium		H.c.p.	3.592	5.675
Dysprosium		H.c.p.	3.585	5.659
Holmium		H.c.p.	3.564	5.631
Erbium		H.c.p**	3.539	5.600
Thulium		H.c.p.	3.530	5.575
Ytterbium		F.c.c.	5.479	-
Lutetium		H.c.p.	3.516	5.570

* F.c.c., face centered cubic; h.c.p., hexagonal close-packed; b.c.c., body. centered cubic.

** Other structures have been reported.

Allotropic forms have been reported for scandium, cerium, lanthanum, praseodymium and possibly erbium. The transitions are sluggish and depend on the previous history of the metal (76). For example, a cerium rod cast and turned showed a mixed face-centred cubic and hexagonal close-packed structure at room temperature. A similar extruded and turned lanthanum rod was also hexagonal close packed with some face centred cubic. A cast and turned praseodymium rod was hexagonal close-packed with a small amount of body centred cubic, and a similar neodymium bar was entirely hexagonal close-packed. The sluggishness of the transition makes it difficult to determine the exact transition temperature. Lanthanum metal has three allotropic forms α La, β La and γ La. The α La form exists from ambient temperature to about 260°C. and is hexagonal close packed. β La form exists from approximately 260° to 864°C. and is face centred cubic. γ La form exists above 864°C. and is body centred cubic. The α - Ce allotrope crystallizes in the face-centred cubic form below - 150°C. and can transform only from γ-ce, the room temperature form. Under high pressure γ -Ce form will transform to α Ce at high temperature e.g. at 7000 Kg./cm^2 and 25°C. transformation will occur. The transformation from α Ce to β Ce occurs below about - 10°C . The β Ce allotrope is hexagonal close packed isomorphous with α La (11).

Constitutional diagrams of La, Ce, Pr and Nd with ~~diff~~ different metals have been studied. Details of some systems of interest in the present investigation are given below (11)

The alloy system La-Al was studied by thermal analysis and metallographic techniques (77). Other studies have been also conducted by Weibke & Schmidt (78) & by Nowotny (79). Five intermediate phases are reported; La_3Al, La_3Al_2, $LaAl_2$ and $LaAl_4$. The $LaAl_4$ undergoes a transformation at 816°C. No solid solubility of Lanthanum in aluminium was detected (78,79). The existence of La_3Al_2 is doubtful.

The alloy system La-Ce has been studied by thermal and metallographic techniques by Vogel & Klose (80) & by Savitskii (81) complete solid solubility of the α, β and γ modifications of cerium and lanthanum exist throughout the diagram.

The phase diagram for the alloy system La-Cu has been determined by thermal and metallographic techniques (82) Vogel & Henmann (83), Nowotny (79). The presence of four and possibly five intermediate phases was established; LaCu, $LaCu_2$, $LaCu_4$, $LaCu_6$ and possibly $LaCu_5$.

Canneri (82) , Vogel & Heumann (84). Weibke & Schmidt (~~1940~~ 78) studied the alloy system of LaMg. Four intermediate phases exist LaMg, $LaMg_2$, $LaMg_3$ and LaMg9. The high temperature modification of lanthanum is stabilized to a lower temperature of alloying with magnesium. The solid solubility of lanthanum in magnesium is about 0.4 weight percentage.

In case of La-Si, the intermediate phase $LaSi_2$ which is body centred tetragonal has been reported (83,82,86).

In case of La-Zr, partial liquid immiscibility exists as determined by metallographic study (87).

Vogel (88) Muthmann & Beck (89) and Vogel & Heumann (83) investigated the Ce-Al alloy system by thermal analysis and metallographic techniques. Four intermediate phases are reported: Ce_3Al_2, CeAl, $CeAl_2$ and $CeAl_4$. $CeAl_4$ undergoes a transformation at 1005°C. The phase Ce_3Al_2 could not be affirmed. Instead the intermediate phase Ce_3Al is reported which has a phase transformation in the solid state at 250°C. Also a CeAl - Ce_3Al eutectic occurs at 29.5 atomic percentage. aluminium and at 645°C. and the Ce-Ce_3Al eutectic occurs at 11 atomic percentage at 580°C. Ce_3Al melts at 655°C. The solid solubility of cerium in aluminium is reported to be less than 0.05 wt. % Ce (90).

Phase diagram of the Ce-Cu system, was determined. Four intermediate phases are reported: Ce-Cu, $CeCu_2$, $CeCu_4$ and $CeCu_6$ (91,92)

The constitution of the alloy system Ce-Mg has been studied by various investigators (93 , 94 , 95) The solid solubility of Ce in magnesium is 0.85 wt. % at 575°C. (11) There are four intermediate phases in the system: CeMg, $CeMg_2$, $CeMg_3$, $CeMg_9$ may undergo a polymorphic change.

In case of Ce-Si system only partial diagram has been established by Vogal (96). The intermediate phases Ce_3Si, Ce_2Si, CeSi and $CeSi_2$ are reported.

3.3 RARE EARTHS AS ALLOYING ELEMENTS

3.3.1 Additions to aluminium and its alloys:

There are conflicting reports about the effects of addition of cerium or cerium misch metal to aluminium and its alloys, (21). Dannison & Tull (22) have observed grain refinement of the metal, while Bowen and Bernstein (23) found little or no refining or coarsening effect especially in aluminium.copper alloys.

Similarly Hodge, Eastwood & Lorig (24) did not find any significant effect of cerium addition on the elevated temperature properties of cast aluminium alloys. Bowen and Bernstein (23) studied the effect of various grain refining elements on duralumin type alloys and have reported no effect on grain size.

Russell (26) and Lorig (25) have carried out elevated temperature testing of complex aluminium alloys with rare earth addition. An aluminium alloy containing 10 to 12 % misch metal (50 % Ce), 1-3 % Si and upto 5 % Cu was patented. The above alloy was claimed to have high temperature strength in the cast condition. Lorig, Boer & Ackerlind (25) studied the effect of misch metal addition to aluminium alloy for elevated temperature service. The new alloy contains 11 % misch metal, 1.5 % Cu, 1.25 % Ni, 1.0 % Mn, 0.3 % Ce and 0.02 % Ti with rest of Al. The elevated temperature properties of this alloy are found out to be superior in the range of (700°F. - 800°F). The room temperature properties were similar to other commercial cast alloys except the strength which was low.

Hodge & Smith (27) studied the effect of rare earths additions to aluminium bath for hot dipping. Hand dipping into an unprotected aluminium bath presents two main difficulties:

i) One of these is that the last edge of the piece to be immersed which is normally the first to emerge from the bath is usually incompletely coated.

ii) Coated section although otherwise satisfactory shows pumples or blocks of oxide or dross distributed irregularly over the surface and thus causing roughness. Coatings tend to be incomplete and rough.

When 1.7 % misch metal is added, complete and attractive coatings were obtained. The authors gave the following two reasons:-

i) . The rare earths so added reduced the surface tension.

ii) The rare earths dissolve the oxide film The iron - aluminium compound which causes lumps to appear on the coated surface without the misch metal addition is dispersed and distributed uniformly throughout the coatings. About 2 % of iron is maximum that can be held in solution in aluminium bath at 135^{o}F. in presence of misch metal.

Bashforth and other coworkers (28) carried out the investigations with a view to study the effect of rare earths additions to molten aluminium under vacuum. Metallographic examination, mechanical and other properties of the treated metal were determined. Both percentage elongation & reduction in area were reported to be improved on addition of small quantities of Lancer Amp-2. On further increase in addition there seems to be a reversal in both the elongation percentage and reduction in area. Decrease in hardness with addition of rare earths mixture was noted. No significant effect was observed with fluoride addition. Metallographic examination with respect to grain structure reveals a general grain refinement with rare earths additions.

Recently Savitskii, Stepnor and Terekhova (97) established that the solubility of Nd in Al is $<$0.2 %. The strengthening effect of Nd addition on aluminium was observed. The hardness increased from 25 to 155 while the plasticity was reported to decrease by 5 - 10 %.

Trehan, Gupte & Nijhawan (10) studied the effect of misch metal on aluminium-magnesium alloy (7 to 10 %) and observed the following effects:

1. Addition of misch metal to Al-Mg (7 to 10 %) alloys, changes their as cast microstructure Al - Mg. phase is progressively eliminated and new phases, essentially containing Ce are formed.

2. The addition of misch metal to Al-Mg(7 to 10 %) alloy imparts to them improved hot workability and increases their hardness.

3. An addition of 2 - 3 % misch metal to Al-Mg (7 to 10 %) alloys has been found to be optimum for their hot workability particularly in the case of the Al-Mg (10 %) alloy.

4. The tensile strength of the cast Al-Mg (7 to 10 %) alloys is reduced by the addition of misch metal upto about 2 to 3 % above which it again shows a rise. The tensile strength of wrought specimens on the other hand continuously rises with an increase in the quantity of misch metal added and reaches values comparable to those for steel or duralumin (30 tons per sq. in.).

3.3.2 Additions to magnesium alloys:

Magnesium alloys containing rare earth metals have been used extensively during the past 10 years in applications requiring the high strength and creep resistance which rare earth metals impart to magnesium at temperatures upto approximately 450°F. - 600°F. They have been used principally for jet engine parts in the form of sand and permanent mould castings of varying size, configuration and complexity. Upto the present time commercial application of rare earth metals in magnesium alloys have been confined entirely to the use of misch metal as the alloying addition largely because of its availability and low cost. The total consumption of misch metal in magnesium industry during the last 10 years has amounted to approximately 250,000 lbs., with a maximum of about 50,000 lbs. being consumed in any one year. The availability during the last few years of some of the lighter rare earth metal in substantial quantities and at reasonable prices has permitted the investigations of these materials as alloying constituents in magnesium. The important outcome of this work has been the finding, that significantly higher strengths at all temperatures are obtained by the use of didymium in place of misch metal as the alloying constituent:

Some of the early investigations carried out in Germany and Great Britain clearly demonstrated the ability of rare earth metals as represented by misch metal to contribute to magnesium higher strength and creep resistance at elevated temperature and work was pursued to the point of developing commercially useful alloys.

Serious development of magnesium misch metal alloys was begun by Dow Chemical Co. in the early forties. Cast and wrought alloys were investigated and the compositional variation of mechanical properties including creep and fatigue was determined over a wide range of misch metal content (98).

Investigators at Batelle Memorial Institute made a thorough study of Mg-misch metal - Mn alloys. From their work they concluded that the optimum composition for casting is magnesium plus 6 % misch metal, 0.8 % Mn, 0.2 % Ni, 0.02 % W, for forging the optimum composition is magnesium plus 2 % misch metal, 1-1.5 % Mn and 0.2 % Ni (99 , 100). Addition of misch metal to magnesium extrusion alloys has been shown by Grube (101) to impart to them better high temperature properties such as high creep and fatigue resistance at 600°F. - 700° F.

Murphy and Payne (102) demonstrated that misch metal and zirconium can be added simultaneously to magnesium and that the grain refining action which zirconium has on magnesium is retained in presence of rare earth metals. The addition of Zirconium not only refines the grain, thus permiting the large intricate castings with no cracking during freezing, but also improves the mechanical properties.

Loenits (103) has evaluated the specific effects of various rare earth metals at room and elevated temperatures on properties of magnesium. All rare earth metals increased the creep resistance of extruded magnesium between 400°and 600°F. The increase depended on the temperature and concentration of the added metal. Alloys containing didymium exhibited the highest tensile and compressive strength at room and elevated temperatures.

Danks (104) studied the effect of adding rare earth metals to magnesium-zirconium alloys. An addition of 3 % misch metal was observed to confer high fatigue and creep resistance on these alloys.

McDonald (105,106) observed that addition of rare earthsto magnesium, produced cast alloys having elevated temperature properties markedly superior to those obtainable with the older alloys. By suitable heat-treatment creep resistance of 8000 psi at 400°F could be developed compared with 1500 psi for magnesium-aluminium-zinc alloys. Optimum concentration of misch metal was found out to be 3 %.

Mellor & Ridley (107) and McDonald (108) have attributed the improvement in the high temperature properties to the formation of fine precipitates in the grain boundaries. Loenitz (103) has found that addition of Ce to magnesium-thorium alloys introduces a magnesium-cerium compound Mg_9Ce into the structure which forms a coarse eutectic with the magnesium solid solution.

Mellor and Ridley (107) have further observed that the presence of manganese in magnesium alloys restrains such coarsening effect of the Mg_9Ce in the eutectic.

Roberts (109) had shown that the high magnification and resolving power of the electron microscope reveal a substantial amount of precipitate in the grain boundaries. Couling and Roberts (110) have however demonstrated that high level of strength and creep resistance of magnesium rare earth metal alloys at elevated temperature is believed to be associated with this grain boundary precipitate which blocks the cyclic grain boundary deformation characterstic of unalloyed magnesium.

Loentis (111) has determined the effects of rare earth additions on the linear coefficient of thermal expansion of commercial alloys. The additions of rare earth metals have no effect on the property.

The damping capacity of magnesium decreases with increasing misch metal additions, but not as rapidly or to as low a level, as with equivalent aluminium additions (111). Loentis and Feisel (112) made a detailed study of the effect of zirconium on the properties of magnesium misch metal and magnesium didymium alloys. A significant increase in the strength of megnesium 3 % misch metal and magensium 3 % didymium alloys at temperatures upto 400°F. was realized with increasing zirconium additions. Zirconium also improves the ductility of these alloys. At 400°F. zirconium has practically no effect on the creep resistance of magnesium misch metal alloys except at very low misch metal content. At 600°F. however alloys containing 3 to 6 % misch metal suffer a significant decrease in creep resistance. Additions of zirconium to magnesium-didymium alloys cause similar decrease in the creep resistance of these alloys at temperature over 400°F.

Although addition of zirconium to extruded magnesium rare earth metal alloys does not lead to significant increase in strength, it does improve ductility sufficiently to warrant consideration for commercial alloys (113). At present there is no commercial extrusion alloy containing rare earth metals.

A complete study (114) of alloys of magnesium with zinc and misch metal and of other zinc containing alloys has shown that a wide variation of mechanical properties can be attained. It was investigated that substantial additions of zinc can be made to magnesium-misch metal alloys, with markedly impairing either their room temperature or elevated temperature strength properties. Similar observations have been made for additions of zinc to magnesium-misch metal - zirconium alloys (102 , 107) (114 , 115). It has also been found that zinc additions to cast Mg-misch metal-Zr alloys improve their general foundry characteristics and more particularly render them completely free of micro porosity thus tending to perfectly pressure tight castings, Furthermore the addition of zinc renders these alloys insensitive to heat-treatment thus making it possible to obtain their optimum properties by aging directly the as cast material.

Aluminium additions to Mg-Ce, Mg-misch metal, Mn alloys have been studied by a number of investigators (98 100,115,116). All have found that aluminium decreases the strength properties of these alloys at all temperatures investigated.

Detailed information on the effects of silver on the properties of magnesium-rare earth metal alloys has not been published. Payne & Bailey (117) have recently found however, that the alloys containing 1 to 3.5 % aluminium plus 1 to 3.5 didymium and 0.4 to 1 % zirconium exhibit an exceptional combination of properties in the forms of both castings and wrought products. Within this composition range the alloy of Mg with 2.5% Ag + 2.0% didymium + 0.7 % Zr has been selected as the optimum for commercial development.

Recently Afanas (118) studied a wide range of alloys of the systems:

Mg - Li - Al (Li upto 20 %)
Mg - Li - Zn (Li upto 26.6 %)
and Mg alloys containing upto 14 % lithium.

The effects of Zr, Ce, Mn and other elements were investigated in the cast, hot pressed and thermally treated states.

Alloys with 3 % misch metal (mixed metal) 1.7 % Mn, 0.2 to 0.7 % Zn (Rods pressed at 320°-350°C.) were found strongest at normal temperature.

Effect of small additions on aging and properties of Au-Mg-Zn-Cu-alloy by Zamotorim and Pal. was studied (119). The following alloys were studied; each containing Zn 7.3, Mg2.7, Mn 0.5, Cu 0.5 and in addition the following (i) no addition, (ii) B 0.025, (iii) B 0.05, (iv) B 0.125 (v) misch metal 0.1 (vi) misch metal 0.25 (vii) misch metal 0.50, (viii) Ag 0.25, (ix) Ag 0.50, (x) Ag 0.75. They reported that misch metal and zirconium make the grain fine, hinder development of recrystallization and increase the plasticity at higher temperatures, after annealing for 5 hours at 350°C and hardening from 470-480°C. Boron and misch metal have no effect while silver increases the strength slightly in hardened state. Zirconium increases the strength in both annealed and hardened conditions. Aging is not effected by any of the additives studied.

Effect of misch metal on aging of magnesium-aluminium-zinc alloys was studied by Vatruba (120). The following kinds of alloys were prepared each containing 1 and 3 Wt% misch metal: Mg-Al containing 1.0, 2.5, 4.0, and 6.0 Wt% Al; Mg-Al-Zn with At. % ratios Al Zn of 20.69, 7.26, 2.42, 0.66 and 0.27 and mg-Zn containing 1.0, 3.0, 3.0 and 7.0 Wt% Zn. Specimens 5 x 8 mm were heated & treated as follows:

(i) annealing 48 hours at 410°C for Mg-Al and Mg-Al-Zn with Al-Zn 7.26 and 20.69; and at 325°C for Mg-Zn and Mg-Al-Zn with Al-Zn 0.27, 0.66 and 2.42, (ii) quenching in cold water after holding 24 hours at the same temperatures as in (i), (iii) quenching and aging at 125°, 150°C, 175°C for 6-240 hours. Afterwards Brinell hardness was determined, with a ball 5 mm in diameter and under a load of 250 Kg.

For each of the alloys, curves of hardnesses. temperature and time of aging were drawn. The aging effect is more intense, the more Zn in the alloy while addition of Al inhibits aging. Alloys that are rich in zinc do not show aging e.g. Mg-Al alloys as well as those with the ratio Al/Zn=20.69. Addition of misch metal raised hardness and decreased the aging effect e.g. deformed and annealed alloys Mg-Al-Zn without misch metal had Brinell Hardness (HB) 30.50, similar alloys with 1% misch metal had 40.60 and with 3% misch metal showed 40.65; similarily, alloys after aging but without misch metal showed maximum at 73HB while those with 3% misch metal at 77HB. The optimum aging effect was obtained in Mg-Al-Zn and Mg-Zn alloys after 48 hours at 175° C.

Effect of rare earth and alkaline earth on the mechanical properties of magnesium allrys of the systems Mg-Mn,Mg-Mn- Ce was studied by Mikherr and Dolgor (121). The possibility was investigated of improving the properties of the M.Al and M-Ag at room and elevated temperature by adding Nd, La, Ba and Sr. It is shown that La and Nd increase the properties of M A 8 from 8 to 12 Kg/mm^2 at 200°C, within 200 hours La, reduces the long term strength of M A 8 from 3.5 to 2 Kgm/mm^2 at 200°C. Nd increases the creep strength of MAl at 200°to 250°C by 1.0-1.5% Kgm/mm^2.

3.3.3 Rare earth alloys of nuclear interest:

As regards nuclear applications of yttrium and the lanthenons following points can be set down (11).

1. Europium, dysprosium, samarium and gadolinum or suitably devised combinations are interesting as poisons for use in reactor control rods. Europium and gadolinum cermets have proved to be usable control materials and their employment is planned in more than one power reactor core.

2. Erbium, dysprosium, europium and possibly lutetium are of interest as possible burnable poison materials. Samarium can be added to this list as a fast burn-up reactivity compensator.

3. Holmium, dysprosium, europium and even scandium might be of use as regional blackners flux-suppressors and similar applications.

4. Yttrium, cerium and alloys of these elements with zirconium would be promising materials for study as hydrogen carriers for solid moderator applications in cores of intermediate and thermal spectrum liquid metal or gas cooled reactors.

5. Applications of these materials as shielding moderators in high efficiency design is also possible. Use of hydrided lanthanons of higher cross section, seems to be alternative interesting path for systems in which homogeneous shields would be desirable.

6. Use of ceria, yttria and lanthana, as diluents for oxide fuelbodies is a promising development area. Mixtures of

burnable poison-type lanthanon oxides as a ternary in such systems might also be axplored.

7. Use of yttrium pipe to carry potential liquid metal fuels for either thermal or fast breeder reactors is already an accomplished fact.

8. Use of the lanthanon metals, and of yttrium as alloy modifiers for zirconium and niobium or their complex alloys is being studied. Interesting developments in this area might lead to fairly large scale utilization of these elements.

9. Miscellaneous small scale applications of the lanthanons in area near the boundary of the nuclear field promise advances in areas where hard radiation may be applied. Europium and possibly scandium when activated in pile could compete with Co^{60} when hard radiation is needed.

10 Several of the lanthanons have been studied as alloy ingradients particularly alloys of dysprosium in zirconium and zircaloy 2. Minor alloy ingradients for property modification also seem quite promising.

11. Large scale reductions in costs of lanthanon oxides and pilot commercial production of lanthanon metals at reasonable prices make the use of naturally occurring mixtures of isotopes of these metals attractive to the nuclear technologists.

12. Use of lanthanon fluorides as coprecipitants for fission product poisons in fused salt fluid fuel reactor systems has been shown to be feasible by studies carried out at Oak Ridge.

Recently because of interest in high thermal neutron cross-sections of gadolinum (76) work has been started on alloys of this metal. Because of the scarcity of the metal alloys were made with the oxide dispersed in titanium or zirconium. They have been saparated both by arc melting and extrusion.

A triple arc melted ingot of titanium containing 4.7 wt. % Gd_2O_3 had a brinell hardness of 285. Alloy containing 90 % Ti, 10 % Gd_2O_3 was resistant to 600°F. water after one month. However an alloy containing 85 % Zr + 5% Sn + 10% Gd_2O_3 was rapidly attacked at 600°F. An alloy prepared by cold compacting evacuating and extruding ZrH_2 plus 10 weight percent Sm_2O_3 disintegrated in 600°F. water.

Zirconium powder with 25 wt. percent Gd_2O_3 and Sm_2O_3 was cold compacted in a cold can then extruded at 1100°F. with a 14:1 ratio. Resulting tensile strength was 57,000 lbs/sq.in. with 0.64 % elongation as compared to 120,000 lbs/sq.in. and 6 % elongation for a zirconium extrusion similarly prepared.

S.4 ADDITIONS TO OTHER NON-FERROUS ALLOYS:

The rare earths and yttrium are but slightly soluble in chromium. None of the rare earths exceeds 0.4 % solubility. Samarium and ytterbium are insoluble in chromium. The effects of all the rare earths on the ability of chromium to withstand high temperatures in air is remarkable (122)

Chromium exhibits a peculiarity among the refractory metals, in that nitrogen is absorbed by chromium at temperatures above red heat and diffuses throughout the metal structure. This is a greater problem than oxidation, since chromium is rendered brittle and has no metallurgical value. Collins, Calkins and McGurty (1961) found that rare earth elements and specially yttrium were effective in improving the resistance of chromium to both nitrogen and oxygen upto 2500°F. Praseodymium, neodymium, gadolinium, erbium, lutetium and yttrium were the most effective in controlling both scalling and nitrogen absorption. Samarium, dysprosium, thulium and ytterbium were found to have low solubility in chromium and therefore had less influence on the high temperature properties. Yttrium also functions as a scavenger and grain refiner for chromium, but it is not effective to the extent that room temperature brittleness is alleviated or the transition temperature lowered.

Although yttrium has no effect on the high temperature oxidation resistance of vanadium, its ~~scavengering~~ scavenging effect and its ability to render massive vanadium ductile in terms of workability and deformation at room temperature are very useful.

Hardness of the vanadium dropped with even minor yttrium additions to such an extent that the original hardness of 221 for unalloyed vanadium dropped to 153 for vanadium alloyed with 1 % yttrium. A series of arc melted buttons containing 0 to 3 % yttrium were cold pressed and cold rolled. It was found (123) that as the yttrium content increased, the ductility increased accordingly. Apparently about 1/2 of the yttrium addition was converted into oxide, as it scavenged the oxygen from the vanadium. The alloy system exhibits a liquid immiscibility gap from 9 to 99.9 wt. percent vanadium. The two liquid regions of the Y - V system is extremely favourable for producing a clean, pure vanadium by the scavenging action of yttrium.

Savitskiy (124) has shown that rare earth metals improve electrical and thermal conductivity of copper alloys and weldability of titanium.

3.3.5 Additions to ferrous alloys:

Morrogh showed that misch metal is one important compound in effecting nodularization of cast iron to the ductile form. Post Sehefistall and Bevaer (125) published the first paper and reported that addition of cerium and lanthanum in the range of about 0.02 and 0.04 % is effective in promoting better hot workability in high frequency Ajax and electric arc melted stainless steel. In other publications the powerful influence of misch metal and that of Lancer Amp addition in steel are sited by them . Lancer=Amp addition was found to be effective for desulphurising steel and increasing the ductility and toughness in steels, increasing fluidity of cast steel and aiding in the control of grain size of steel. Increase in corrosion resistance of many alloys by such addition has also been reported.

Knapp (126), Jackson (127) and Bearer & Lamphier have also studied hot workability of rare earth stainless steels and have confirmed the original results obtained by Post & coworkers . They attribute absence of hot -shortness of high alloy steels by rare earth addition, to the increased ductility at the rolling temperature. Oxidation resistance of steel and non-ferrous metals was found out to be improved.

Russell (128) Snellman (129) and Gautshi & Langenberg (130) studied the desulphurization of steel by rare earths addition and have found that rare earths are strong desulphurizers, but efficiency of desulphurization depends on many factors like initial sulphur content, mode of rare earths additions, and the holding time.

Beaver (131) in his further work has reported superior tensile and impact properties, improved oxidation resistance for rare earths added stainless steels. The superior mechanical properties of rare earth added stainless steels were supposed to be due to grain refinement and inclusion modifications.

Evan (132) has reported that Ni-Cr alloys with rare earths addition had a longer life, and concluded that it was due to improved adhesion of the oxide scale, caused by keying on the oxides in the protective film.

Trombe (133) has reported superior oxidation properties for stainless steel and Ni-Cr alloys. In other publication (134,81) it has been reported that rare earths addition improves the resistance of steel to high temperature corrosion.

Cheetham (136) has described important advantages gained by rare earth additions in the production of alloy steel, stainless steel, high speed steels and silicon steels for electrical applications. These include superior mechanical properties such as higher impact value at normal and low temperature, greater reduction in area percentage, better resistance to corrosion and oxidation at elevated temperatures, easier cogging and fogging of high speed steel, superior electrical qualities in silicon steels and an increase in grain growth temperature.

Gautschi and Langenberg (130) did not find any change in size, shape, amount or distribution of non-metallic inclusions between the treated and untreated heats.

In other publications the useful effects of rare earths addition,such as improved ductility, superior tensile and impact properties, grain refinement, effective deoxidation and lower oxygen content, scale suppression and prevention of segregation have been tabulated. The American Metallurgical Product Co. has reported an improvement of impact values at low temperatures by 60 % with the use of Lancer-Amp addition.

Becker & coworker (31) have observed that small amounts of boron and zirconium picked up from magnesia or zirconium crucibles improve the high temperature properties of vacuum melted Ni-Cr-Co alloys and as the rare-earths contain some boron and zirconium, part of the improvement could be due to the passage of these elements through rare earths addition.

Farrel & Rosenthal (135) have reported that small boron addition reduces the susceptibility of strainless steel to intergranular corrosion.

Fink (136) studied the effect of niobium, misch metal and vanadium additions on alloy 20, which exhibits high resistance to sulphuric acid attack. The evidence available indicates that it may be the result of: (i) more thorough deoxidation, (ii) promotion of nucleation and (iii) increase in strength at the grain boundaries.

Sehgal & Eppelsheimer (137) reported the possibility of desulphurization by rare earths addition in the induction melted heats. Oxides of rare earth were more effective than fluorides and Lancer-Amp. They have shown that oxides and flourides of rare earths and Lancer-Amp. tend to minimise the size of inclusions and distribute them more uniformly. Rare earths oxides and Lancer-Amp. addition appear to refine the grain size of low alloy steel, while rare earth fluoride additions result in grain coarsening.

In his further work Sehgal (31) has observed the following effects in low alloy and stainless steel.

i) Refinement of grain size, reduction in number of inclusions and improvement in forgeability.

ii) Improvement in high temperature tensile strengths and oxidation resistance by Lancer-Amp 23, Lancer-Amp 2 and ~~x~~ rare earths fluorides addition .

iii) Decrease in the corrosion resistance of stainless steel by small additions of Lancer-Amp. 23 and Lancer-Amp 2 additions.

Work conducted by and for General Electric Company (138) has revealed some extraordinary effects of yttrium and rare earths on the properties of ferrous base alloys. Yttrium and the rare earths are only slightly soluble in iron base alloys. Yet this slight solubility has a notable effect on certain properties of Fe-Cr system such as workability , grain refinement, resistance to recrystallization at elevated temperatures and especially oxidation resistance.

Oxidation tests carried out in case of the type 446 composition (75 % Fe, 25 % Cr) resulted in 50 % improvement in short time oxidation rate. It was found that the addition of 1 % yttrium is as effective as the addition of 5 % aluminium in improving the oxidation resistance of the Fe-Cr base metal. A fine grained structure is retained in the yttrium containing material after prolonged heating.

The combination of 25 % Cr, 1 % Y and 3 % Al provides an oxide coating at temperature upto 2600°F. which is enamel like non-spalling and thermally shock resistant. For comparison the alloys were heated in air at 2300°F. for 100 hours. The Fe-Cr-Al alloy has a powder layer of non-adherent aluminium oxide whereas the sample with 1 % yttrium addition shows a smooth enamel coating. Similar oxidation resistance is obtained by additing 1 % thorium to the 30 % Cr - 1 % yttrium alloy, but the oxide film is not as smooth as that of Al-Y combination.

Only minor improvements were noted in the oxidation resistance of austenitic stainless steel with yttrium additions (B38). Alloys of higher chromium such as type 310 showed some improvements but the standard 18-8 compositions did not benefit from yttrium addition.

The new direction of Soviet work studies the application of individual rare earth metals, the effect of rare earth-elements on alloying properties of common metals, and the optimum combination of rare earth metals to improve the properties of industrial alloys. A review of the use of rare earths in the Soviet Union is contained in a recent monograph (124) on rare metals by Savitskiy. Because of their chemical activity and the stability of their oxides and sulphides rare earths are used in Russia as reducing agents, desulphurizers, modifiers and hardening agents for steel& alloys. In Fe-Al alloys additions from 0.25 to 0.5 % cerium diminish the grain size, shift the magnetic conversion time and improve the ability to be hot rolled. In cast iron the addition of 0.2 % ferrocerium increases the mechanical properties by a factor of two and makes possible large cast iron crankshafts with better properties than those of forged crankshaft for tractor diesel engines.

Recently Savitskill (139) analysed systematically and extensively the effect of rare earths metals on the properties of heat resistant alloys from the point of view of their modifying action, removal of oxygen, nitrogen, hydrogen and carbon impurities, improved structure and oxidation resistance of surface layer, variation in the mechnism of plastic deformation of alloy matrix, lowering of recrystallization temperature and strengthening action of alloying elements.

Singh and Sharan (14) have recently shown that addition 0.15 % misch metal improves the machinability and mechanical properties of grey cast iron. Sharan, Gaindhar and Narayan (15) carried out the systematic investigation to study the effects of mischmetal on the malleabilization cycle, micro structure, machinability, mechanical properties (shear, compressive strength and hardness) and corrosion behaviour in HCl, HNO_3, and NaOH media of malleable cast iron. Following beneficial effects of misch metal have been established.

(i) Reduction in malleabilization time on treatment with about 0.2 % misch metal.

(ii) Improvement in machinability, shear and compressive strength by optimum addition of 2 % misch metal.

(iii) Large improvement in corrosion resistance of 0.05 % misch metal for 5 % HNO_3 and 50 % NaOH and at 0.1 % for 5 % HCl media.

Khan (16) extended the work further to study the effect of misch metal on alloyed cast iron and has shown the beneficial effect of the addition to cast iron from mechinability and strength consideration.

The need is, however, emphasized for wider and more detailed investigation of these problems partiaularly as regards the development of a physico-chemical theory of mechanism by which rare earth metals influence various types of alloys.

CHAPTER IV

HIGH TEMPERATURE X-RAY DIFFRACTION STUDY OF ALUMINIUM AND ITS ALLOYS.

CHAPTER IV

HIGH TEMPERATURE X-RAY DIFFRACTION STUDY OF ALUMINIUM AND ITS ALLOYS

4.1 GENERAL INTRODUCTION.

The review of the previous work has indicated that very little work on the study of lattice parameter of aluminium alloys at high temperatures has been reported. The present investigation is an attempt to determine as accurately as possible the change in lattice spacing with temperature and composition under equilibrium conditions of aluminium-copper and aluminium-silver alloy systems. The equations representing the change of lattice parameter with temperature for pure aluminium and other aluminium alloys were determined by computers. The densities and co-efficients of thermal expansion at different temperatures were found in single phase region. The solid solubility limit line for aluminium-copper system between 450° and 548° C has also been determined.

4.2 EXPERIMENTAL PROCEDURE.

4.2.1 Preparation and chemical analysis of alloy :

Aluminium of 99.9% purity, electrolytic copper and silver of 99.9% purity formed the basis of test alloys. Different alloys were prepared and analysed as per details given below:-

4.2.1.1 Aluminium-copper alloys :

At first hardener alloy of about 50% copper and 50% aluminium was made by melting in a graphite crucible in an oil fired furnace. This was then chemically analysed.

Other alloys were then made from this hardener alloy by adding weighed amount of aluminium. Alloys were melted in a vertical furnace in an argon atmosphere. Stirring was done to provide homogenity in the melt. Casting was done in cast iron metal mould, and then homogenised at 500° C in an electric furnace for about 20 hours to eliminate the effect of coring if any.

The iodometric method was used for copper estimation. Copper was precipitated as copper-sulphide and was ignited to oxide and fumed to evaporation with H_2SO_4. The solution was then diluted, neutralised with Na_2CO_3 and the iodometric titration was carried out in a weakly acidic medium (acetic Acid) against standard sodium thiosulphate solution using starch as indicator. Duplicate samples were taken for each alloy. Results of chemical analysis are giveh in table 4.01 .

4.2.1.2 Aluminium-silver alloys :

A master alloy of about 50 percent aluminium and 50 percent silver was made in a graphite crucible in argon atmosphere controlled furnace. The other alloys were then made from this master alloy by adding weighed amounts of

TABLE - 4.01

CHEMICAL ANALYSIS OF ALLOYS

S.No.	Aluminium copper alloy		Aluminium silver alloy	
	Desired composition	Actual analysis	Desired composition	Actual analysis
1	2 %	1.91 %	10 %	7.5 %
2	4 %	4.35 %	20 %	19.9 %
3	10 %	9.92 %	-	-
4	12 %	12.18 %		
5	14 %	13.79 %		

aluminium and cast in cast iron moulds. Alloys were homogenised at 500° C in an electric furnace for about 20 hours.

The alloys were chemically analysed by chloride method. The drillings were dissolved in nitric acid and silver was precipitated as silver chloride by adding dil. HCl. Tartaric acid was added to avoid oxidation. The silver chloride precipitated was filtered and weighed and the amount of silver was calculated from it. Results of chemical analysis, based on duplicate samples are given in table 4.01 .

4.2.2 X-ray unit :

The Debye Scherrer photographs were taken in a 19 cm. diameter Unicam high temperature camera. The instrument comprises of four main parts, the camera, the evacuation chamber, the furnace, and the main base. The description of these parts is given briefly.

4.2.2.1 Camera :

The camera consists of a circular ring base which carries a short vertical cylinder, the outer diameter of which is 19 cm., Round this cylinder the two pieces of the photographic film may be clamped by means of steel spring straps . The straps terminate in brass strips which may be pressed under spring loaded steel balls to hold them in position. The ends of each piece of film extended beyond the knife-edges are incorporated in the cylinder supporting the film. The aperture which limits the size of the incident beam of X-ray is carried on the circumference of the film supporting cylinder and is 1.25 x 1.5 mm.

A light-tight cover fits over the base. The cylinderical side of this cover tapers to a thin edge which fits into a groove, the correct position of the cover relative to the base is secured by a projection which fits into a gap on the bottom edge. On the inlet side of the cover there is provision for a filter to separate white radiations from the characteristic radiations. All light must be excluded from the aperture into the camera and if filter proves inefficient for this purpose, black paper may be used. On the outlet side of the cover is removable fluorescent screen.

4.2.2.2 Evacuation chamber :

This is necessary to prevent cooling of the furnace by convection and to prevent oxidation of certain specimens at high temperatures. It consists of a cylinder closed at the top and open at the bottom. On the side is cut an opening for the passage of the reflected X-rays covered by a cellophane which is held in place by strong wires and rendered air tight with secotina. The chamber is cooled by a stream of water passing through the pipes at the rear of the main base.

4.2.2.3 Furnace :

The furnace is constructed in the form of two short cylinders mounted one above the other with a small gap between them. The upper and lower portions are joined together by metal parts which support the inlet and outlet guards to prevent diffusely scattered radiation from reaching the photographic film. The heating wires are 10 percent platinum - rhodium alloy and are wound to give a large region with a uniform temperature at the centre of the furnace. The temperature is measured by two thermocouples, one mounted radially at the top of the lower half, and the other mounted axially in the upper half of the furnace. The second thermocouple may be adjusted in position and so brought near to the specimen. The furnace operates safely upto a temperature of about 1000° C. The furnace is nickel plated on all the metallic parts to provide the surface, resistant to oxidation even when hot.

4.2.2.4 Main base :

It consists of a circular casting on the top of which are locating pins. These define the positions of the camera, the tube into which the furnace fits, and the specimen holder. On the outside of the base, is the control panel which carries motor and fuses for each half of the furnace, the terminals for connecting the furnace to the control unit, and a plug for connecting the temperature indicator with one or other of the two thermocouples. A mains-fuse is mounted on the control unit. The evacuation pipe comes out at the rear of the base between the inlet & outlet of the cooling water ring. Levelling toes & a spirit level are the parts of the main base.

4.2.2.5 Alignment of the camera in the X-ray beam :

Approximate beam alignment is obtained before connecting the necessary equipment as the position of the later depends upon the location of the camera. During alignment the X-ray beam is viewed through the fluorescent screen at the exit slit covered by a protective lead glass window. The specimen is kept in position to make this adjustment. The pivoted beam trap should cover the fluorescent screen during exposure.

4.2.2.6 Mounting of the specimen :

Before mounting the specimen the camera is dismantled in the following manner.

i) The camera ring is removed first by lifting directly upwards the camera is placed on a flat surface, care being taken to avoid the damage of the black paper screen.

ii) The evacuation chamber is next lifted vertically clear of the furnace assembly. Cellophane window is protected from damage.

iii) Furnace couplings are withdrawn and the furnace is lifted vertically. The lead insulated connections are handled with extreme care.

iv) If the sample is not in the form of a wire but in the form of a powder, it is inserted in a fine capillary. Silica, alumina, or platinum capillaries are generally used.

4.2.2.7 Centering of the Specimen :

Proper specimen alignment is done by the microscope attachments which can easily be fixed to the main base with a clamping screw.

The microscope is adjusted at the factory so that the vertical crosswire is accurately in line with the main axis. Any error is corrected by the screw adjustment provided on the microscope bracket. The specimen is brought in exact alignment with the main axis of the camera by a series of approximations using the four screws mutually opposed at right angles.

4.2.2.8 Operation of the unit :

The following steps have to be carefully followed in order to take a photograph with the high temperature powder camera.

i) The vacuum pump is connected to the pumping out tube.

ii) Thermocouples are connected to the temperature indicator by the light weight cable coming from the lower part of the front control panel.

iii) Control units are connected to the camera at the two control terminals. The units are connected to the main supply through a 10 ampere plug.

iv) Before raising the temperature of the furnace, the cooling water must be circulated and a partial vacuum applied. Final pumping out is completed during the heating period.

The variac is set to zero before begining the run up as the cold resistance of the furnace is of the order of 1.0 ohm. Gradual increase and decrease of furnace temperature is necessary to avoid damage to the refractory linings.

4.2.3 Knife - edge calibration of the high temperature Unicam camera :

The adjustments of the knife-edges in each half of the camera are made precisely at the factory. The ends of each film extend beyond the knife - edges, which are incorporated in the cylinder supporting the film shadows of knife edges formed by reflected X-rays, give fluidiced lines corresponding to 5^{o} and 85^{o} C respectively. The knife - edge calibration was checked by taking X-ray diffraction photographs of spectroscopically pure silver wire at room temperature. The wire was perfectly annealed at 400^{o} C in camera under vacuum for 10 hours to relieve the stresses before taking the photographs.

Both the films for pure silver were measured separately and lattice parameter was calculated for all the lines. Nelson Riley's extrapolation curve was then plotted separately, for each film. Extrapolation graphs were drawn on large graph sheets so that each small division along the ordinate represented 0.00002 A^0. Very precise values of the lattice parameter were then found out. Values were compared with the accurate values of lattice parameter already reported.

The photographs of the extrapolation curves are given in Fig. 4.01 and 4.02. X-ray diffraction photographs are given in Plates 4.01 and 4.02, & observations are given in tables 4.02 & 4.03.

The values of lattice parameter determined, using the knife edge calibration were found in agreement with the accurately reported values of lattice parameter of pure silver at room temperature. This indicated that the knife edges were perfectly fixed in the camera.

4.2.4 Temperature calibration of the high temperature Unicam camera :

The exact mean specimen temperature may be different from the hot junction temperature of the thermocouple actually recorded. Therefore, exact mean temperature should be determined from the recorded temperature.

In these experiments the thermocouple marked T_2 was used throughout the investigation. The thermocouple was calibrated for mean specimen temperature by the author by using pure silver specimen, the lattice parameters of which were known at different temperatures very precisely by Owen and Roberts. Their results were used for calibration.

The sample of silver was used in the form of a wire. The sample was annealed in vacuum at 700° C for 12 hours. The X-ray diffraction photographs were taken at 140°, 240° and 450°C as recorded by the thermocouple T_2. The lattice parameters at different temperatures were determined by using Nelson and Riley's extrapolation method. The results are given in tables . 4.04 to 4.07 and Figures . 4.03 to 4.05 and plates 4.03 to 4.05. From the values of lattice parameter so determined and standard values given by Owen and Roberts, (141) the correct mean specimen temperature to which the value should correspond, was found out. The curve was plotted between observed thermocouple temperature T_2 and the mean specimen temperature (Figure 4.06).

X- RAY DIFFRACTION PHOTOGRAPHS OF PURE Ag AT DIFFERENT TEMPERATURES 20mA, 40 KVP, 12 hrs.

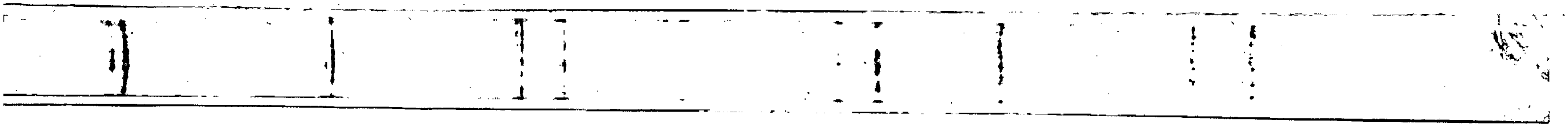

PLATE 4.01 (FIRST FILM) TEMP. 22°c

PLATE 4.02 (SECOND FILM) TEMP. 22°c

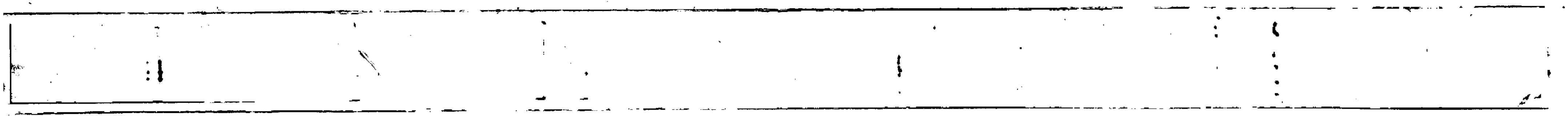

PLATE 4.03 TEMP. 140°c

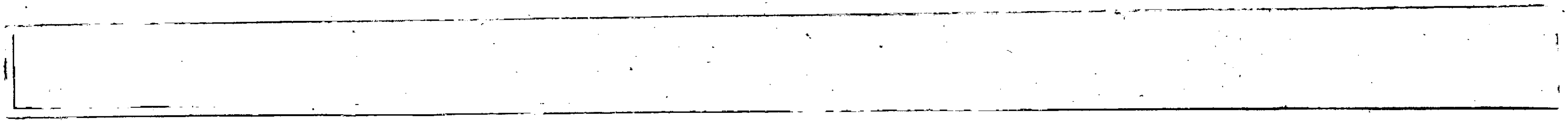

PLATE 4.04 TEMP. 240°c

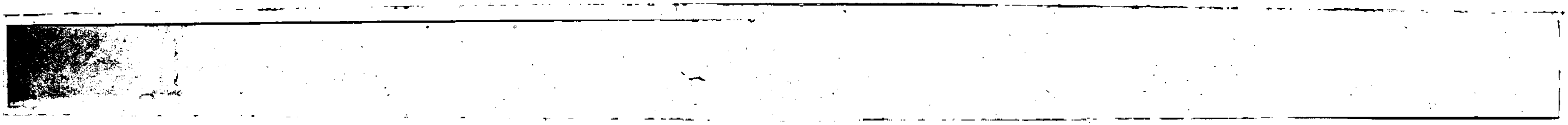

PLATE 4.05 TEMP. 450°c

LATTICE PARAMETER VS EXTRAPOLATION FUNCTION FOR PURE Ag AT DIFFERENT TEMPERATURES

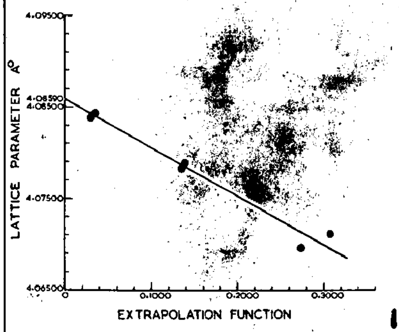

FIG.4.01 PURE SILVER FILM No.1
AT 22°C

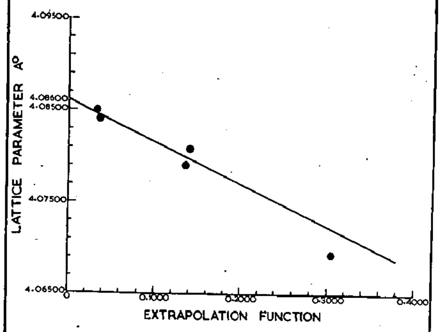

FIG.4.02 PURE SILVER FILM No.2
AT 22°C

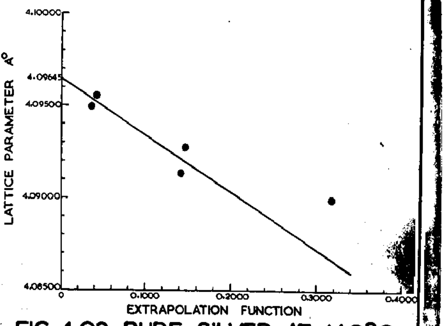

FIG.4.03 PURE SILVER AT 140°C

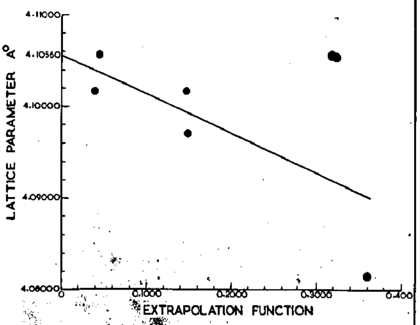

FIG.4.04 PURE SILVER AT 240°C

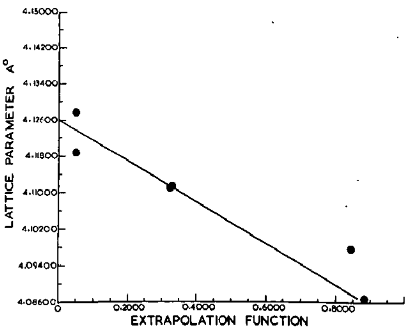

FIG.4.05 PURE SILVER AT 450°C

TABLE - 4.02

X-RAY DIFFRACTION RESULTS OF PURE SILVER AT 22°C

Film No.1 (Left Side)

Indices	Wave length	θ	Lattice parameter A. U.	Nelson & Riley's function	Extrapolated value
420	α_1	57.83	4.06934	0.3080	
420	α_2	58.05	4.06974	0.2950	4.08590
422	α_1	67.68	4.07895	0.1390	
422	α_2	68.06	4.07860	0.1350	
511	α_1	78.50	4.08437	0.0350	
511	α_2	79.25	4.08391	0.0305	

TABLE - 4.03

X-RAY DIFFRACTION RESULTS OF PURE SILVER AT 22° C

Film No. 2 (Right side)

Indices	Wave length	θ	Lattice Parameter A. U.	Nelson & Riley's Function	Extrapolated Value
420	α	57.98	4.07019	0.3030	
422	α_1	67.62	4.08080	0.1400	
422	α_2	68.03	4.07905	0.1350	4.0860
511	α_1	78.34	4.08406	0.0360	
511	α_2	79.20	4.08500	0.0330	

TABLE 4.04

X-RAY DIFFRACTION RESULTS OF PURE SILVER AT 140° C

Indices	Wave length	θ	Lattice parameter A. U.	Extrapolation function	Extrapolated value A. U.
420	α	57.39	4.09000	0.3170	
422	α_1	67.22	4.09277	0.1460	
422	α_2	67.51	4.09570	0.1420	4.09645
511	α_1	77.74	4.09558	0.0400	
511	α_2	78.49	4.09496	0.0350	

TABLE - 4.05

X-RAY DIFFRACTION RESULTS OF PURE SILVER AT 240° C

Indices	Wave length	θ	Lattice Parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α_2	55.51	4.08163	0.3600	
420	α_1	57.05	4.10534	0.3250	
420	α_2	57.27	4.10547	0.3200	
422	α_1	67.08	4.09700	0.1476	4.10560
422	α_2	67.15	4.10161	0.1465	
511	α_1	77.13	4.10562	0.0440	
511	α_2	77.81	4.10160	0.0370	

TABLE - 4.06

X-RAY DIFFRACTION RESULTS OF PURE SILVER AT 450° C

Indices	Wave length	θ	Lattice Parameter A. U.	Extrapolation function	Extrapolated value A. U.
311	α	40.67	4.09868	0.82428	
422	α_1	56.92	4.11111	0.3280	
422	α_2	57.15	4.11088	0.3230	
511	α_1	76.35	4.11859	0.0495	
511	α_2	76.46	4.12733	0.0485	4.1260 A°

TABLE - 4.07

CALIBRATION OF THERMOCOUPLE T_2 TEMPERATURE FOR MEAN SPECIMEN TEMPERATURE

S.No.	Observed thermocouple T_2 °C	Lattice parameter A. U.	Lattice parameter K X units	Corrected mean specimen temperature
1	140	4.09645	4.0881	152° C
2	240	4.10560	4.09732	257° C
3	450	4.12600	4.11768	494° C

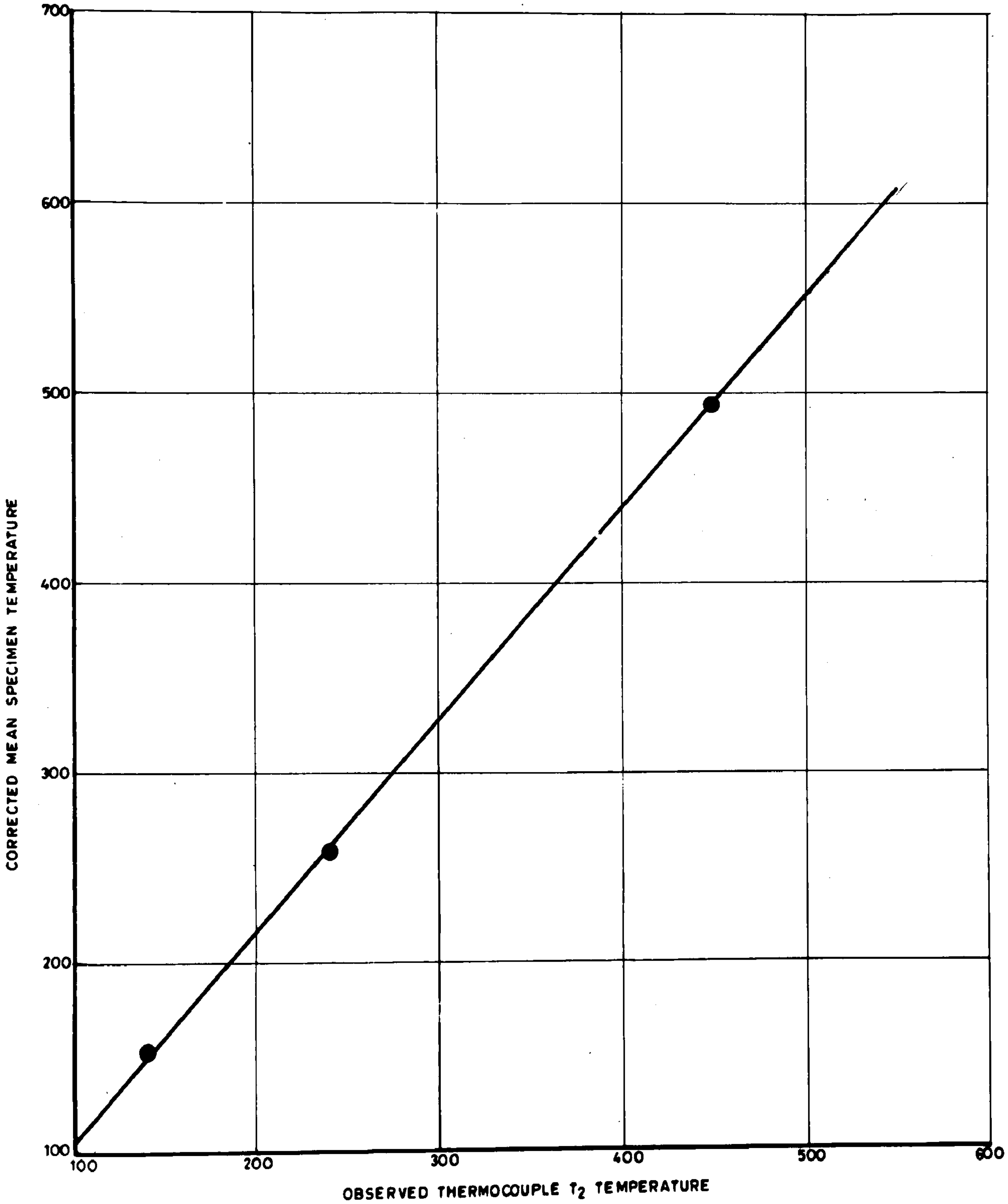

FIG. 4.06-CALBRATION CURVE FOR THERMOCOUPL T_2 TEMPERATURE FOR MEAN SPECIMEN TEMP

4.2.5 X-ray technique and lattice spacing measurement :

Debye Scherrer photographs were taken using radiation obtained from a Raynax demountable unit. Following factors were used.

Copper K_α radiation was used for taking X-ray diffraction patterns and nickel was used as filter at the window.

Though all the lines have been measured, only high angle lines 511, 422, 420, 331 and 400 both α_1 and α_2 have been used for the calculation of the lattice parameter. Films were measured on a film measuring rule giving an accuracy upto 0.005 cm. Each film was measured twice. One reading was taken from right to left and then film was reversed and another reading was taken. Same procedure was followed for another film. Two photographs were taken for each temperature in all the cases. The θ values reported in the table are the mean of the four values taken from the two photographs. Lattice parameter was calculated for each line.

Curve has been plotted between Nelson & Riley's extrapolation function and lattice parameter and actual value has been calculated by extrapolating the values to $\theta = 90^\circ$ C. Each division in the ordinate had an accuracy of 0.00002 A°. Reduced photographs of the curve, lattice parameter vs. extrapolation function have been given for reference.

4.2.6 Method for determining the density and coefficient of thermal expansion :

In the case of pure aluminium density and coefficient of thermal expansion were calculated at all temperatures . In case of aluminium copper and aluminium silver alloys, coefficient of thermal .expansion and densities were calculated in single phase region only when whole of the Cu or silver remains in solid solution. Thus in case of Al-Cu. alloy, values were reported at 525° and 548° C while in case of aluminium silver alloys, values were reported at 450°, 475°, 500° and 525° C.

From the lattice parameter values, the density for each alloy was calcullated by the given formula ;

$$dt = \frac{m\,M.N}{a_t^3}$$

where m = mass of atom at unit atomic weight.

M = composite atomic weight

N = no. of atoms/unit cell

a_t = lattice parameter measurement at temperature t° C

The composite atomic weight was calculated by the given formula ;

$$M = \frac{\text{Wt. \% A.X atomic wt. of A + wt.\% B x atomic wt. of B}}{100}$$

Thus in case of pure aluminium, M was replaced by the atomic weight of aluminium for calculating the densities.

The coefficients of thermal expansion were calculated in case of pure aluminium with reference to lattice parameter at 0° C while in case of aluminium-copper and aluminium-silver alloys, the coefficient of thermal expansion at a particular temperature was determined with reference to the lattice parameter at that temperature. Thus the following formulae were used in case of aluminium and its alloys for calculating the values of coefficient of thermal expansion.

$$\text{Aluminium} \quad \alpha_t = \frac{1}{ao}\frac{da}{dt}$$

$$\text{Aluminium alloy} \quad \alpha_t = \frac{1}{aT}\frac{da}{dt}$$

da/dt, a_o and aT were determined from equation of the lattice spacing vs temperature curve for a given metal or alloy.

4.3 EXPERIMENTAL RESULTS .

4.3.1 Pure aluminium :

High temperature X-ray diffraction results are given in Figs. 4.07 to 4.11, plates 4.06 to 4.10 and tables 4.08 to 4.12.

4.3.2 Aluminium copper alloys :

High temperature X-ray diffraction results are given in Figs. 4.12 to 4.36, tables 4.13 to 4.37 and plates

4.11 to 4.35 .

4.3.3 Aluminium silver alloys :

High temperature X-ray diffraction results are given in Figs - 4.37 to 4.44, tables 4.38 to 4.45 and plates 4.36 to 4.43.

TABLE - 4.08

X-RAY DIFFRACTION RESULTS OF PURE ALUMINIUM AT 30° C

Indices	Wave length	θ	Lattice parameter A.U	Extrapolation function	Extrapolated value A. U.
331	α_1	56.11	4.04467	0.3460	4.05090
331	α_2	56.33	4.04458	0.3410	
420	α_1	58.37	4.04613	0.2970	
420	α_2	58.61	4.04537	0.2920	
422	α_1	68.77	4.04848	0.1255	
422	α_2	69.15	4.04824	0.1205	
511	α_1	81.22	4.04987	0.0200	
511	α_2	82.05	4.05143	0.0165	

TABLE - 4.09

X-RAY DIFFRACTION RESULTS OF PURE ALUMINIUM AT 271° C

Indices	Wave length	θ	Lattice Parameter A.U.	Extrapolation function	Extrapolated value A. U .
331	α_1	55.45	4.07679	0.362	4.08450
331	α_2	55.63	4.07809	0.358	
420	α_1	57.61	4.07945	0.313	
420	α_2	57.89	4.07753	0.306	
422	α_1	67.59	4.08207	0.1400	
422	α_2	67.88	4.08376	0.1360	
511	α_1	78.54	4.08390	0.0350	
511	α_2	79.28	4.08379	0.0330	

TABLE - 4.10

X-RAY DIFFRACTION RESULTS OF PURE ALUMINIUM AT 382° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.26	4.08974	0.366	
420	α_1	57.39	4.09000	0.317	4.09510
420	α_2	57.62	4.08839	0.313	
422	α_1	67.24	4.09233	0.145	
422	α_2	67.57	4.09282	0.140	
511	α_1	77.83	4.09440	0.039	
511	α_2	78.51	4.09445	0.035	

TABLE - 4.11

X-RAY DIFFRACTION RESULTS OF PURE ALUMINIUM AT 494° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	57.14	4.10110	0.3240	
420	α_2	57.35	4.10172	0.3180	4.10698
422	α_1	66.87	4.10355	0.1500	
422	α_2	67.20	4.10350	0.1460	
511	α_1	77.08	4.10661	0.0440	
511	α_2	77.75	4.10593	0.0395	

TABLE 4.12

X-RAY DIFFRACTION RESULTS OF PURE ALUMINIUM AT 606.5°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α_1	54.69	4.11506	0.3790	4.12094
331	α_2	54.99	4.11062	0.3720	
420	α_1	56.83	4.11545	0.3300	
420	α_2	57.09	4.11406	0.3240	
422	α_1	66.41	4.11760	0.1570	
422	α_2	66.72	4.11820	0.1520	
511	α_1	76.29	4.11970	0.0490	
511	α_2	76.86	4.12037	0.0460	

X-RAY DIFFRACTION PHOTOGRAPHS OF PURE Al AT DIFFERENT TEMPERATURES 20mA, 40 KVP, 6 hrs.

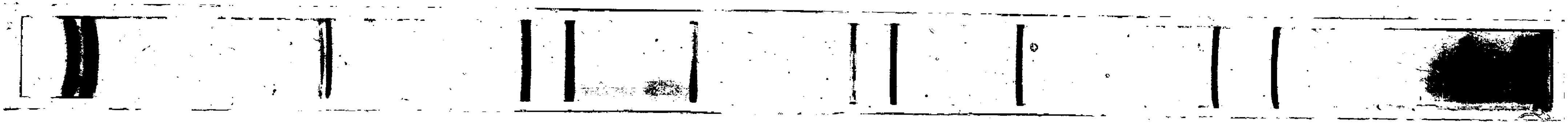

PLATE 4.06 TEMP. 30°c

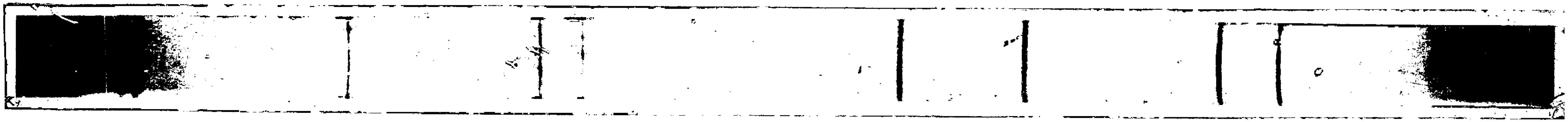

PLATE 4.07 TEMP. 271°c

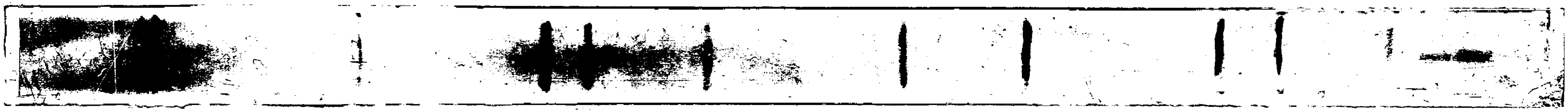

PLATE 4.08 TEMP. 382°c

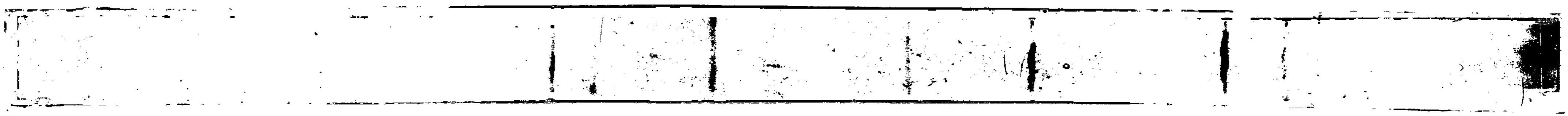

PLATE 4.09 TEMP. 494°c

PLATE 4.10 TEMP. 606.5°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR PURE Al AT DIFFERENT TEMPERATURES

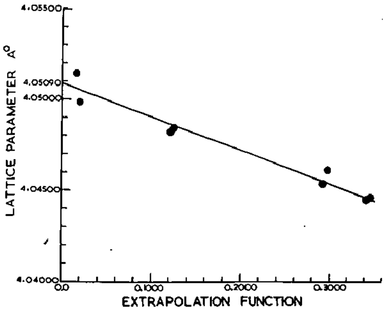

FIG.4.07 ALUMINIUM AT 30°C

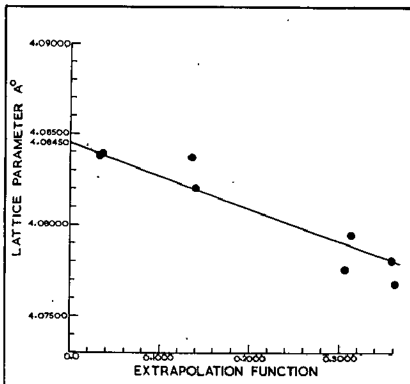

FIG.4.08 ALUMINIUM AT 271° C

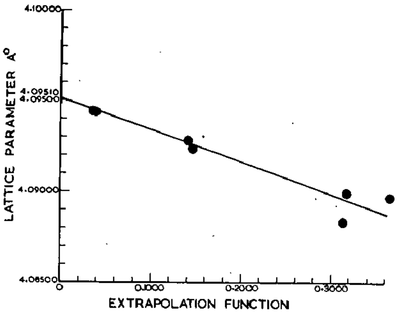

FIG.4.09 ALUMINIUM AT 382°C

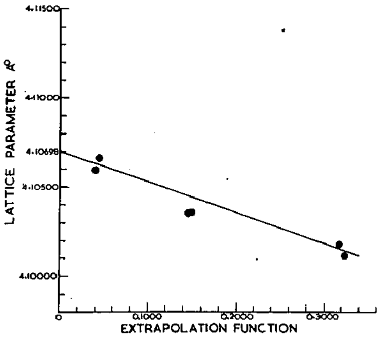

FIG.4.10 ALUMINIUM AT 494° C

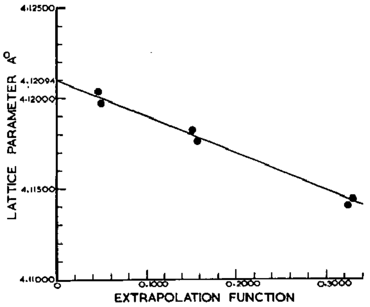

FIG.4.11 ALUMINIUM AT 606.5°C

TABLE - 4.13

X-RAY DIFFRACTION RESULTS OF Al + 1.9 Wt.% Cu.AT 30°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value$ A.U.
331	α_1	56.12	4.04427	0.3460	4.04938
331	α_2	56.33	4.04453	0.3410	
420	α_1	58.39	4.04537	0.2960	
420	α_2	58.62	4.04501	0.2920	
422	α_1	68.77	4.04853	0.1250	
422	α_2	69.17	4.04774	0.1200	
511	α_1	81.30	4.04898	0.0200	
511	α_2	82.29	4.04914	0.0150	

TABLE 4.14

X-RAY DIFFRACTION RESULTS OF Al + 1.9% Cu AT 271° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value . A.U.
331	α_1	55.61	4.06868	0.3580	
331	α_2	55.82	4.06873	0.3530	4.07580
420	α_1	57.83	4.06898	0.3080	
420	α_2	58.09	4.06863	0.3020	
422	α_1	67.91	4.07239	0.1360	
422	α_2	68.26	4.07278	0.1310	
511	α_1	79.19	4.07491	0.0310	
511		DIFFUSED			

X-RAY DIFFRACTION PHOTOGRAPHS OF Al+1.9 wt.% Cu AT DIFFERENT TEMPERATURES 20 mA, 40 KVP, 6 hrs.

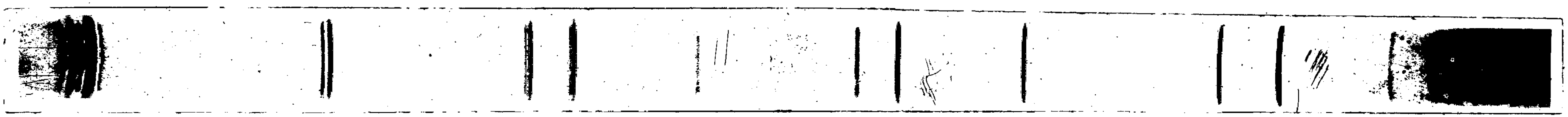

PLATE 4.11

TEMP. 30°c

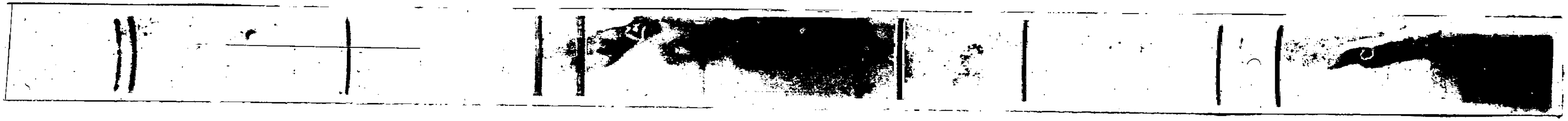

PLATE 4.12

TEMP. 271°c

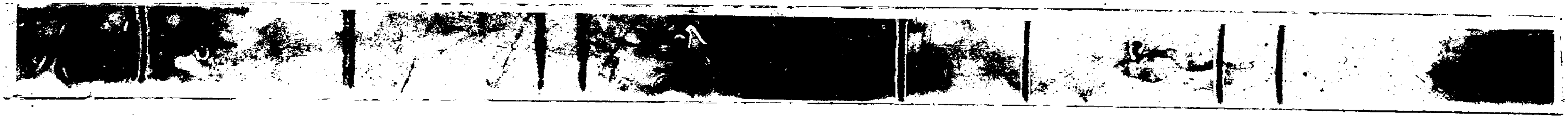

PLATE 4.13

TEMP. 382°c

PLATE 4.14

TEMP. 494°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al+1.9 wt.% Cu AT DIFFERENT TEMPERATURES

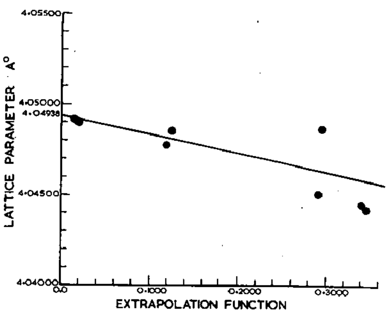

FIG.4.12 ALUMINIUM + 1.9 WEIGHT % COPPER AT 30° C

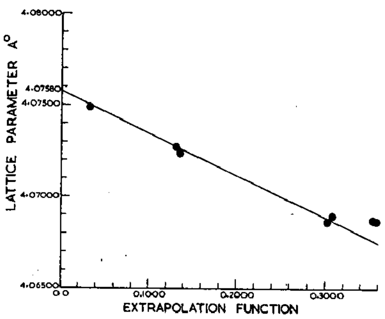

FIG.4.13 ALUMINIUM + 1.9 WEIGHT % COPPER AT 271° C

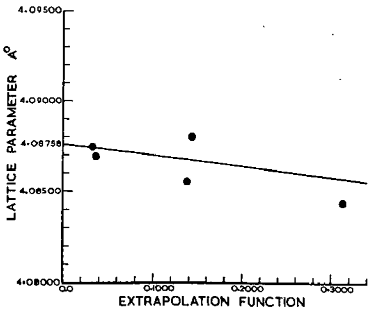

FIG.4.14 ALUMINIUM + 1.9 WEIGHT % COPPER AT 382° C

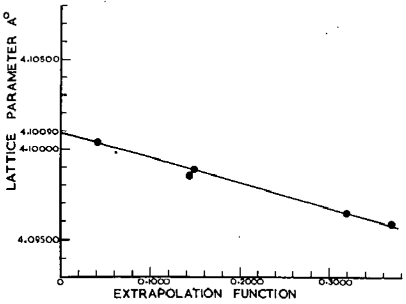

FIG.4.15 ALUMINIUM + 1.9 WEIGHT % COPPER AT 494° C

TABLE - 4.17

X-RAY DIFFRACTION RESULTS OF Al + 4.35 % wt. Cu ALLOY AT 25°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	56.27	4.04153	0.342	4.04760
420	α	58.51	4.04300	0.294	
422	α_1	68.88	4.04549	0.123	
422	α_2	69.28	4.04481	0.119	
511	α_1	81.45	4.04742	0.019	
511	α_2	82.43	4.04763	0.011	

TABLE - 4.18

X-RAY DIFFRACTION RESULT OF Al + 4.35% Wt. Cu alloy at 271°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α	57.94	4.06823	0.306	
422	α_1	67.98	4.07067	0.135	4.07290
422	α_2	68.37	4.06970	0.130	
511	α_1	79.36	4.07252	0.029	
511	α_2	80.17	4.07231	0.025	

TABLE - 4.17

X-RAY DIFFRACTION RESULTS OF Al + 4.35 % wt. Cu ALLOY AT 25°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	56.27	4.04153	0.342	4.04760
420	α	58.51	4.04300	0.294	
422	α_1	68.88	4.04549	0.123	
422	α_2	69.28	4.04481	0.119	
511	α_1	81.45	4.04742	0.019	
511	α_2	82.43	4.04763	0.011	

TABLE - 4.18

X-RAY DIFFRACTION RESULT OF Al + 4.35% Wt. Cu alloy at 271°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α	57.94	4.06823	0.306	
422	α_1	67.98	4.07067	0.135	4.07290
422	α_2	68.37	4.06970	0.130	
511	α_1	79.36	4.07252	0.029	
511	α_2	80.17	4.07231	0.025	

TABLE - 4.19

X-RAY DIFFRACTION RESULTS OF Al + 4.35 % Wt.ALLOY AT 382°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.55	4.07535	0.359	
420	α	57.76	4.07641	0.310	
422	α_1	67.64	4.08042	0.140	4.08157
422	α_2	68.08	4.07807	0.134	
511	α_1	78.76	4.08078	0.034	
511	α_2	79.46	4.08130	0.029	

TABLE - 4.20

X-RAY DIFFRACTION RESULTS OF Al + 4.35% Wt.Cu. ALLOY AT 437.5°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α	57.53	4.08661	0.315	4.09012
422	α_1	67.39	4.08802	0.143	
511	α_1	78.14	4.08977	0.037	
511	α_2	78.85	4.08962	0.033	

TABLE - 4.21

X-RAY DIFFRACTION RESULTS OF Al. + 4.35% Wt.Cu. ALLOY AT 494° C.

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.30	4.08721	0.365	
420	α	57.41	4.09188	0.317	4.09296
422	α_1	67.32	4.08978	0.144	
422	α_2	67.58	4.09200	0.142	
511	α_1	77.97	4.09240	0.038	
511	α_2	78.64	4.09260	0.034	

TABLE - 4.21

X-RAY DIFFRACTION RESULTS OF Al + 4.35% Wt.Cu. ALLOY AT 494° C.

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.30	4.08721	0.365	
420	α	57.41	4.09188	0.317	4.09296
422	α_1	67.32	4.08978	0.144	
422	α_2	67.58	4.09200	0.142	
511	α_1	77.97	4.09240	0.038	
511	α_2	78.64	4.09260	0.034	

X- RAY DIFFRACTION PHOTOGRAPHS OF Al + 4.35 wt.% Cu AT DIFFERENT TEMPERATURES 20 mA, 40 KVP, 6 hrs.


PLATE 4.15

TEMP. 25°c


PLATE 4.16

TEMP. 271°c

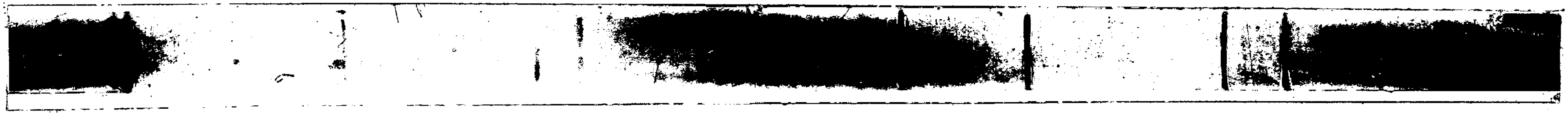
PLATE 4.17

TEMP. 382°c


PLATE 4.18

TEMP. 437.5°c

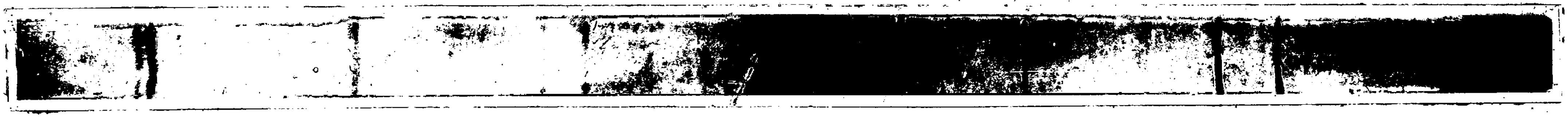
PLATE 4.19

TEMP. 494°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al + 4.35 wt.% Cu AT DIFFERENT TEMPERATURES

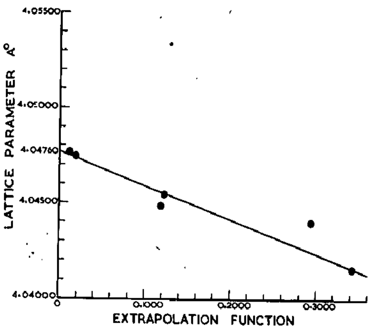

FIG.4.16 ALUMINIUM + 4.35 WEIGHT % COPPER AT 25° C

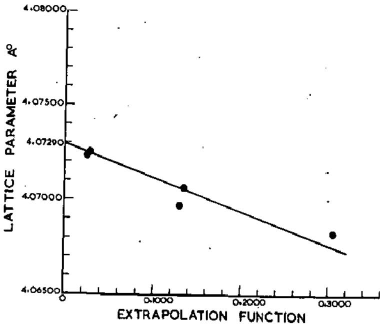

FIG.4.17 ALUMINIUM + 4.35 WEIGHT % COPPER AT 271°C

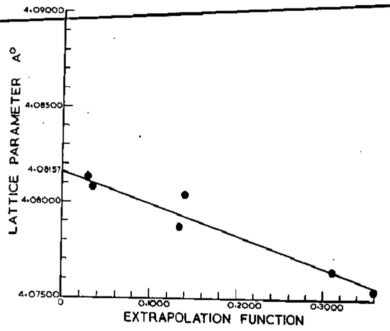

FIG.4.18 ALUMINIUM + 4.35 WEIGHT % COPPER AT 382°C

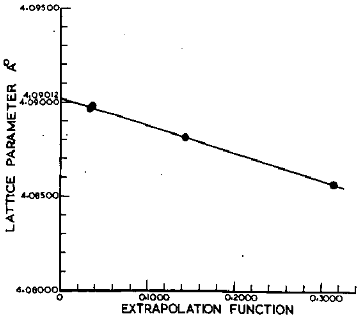

FIG.4.19 ALUMINIUM + 4.35 WEIGHT % COPPER AT 437.5°C

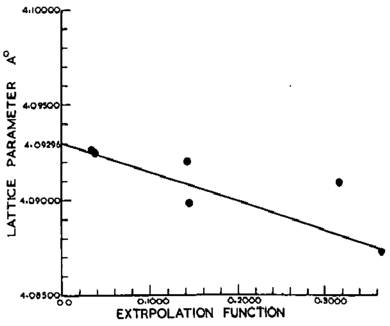

FIG.4.20 ALUMINIUM + 4.35 WEIGHT % COPPER AT 494° C

TABLE -4.22

X-RAY DIFFRACTION RESULTS OF Al + 5% Wt.Cu ALLOY AT 25° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation functio	Extrapolated value A.U.
331	α	56.26	4.04139	0.343	
420	α	58.51	4.04296	0.294	4.04712
422	α_1	68.90	4.04461	0.123	
422	α_2	69.30	4.04393	0.119	
511	α_1	81.50	4.04680	0.019	
511	α_2	82.52	4.04683	0.015	

TABLE - 4.23

X-RAY DIFFRACTION RESULTS OF Al + 5.% Wt.Cu ALLOY AT 161°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.77	4.06481	0.332	4.06698
420	α	58.01	4.06481	0.304	
422	α_1	68.11	4.06671	0.134	
422	α_2	68.51	4.06560	0.129	
511	α_1	79.76	4.06850	0.027	
511	α_2	80.77	4.06509	0.022	

TABLE - 4.24

X-RAY DIFFRACTION RESULTS OF Al +5% Wt.Cu ALLOY AT 382° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.51	4.07697	0.360	4.08344
420	α	57.75	4.07681	0.310	
511	α_1	78.65	4.08281	0.034	
511	α_2	79.35	4.08276	0.029	

TABLE - 4.25

X-RAY DIFFRACTION RESULTS OF Al + 5% Wt.Cu ALLOY AT 494°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
422	α_1	67.49	4.08508	0.141	4.09060
	α_2	67.85	4.08459	0.137	
511	α_1	78.22	4.08853	0.037	
511	α_2	78.83	4.08988	0.033	

TABLE - 4.24

X-RAY DIFFRACTION RESULTS OF Al +5% Wt.Cu.ALLOY AT 382° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.51	4.07697	0.360	4.08344
420	α	57.75	4.07681	0.310	
511	$α_1$	78.65	4.08281	0.034	
511	$α_2$	79.35	4.08276	0.029	

TABLE - 4.25

X-RAY DIFFRACTION RESULTS OF Al + 5% Wt.Cu ALLOY AT 494°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α	57.75	4.07681	0.310	
422	$α_1$	67.49	4.08508	0.141	4.09060
422	$α_2$	67.85	4.08459	0.137	
511	$α_1$	78.22	4.08853	0.037	
511	$α_2$	78.83	4.08988	0.033	

X-RAY DIFFRACTION PHOTOGRAPHS OF Al + 5 wt.% Cu AT DIFFERENT TEMPERATURES 20 mA, 40 KVP, 6 hrs

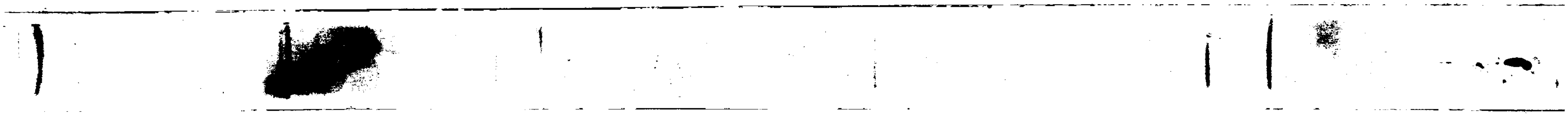

PLATE 4.20 TEMP. 25°c

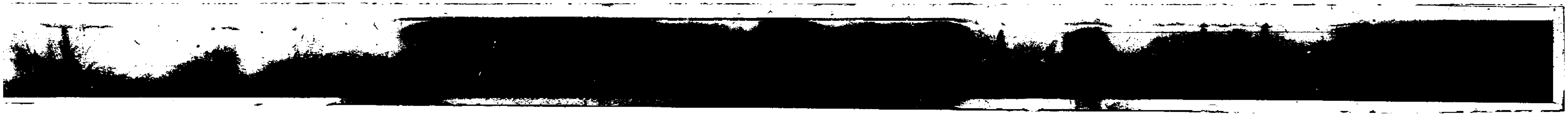

PLATE 4.21 TEMP. 161°c

PLATE 4.22 TEMP. 382°c

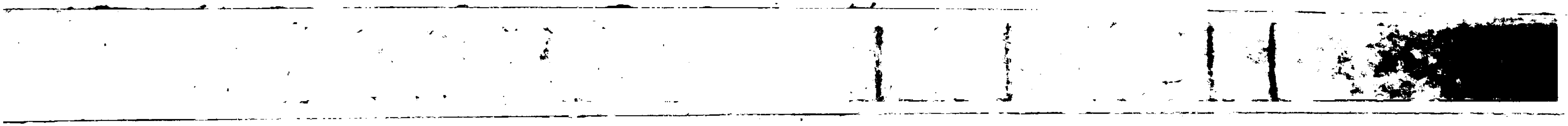

PLATE 4.23 TEMP. 494°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al +5.0 wt.% Cu AT DIFFERENT TEMPERATURES

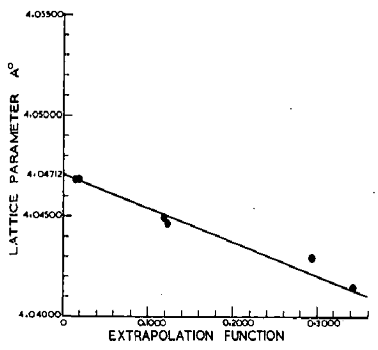

FIG.4.21 ALUMINIUM + 5 WEIGHT % COPPER AT 25° C

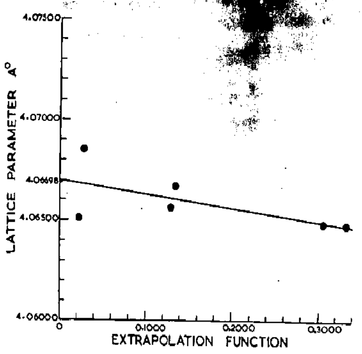

FIG.4.22 ALUMINIUM + 5 WEIGHT % COPPER AT 161° C

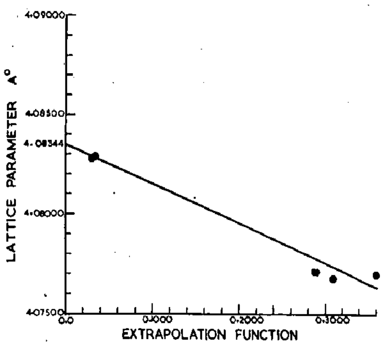

FIG.4.23 ALUMINIUM + 5 WEIGHT % COPPER AT 382°C

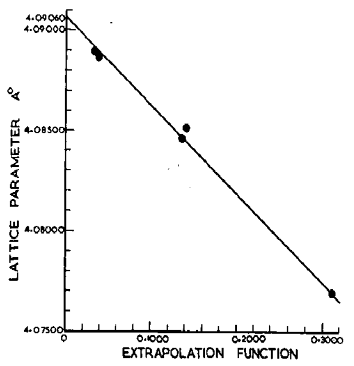

FIG. 4.24 ALUMINIUM + 5 WEIGHT % COPPER AT 494° C

TABLE - 4.26

X-RAY DIFFRACTION RESULTS OF Al + 9.92% Wt.Cu ALLOY AT 25° C.

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	58.29	4.04768	0.2980	
420	α_2	58.77	4.03790	0.2880	
422	α_1	69.08	4.04005	0.1220	4.04668
422	α_2	69.29	4.04608	0.1190	
511	α_1	81.54	4.04606	0.0185	
511	α_2	82.67	4.04659	0.0140	

TABLE - 4.27

X-RAY DIFFRACTION RESULTS OF Al + 9.92% Wt.Cu AT 271°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	57.84	4.06684	0.3080	
422	α_1	67.91	4.07500	0.1360	4.07578
422	α_2	68.27	4.07299	0.1310	
511	α_1	79.15	4.07533	0.0310	
511	α_2	79.90	4.07548	0.0270	

TABLE - 4.28

X-RAY DIFFRACTION RESULTS OF Al + 9.92% Wt.Cu AT 382° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α	57.25	4.07170	0.3200	
422	α_1	67.45	4.08300	0.1420	
422	α_2	67.88	4.08347	0.1365	
511	α_1	78.40	4.08567	0.0350	4.08664
511	α_2	79.11	4.08567	0.0310	

TABLE - 4.29

X-RAY DIFFRACTION RESULTS OF Al + 9.92% Wt.Cu AT 494° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	57.45	4.08638	0.3160	
420	α_2	57.54	4.09264	0.3140	
422	α_1	67.15	4.09468	0.1465	4.09630
422	α_2	67.49	4.09473	0.1410	
511	α_1	77.85	4.09580	0.0380	
511	α_2	78.41	4.09590	0.0350	

X-RAY DIFFRACTION PHOTOGRAPHS OF Al+9.92 wt.% Cu AT DIFFERENT TEMPERATURES 20 mA, 40 KVP, 6 hrs

PLATE 4.24 TEMP. 25°c

PLATE 4.25 TEMP. 271°c

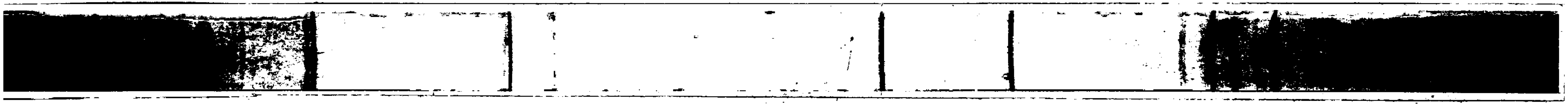

PLATE 4.26 TEMP. 382°c

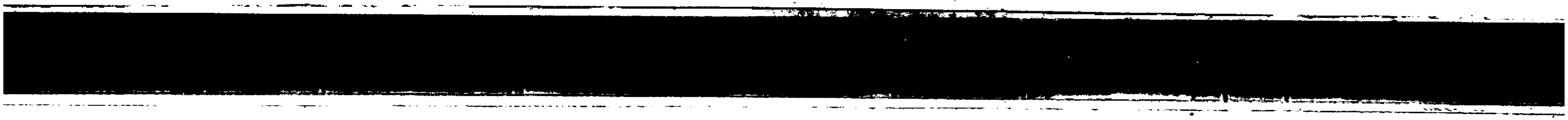

PLATE 4.27 TEMP. 494°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al+9.92 wt.% Cu AT DIFFERENT TEMPERATURES

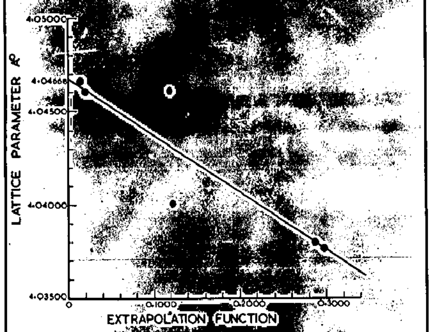

FIG.4.25 ALUMINIUM + 9.92WEIGHT% COPPER AT 25°C

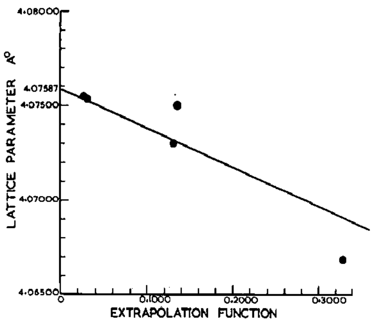

FIG.4.26 ALUMINIUM + 9.92 WEIGHT% COPPER AT 271°C

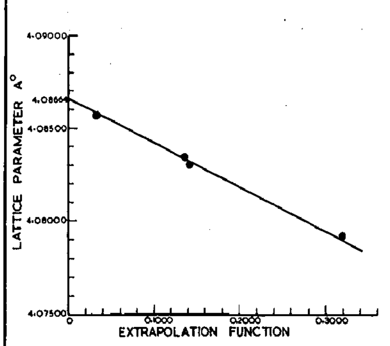

FIG.4.27 ALUMINIUM + 9.92WEIGHT% COPPER AT 382°C

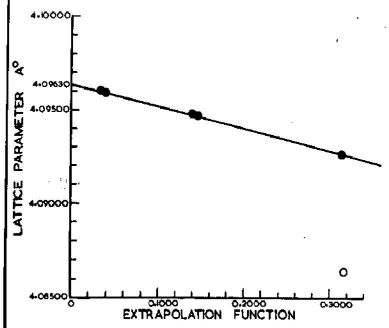

FIG.4.28 ALUMINIUM + 9.92WEIGHT% COPPER AT 494°C

TABLE - 4.30

X-RAY DIFFRACTION RESULTS OF Al + 12.18% Wt.Cu. AT 25°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	56.18	4.04250	0.3770	
420	α	58.50	4.04622	0.2940	4.04630
422	α_1	68.89	4.04520	0.1200	
422	α_2	69.27	4.04501	0.1190	
511	α_1	81.58	4.04607	0.0180	
511	α_2	82.62	4.04592	0.0140	

TABLE - 4.31

X-RAY DIFFRACTION RESULTS OF Al + 12.18% Wt.Cu. AT 271°C

331	α	55.63	4.07134	0.3575	
420	α	57.87	4.07153	0.3070	4.07454
422	α_1	67.93	4.07195	0.1360	
422	α_2	68.28	4.07229	0.1310	
511	α_1	79.23	4.07418	0.0310	
511	α_2	80.00	4.07424	0.0260	

TABLE - 4.32

X-RAY DIFFRACTION RESULTS OF Al + 12.18% Wt.Cu., AT 382° C.

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.56	4.07500	0.3630	
420	α	57.82	4.07341	0.3085	
422	α_1	67.61	4.08116	0.1400	4.08520
422	α_2	67.96	4.08136	0.1355	
511	α_1	78.50	4.08437	0.0350	
511	α_2	79.22	4.08447	0.0310	

TABLE - 4.33

X-RAY DIFFRACTION RESULTS OF Al + 12.18% Wt.Cu.

AT 494° C.

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.32	4.08443	0.3650	
420	α	57.50	4.08772	0.3150	
422	α_1	67.31	4.09001	0.1440	4.09462
422	α_2	67.63	4.09106	0.1400	
511	α_1	77.91	4.09320	0.0390	
511	α_2	78.55	4.09398	0.0345	

X-RAY DIFFRACTION PHOTOGRAPHS OF Al+12.18 wt% Cu AT DIFFERENT TEMPERATURES 20mA,40KVP,6hrs


PLATE 4.28 TEMP. 25°c


PLATE 4.29 TEMP. 271°c


PLATE 4.30 TEMP. 382°c

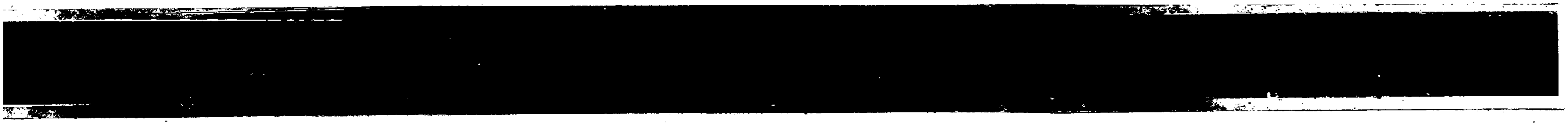
PLATE 4.31 TEMP. 494°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al + 12.18 wt.% Cu AT DIFFERENT TEMPERATURES

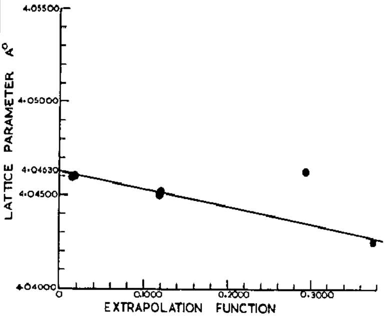

FIG.4.29 ALUMINIUM + 12.18 WEIGHT %
COPPER AT 25°C

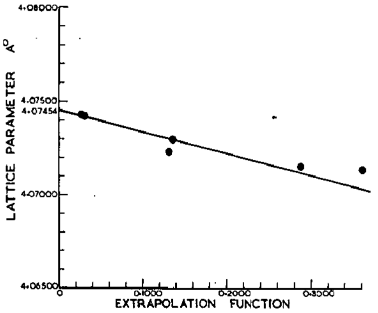

FIG.4.30 ALUMINIUM + 12.18 WEIGHT %
COPPER AT 271°C

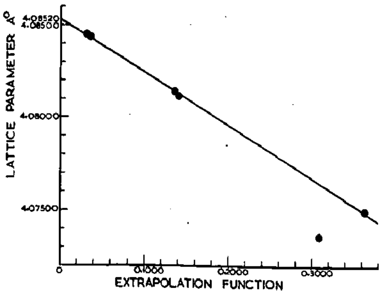

FIG.4.31 ALUMINIUM + 12.18 WEIGHT %
COPPER AT 382° C

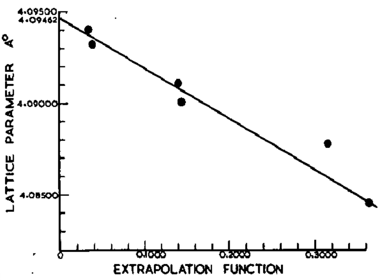

FIG.4.32 ALUMINIUM + 12.18 WEIGHT %
COPPER AT 494°C

TABLE - 4.34

X-RAY DIFFRACTION RESULTS OF Al + 13.79% Wt.Cu AT 30° C.

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	58.42	4.04354	0.2960	
420	α_2	58.67	4.04314	0.2900	
422	α_1	68.85	4.04618	0.1240	4.04860
422	α_2	69.22	4.04615	0.1200	
511	α_1	81.34	4.04856	0.0200	
511	α_2	82.36	4.04805	0.0150	

TABLE - 4.35

X-RAY DIFFRACTION RESULTS OF Al + 13.79% Wt.Cu AT 271° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.69	4.06886	0.3560	
420	α	57.93	4.06854	0.3060	
422	α_1	67.91	4.07220	0.1360	4.07598
422	α_2	68.25	4.07298	0.1315	
511	α_1	79.11	4.07574	0.0310	
511	α_2	79.99	4.07476	0.0260	

TABLE - 4.36

X-RAY DIFFRACTION RESULTS OF Al + 13.79% Wt.Cu AT 382° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	57.67	4.07645	0.3120	4.08440
420	α_2	57.90	4.07645	0.3060	
422	α_1	67.64	4.08038	0.1395	
422	α_2	67.97	4.08116	0.1360	
511	α_1	78.82	4.08360	0.0330	
511	α_2	79.34	4.08339	0.0300	

TABLE - 4.37

X-RAY DIFFRACTION RESULTS OF Al + 13.79% Wt.Cu AT 549° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	56.97	4.08544	0.3270	4.09572
422	α_1	67.24	4.09204	0.1450	
422	α_2	67.56	4.09248	0.1405	
511	α_1	77.71	4.09351	0.0400	
511	α_2	78.47	4.09465	0.0350	

X-RAY DIFFRACTION PHOTOGRAPHS OF Al+13.79 wt.% Cu AT DIFFERENT TEMPERATURES 20mA 40KVP 6hrs

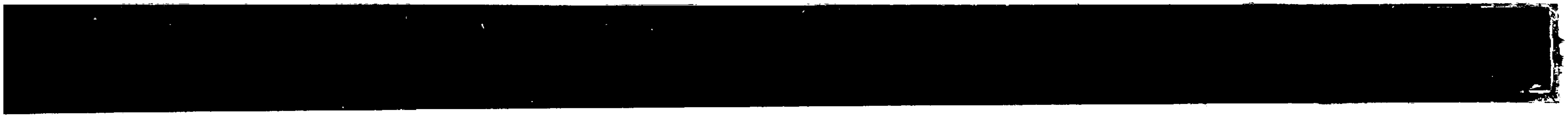

PLATE 4.32 TEMP. 30°c

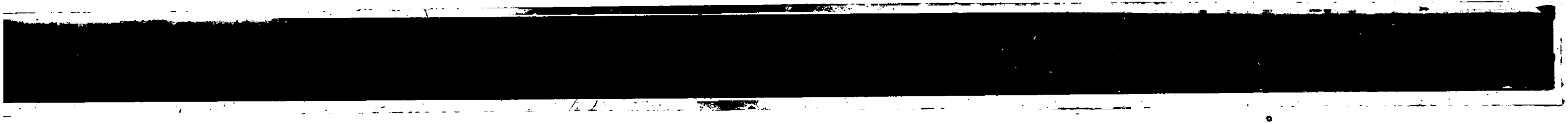

PLATE 4.33 TEMP. 271°c

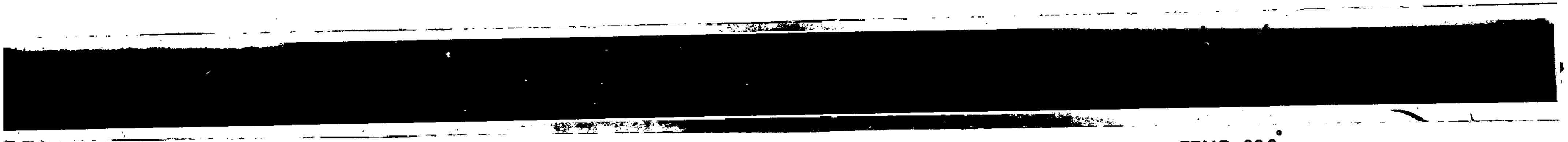

PLATE 4.34 TEMP. 382°c

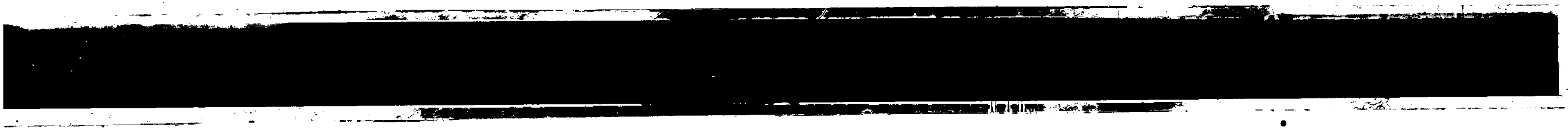

PLATE 4.35 TEMP. 494°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al + 13.79 wt.% Cu AT DIFFERENT TEMPERATURE

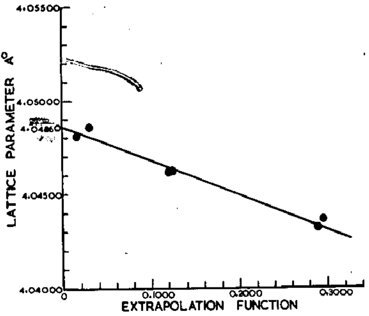

FIG. 4.33 ALUMINIUM + 13.79 WEIGHT % COPPER AT 30° C

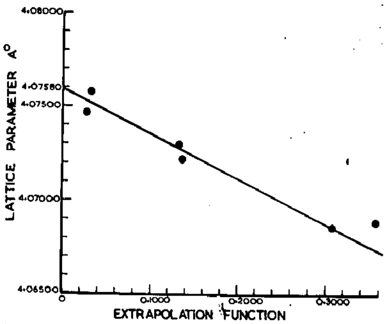

FIG. 4.34 ALUMINIUM + 13.79 WEIGHT % COPPER AT 271° C

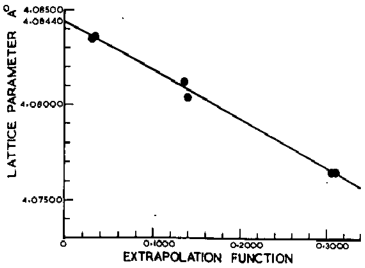

FIG. 4.35 ALUMINIUM + 13.79 WEIGHT % COPPER AT 382° C

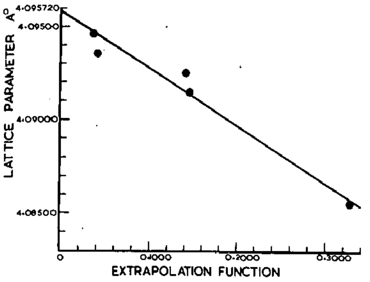

FIG. 4.36 ALUMINIUM + 13.79 WEIGHT % COPPER AT 549°C

TABLE 4.38

X-RAY DIFFRACTION RESULTS OF Al + 7.5% Ag AT 34° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α_1	58.46	4.04036	0.2950	
420	α_2	58.67	4.04206	0.2900	
422	α_1	68.82	4.04775	0.1250	4.05110
422	α_2	69.24	4.04594	0.1200	
511	α_1	81.22	4.05101	0.0210	
511	α_2	82.36	4.05090	0.0145	

TABLE - 4.39

X-RAY DIFFRACTION RESULTS OF Al + 7.5% Ag AT 161°C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	56.28	4.06131	0.306	
420	α_1	57.91	4.06599	0.302	
420	α_2	58.14	4.06608	0.301	
422	α_1	68.04	4.06891	0.135	4.07342
422	α_2	68.42	4.06808	0.130	
511	α_1	79.44	4.07138	0.029	
511	α_2	80.03	4.07387	0.026	

TABLE - 4.40

X-RAY DIFFRACTION RESULTS OF Al + 7.5% Ag at 271° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α_1	55.45	4.07727	0.3610	4.08254
420	α_1	57.74	4.07390	0.3095	
420	α_2	57.91	4.07592	0.3060	
422	α_1	67.66	4.07994	0.1395	
422	α_2	68.03	4.07905	0.1350	
511	α_1	78.66	4.08223	0.0290	
511	α_2	79.44	4.08146	0.0240	

TABLE - 4.41

X-RAY DIFFRACTION RESULTS OF Al + 7.5% Ag AT 382° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
222	α	40.91	4.07801	0.5370	4.09772
331	α	55.19	4.09344	0.3670	
420	α	57.34	4.09524	0.3190	
422	α	67.13	4.09630	0.1460	
511	α_1	77.65	4.09720	0.0405	
511	α_2	78.30	4.09746	0.0360	

X-RAY DIFFRACTION PHOTOGRAPHS OF Al + 7.5wt % Ag AT DIFFEREN TEMPERATURES 12mA, 40 KVP, 20hrs.

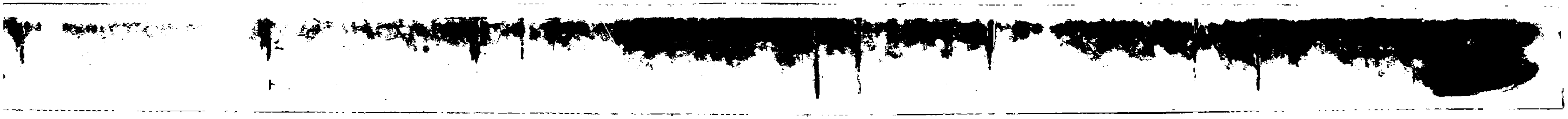

PLATE 4.36 TEMP. 34°c

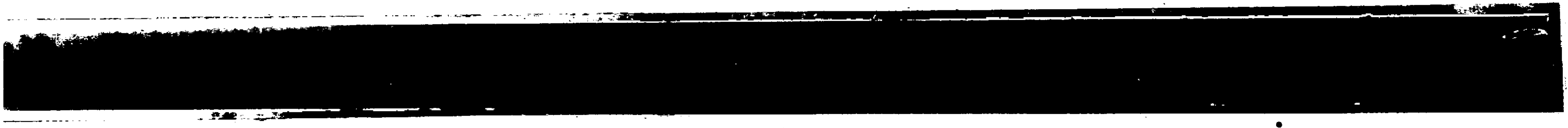

PLATE 4.37 TEMP. 161°c

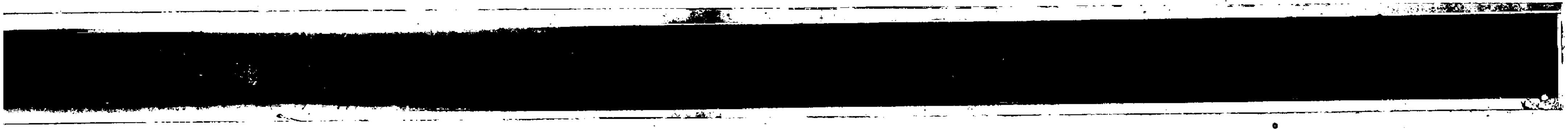

PLATE 4.38 TEMP. 271°c

PLATE 4.39 TEMP. 382°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al+7.5 wt.% Ag AT DIFFERENT TEMPERATURES

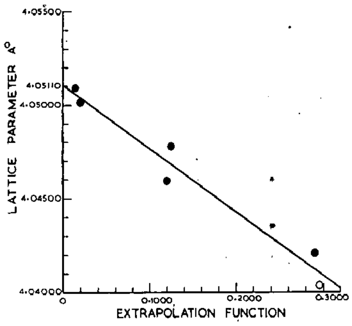

FIG. 4.37 ALUMINIUM +7.5 WEIGHT % SILVER AT 34° C

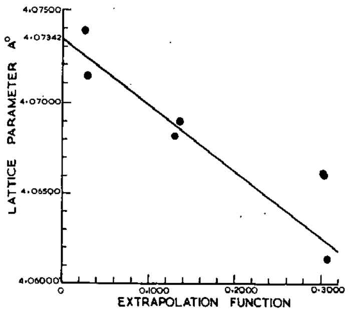

FIG. 4.38 ALUMINIUM +7.5 WEIGHT % SILVER AT 161° C

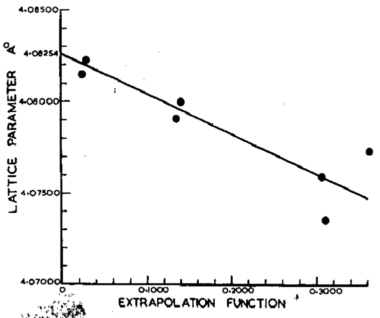

FIG. 4.39 ALUMINIUM +7.5 WEIGHT % SILVER AT 271°C

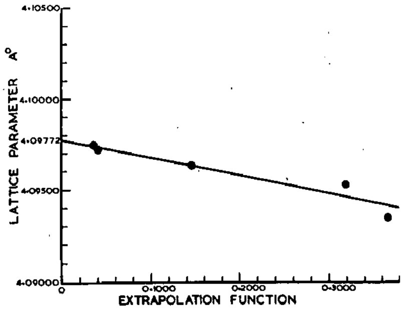

FIG. 4.40 ALUMINIUM+7.5 WEIGHT % SILVER AT 382° C

TABLE - 4.42

X-RAY DIFRRACTION RESULTS OF Al + 19.9% Ag AT 30° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
222	α	41.44	4.03547	0.8170	
331	α	56.23	4.04281	0.3440	
420	α	58.48	4.04480	0.2940	
422	α	68.82	4.04643	0,1250	4.04955
511	α_1	81.31	4.04888	0.0200	
511	α_2	82.25	4.04940	0.0155	

TABLE - 4.43

X-RAY DIFFRACTION RESULTS OF Al + 19.9% Ag at 271° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
220	α	32.54	4.05440	1.2860	
311	α	39.06	4.05842	0.9230	
222	α	41.03	4.06841	0.8310	4.07632
331	α	55.63	4.07134	0.3570	
511	α_1	79.10	4.07590	0.0310	
511	α_2	79.86	4.07611	0.0270	

TABLE - 4.44

X-RAY DIFFRACTION RESULTS OF Al + 19.9% Ag at 382° C

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
420	α	57.53	4.08656	0.3140	4.09045
422	α_1	67.38	4.08831	0.1430	
422	α_2	67.71	4.08836	0.1390	
511	α_1	78.14	4.08980	0.0365	
511	α_2	78.82	4.08998	0.0330	

TABLE - 4.45

X-RAY DIFFRACTION RESULTS OF Al + 19.9% Ag at 494° C.

Indices	Wave length	θ	Lattice parameter A.U.	Extrapolation function	Extrapolated value A.U.
331	α	55.06	4.09968	0.3710	4.10392
420	α_1	57.23	4.10024	0.3210	
422	α_2	67.07	4.10256	0.1470	
511	α_1	77.27	4.10344	0.0430	
511	α_2	77.90	4.10350	0.0380	

X-RAY DIFFRACTION PHOTOGRAPHS OF Al+ 19.9 wt. Ag AT DIFFERENT TEMPERATURES 12 mA, 40 KVP, 20 hrs.

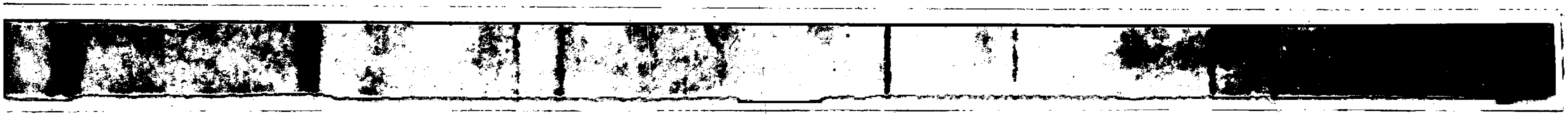

PLATE 4.40 TEMP. 30°c

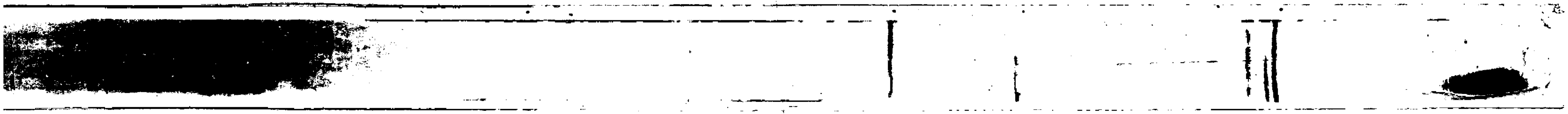

PLATE 4.41 TEMP. 271°c

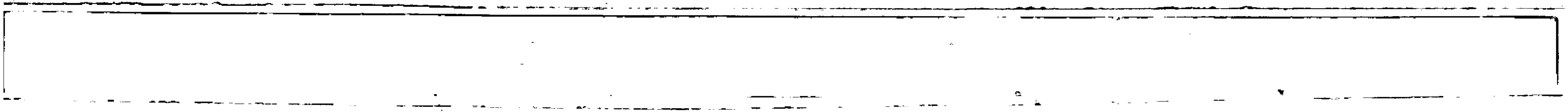

PLATE 4.42 TEMP. 382°c

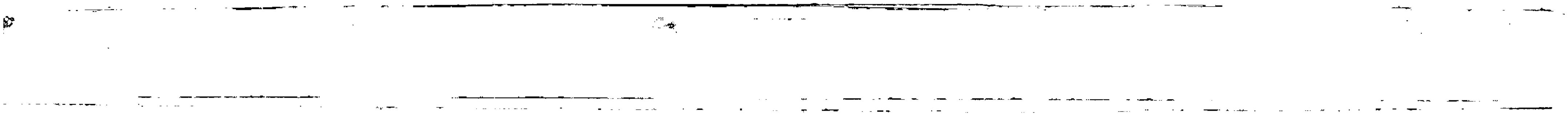

PLATE 4.43 TEMP. 494°c

LATTICE PARAMETER VS. EXTRAPOLATION FUNCTION FOR Al + 19.9 wt.% Ag AT DIFFERENT TEMPERATURES

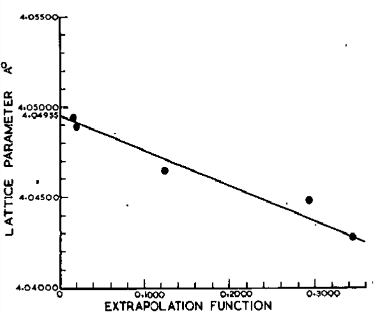

FIG.4.41 ALUMINIUM + 19.9 WEIGHT % SILVER AT 30°C

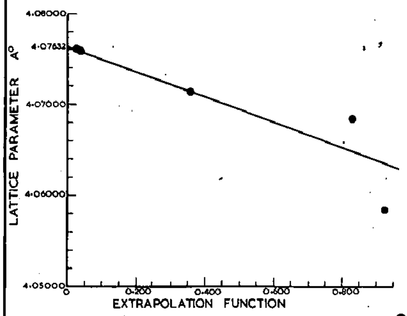

FIG.4.42 ALUMINIUM + 19.9 WEIGHT % SILVER AT 271°C

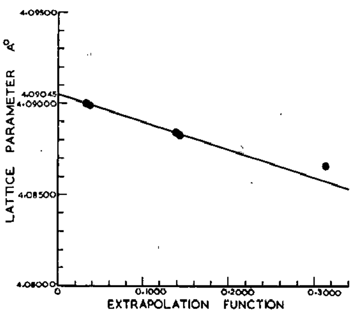

FIG 4.43 ALUMINIUM + 19.9 WEIGHT % SILVER AT 382° C

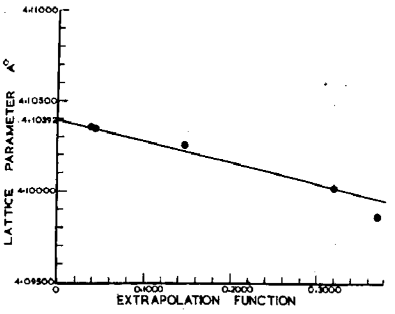

FIG.4.44 ALUMINIUM + 19.9 WEIGHT % SILVER AT 494°C

4.4 DISCUSSION OF THE RESULTS.

4.4.1 Pure Aluminium :

The salient features of the results are as follows :

i) There is regular increase in the lattice spacing with temperature. The relation appears to be nearly linear (Fig. 4.45)

ii) The equation

$$a = 4.04711 + 0.140783 \times 10^{-3} T - 0.339197 \times 10^{-7} T^2$$

as determined by computer, represents the relation between lattice parameter (a) and temperature (T^oC)

iii) The values of lattice spacing at various temperatures as determined from equation are in close agreement with the experimental values (Fig. 4.45).

iv) With increase in temperature the solid density as determined by X-ray, goes on decreasing gradually for lower temperatures, beyond which it decreases at an appreciable rate (Fig. 4.45).

v) There is a slight decrease in coefficient of thermal expansion ($\frac{1}{a_o} \frac{da}{dt}$) with increase in temperature upto 548°C (Fig. 4.45).

Lattice spacings of aluminium at elevated temperatures were measured by several workers. In the high temperature range the measurement of Ellwood and Silcock (8) and Wilson (7,140) are in agreement. The results of Ener are in good agreement with the measurement of Wilson (71,140) below

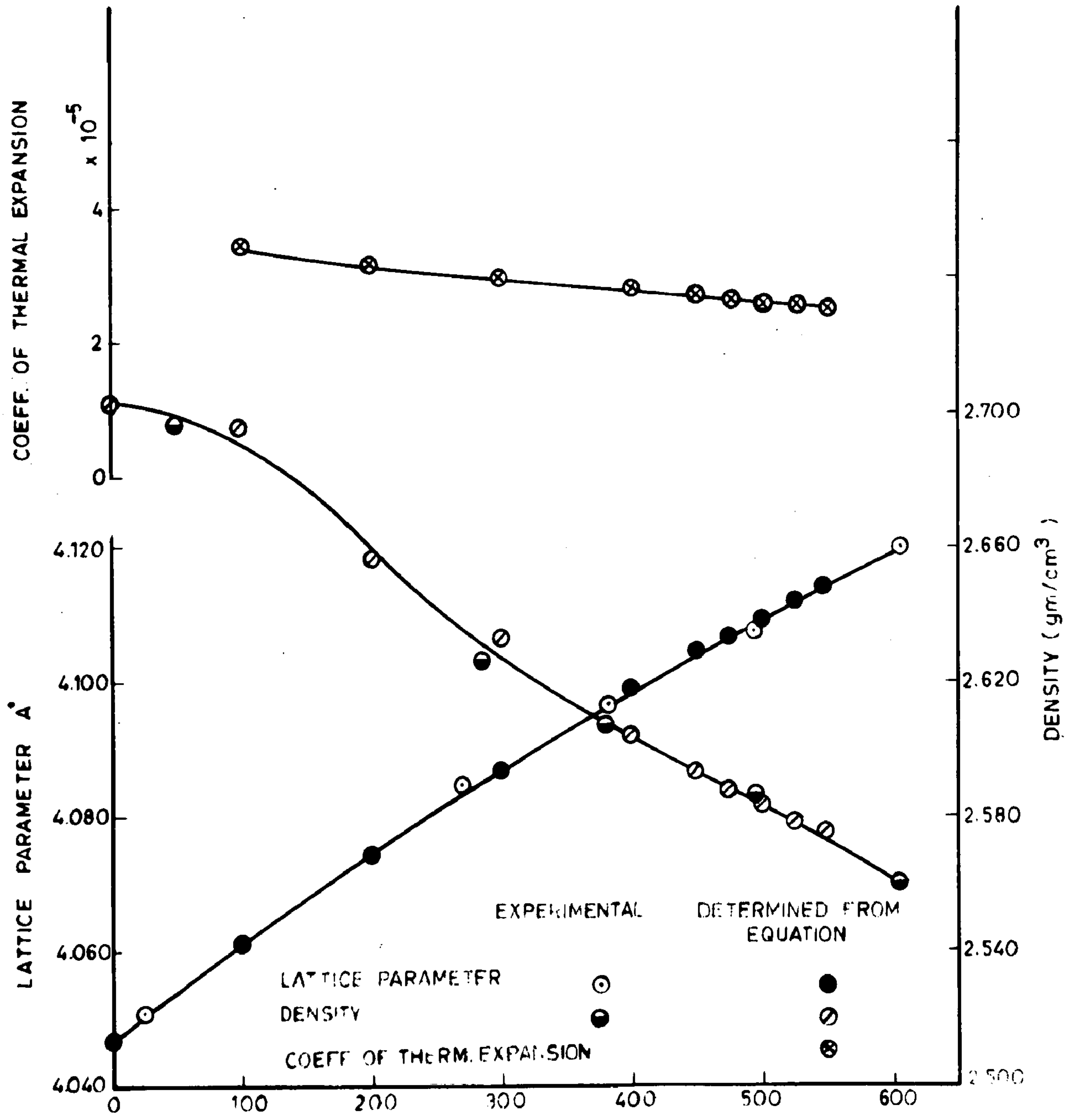

FIG.4.45_ LATTICE PARAMETER, THERMAL EXPANSION AND DENSITY OF PURE ALUMINIUM AT DIFFERENT TEMPERATURES

TABLE 4.46

COEFFICIENT OF THERMAL EXPANSION AND DENSITY OF ALUMINIUM AT DIFFERENT TEMPERATURES.

S.No.	Temperature 0^{o} C	Lattice parameter A^{o}	Coefficient of thermal expansion $\frac{1}{a_{o}} \frac{da}{dT}$ **	Density gm/cc.
1	0	4.0471		2.7030
2	100**	4.0608	$3.39x10^{-5}$	2.6758
3	200**	4.0739	$3.14x10^{-5}$	2.6559
4	271*	4.08450	-	2.6296
5	300**	4.0863	$2.97x10^{-5}$	2.6334
6	382*	4.0951	-	2.6092
7	400**	4.0980	$2.81x10^{-5}$	2.6036
8	450**	4.1036	$2.72 x10^{-5}$	2.5930
9	475**	4.1063	$2.68x10^{-5}$	2.5879
10	494*	4.1069	-	2.5866
11	500**	4.1090	$2.64 x10^{-5}$	2.5829
12	525**	4.1117	$2.59 x 10^{-5}$	2.5777
13	548**	4.1141	$2.56 x 10^{-5}$	2.5756
14	606.5*	4.1209	-	2.5604

* Experimentally determined values.

** Calculated values from equation determined by computer.

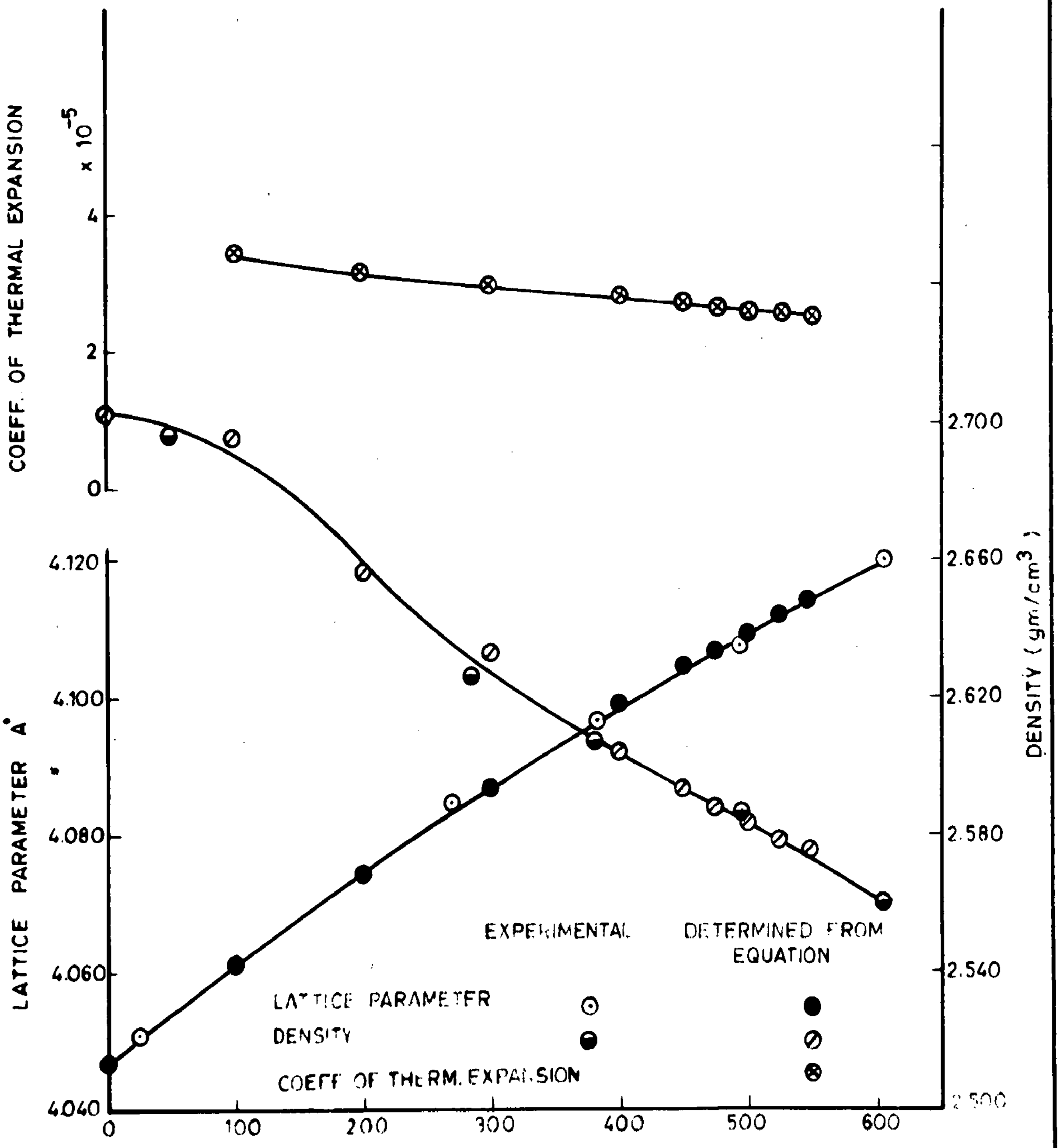

FIG.4.45_ LATTICE PARAMETER, THERMAL EXPANSION AND DENSITY OF PURE ALUMINIUM AT DIFFERENT TEMPERATURES

200° C, but at higher temperatures greater expansion was found. Lattice spacing as determined in the present investigation at 0° C was found out to be 4.0471 A°. The result is in agreement with the result of Wilson within ± 0.001 A°, who found out the lattice spacing at 0° C to be 4.0472 A°. Thus the values of lattice spacings agree with the results of Wilson (71,140) in lower temperature range. However, for higher temperatures the results of lattice spacing are slightly higher. Our results confirm results of Ener (42) for higher temperatures. The powders were not annealed by Wilson after filing. It may be emphasized that the specimen used in accurate determination of expansion coefficient should be annealed at a temperature at least as high as that of the highest exposure for a time relatively long compared with the exposure time to complete all recovery recrystallization and grain growth processes that can take place upto temperatures of highest exposure, because these may be accompanied by a change of lattice parameter. This may account for slightly lower values reported by Wilson (71) at elevated temperatures.

Densities calculated at different temperatures are given in ta-ble 4.46 and Fig. 4.45 . There is gradual decrease in density with temperature upto about 250° C. Beyond 250° C there is change in slope, and density decreases at an appreciable rate. Perryman has reported the density Dm = 2.6984 ± 0.0003 gm/cm^3 at 20° C while Taylor (142) has reported 2.6981 gm/cm^3. In the present investigation the density calculated at 20° C comes out to be 2.6980 gm/cm^3 . This is in agreement with the previously reported values at 20° C.

At about 260° C, change in slope of lattice parameter vs temperature curve, for aluminium is observed (Fig. 4.45). The anomalous behaviour is reported to be more apparent in macroscopic measurement of its expansion and for which no particular quantitative explanation has been made. Though the above anomaly could not be detected in lattice parameter curve in the present investigation, it is apparent in the density curve. It is understood that probably this may be due to the fact that vacancies start generating at a very fast rate beyond this temperature.

Coefficients of thermal expansion has been reported by Wilson and others at elevated temperatures. This has been calculated by finding the change in lattice parameter for a small change in temperature and is based on the assumption, that at temperatures above the Debye characteristic temperature, the variatio-n of its thermal expansion coefficient as a function of temperature will not depart much from linearity over intervals of 50° C - 100° C and is a good approximation over such intervals.

In the present investigation, second degree equation has been determined by the computer and then da/dt at a particular temperature has been found out by differentiating . This method of determination of slope should necessarily be more accurate. Nelson and Riley's extrapolation method has been used for precision determination of lattice parameter. The wire has been annealed properly at the temperature of highest exposure to complete all recovery recrystallization and grain growth processes. The position of the sample was not disturbed. And all the readings for different temperatures were taken with the same sample so that the particle size may not have any effect on the reading. Knife-edge calibration has been made by taking pure silver picture and thermocouple has also been calibrated for mean specimen temperature, It is claimed that coefficient of thermal expansion values so determined should necessarily be more accurate than the previously reported values.

4.4.2 Aluminium-copper alloys :

The results are given in tables 4.47 to 4.49 and figs 4.46 to 4.53. The salient features of the results are as follows :-

i) The lattice parameter increases with increase in temperature in all the cases.

ii) The lattice parameter vs temperature curve shifts vertically in the downward direction with increase in copper concentration from 0 to 1.9% and the values are lower for all temperatures. The slope of the curve decreases for lower temperatures and becomes parallel to the pure aluminium curve beyond about 400 ^{o}C (Fig.4.46)

iii) With increase in copper concentration from 1.9 to 4.3% same curve is obtained for lower temperatures. However, for higher temperatures the slope has decreased appreciably in case of 4.35% Cu alloy (Fig. 4.46).

iv) For 5% copper alloy the curve is common to 4.35% Cu alloy upto about 450° C beyond which it has appreciably decreased and lattice parameter becomes nearly constant with increase in temperature.

v) The values of lattice parameter as determined from equations (Table 4.47) and the experimental values agree well with each other.

vi) Density values of aluminium calculated at 525° C and 548° C and the slopes of the curves increase with increase in copper concentration.

vii) Coefficient of thermal expansion at 525° and 548° C calculated from equations remain almost unchanged upto about 4 % copper, beyond which it decreases at an appreciable rate. No detectable effect is observed due to increase in temperature from 525° to 548° C.

viii) Values of lattice spacings determined from equations, at different temperatures decrease linearly with increase in copper concentration, the slope remaining almost the same for all temperatures, and then becomes constant in the two phase region. The point of intersection of the inclined and horizontal portion gives the solubility limit at that temperature.

ix) The solubility limit line of copper in aluminium between 475° and 548° C is in agreement with the previous results (4.53). The maximum solid solubility of copper in aluminium at eutectic temperature is reported to be 5.7 wt % copper while at 260° C, it is between 0.1 to 0.2% . Thus the solubility decreases drastically with decrease in temperature. The relation between lattice spacing and composition has been determined by Axon and Hume-Rothery(9), Ellwood and Silcock(8), Sharan and Swaroop (143) Gilejaev and Trasova (49) and Dorn and others (45). Lattice spacing has been found to decrease linearly with increase in concentration.

Axon and Hume Rothery (33) have reported on the basis of density measurement, that there are probably no vacant lattice sites formed in the alloy containing less than 2 atomic % copper, though discrepency in the values of the alloy containing 2.17% copper was observed, but according to them the difference was due to enclosed pores. Sharan and Swaroop (143) have indicated the generation of vacancies at about 1.9 atomic percentage at an enormous rate. However, due to less accuracy in the back reflection method the result could not be confirmed. The above measurements were based on the assumption that vacancies are retained at ordinary temperature by rapid quenching. It can be seen that by quenching even at very rapid rate, the condition at elevated temperature might not be retained, and the vacancies generated, if any at high temperatures might not appear as apparent during quenching.

TABLE 4.47 LATTICE PARAMETER* OF ALUMINIUM COPPER AND ALUMINIUM SILVER ALLOYS IN EQUILIBRIUM AT DIFFERENT TEMPERATURES AND EQUATIONS** OF LATTICE PARAMETER (a) VS. TEMPERATURE (T)

Alloy	Lattice paramter (Å) at different temperature A.U.					Equation
Pure Al	4.05090 at 30°C	4.08450 at 271°C	4.09510 at 382°C	4.10698 at 494°C	4.12094 at 606.5°C	$a = \pm 0.33921118 \times 10^{-7} T^2 + 0.1407848 \times 10^{-3} T + 4.047116$ Max^m.error 0.1727×10^{-2} at 271°C.
Al +1.9%Cu	4.04938 at 30°C	4.07580 at 271°C	4.08758 at 382°C	4.10090 at 494°C	-	$a = -0.22127374 \times 10^{-7} T^2 + 0.9722046 \times 10^{-4} T + 4.0465823$ Max^m. error $= 0.105520 \times 10^{-2}$ at 271°C
Al +4.35%Cu	4.04760 at 25°C	4.07290 at 271°C	4.08157 at 382°C	4.09012 at 437.5°C	4.09216 at 494°C	$a = -.14617815 \times 10^{-7} T^2 \pm 0.10546260 \times 10^{-3} T + 4.0449977$ Max^m.error $= 0.3955 \times 10^{-3}$ at 271°C
Al +5%Cu	4.04712 at 25°C	4.06698 at 161°C	4.08344 at 382°C	4.09060 at 494°C	-	$a = -0.12764805 \times 10^{-6} T^2 + 0.15618447 \times 10^{-3} T + 4.0438590$ Max^m.error $= 0.12841 \times 10^{-2}$ at 161°C

continued........./

Alloy	Lattice parameter Å at different temperature					Equation
Al +9.92% Cu	4.04668 at 25°C	4.07578 at 271°C	4.08664 at 382°C	4.09630 at 494°C	-	$a = -0.5564743 \times 10^{-7} T^2 + 0.13464586 \times 10^{-3} T + 4.0433543$ Max^m.error $= 0.23500 \times 10^{-4}$ at 271°C
Al +12.18%Cu	4.04630 at 25°C	4.07454 at 271°C	4.08520 at 382°C	4.09462 at 494°C	-	$a = -0.527789 \times 10^{-7} T^2 + 0.13042285 \times 10^{-3} T + 4.0430741$ Max^m.error $= 0.240 \times 10^{-5}$ at 271° C
Al +13.79%Cu	4.04860 at 30°C	4.07598 at 271°C	4.08444 at 382°C	4.09572 at 549°C	-	$a = -0.75746218 \times 10^{-7} T^2 + 0.13423446 \times 10^{-3} T + 4.0447189$ Max^m.error $= 0.4465 \times 10^{-3}$ at 271°C
Al +7.5%Ag	4.05110 at 34°C	4.07342 at 161°C	4.08254 at 271°C	4.09772 at 382°C	-	$a = -0.99204616 \times 10^{-7} T^2 + 0.17071953 \times 10^{-3} T + 4.0461148$ Max^m.error $= 0.23909 \times 10^{-2}$ at 161°C
Al +19.9%Ag	4.04955 at 30°C	4.07632 at 271°C	4.09045 at 382°C	4.10392 at 494°C	-	$a = -0.22587709 \times 10^{-7} T^2 + 0.10579341 \times 10^{-3} T + 4.0463082$ Max^m.error $= 0.3170 \times 10^{-3}$ at 271°C

* Experimental data ** Determined by computer.

TABLE 4.48 LATTICE PARAMETERS OF ALUMINIUM COPPER & ALUMINIUM SILVER ALLOYS IN EQUILIBRIUM AT DIFFERENT TEMPERATURES.

Alloy	Lattice parameter * Å AT								
	100°C	200° C	300°C	400°C	450°C	475°C	500°C	525°C	548°C
Pure aluminium	4.0608	4.0739	4.0863	4.0980	4.1036	0.1063	4.1090	4.1117	4.1141
Al+1.9% Cu	4.0565	4.0669	4.0777	4.0890	4.0948	4.0978	4.1007	4.1038	4.1065
Al+4.35% Cu	4.0554	4.0655	4.0753	4.0848	4.0895	4.0918	4.0941	4.0966	4.1023
Al+5% Cu	4.0582	4.0700	4.0792	4.0859	4.0883	4.0892	4.0900	4.0907	4.0911
Al+9.92% Cu	4.0562	4.0680	4.0787	4.0882	4.0912	4.0947	4.0967	4.0987	4.1004
Al+12.18% Cu	4.0556	4.0686	4.0774	4.0868	4.0911	4.0931	4.0951	4.0970	4.1013
Al+13.79% Cu	4.0589	4.0685	4.0782	4.0863	4.0898	4.0914	4.0929	4.0943	4.0955
Al+7.5% Ag	4.0622	4.0763	4.0884	4.0985	4.1025	4.1048	4.1067.	4.1084	–
Al+19.9% Ag	4.0571	4.0684	4.8007	4.09284	4.0985	4.1016	4.1047	4.1081	–

* Calculated from equations.

TABLE - 4.49

COEFFICIENT OF THERMAL EXPANSION AND DENSITY OF ALUMINIUM-COPPER ALLOYS AT ELEVATED TEMPERATURES.

S.No.	% Cu by weight	Temperature 525° C		Temperature 548° C	
		$\frac{1}{a_T} \frac{da}{dT}$	Density gm/cc	$\frac{1}{a_T} \frac{da}{dT}$	Density gm/cc
1	0.00%	2.56×10^{-5}	2.578	2.52×10^{-5}	2.5760
2	1.91%	2.92×10^{-5}	2.6587	2.92×10^{-5}	2.6340
3	4.35 %	2.19×10^{-5}	2.7590	2.18×10^{-5}	2.7475
4	5.00%	0.54×10^{-5}	2.7940	0.42×10^{-5}	2.7930

TABLE - 4.50

MAXIMUM SOLUBILITY OF COPPER IN ALUMINIUM AT DIFFERENT TEMPERATURES AS DETERMINED BY X-RAY METHOD

t° C	1/T° A x10⁴	Solubility* in wt.%	Log Solubility in atomic % .	Solubility at Eutectic temp. from the curve logs vs 1/T (method ii)
450	14.225	2.6	0.05038	
475	13.370	3.25	0.14796	
500	12.937	3.97	0.2397	5.7 wt% ***
525	12.531	4.97	0.3373	
548	12.18	5.70	0.39913	

* Solubility as determined by Method (i).

** Maximum solubility at eutectic temp. by method (ii).

*** Maximum solubility at eutectic temp. by method (ii) by extrapolating the curve log S vs 1/T to the eutectic temperature.

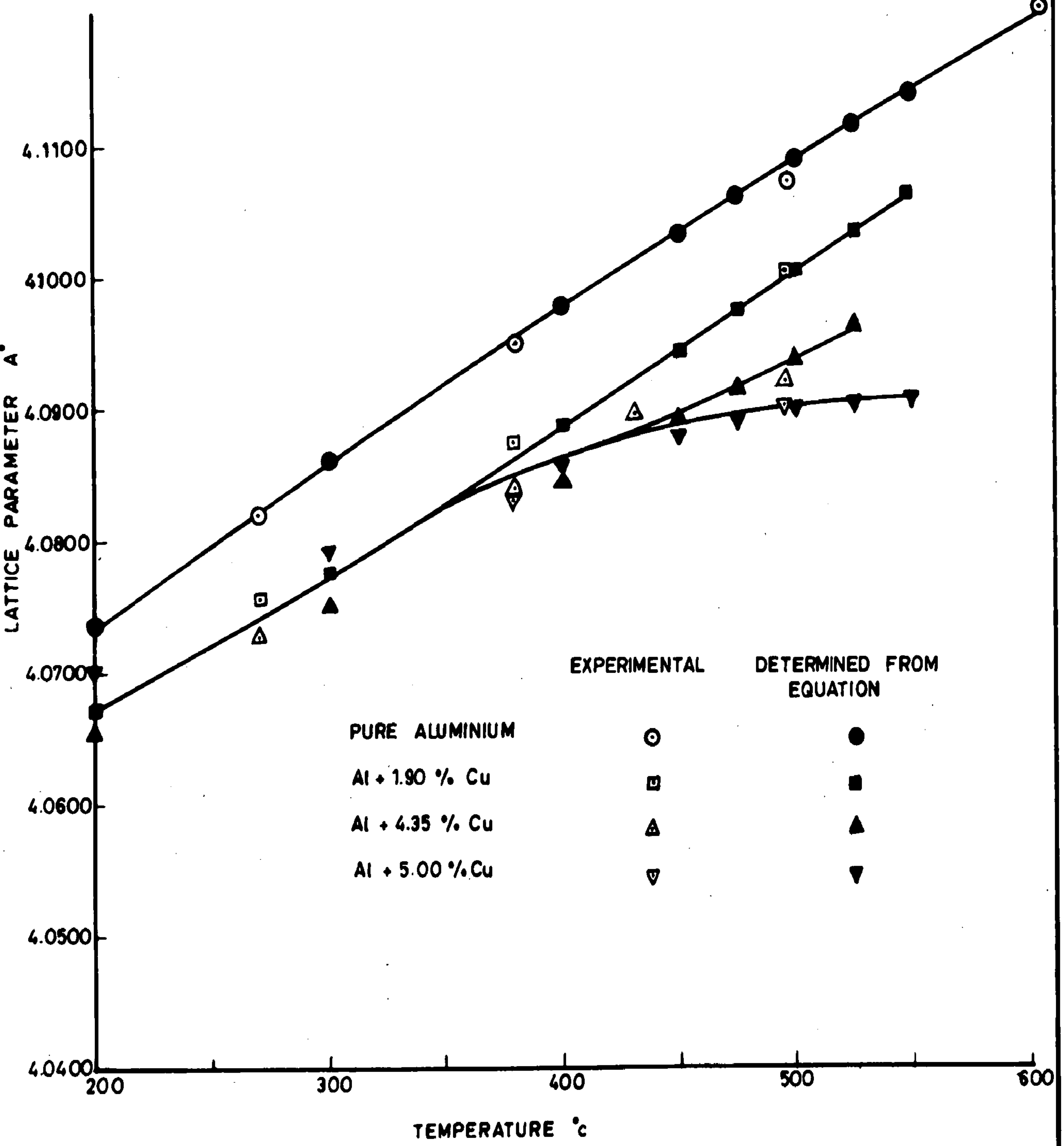

FIG. 4.46_ TEMPERATURE VS. LATTICE PARAMETER IN EQUILLIBRIUM FOR Al-Cu. ALLOYS

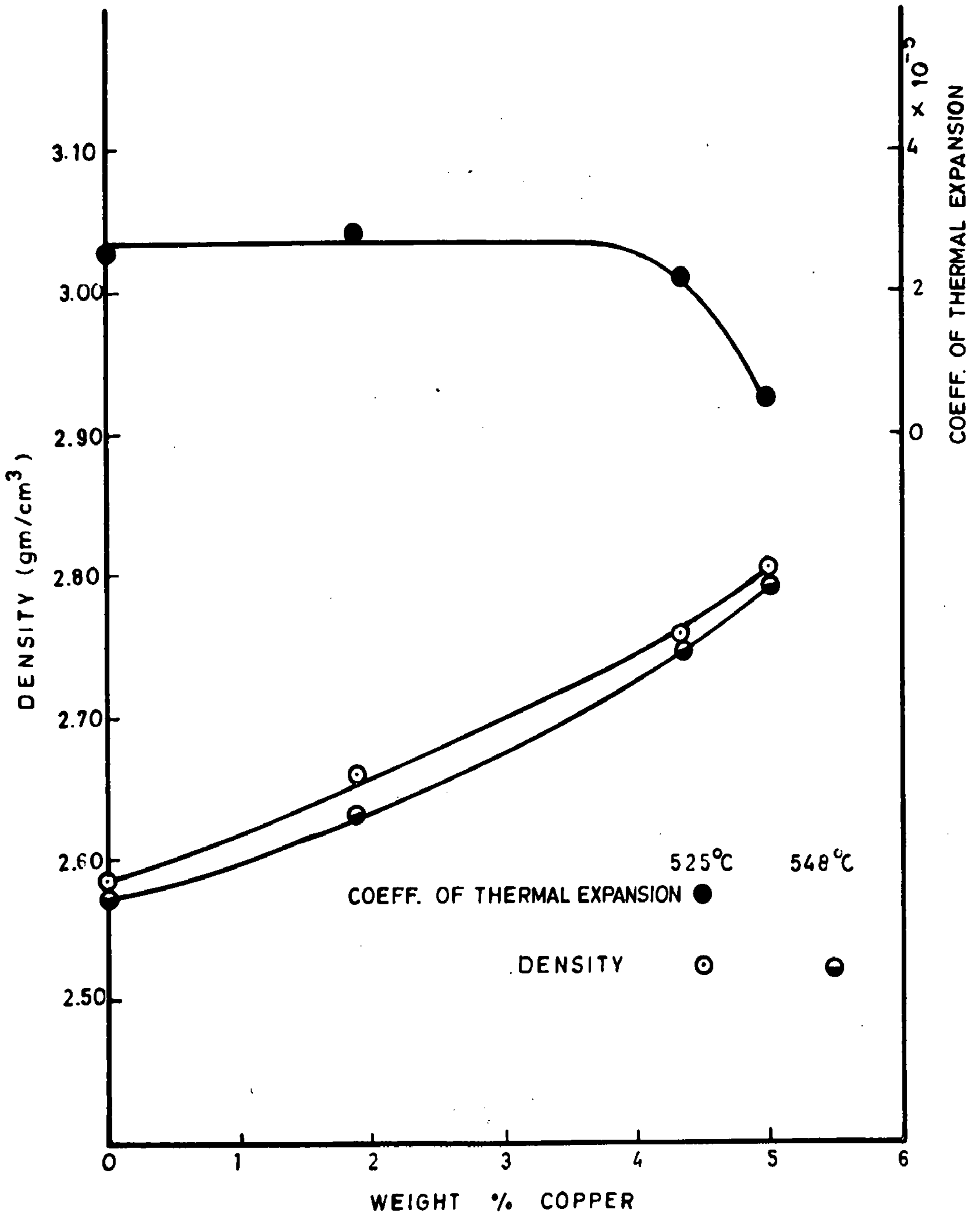

FIG. 4.47_EFFECT OF COPPER ON DENSITY AND COEFFICIENT OF THERMAL EXPENSION OF PURE ALUMINIUM.

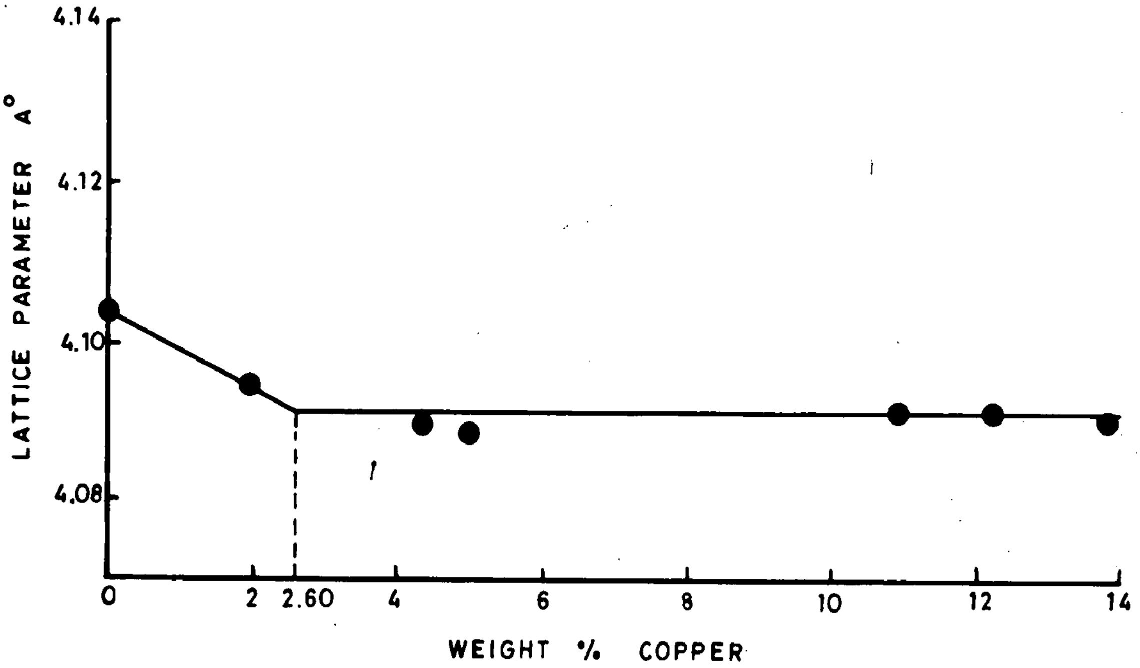

FIG. 4. 48 _ COPPER(% wt) VS. LATTICE PARAMETER OF Al. AT 450 °C

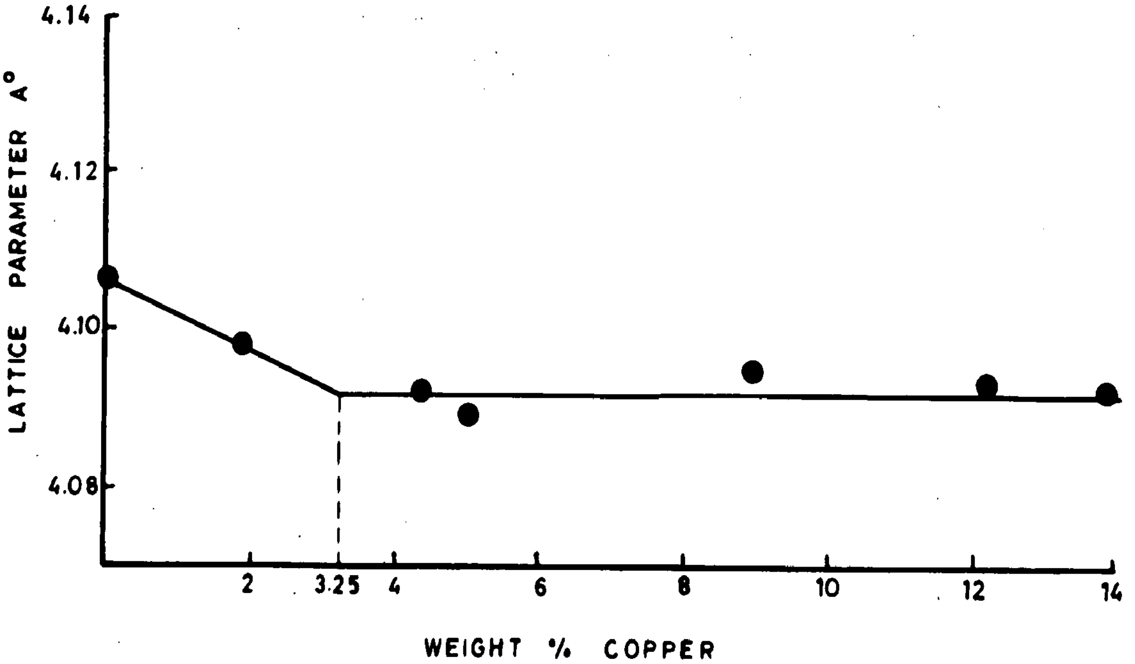

FIG. 4.49 _ COPPER(%wt) VS. LATTICE PARAMETER OF Al. AT 475 °C

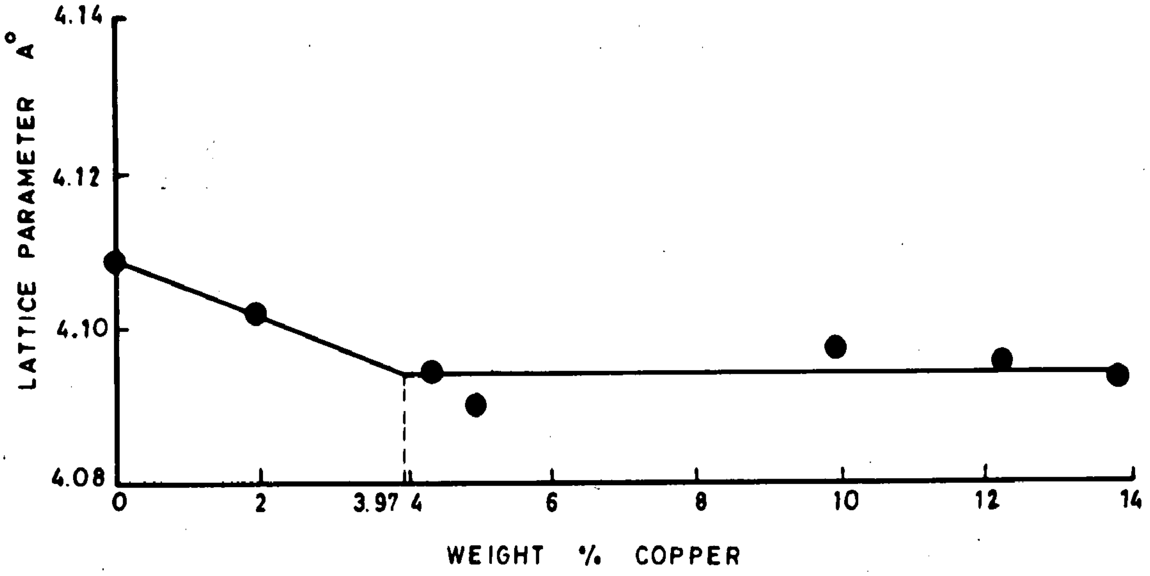

FIG. 4.50_ COPPER (%wt) VS. LATTICE PARAMETER OF AL. AT 500 °C

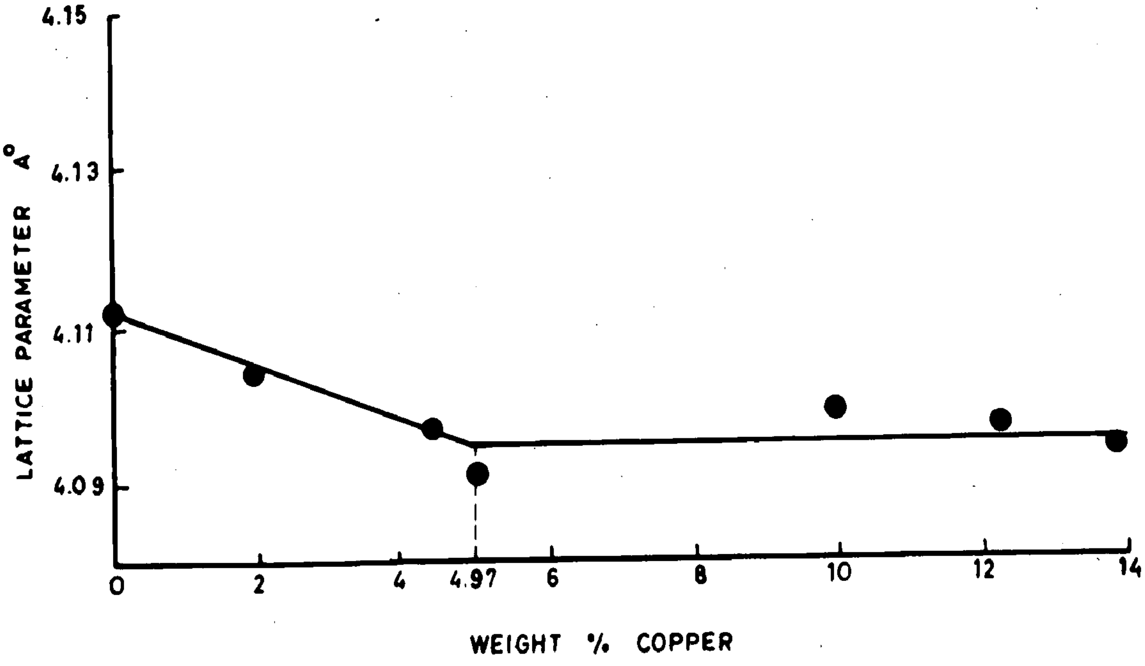

FIG. 4.51_ COPPER (% wt) VS. LATTICE PARAMETER OF Al AT 525 °C

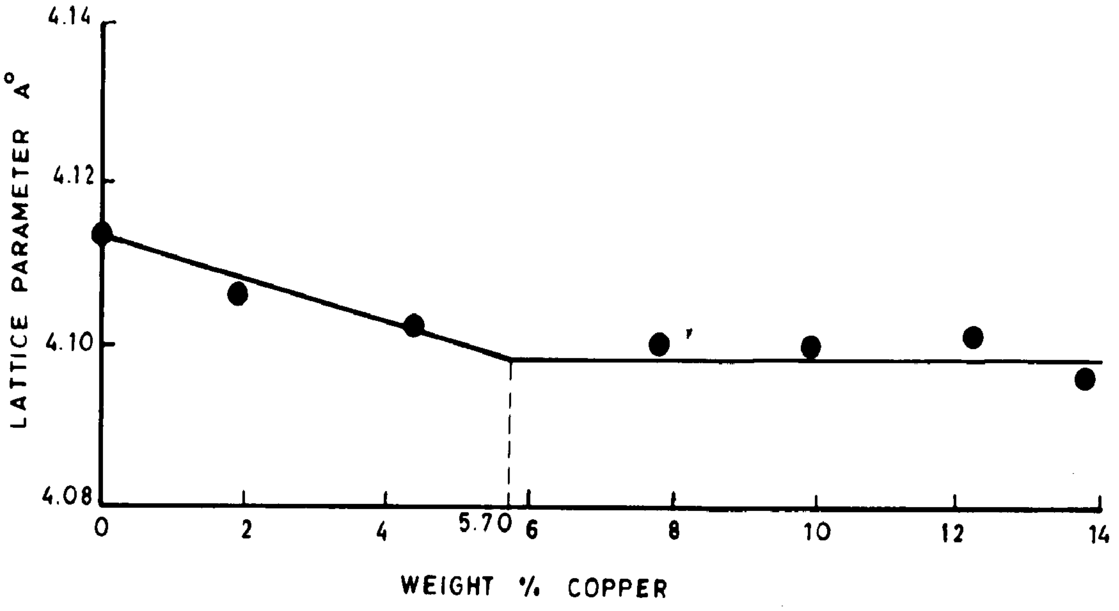

FIG. 4.52 _ COPPER (% wt) VS. LATTICE PARAMETER OF ALUMINIUM AT 548 °C

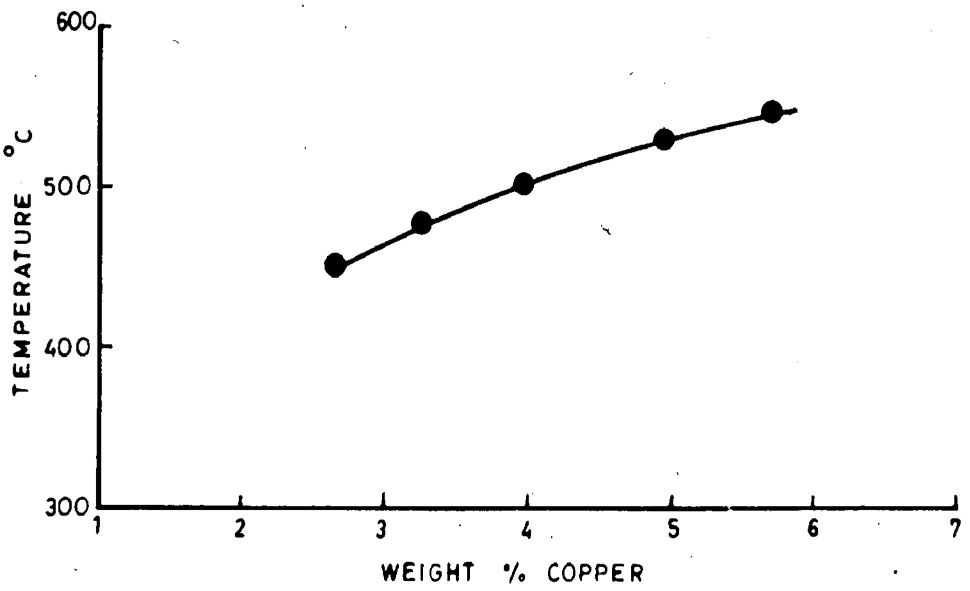

FIG. 4.53 _ SOLID SOLUBILITY LIMIT OF COPPER IN ALUMINIUM.

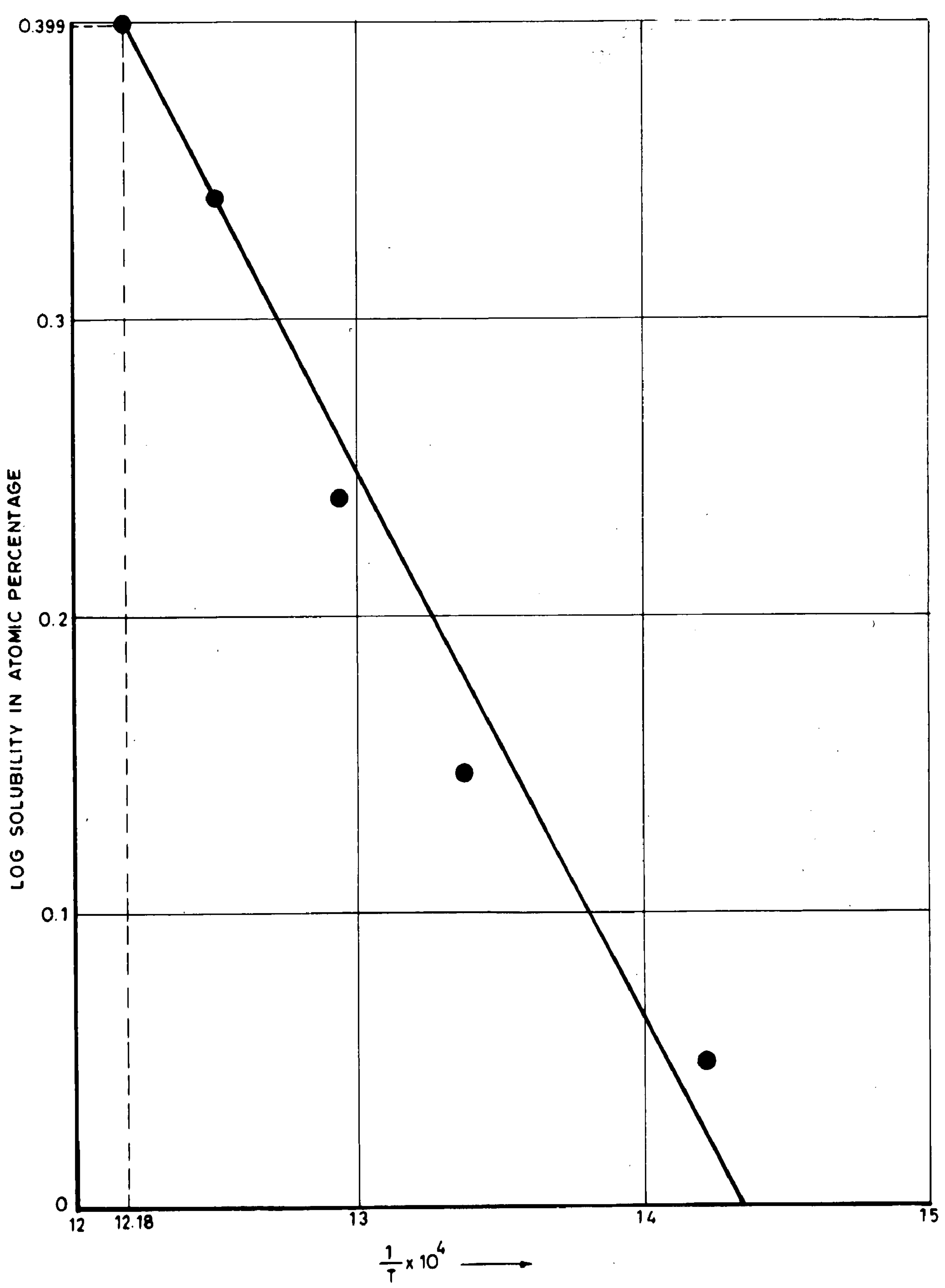

FIG. 4.54 – LOG SOLUBILITY OF Cu IN ATOMIC PERCENTAGE VS. 1/ABSOLUTE TEMP. (T) (SOLUBILITY OF Cu IN Al. DETERMINED AT EUTECTIC TEMP. 5.7 wt.%)

From the results (Figs. 4.46, 4.47, Tables 4.47 and 4.48) it can be seen that, as normally expected due to thermal expansion, lattice parameter has increased with temperature in all the cases, except that the slope of the curve is different for different compositions and temperatures. The results of lattice spacing, coefficient of thermal expansion, and density can be understood on the basis of the following considerations.

i) At a given temperature, the increase in concentration will result in decrease in lattice spacing (8,9,44,143); lattice parameter of copper being less than that of aluminium.

ii) Rise in temperature for a given composition upto the maximum solid solubility limit, will result in increase in concentration due to increased solubility of copper in aluminium tending to decrease the lattice parameter.

iii) With increase in temperature, the difference in spacing of the two elements becomes greater owing to relatively greater thermal expansion of aluminium as compared to that of copper. It is, therefore, expected that changes in spacing caused by the addition of unit quantity of copper should increase with temperature, assuming ideal solution.

iv) Possibilities of generation of more and more vacancies increase with rise in temperature, tending to increase the lattice parameter, the effect being negligibly smaller for lower temperatures.

It can easily be seen that effect due to vacancies will be negligibly small at low temperature and can be ignored. Since lattice parameter has been found out in equilibrium, the value represents the terminal solid solution lattice spacing at a particular temperature. As solubility is very small at low temperature and increases tremendously with rise in temperature, the increase in temperature in case of 1.9% copper alloy should bring about increase in concentration resulting in more and more contractional effect upto 400° C due to above reasons. Beyond 400° C effect due to (ii) vanishes and the contractional effect is produced only due to (iii) with rise in temperature. Since the copper-concentration is small the contractional effect so produced at higher temperatures will be less. It appears that the effect due to vacancies beyond 400° C predominates over the contractional effect produced resulting in slight increase of slope.

For the alloys containing higher concentrations (4.35 and 6.00 wt.% Cu.) the curve is common with that of 1.90% upto about 400° C) due to same reasons, since the same amount is in solid solution in all the cases. For higher temperatures because of tremendous increase in solubility with increase in temperature, the concentration will increase at a fast rate with rise in temperature, bringing about larger and larger contractional effect. The effect will further increase at higher temperatures due to reason (iii). It appears that larger contractional effect produced due to above reasons predominates over the effect due to vacancies, resulting in decrease in slope of the lattice spacing vs. temperature curve for all temperatures in both the cases. The effect is more pronounced in case of 5% copper alloy where the slope has appreciably decreased.

The values of the coefficient of thermal expansion of alloys with reference to the lattice parameter at 548° C are plotted against copper weight percentage. It can be seen that due to reasons (i) and (iii) larger contractional effect should be produced with increase in concentration. It is, therefore, expected that coefficient of thermal expansion should decrease with increase in concentration, if the effect due to vacancies is ignored. Ellwood and Silcock have indicated a regular decrease with increase in concentration. In the present investigation it can easily be seen that coefficient of thermal expansion is found to remain almost unchanged upto about 4% beyond which it decreases at an appreciable rate with increase in concentration. At 548° C there is possibility of generation of larger amount of vacancies which will have a tendency to increase the lattice parameter and coefficient of thermal expansion. It is interesting to note that probably for lower concentration, contractional effect produced due to (i) and (ii) becomes equal to expansion effect produced due to (iv). For higher concentrations however, larger contractional effect produced due to large increase in concentration, predominates over the effect due to vacancies resulting in decrease of coefficient of thermal expansion with increase in concentration. In Figs.4.48 to 4.52 are plotted the values of lattice spacing vs. composition curve for different temperatures. The point of intersection of the horizontal and inclined portion gives the solubility limit at that temperature. The solid solubility line between 475° and 548° C is given in Fig. 4.53. The maximum solubility at the eutectic temperature has been found out by the two methods.

i) By finding out the lattice parameter with the help of equations as determined by computer in single and double phase region, and finding out the point beyond which the lattice parameter becomes unchanged.

ii) By plotting log solubility (in atomic %) against $(\text{absolute temperature})^{-1}$ and then extrapolating to that of eutectic temperature.

Literature survey indicates that there is some controversy about the value of maximum solubility limit at the eutectic temperature. Raynor has reported the value to be 5.7 weight percentage while Ellwood and Silcock (8) have reported the value to be 5.85 weight percentage as determined by method (ii). In the present investigation values as determined by both of the above methods was found out to be 5.7 weight percentage, confirming the result due to Raynor. The solubility limit line between 475° and 548° C as determined in the present investigation is in agreement with the previously reported result, within ± 0.05 weight percentage.

Densities have been calculated at 548° and 525° C and are plotted against weight percentage copper. The density at 525° C increases with increase in percentage copper. The rate however, goes on increasing with increase in percentage. With increase from 525° to 548° C the same nature of the curve is obtained except that rate of increase of density increases more rapidly with increase in percentage. The increasing rate of increase of density with increase in weight percentage can also be understood with the same reasons. With increase in concentration, more and more contractional effect in the lattice is produced resulting in corresponding decrease in volume per unit mass and increase of density . This explains why the slope goes on increasing with increase in concentration . Because of difference of coefficient of thermal expansion of aluminium and copper the contractional effect produced by increase in concentration due to reason (iii) increases in case of higher temperature. The rate of increase of density at 548° C , therefore, increases more rapidly with increase in percentage as compared to that of 525° C.

A general understanding is thus gained of the effects of copper addition on the lattice parameter, coefficient of thermal expansion and density values, a more clear picture can however, only emerge out by the study of actual vacancies at higher temperatures and by extending work on few more alloys between 0 and 1.9 weight percentage. Thus, though, there is a good scope of further work in the above direction the work had to be limited to the above study only in the present investigation reported, because of having no facility of X-ray work in the Metallurgy department of this University.

4.4.3 Aluminium-silver alloys :

The results are given in Figs.4.55 to 4.57 , tables 4.47 , 4.48 and 4.50. The salient features of results are as follows :-

i) The lattice parameter increases with increase in temperature for both Al + 7.5 % Ag and Al + 19.9% Ag except that the values for 19.9% silver are lower than that of 7.5% silver alloy (Table 4.48).

ii) With increase in silver concentration the lattice parameter is found to decrease for all temperatures in single phase region. The rate, however, decreases with increase in temperature.

iii) The density goes on increasing almost linearly, with increase in silver concentration. With increase in temperature, the curve shifts slightly upwards in vertical direction (Fig. 4.57).

iv) The density is found to decrease linearly with increase in temperature. With increase in silver concentration the curve shifts vertically in the upward direction, the slope remaining almost the same.

v) The coefficient of thermal expansion for 7.5% silver alloy is lower and for 19.9% silver alloy is higher than for pure aluminium. The value for 7.5% silver and 19.9% silver alloy decreases and increases respectively with rise in temperature.

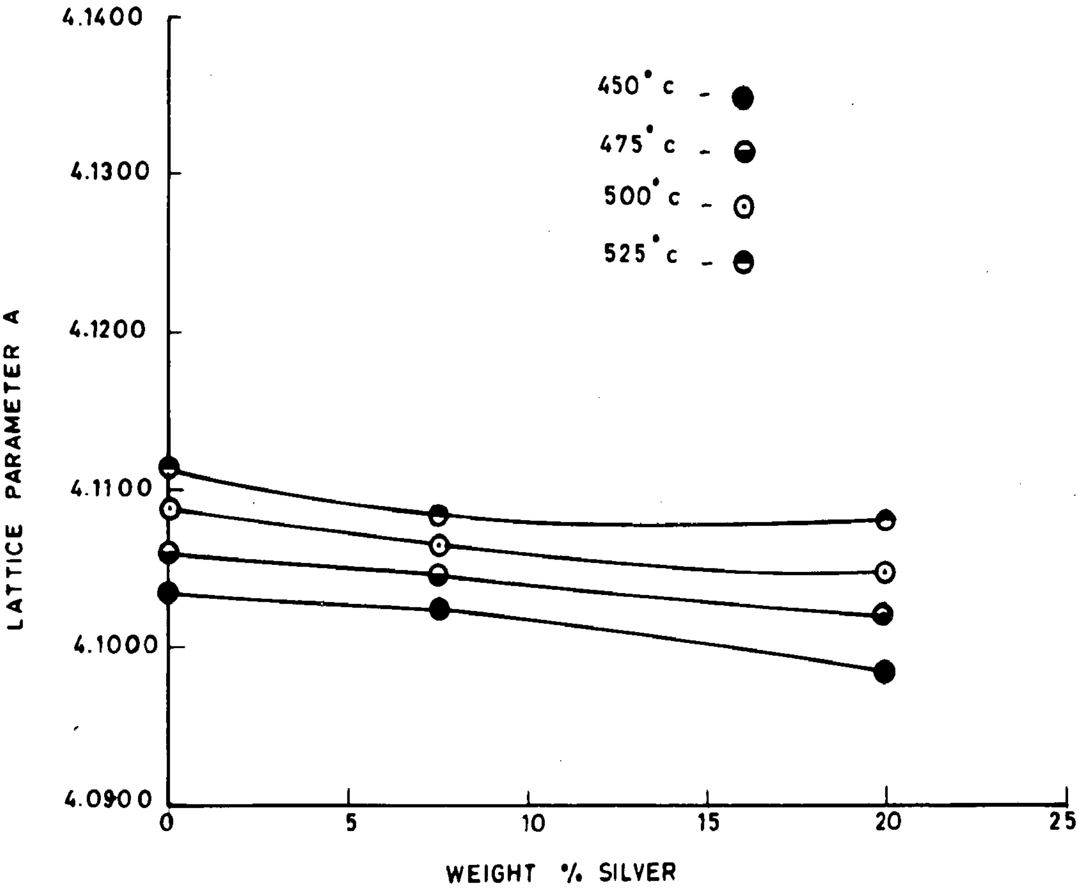

FIG.4.55. EFFECT OF SILVER ON LATTICE PARAMETER OF ALUMINIUM AT DIFFERENT TEMPERATURES

TABLE 4.51

COEFFICIENTS OF THERMAL EXPANSION AND DENSITIES OF ALUMINIUM SILVER ALLOYS AT DIFFERENT TEMPERATURE *

S.No.	Temp. °C	0% Ag		7.5% Ag		19.9% Ag	
		$\frac{1}{aT}\frac{da}{dT}$	Density gm/cc	$\frac{1}{aT}\frac{da}{dT}$	density gm/cc	$\frac{1}{aT}\frac{da}{dT}$	density gm/cc
1	450	-		-	3.180	-	4.155
2	475	2.64×10^{-5}		1.86×10^{-5}	3.176	3.15×10^{-5}	4.145
3	500	2.61×10^{-5}		1.74×10^{-5}	3.172	3.12×10^{-5}	4.135
4	525	2.56×10^{-5}		1.64×10^{-5}	3.168	2.65×10^{-5}	4.125

* Calculated from equations as determined by computer.

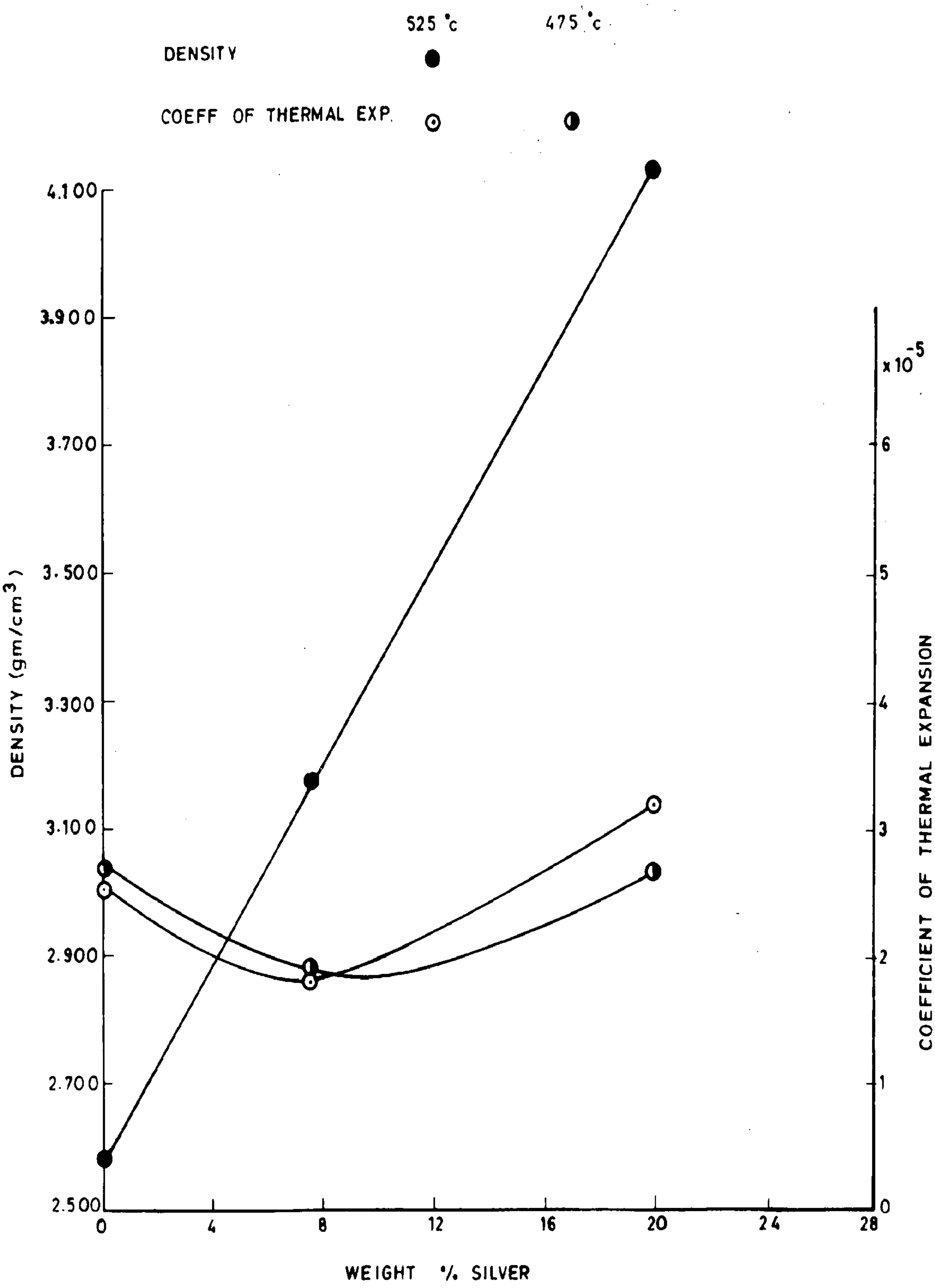

FIG.4.56.EFFECT OF SILVER ON DENSITY AND COEFFICIENT OF THERMAL EXPANSION OF ALUMINIUM.

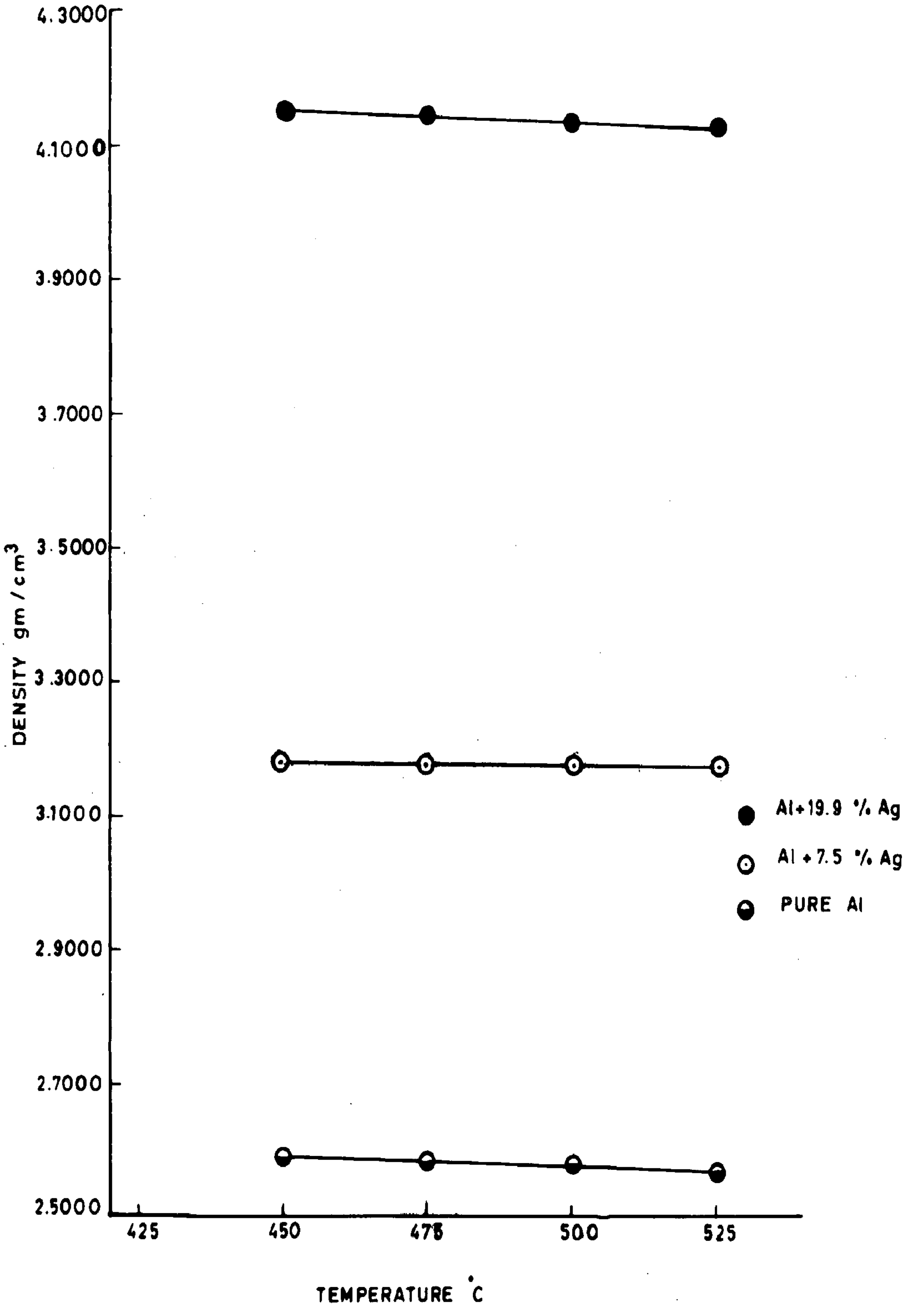

FIG.4.57.EFFECT OF TEMPERATURE ON DENSITY OF Al & Al-Ag ALLOYS

There are conflicting reports about the effects of silver on the lattice parameter of aluminium in solid solution. Gulizaev and Trusova (44) and Dorn (45) have reported increase in lattice spacing with increasing silver concentration, while Axon and Hume Rothery and Ellwood have reported no change in lattice spacing upto 6 atomic % silver. Beyond 6 atomic % Ellwood has shown increase in lattice spacing to a constant value at 14 atomic % silver and according to him the increase is associated with the filling up of the Brillouin zone. Hull and Axon's results show slight decrease in lattice spacing upto 14 atomic % silver.

Most of the work reported are based on the room temperature measurements after rapid quenching. Since solubility decreases drastically with decrease in temperature, very very rapid quenching is desirable to prevent the precipitation of the second phase and to retain conditions of elevated temperature. This may be the reason of conflicting reports by different workers.

In the most recent investigation by Sharan, Narayan and Kalra (144) results of lattice spacing of Hull and Axon have been confirmed. Very severe quenching from 500° C was done to prevent the precipitation of the second phase. Slight contractional effect is observed with silver addition. Their results have indicated the generation of definite vacancies beyond 7.5% addition, the amount of which increased with further addition.

From the above considerations, it can be seen that increase in silver concentration should bring about no or little contractional effect on the lattice spacing of aluminium. The results of lattice spacing vs. composition at different temperatures in single phase region are given in Fig. 4.55. Little contractional effect is produced with silver additions for all temperatures. The results are in agreement with the results of Hull and Axon (64), Sharan and others (144) who have studied the lattice spacings of quenched alloys. With increase in temperature it is apparent that the slope decreases beyond 7.5% .

It is considered that with silver additions beyond 7.5 % , vacancies are generated with alloying. The effect may not be apparent at ordinary temperature but the amount should definitely increase with increase in temperature. The generation of vacancies should bring about increase in the lattice spacing. Thus a part of the contractional effect produced is balanced by the expansion due to generation of vacancies at elevated temperature resulting in corresponding decrease in the slope of the lattice spacing vs. composition curve. The coefficient of thermal expansion of 19.9% silver alloy is higher than that of pure aluminium and the value is found to increase with increase in temperature. The above fact can also be understood with the same reasons. With increasing temperature, the increasing amount of vacancies will lead to higher and higher values of thermal expansion.

Since the vacancies are reported to be generated beyond 7.5%, naturally the effect of vacancies will be negligibly small for additions upto 7.5 % silver. The contractional effect with silver addition will predominate over the very small effect due to vacancies leading to lower value of thermal expansion than that of aluminium. Since the coefficient of thermal expansion of silver is less than that of pure aluminium, the contractional effect produced should be found to increase with temperature, assuming the effect of vacancies to be negligibly small for lower concentrations.

Although a general understanding is thus gained of the effect of silver addition on the lattice spacing and thermal expansions of aluminium at elevated temperatures, a clear picture can only be obtained by the study of actual vacancies at elevated temperature and not by quenching and carrying out work on few more alloys. The work could not be extended in this direction because of having no facility for high temperature X-ray work in our new metallurgy department of Roorkee University. Most of the work reported here has been carried out at the Metallurgical Engineering Department of Banaras Hindu University.

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CHAPTER V

EXPERIMENTAL TECHNIQUE FOR THE EVALUATION OF MACHINABILITY, METALLOGRAPHIC STRUCTURE AND MECHANICAL PROPERTIES.

CHAPTER V

EXPERIMENTAL TECHNIQUE FOR THE EVALUATION OF MACHINABILITY, METALLOGRAPHIC STRUCTURE AND MECHANICAL PROPERTIES:

5.1 METALLOGRAPHIC STUDY

For study of microstructure cylindrical specimen of about 1" diameter were polished by emery paper down to 000 grade, hand polished on selvet cloth and atched by Kellar's reagent. The photomicrographs were taken on the Vicker's Projection Microscope.

5.2 HARDNESS TESTING

Hardness was measured on polished specimens using a Vicker's Hardness Testing Machine with 10 Kg load. Hardness values were determined as the average of 10 values taken at different regions of the same sample.

5.3 MECHANICAL TESTING AT ROOM TEMPERATURE

Tensile tests of specimen, of appropriate specification were made by machining the cylindrical cast ingots and tested in Hounsfield tensometer. Sample of the alloys treated with tantalum and individual rare earth elements were tested in Avery Tensile testing machine. Four samples were taken in each case.

5.4 TECHNIQUE FOR EVALUATION OF MACHINABILITY

5.4.1 General introduction:

Considerable efforts have been made to investigate the

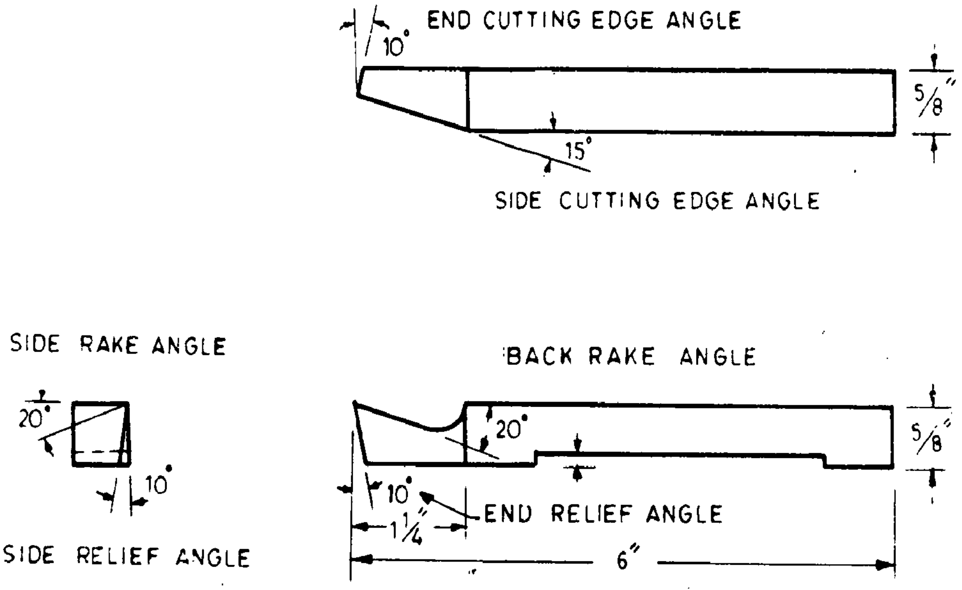

END CUTTING EDGE ANGLE
10°
5/8"
15°
SIDE CUTTING EDGE ANGLE
SIDE RAKE ANGLE
20°
10°
SIDE RELIEF ANGLE
BACK RAKE ANGLE
20°
5/8"
10°
END RELIEF ANGLE
1¼"
6"

FIG. 5.01

machinability of metals by investigators like Ernst, Boston, Leyensetter, Woldman,Gibbons, Takeyama and others. However the definition put forward by Woldmann and Gibbons that "The most machinable metal is one which will permit the fastest removal of the large amount of material per grind of tool (without resharpening of the tools) with satisfactory finish" sounds pretty satisfactory (145).

The different methods for evaluation of machinability were suggested by Herbert (146) Boston (147) Soremsan & Gates (148), Kessener (149), Keneford (150) and Schmidt (151). From the methods suggested above it can be easily seen that there is no direct method of determining the ease and difficulty of removing metal and use of individual methods have always been made to evaluate the machinability of metals. The most important factors which are these days considered for evaluation of machinability of the metals are:

(i) Tool life
(ii) Chip formation and shear plane angle Ø
(iii) Cutting forces and power consumption.
(iv) Surface finish.

The relative importance of these four factors depend, upon the kind of machining operations. This can be tabulated in the form as below :-

Table 5.01

Order	Rough operation	Finishing operation	Automatic M/C Tools
1.	Tool life	Surface finish	Types of chips
2.	Power consumption.	Types of chips	Surface finish
3.	Types of chips	Tools life	Tool life
4.	Surface finish	Power consumption.	Power consumption.

5.4.2 Consideration for the selection of calorimetric methods for power consumption and chip formation criteria for evaluation of machinability.

5.4.2.1 Calorimetric method.

Various methods have been employed in the past for the measurement of power consumption. Some of the important instruments,used for the measurement are:

(i) Dynamometer
(ii) Ammeter
(iii) Wattmeter
(iv) Calorimeter

(i) Dynamometer: A dynamometer is a device which indicates the cutting forces at the tool & is the most reliable device because the mechanical efficiency of motor machine system is eliminated and it gives a picture about what happens at the cutting edge of the tool.

ii) Ammeter: Ammeter in the line may be of sufficient value where the line voltage is fairly constant, as an indicator of power developed by motor.

iii) Wattmeter: It is a more satisfactory device than an ammeter since it automatically compensates for voltage fluctuation. This records horse-power developed by the motor during cutting and tare horse-power when not cutting.

iv) Calorimeter: A calorimeter has been used by by Schmidt and others (151) to determine the heat balance of total work done in the metal cutting process. The set up is simple and has been found very successful for lighter metals like aluminium and magnesium, during the machining of which develops small cutting forces.

Discussing the four methods, the wattmeter has been found to be commonly used and is sufficiently accurate for production set ups. But discrepencies caused by varying drive motor efficiency , lubrication and friction factors in gears and bearings render this method unsatisfactory for precise determination of power required by the cutting tool.

These sources of errors can very well be eliminated by using dynamometer but such instruments require expert design and calibration and must be checked frequently, especially when used at high cutting speeds. Sources of small erros, relatively difficult to exclude in investigations of this sort are :

i) Variation in work - piece material which is actually never homogeneous.

ii) Change in cutting edge angles caused by wear which begins as soon as cutting starts.

Ammeter method is not a very good method and depends on the value of line voltage which is always fluctuating.

In the present investigation the calorimetric method has been used for the determination of power consumption due to following considerations:

i) It is very much reliable and successful method for aluminium and its alloys.

ii) It is very sensitive, and even a very small difference of rise in temperature can be measured, whereas Wattmeter and dynamometer methods are not so sensitive for small changes in machinability.

iii) It does not require expensive cooling provisions which are necessary in other methods.

iv) Very small test pieces are required.

v) Experimental set-up is simple and does not require special training for handling.

5.4.2.2 Chip formation study.

Study of chip - formation and shear plane angle Ø have been taken as another criteria for evaluating machinability. In practice shear angle is variable and its value depends upon the cutting conditions and the material being cut. As can be seen if the angle is small, the path of shear will be long, the chip will be thick and the force required to remove the layer of metal will be high. If the angle is large, the path of shear will be short, the chip will be thinner and the force required to remove the layer of metal will be small. A metal requiring less force for cutting can obviously be classed as easily machinable metal. Thus the shear angle is an important geometrical quantity in the cutting of the metal, which can be used as the criterian for machinability of metals. Due to the above reasons measurement of chip thickness ratio and shear plane angle Ø has been made in all the cases for evaluating the machinability in the present investigation.

5.4.3 Principle of calorimetric method and chip formation study.

5.4.3.1 Principle of calorimetric method for measuring power consumption in drilling:- To measure the power consumption, the tool (drill) and the work - piece are submerged in a specific quantity of distilled water. The test pieces are tubular having outer diameter less than the diameter of the drill. This eliminates the friction and squeezing action at chisel edge point and makes the cutting action similar to that of single point tool. Chips do not flow through the clearance space of the drill, but they fall down into the water. Thus by measuring the difference in temperature before & after the drilling operations the specific power consumption can be calculated. This gives the power required directly at the cutting tool.

5.4.3.2 Principal of Chip Formation Study:-

A chip is a portion, which is sheared from a body by the cutting tool during machining. The basic mechanism of chip formation is by shear deformation. Let us consider the case of a continuous chip (Fig. 5.02) The metal deforms by shear in a narrow zone extending from the cutting edge to the work surface. For mathematical analysis the shear zone has been treated as a single plane, commonly known as shear plane. It is usually indicated by a straight line which makes an angle ϕ with the direction of tool travel. As the metal lying ahead of the tool reaches this plane, it is displaced by shear to form the chip, which then slides up the face of the tool. In the case of continuous chips, (Fig. 5.03) this shear occurs without fracture of metal and so continuous chip is formed. In the case of discontinuous chips(Fig.5.04) the metal is not able to undergo the required amount of shear without fracture and thus fracture does occur intermittently along the shear plane, breaking the chips up into segments.

In the case of the built up edge chips (Fig. 5.05) the metal shears without fracture, but the resistance to the sliding of the chip up the tool surface produces additional shearing deformations in the chip; that portion of the chip metal adjacent to the tool face shears away from the body of the chip and remains as the built up edge. As the cutting proceeds the built up edge constantly changes in size. It builds up because of more metal shearing away from the flowing chip and the work material passing beneath, then it decreases in size again as this extra metal bolughs off in 'the form of fragments left on the chip and finished surface. (152)

For the sake of analysis the metal machining processes have been classified into two systems viz:

CHIP FORMATION AND TYPES OF CHIPS

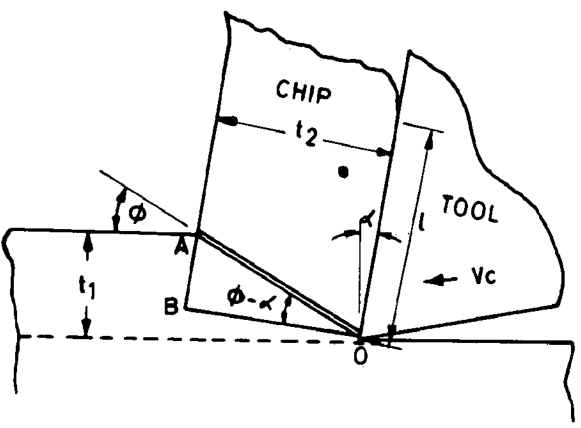


FIG.5.02_GEOMETRY OF CHIP FORMATION

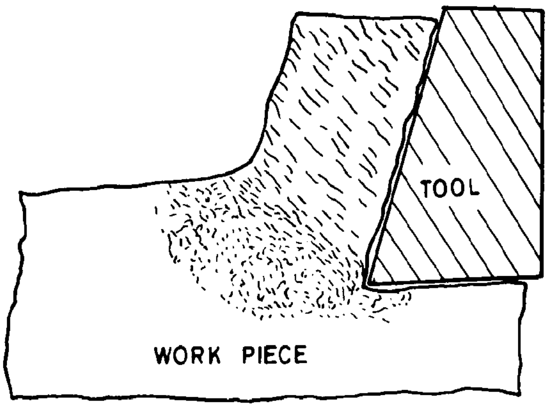


FIG.5.03_CONTINUOUS CHIP

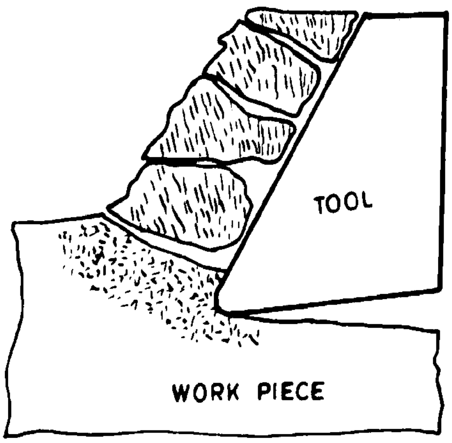


FIG.5.04_ DISCONTINUOUS CHIP

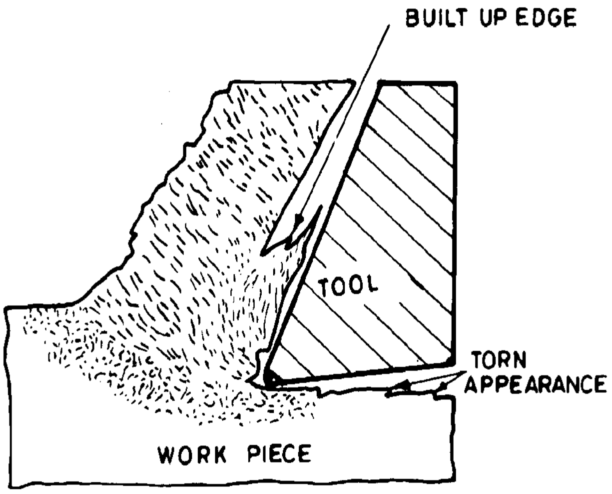


FIG.5.05_CONTINUOUS CHIP WITH BUILT UP EDGE

i) Orthogonal cutting system.

ii) Oblique cutting system.

In the case of orthogonal cutting, the tool is set with its cutting edge perpendicular to the direction of tool travel and only a single cutting edge is active. The width of the cutting edge has been assumed a bit greater than width of the work-piece. (Fig. 5.06). In oblique cutting as shown in fig. (5.07) the cutting edge of the tool is inclined at an angle 'ν' called the inclination angle. This inclination of the cutting edge introduces additional effects not found in orthogonal cutting. The chip does not flow in a direction perpendicular to the cutting edge. The geometry of chip formation in this case is illustrated in(fig.5.08) in which ν measures the inclination of the cutting edge. As a result of this angle, the chip flows up the tool face with a side-ward motion measured by angle β. The two basic tool angles play important roles in the process of machining. These are the rake angle α and the inclination angle ν. In practice the effective rake angle of the orthogonal cutting system is virtually equivalent to those called velocity or true rake angle of a cutting tool. The velocity rake angle is the slope of the tool face with respect to a plane perpendicular to the direction of tool travel measured in a plane which both contains the tangential velocity vector and is perpendicular to the generated surface at that point.

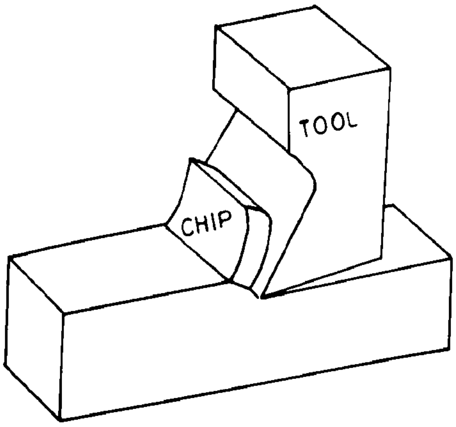


FIG. 5.06_A MODEL OF ORTHOGONAL CUTTING

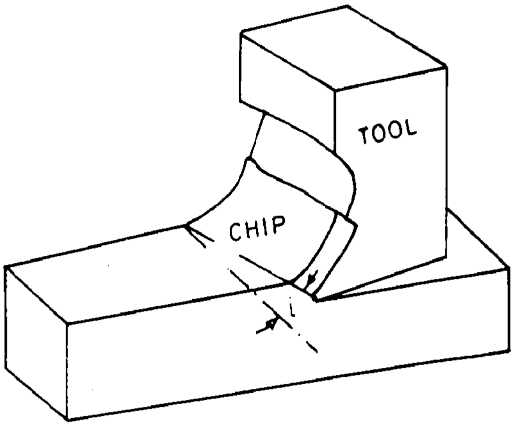


FIG. 5.07_A MODEL OF OBLIQUE CUTTING

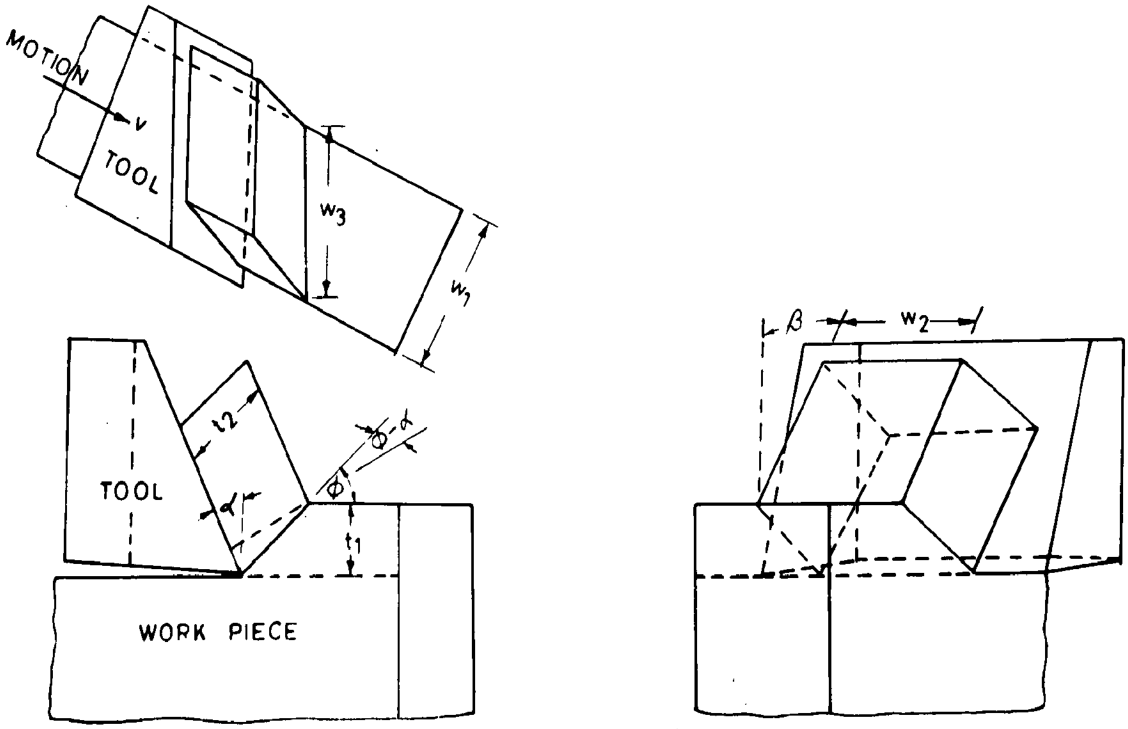


FIG. 5.08_GEOMETRY OF CHIP FORMATION IN OBLIQUE CUTTING

It is a function of back rake, side rake and side cutting edge angles of a cutting tool. This angle may be calculated from its component angles by means of equation (153)

$$\tan t' = \tan \gamma \cos C + \tan \alpha \sin C \qquad (5.01)$$

where t = - velocity rake angle of cutting tool

α - back rake angle

γ - side rake angle

C - side cutting edge angle

Merchant has independently deduced an expression for shear angle in terms of rake angle, chip thickness, and depth of cut (i.e. feed) and the expression can be given as

$$\tan \phi = \frac{\frac{t_1}{t_2} \cos t}{1 - \frac{t_1}{t_2} \sin t} \qquad (5.02)$$

5.4.4 Calorimetric set up:-

For the set up the complete calorimeter was designed and constructed in the workshop based on the method developed by Schmidt. The details are shown in fig. 5.09. The calorimeter was made of $\frac{1"}{16}$ thickness sheet. The calorimeter was covered with a lid at the top. The whole of the calorimeter was covered with glass wool and then by 1/8" ply-wood sheet to prevent conduction of heat outside; and was made completely heat insulated. The set up was made on "Ex-Cell-O Room Turret Milling

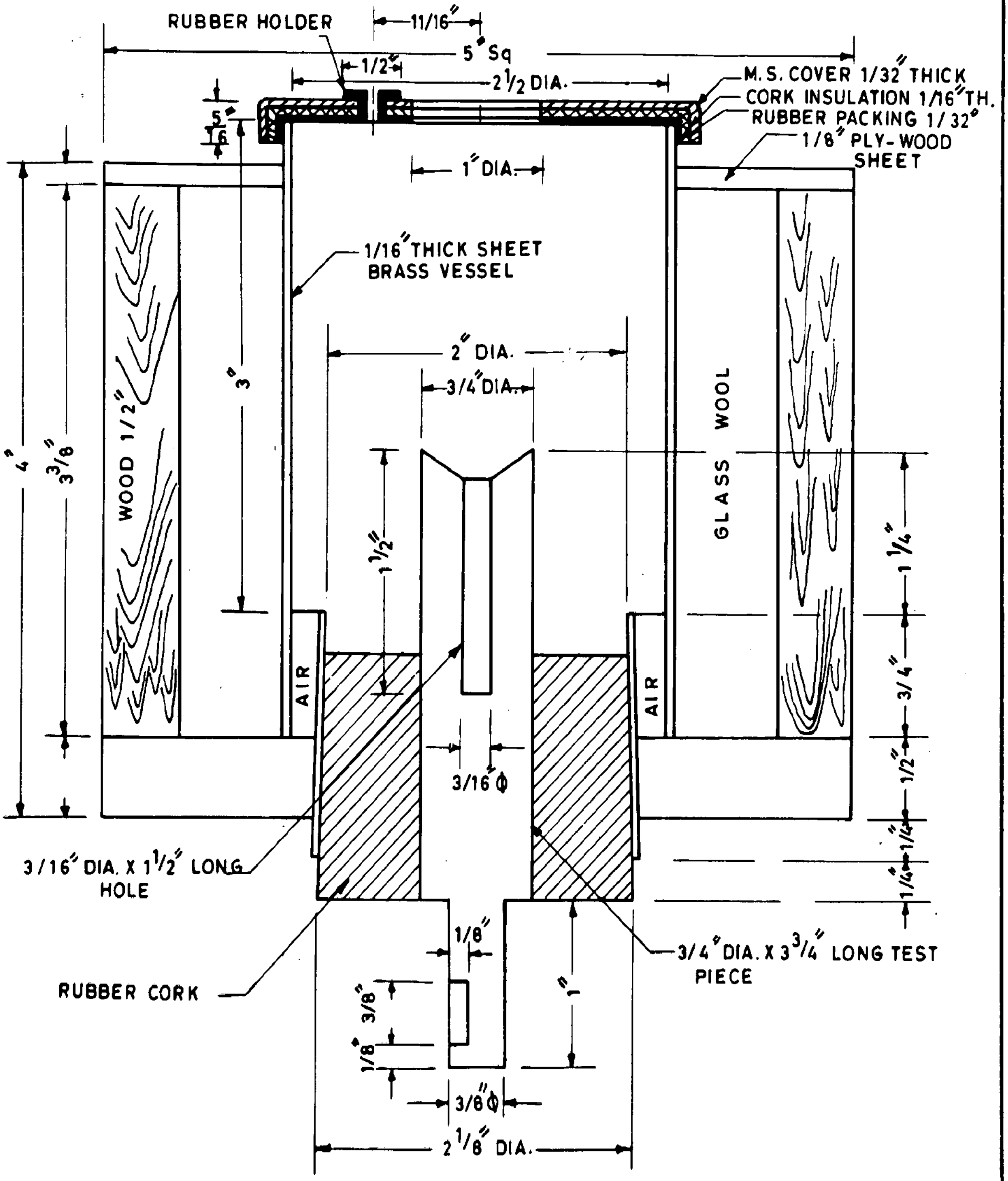


FIG. 5.09_ DRILLING CALORIMETER

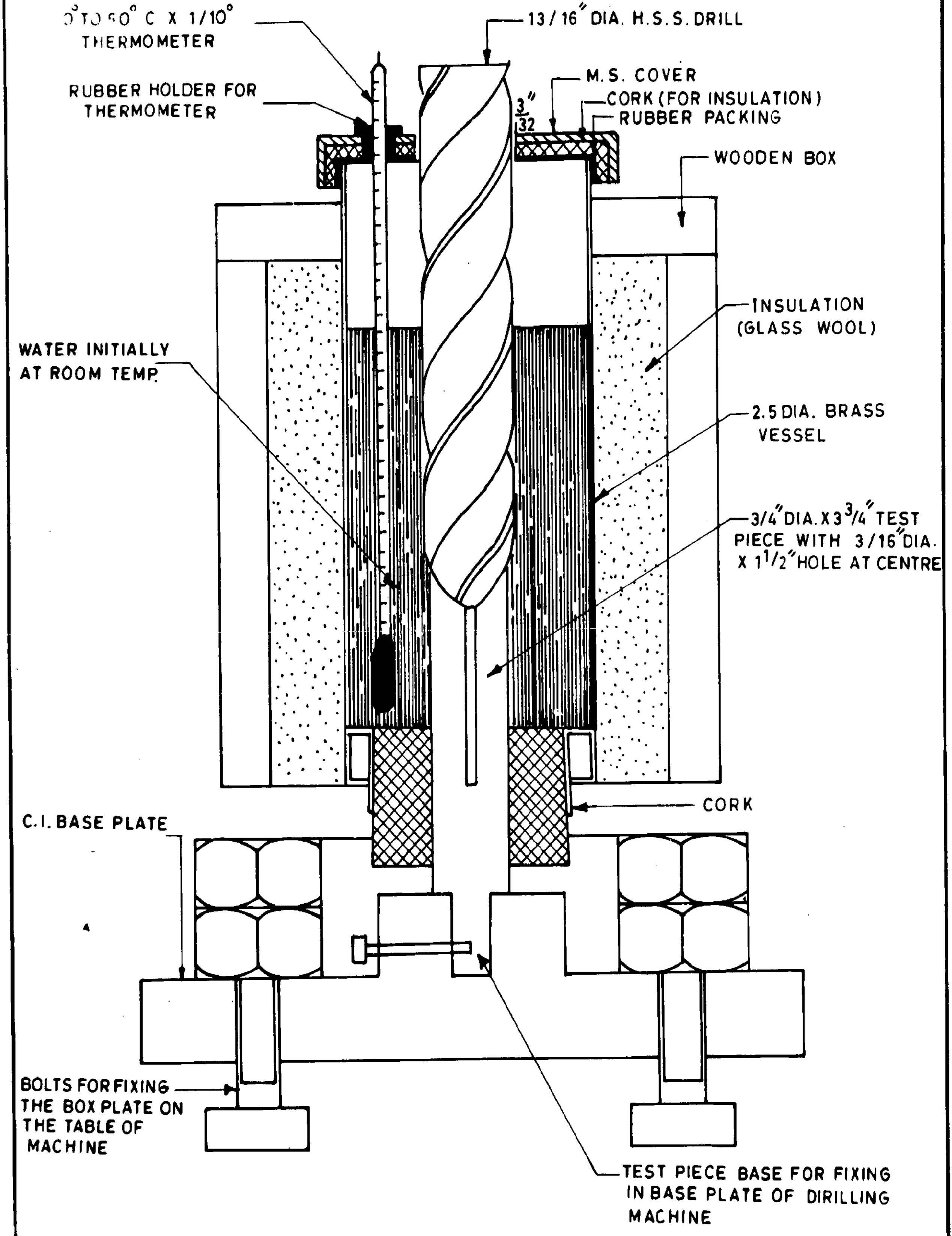


FIG. 5.10_ CALORIMETRIC SET-UP

Machine", having variable speed and feed arrangements. The details of the calorimeter along with the fixture, & specimen are shown in fig. No. (5.10)

5.4.5 Experimental Procedure:

5.4.5.1 Preparation of test pieces:

For each alloy required number of test pieces having the shape and dimension as shown in fig. (5.09), were machined from already cast rods for calorimetric tests .

For chip formation study 1" dia and 5-3/8" long cast rods were used. A very little cut was given just to remove the surface layer before the actual tests were performed. A rod of 15/16" dia was actually used for collecting the chips.

5.4.5.2 Determination of water equivalent of the instrument with test piece and drill.

The water equivalent of the apparatus was calculated considering the total set up and specimen with the drill partly immersed as one body and applying the method as given by the formula:

$$m_1 S (t_2 - t_1) + W (t_2 - t_1) = m_2 S (T - t_2)$$

where W is the water equivalent of calorimeter when taken as one body;

m_1 = mass of water initially in the calorimeter;

t_1 = Initial temperature of water;

t_2 = Final temperature of total water in calorimeter after adding hot water of mass m_2 and temperature T.

S = Specific heat of water.

The value of W was taken as average of three values. The water equivalent was calculated with all the four alloy specimens seperately and are given in table No.5.02. There appeared to be no effect with small variation in composition due to addition of rare earth metals on water equivalent.

Table No. 5.02

S.No.	Name of the alloy	Water equivalent gms.
1.	Pure Aluminium	37.18
2.	Aluminium + 8% Mg.	36.3
3.	Alloy 195 (4.5 % cu.)	37.7
4.	Alloy (122) (Al 10% Cu +0.2% Mg)	38.4
5.	Aluminium Silcon alloy 1 type (A - 13)	36.9

5.4.5.3 Procedure for determining power consumption and calibration of the calorimeter for actual power consumption with wattmeter.

The set up was made on "Ex-cell-o-Ram Turret Milling machine" having variable speed and feed arrangements. Same speed and feed were used as in the actual experiment. Following procedure was adopted:

i) The specimen was fixed in its position after measuring its diameter.

ii) The calorimeter, set along with the specimen was then fixed with the help of attachment as shown on the platform of drilling machine.

iii) 100 cc of distilled water was taken , the drill and specimen were kept submerged in water.

iv) Then the initial temperature t_1 of the water was noted.

v) Required amount of feed and speed were adjusted.

vi) Time in sec. for cutting upto 1" was noted. The final temperature t_2 after cutting was also noted.

vii) The whole set was allowed to cool for the same time as was required for cut. The temperature t_3 was noted. The difference of t_2 and t_3 divided by 2 when added to final temperature t_2 gave the correct temperature t_4 after radiation correction.

viii) The power was also measured with the help of wattmeter kept in the circuit, The watts consumed were measured on no load and while drilling the specimen. The difference of the two readings gave the watts consumed,

ix) The power consumed with the help of calorimeter was calculated as given below.

Volume removed per minute Vm in cubic inches

= Effective area x Feed Rate / min.

$$= \pi/4 \left[D^2 - (3/16)^2\right] \times \text{R.P.M.} \times \text{Feed /Rev} \qquad (5.3)$$

The quantity of heat caused by the combination of friction and deformation during the cutting of 1" length of test piece and unit HP were determined by finding out the rise in temperature $(t_4 - t_1)$ of the water surrounding the tool and work piece, the total water equivalent and volume of metal removed per minute

$$\text{B.T.U/Min} = (\text{Total water equivalent}) \times \text{Feed rate per min} \times \text{rise in temperature } (t_4 - t_1) \text{ in degree F} / 453.6 \quad (5.4)$$

$$\text{H.P.} = \text{B.T.u / min} \times \frac{778.7}{33000} \quad (5.5)$$

The horse power so calculated was compared with the actual horse power measured by wattmeter and %fraction was determined. Observations are given in table 5.03 . This fraction was found out to be 95%.

Experiment was performed for all the test pieces using the procedure as described and power consumption was calculated by the rise in temperature (t_4-t_1). The H.P. was then calculated as shown in (54) and (55). The HP was then corrected for the actual horse power using the fraction as determined in previous paragraph. Volume of metal removed per minute in cubic inches was then determined by the formula

$$\pi/4 \ (D^2-d^2) \times \text{R.P.M.} \times \text{Feed/Rev} \quad (5.6)$$

The Volume was converted into cubic-cm unit. H.P so determined divided by volume removed per minute gave the values of HP unit. Duplicate samples were taken in all the cases, and the average values have been reported in the tables in preceeding chapters. The two readings of HP-unit taken for two different test pieces for the same alloy were found to be within ± .0005 HP.

Table No. 5.03 Calibration of calorimeter for actual power consumption.

Volume of water = 100 cc

S.No.	Name of the alloy	Water equivalent	Rise in Temperature (after reading correction.)	H.P.	$W_1 - W_2$	H.P by Wattmeter.	% fraction	Average.
1.	Pure Aluminum	137.18 gm	13.6°C	0.655	915-400	0.690	95%	
2.	Al+8% Magnesium	136.3 gm	11.7°C	0.560	844-400	0.590	95%	
3.	Al+4.5% Cu	137.7 gm	11.4°C	0.551	825-400	0.569	97%	95%
4.	Alloy 122	138.4 gm	11.5°C	0.559	850-400	0.603	93%	
5.	Al +13% si	136.9 gm	10.2°C	0.491	785-400	0.516	95%	

5.4.5.4 Procedure for chip formation study.

The set up was made on Harihar model lathe machine . High speed tool was used. Tool geometry as used, is shown in Fig. 5.0 1. The chip thickness was measured with the help of tool maker's microscope having least count of .05 mm. At least 50 chips were measured in each case. The chip thickness ratio and shear plane angle ϕ were calculated by the equation 5.02.

5.4.5.6 Cutting conditions.

5.4.5.5.1 Calorimetric method

Speed 1000 r.p.m.

Time for cutting 1" length = 16 secs.

$$\text{Feed} = \frac{60}{16\text{x}1000} = 0.00375\text{ " per minute}$$

Drill 13/16" dia high speed steel drill.

5.4.5.5.2

Chip formation study

Depth of cut 3/32"

$$\text{Diameter of face plate } D = 7\frac{21}{32}\text{ "}$$

Surface speed = 550 feet / min

Time of the travel for 1" = 1 min 4 secs.

$$\text{Therefore } 550 = \frac{\pi DN}{12}$$ Where N is the number of revolutions / min.

N = 274 r.p.m.

Feed = Distance travelled

per revolution = 0.087 mm

Tool geometry.- Tool geometry is given in Fig. 5.01.

End cutting edge angle = 10°

C = side cutting edge angle = 15°.

Y = side rake angle = 20°

Side relief angle = 10°

End relief angle = 10°

Back rake angle = 20°

r = cutting ratio = t_1/t_2

$$\tan t = \tan \gamma \cos C + \tan \alpha \sin C$$

$$t = 32° \ 51.5'$$

Therefore $$\phi = \tan^{-1} \frac{r \cos t}{1 - r \sin t}$$

where $$r = \frac{0.087 \text{ mm.}}{t_2 \text{ in mm.}}$$

5.5 PURITY OF BASE METALS AND ALLOYING ELEMENTS AND COMPOSITION OF MIXTURES OF RARE-EARTH ELEMENTS USED IN THE PREPARATION OF ALLOYS.

Aluminum of 99.5 % purity electro-lytic copper of 99.9%, purity, magnesium of 99.8% purity, and **silicon** of 99.3% purity formed the basis of test alloys.

Misch metal containing 50-52% cerium, 20-22% lanthanum, 15 - 17% neodymium and 10-12% of other rare earth elements was used throughout the investigation .Lancer- amp 2 used was straight misch metal containing 5% iron and 10% zirconium, Rare - earth fluroides was mixture of fluorides of all rare - earths present in the same proportion as present in misch metal. Cerium of fluka quality and of 99.5% purity, lanthanum of light make, neodymium and tantalum of 99% purity were used in all the cases.

CHAPTER VI

EFFECTS OF TANTALUM, RARE EARTH ELEMENTS AND MIXTURE OF RARE EARTH ELEMENTS ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM AT ORDINARY TEMPERATURE.

CHAPTER VI

EFFECTS OF TANTALUM, RARE-EARTH ELEMENTS AND MIXTURES OF RARE-EARTH ELEMENTS ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM AT ORDINARY TEMPERATURE.

6.1 GENERAL INTRODUCTION.

A systematic study has been made with a view to explore the possibility of beneficial additions of tantalum and rare earths on aluminium from strength and machinability considerations. The rare earths have been used in the form of mixtures viz., misch metal, lancer-amp 2, and rare earth fluorides. Effects of individual rare earth metals viz. cerium, lanthanum and neodymium on the microstructure, machinability, and mechanical properties have also been investigated. The investigation will help to develop the suitable applications of these elements as alloying, as well as to give the more clear picture of the causes of the effects of additions of rare earth mixtures based on the effects of the individual elements contained in them.

Review of the literature has indicated that although there are little but conflicting reports about the effects of addition of misch metal on the grain size and mechanical properties, no work has been done in the direction of machinability. Practically no attempt seems to have been made to investigate the effects of tantalum and individual rare earth metals Ce, La and Nd on the machinability and mechanical properties of aluminium.

6.2 EXPERIMENTAL PROCEDURE.

6.2.1 Preparation of aluminium alloy with rare earths and tantalum :

The required amount of aluminium of 99.5% purity was taken in the graphite crucible and melted in temperature controlled vertical muffle furnace. When the temperature was raised to 800° C the crucible was taken out. The required amount of rare-earths, wrapped in aluminium foil, was dipped inside the melt and stirred. In the case of tantalum, the molten metal was superheated to 1000° C and tantalum wrapped in aluminium foil was added. The liquid aluminium was kept in the furnace for about 15 minutes for homogenisation. The casting was done at 700° C in cast iron metal moulds.

6.3 EXPERIMENTAL RESULTS .

6.3.1 Effects of Tantalum additions:

Results are given in Figs.6.01 & 6.02, table 6.01 and plates 6.01 to 6.08 .

6.3.2 Effects of cerium additions :

Results are given in figs. 6.03 &. 6.04, table 6.02 and plates 6.09 to 6.12.

6.3.3 Effects of lanthanum additions:

Results are given in figs.6.05 & 6.06, table 6.03 and plates 6.13 to 6.16.

6.3 .4 Effects of neodymium additions:

Results are given in figs. 6.07 & 6.08, table 6.04 and plates 6.17 to 6.19.

6.3.5 Effects of misch metal additions:

Results are given in figs. 6.09 & 6.10, table 6.05 and plates 6.20 to 6.25.

6.3.6 Effectsof lancer amp 2 and rare earth fluorides additions :

Results are given in figs. 6.11 and 6.12, table 6.06 and plates 6.26 to 6.35.

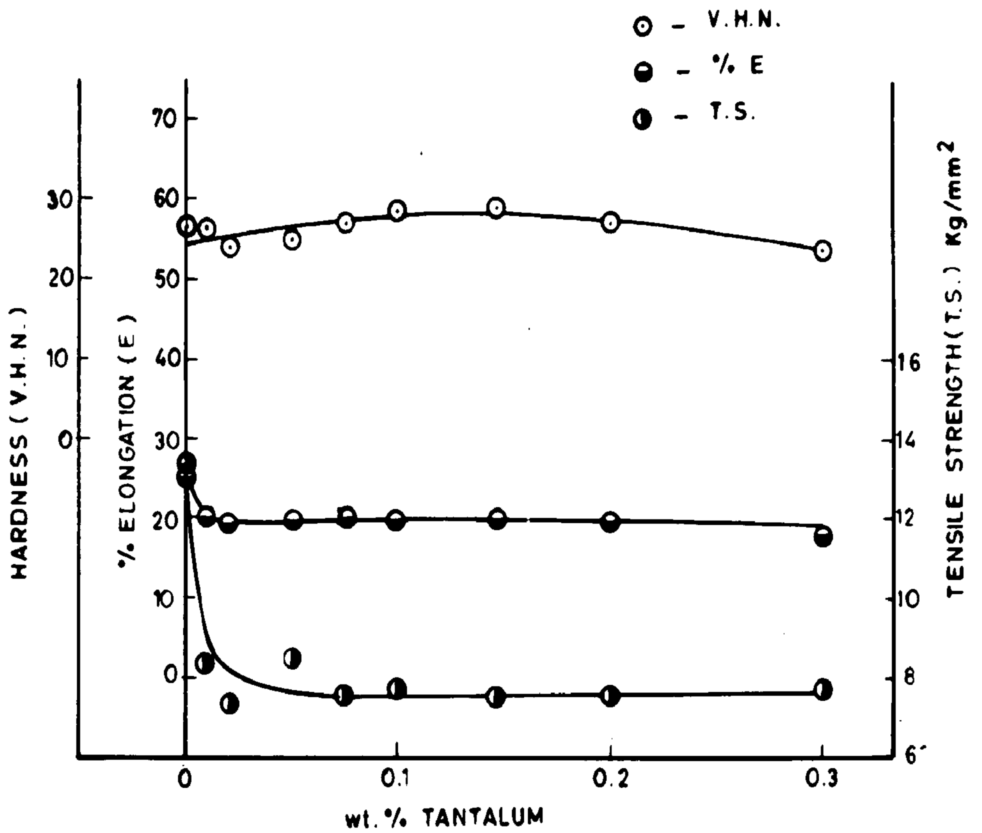


FIG.6.01_EFFECT OF TANTALUM ON THE MECHANICAL PROPERTIES OF ALUMINIUM

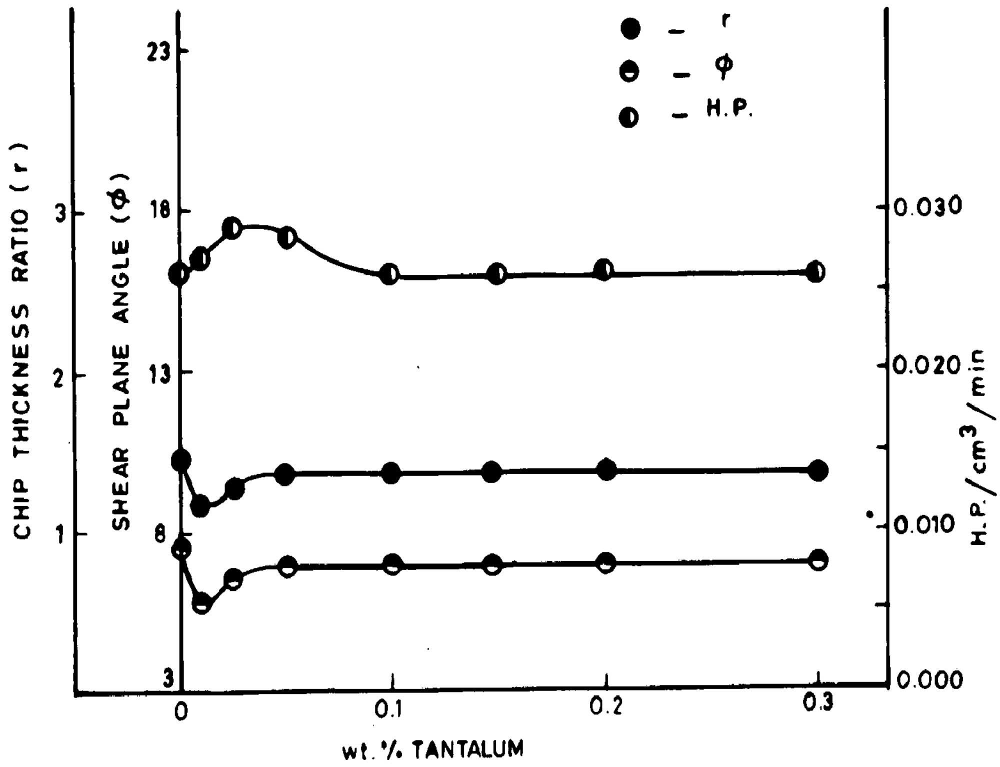


FIG.6.02_EFFECT OF TANTALUM ON THE MACHINABILITY OF ALUMINIUM

TABLE - 6.01

EFFECTS OF TANTALUM ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALUMINIUM

S. No.	% Alloying Element	Tensile Strength kgm/mm^2	% elongation	Hardness V.H.N.	Chip thickness ratio	Shear plane angle ϕ^o	HP/cm^3 /min
1	0.000	13.3	25.0	27	0.145	7°32'	0.0259
2	0.010	8.3	20.0	26	0.113	5°46'	0.0271
3	0.025	7.4	18.8	24	0.125	6°25'	0.0291
4	0.050	8.5	20.0	25	0.134	6°55'	0.0230
5	0.075	7.6	20.0	27	0.131	6°45'	0.0278
6	0.100	7.7	20.0	28	0.134	6°55'	0.0261
7	0.150	7.5	20.0	29	0.134	6°55'	0.0261
8	0.200	7.7	20.0	27	0.134	6°55'	0.0261
9	0.300	7.6	17.5	24	0.134	6°55'	0.0261

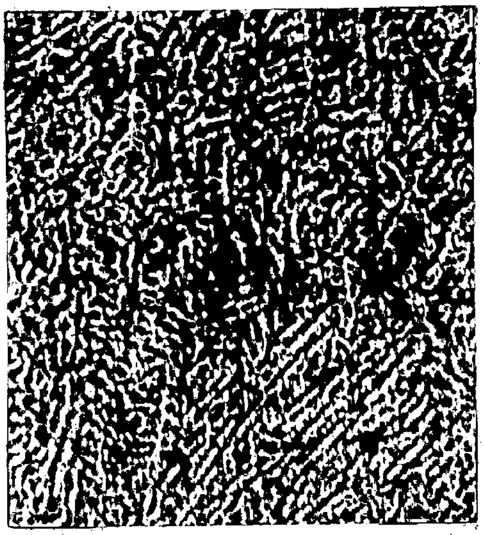


PLATE 6.01 70 X

PURE ALUMINIUM

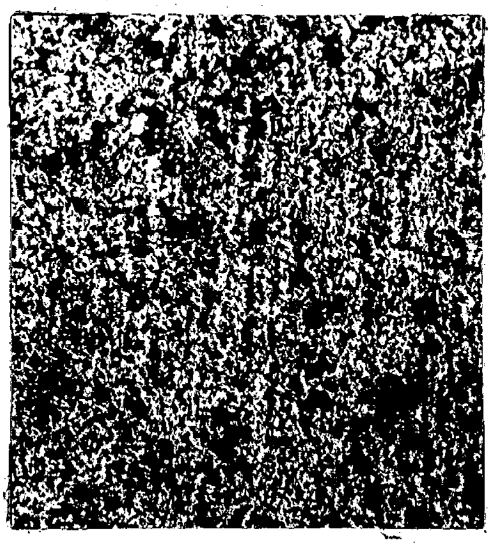


PLATE 6.02 70 X

Al + 0.01 wt.% Ta

PLATE 6.03 70 X

Al +0.025 wt. % Ta

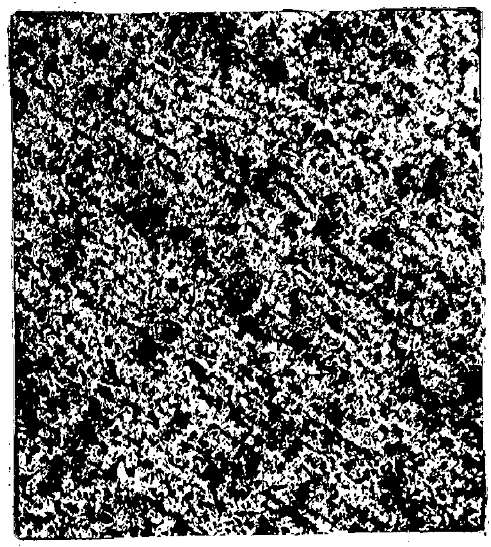


PLATE 6.04 70 X

Al + 0.05 wt. % Ta

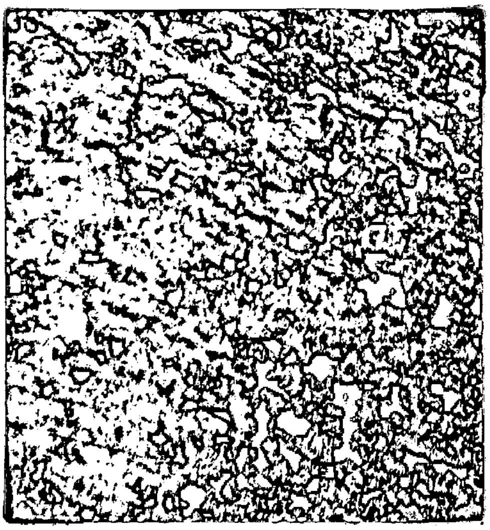


PLATE 6.05 70 X

Al + 0.075 wt.% Ta

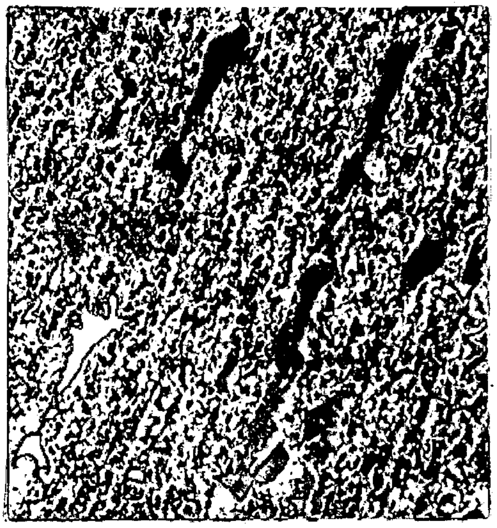


PLATE 6.06 70 X

Al + 0.1 wt. % Ta

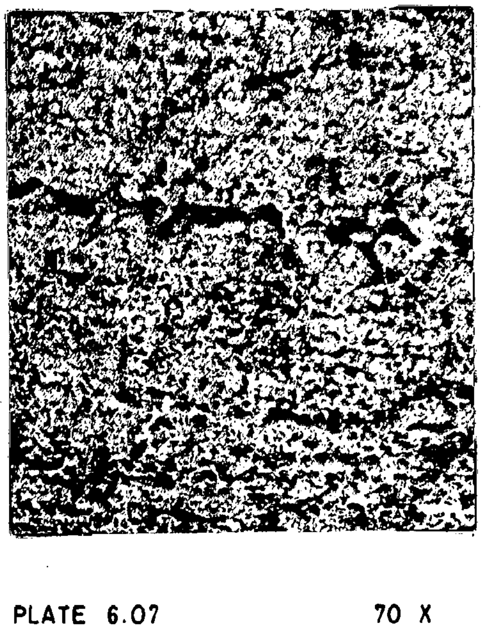


PLATE 6.07 70 X

Al + 0.15 wt. % Ta

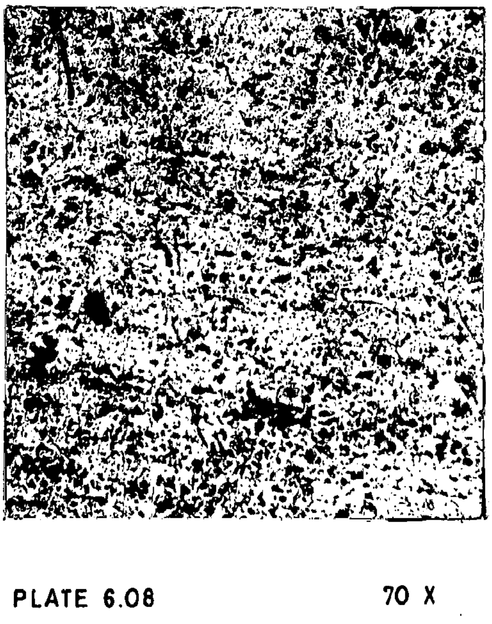


PLATE 6.08 70 X

Al + 0.3 wt. % Ta

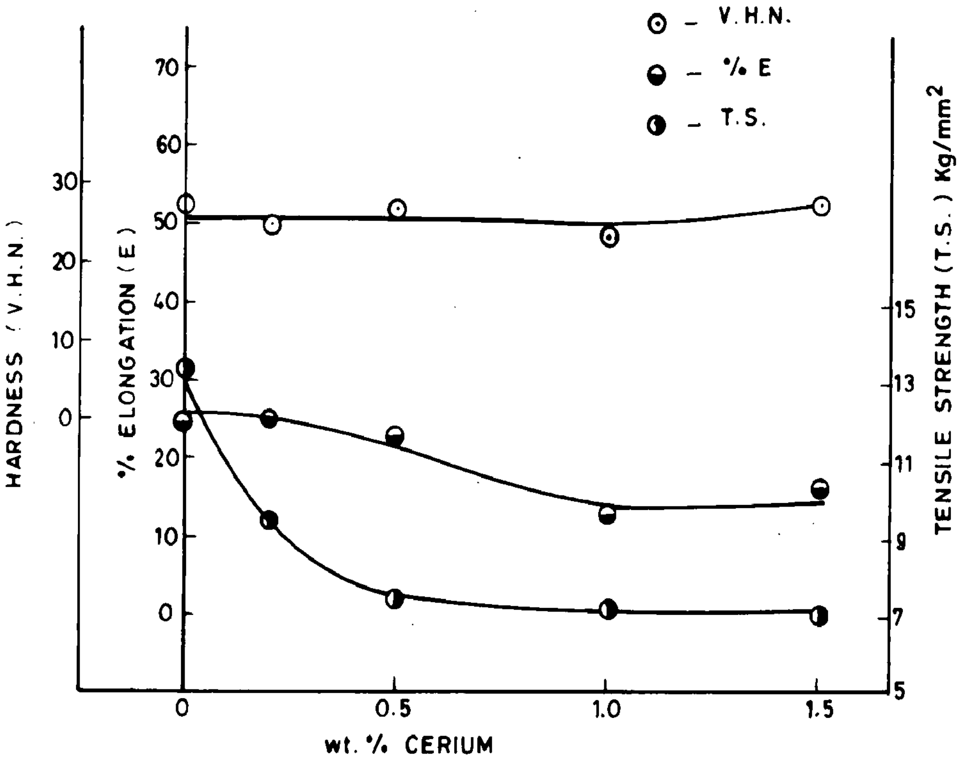


FIG.6.03_EFFECT OF CERIUM ON THE MECHANICAL PROPERTIES OF ALUMINIUM

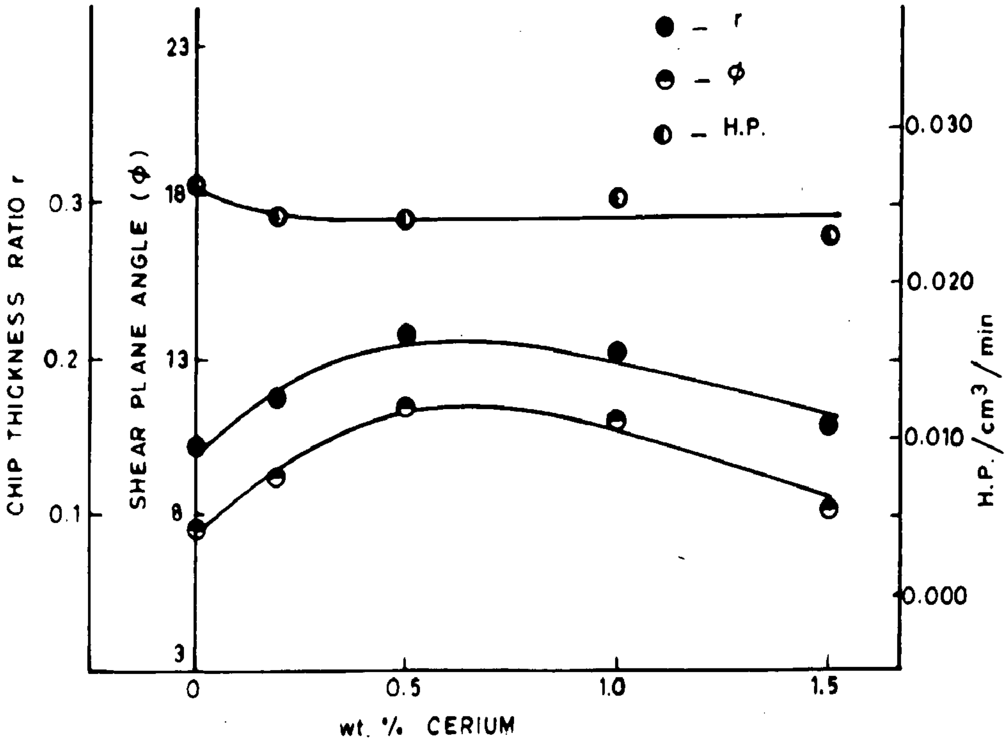


FIG.6.04_EFFECT OF CERIUM ON THE MACHINABLITY OF ALUMINIUM

TABLE 6.02

EFFECTS OF CERIUM ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALUMINIUM

S. no.	% Alloying element	Tensile strength kgm/mm^2	% elongation	Hardness VHN.	Chip thickness ratio	Shear plane angle Ø°	HP/cm^3/min.
1	0.0	13.3	25.0	27	0.145	7°32'	0.0259
2	0.2	7.5	24.0	25	0.174	9°11'	0.0240
3	0.5	7.4	23.8	27	0.213	11°26'	0.0240
4	1.0	7.4	13.8	24	0.206	11°02'	0.0250
5	1.5	7.1	17.5	28	0.157	8°12'	0.0232

TABLE 6.03

EFFECTS OF LANTHANUM ADDITIONSON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALUMINIUM.

S. No.	% Alloying element	Tensile strength kgm/mm^2	% elongation	Hardness VHN.	Chip thickness ratio	Shear plane angle Ø°	HP/cm^3/min.
1	0.00	13.3	25.0	27	0.145	7°32'	0.0259
2	0.05	7.8	17.5	22	0.150	7°49'	0.0241
3	0.10	7.8	20.0	20	0.150	7°49'	0.0253
4	0.20	7.8	25.0	25	0.146	7°32'	0.0258
5	0.30	6.7	30.0	21	0.141	7°19'	0.0272

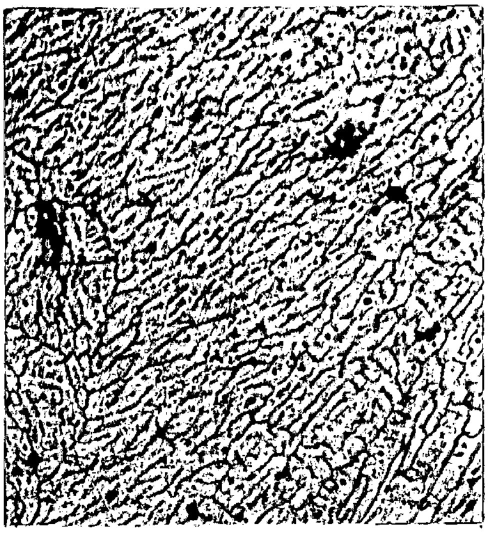


PLATE 6.09 70 X

Al +0.2 wt.% Ce

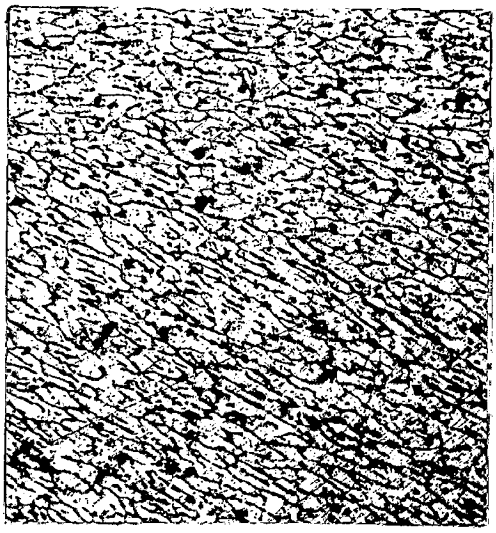


PLATE 6.10 70 X

Al +0.5 wt.% Ce

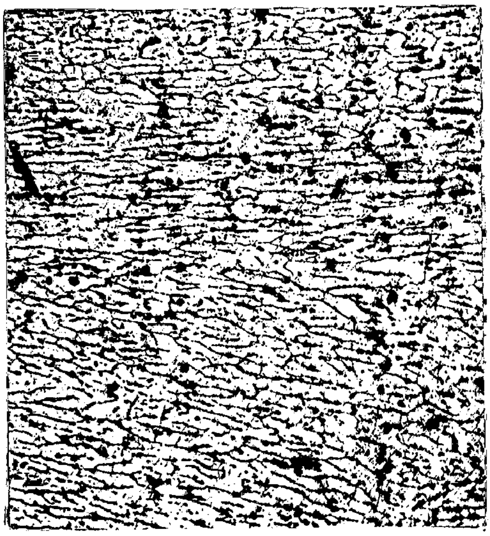


PLATE 6.11 70 X

Al +1.0 wt.% Ce

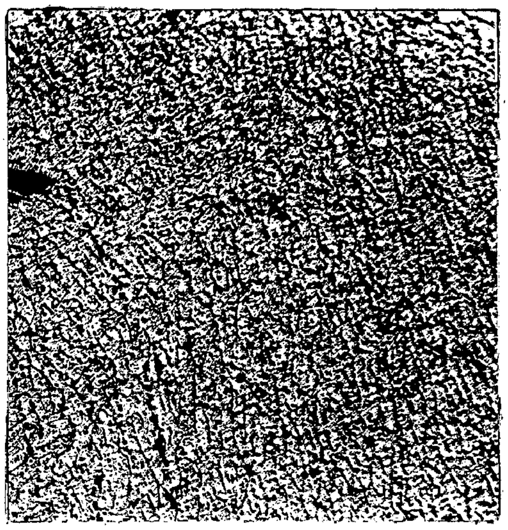


PLATE 6.12 70 X

Al+1.5 wt.% Ce

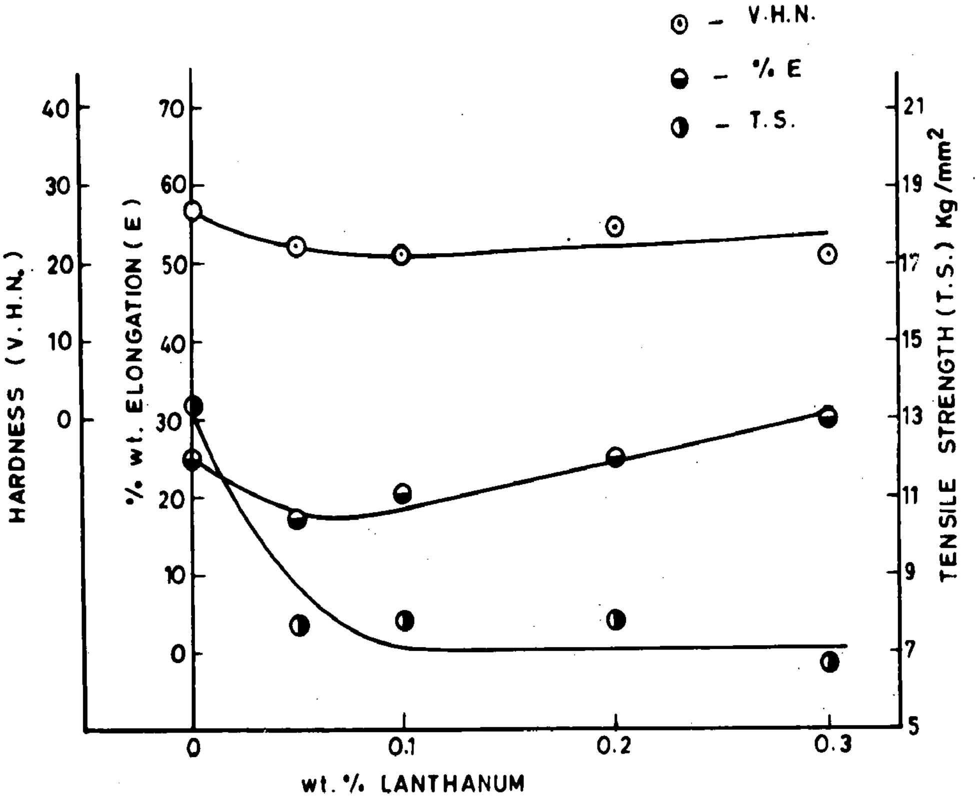


FIG.6.05_EFFECT OF LANTHANUM ON THE MECHANICAL PROPERTIES OF ALUMINIUM

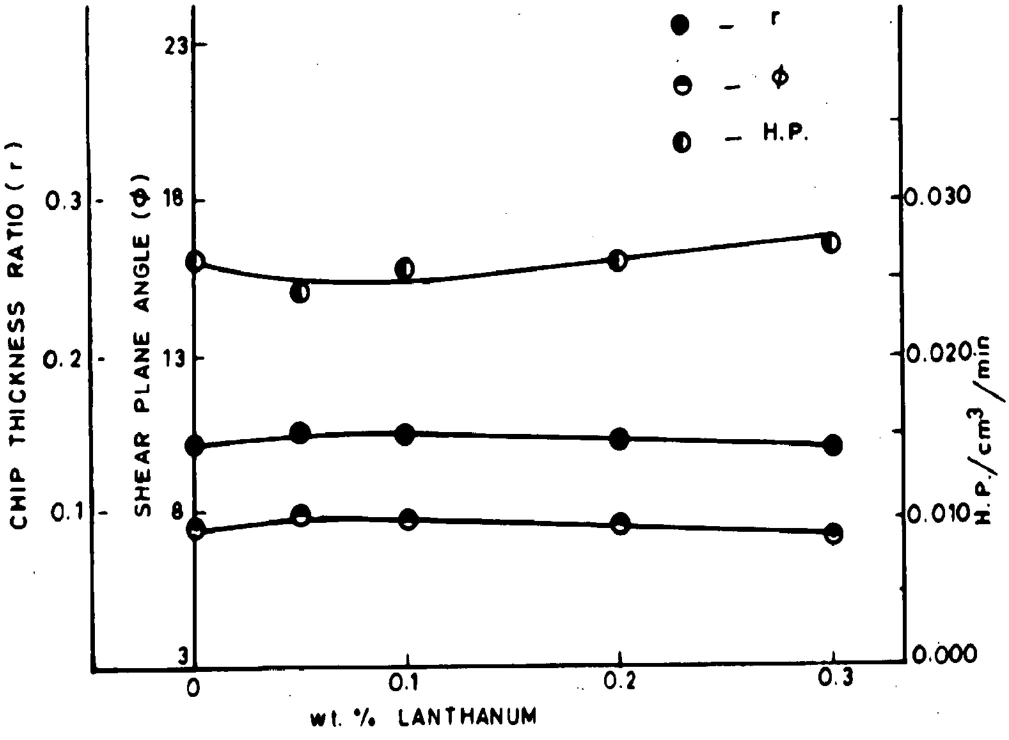


FIG. 6.06_EFFECT OF LANTHANUM ON THE MACHINABILITY OF ALUMINIUM

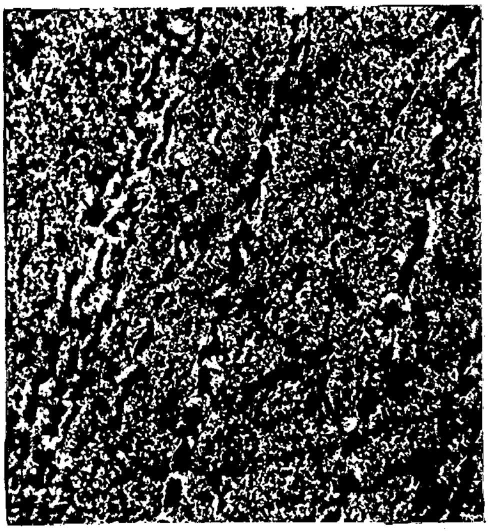


PLATE 6.13 70 X

Al + 0.05 wt. % La

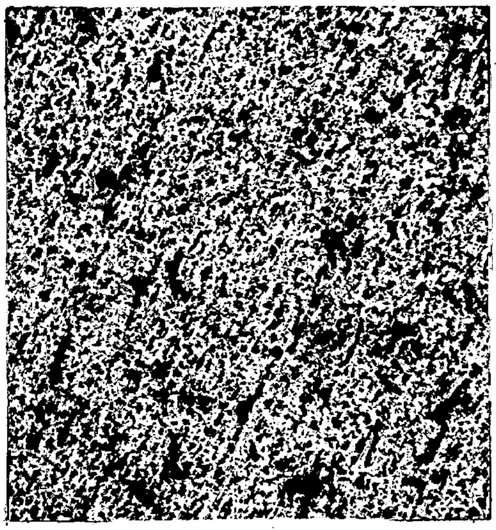


PLATE 6.14 70 X

Al + 0.1 wt. % La

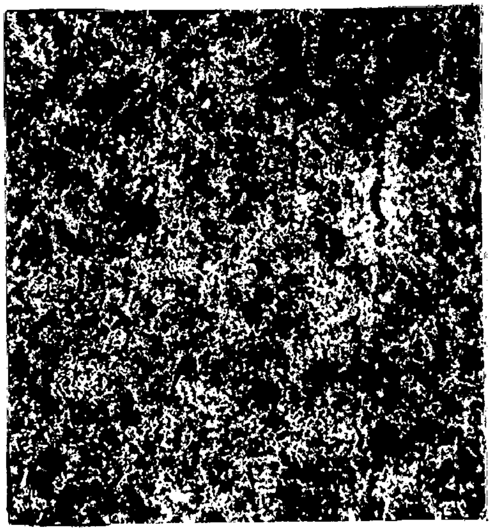


PLATE 6.15 70 X

Al + 0.2 wt. % La

PLATE 6.16 70 X

Al + 0.3 wt. % La

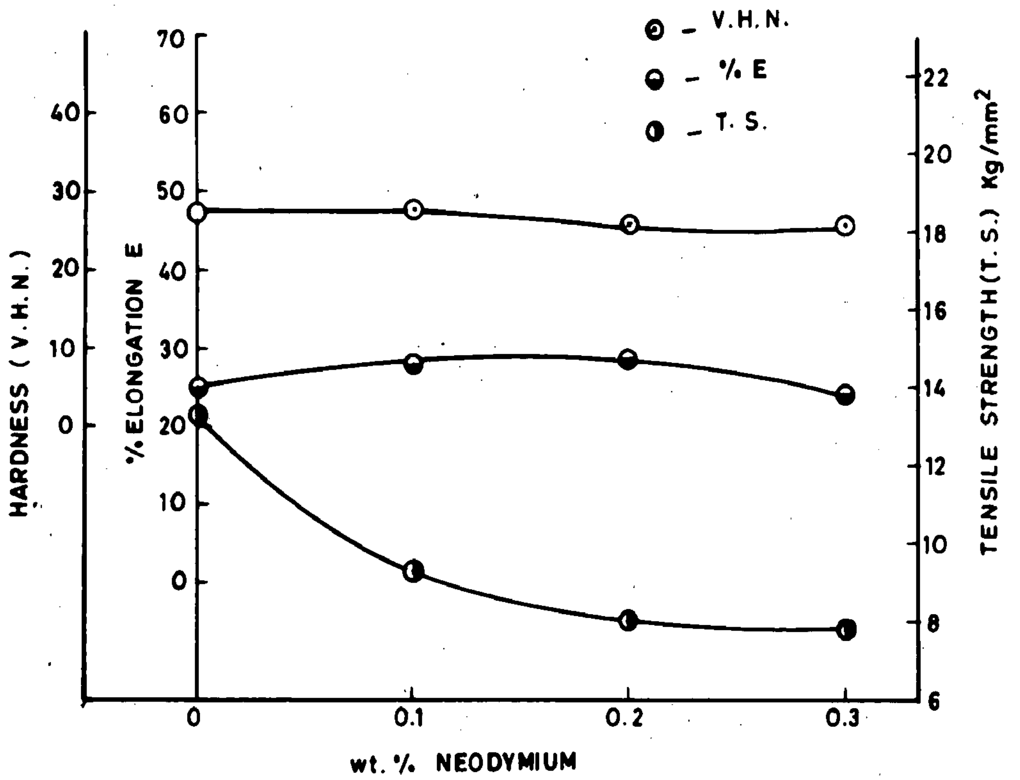


FIG.6.07.EFFECT OF NEODYMIUM ON THE MECHANICAL PROPERTIES OF ALUMINIUM

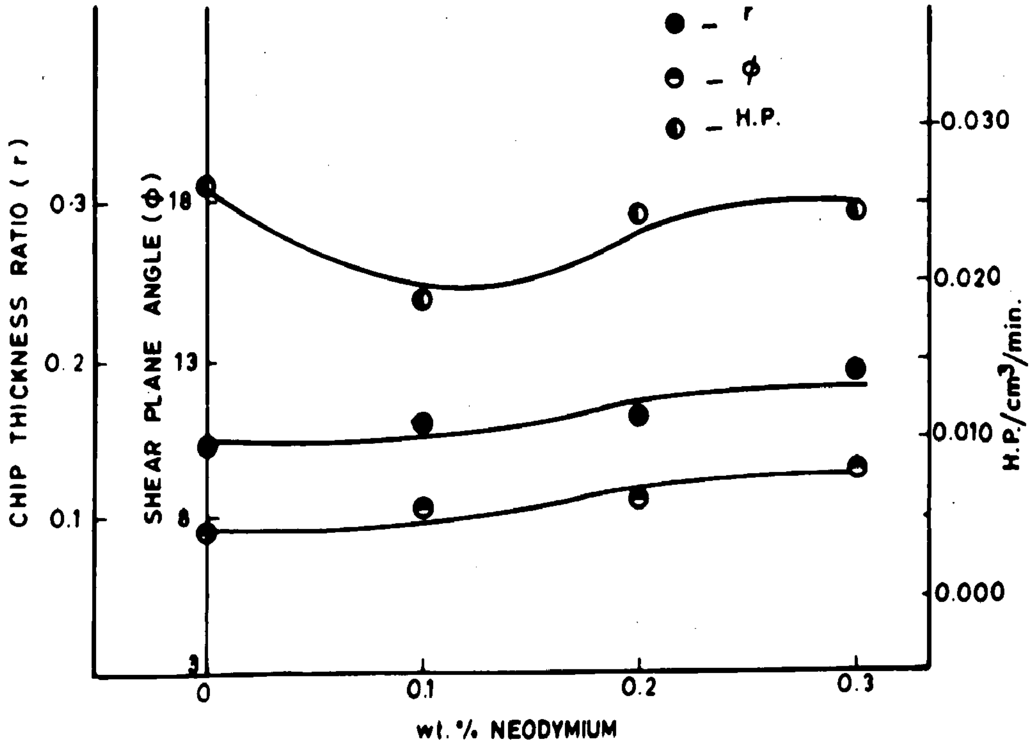


FIG.6.08_EFFECT OF NEODYMIUM ON THE MACHINABILITY OF ALUMINIUM

TABLE - 6.04

EFFECTS OF NEODYMIUM ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALUMINIUM.

S. No.	% Alloying element	Tensile strength kgm/mm^2	% elongation	Hardness VHN	Chip thickness ratio	Shear plane angle ϕ°	HP/cm^3/min.
1	0.00	13.3	25.0	27	0.145	$7^\circ 32'$	0.0259
2	0.10	8.5	28.0	28	0.158	$8^\circ 16'$	0.0188
3	0.20	8.0	29.0	26	0.163	$8^\circ 33'$	0.0242
4	0.30	7.8	26.0	26	0.181	$9^\circ 34'$	0.0243

TABLE 6.05

EFFECTS OF MISCH METAL ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALUMINIUM

S. No.	% Alloying element	Tensile strength kgm/mm^2	% elongation	Hardness VHN	Chip thickness ratio	Shear plane angle ϕ°	HP/cm^3/min.
1	0.00	13.3	20.0	27.0	0.145	$7^\circ 32'$	0.0259
2	0.20	12.3	24.0	41.0	-	-	-
3	0.50	12.0	27.0	44.0	0.335	$20^\circ 17'$	0.0188
4	0.75	10.6	19.5	44.0	-	-	-
5	1.50	11.1	13.5	44.0	0.410	$23^\circ 54'$	0.0211
6	2.00	10.9	12.0	45.0	-	-	-
7	2.50	10.9	12.0	44.0	0.418	$24^\circ 26'$	0.0200
8	3.00	10.9	12.0	47.0	0.420	$24^\circ 34'$	0.0198

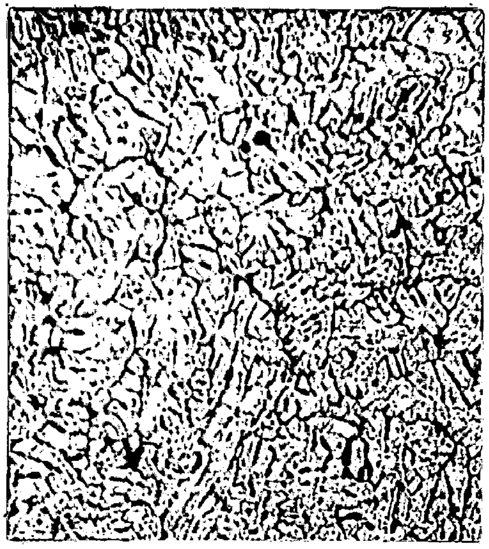


PLATE 6.17 70 X

Al + 0.1 wt. % Nd

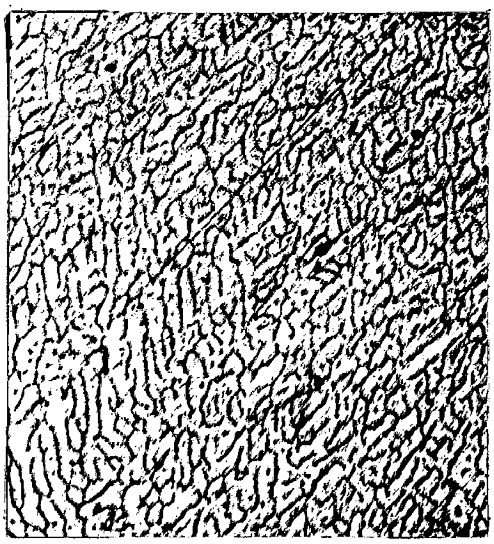


PLATE 6.18 70 X

Al + 0.2 wt. % Nd

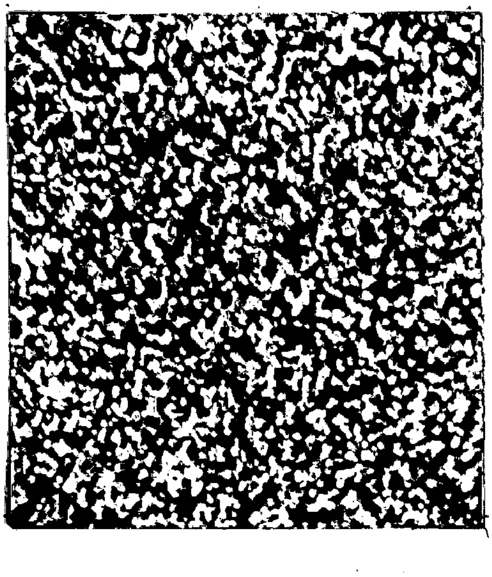


PLATE 6.19 70 X

Al + 0.3 wt. % Nd

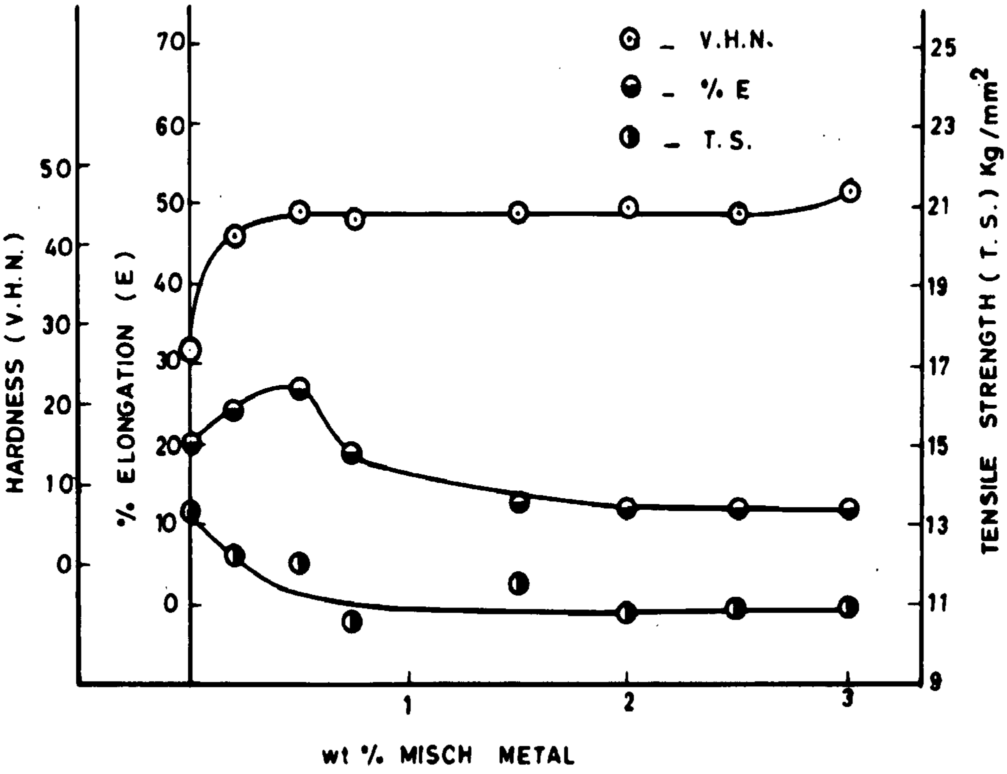


FIG.6.09_FFFECT OF MISCH METAL ON THE MECHANICAL PROPERTIES OF ALUMINIUM.

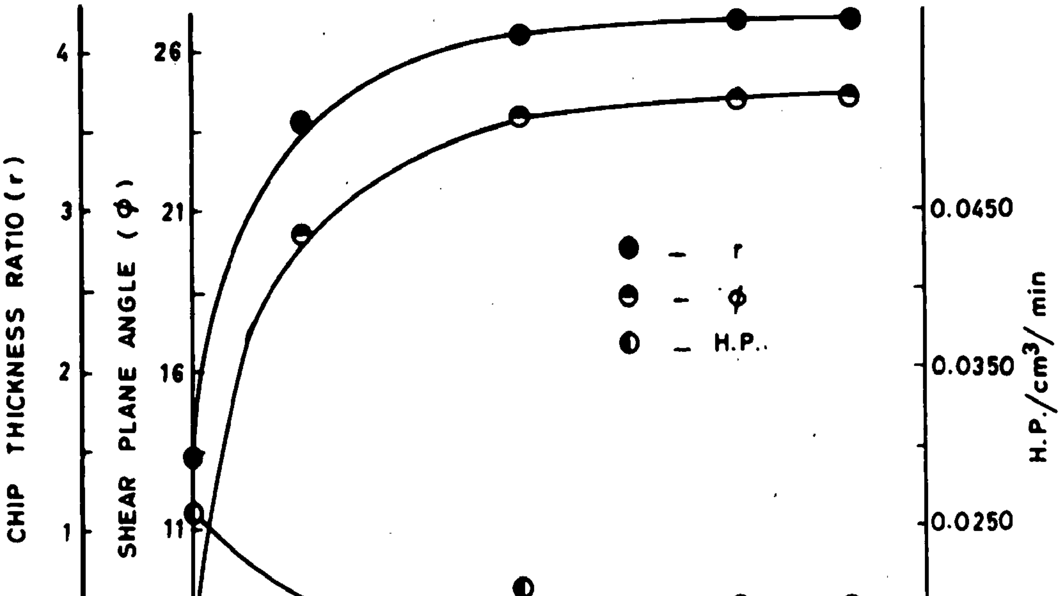


FIG. 6.10_EFFECT OF MISCH METAL ON THE MACHINABILITY OF ALUMINIUM

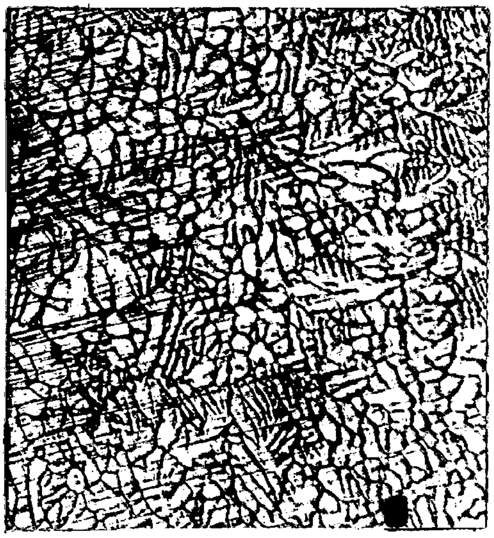
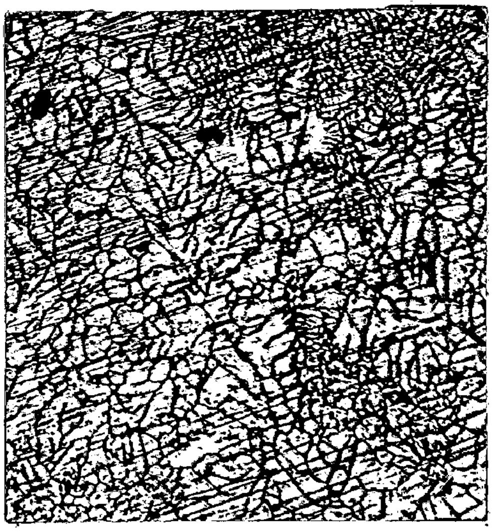


PLATE 6.20 106 X

PURE ALUMINIUM

PLATE 6.21 106 X

Al + 0.5 wt. % MISCH METAL



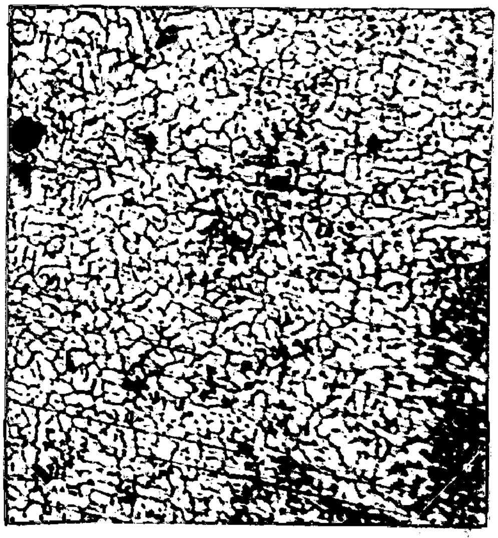


PLATE 6.22 106 X

Al + 0.75 wt. % MISCH METAL

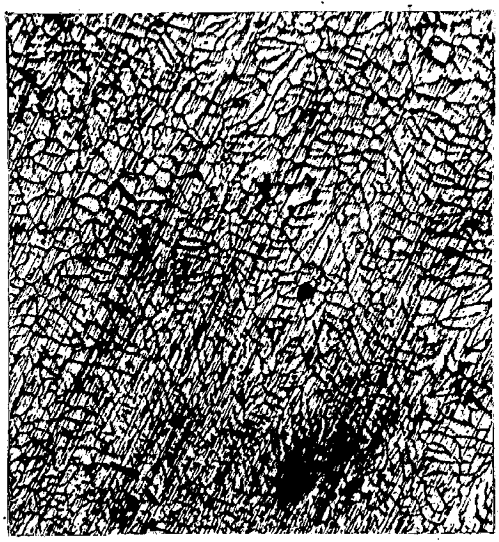


PLATE 6.23 106 X

Al + 1 wt. % MISCH METAL

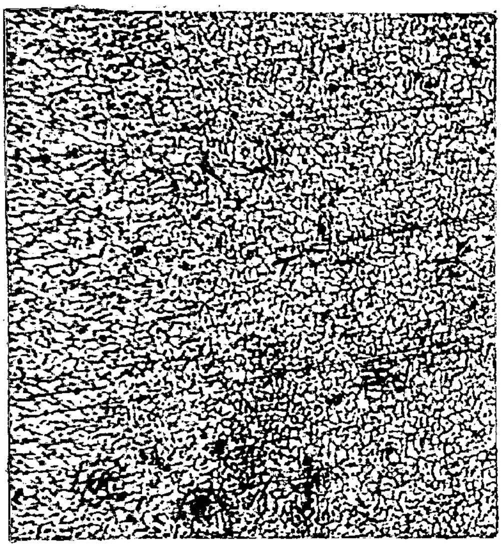


PLATE 6.24 106 X

Al + 2 wt. % MISCH METAL

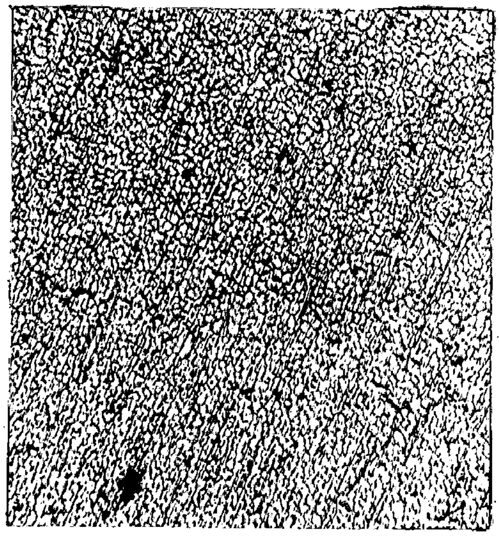


PLATE 6.25 106 X

Al + 3 wt. % MISCH METAL

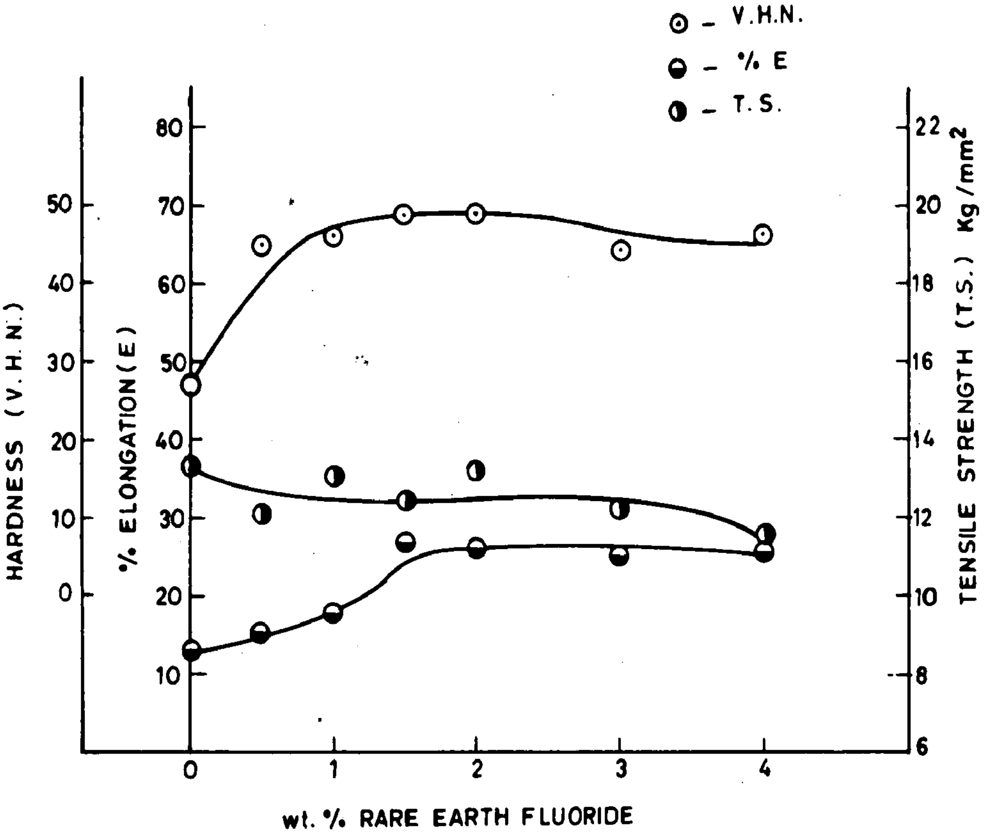


FIG.6.11_EFFECT OF RARE EARTH FLUORIDE ON THE MECHANICAL PROPERTIES OF ALUMINIUM

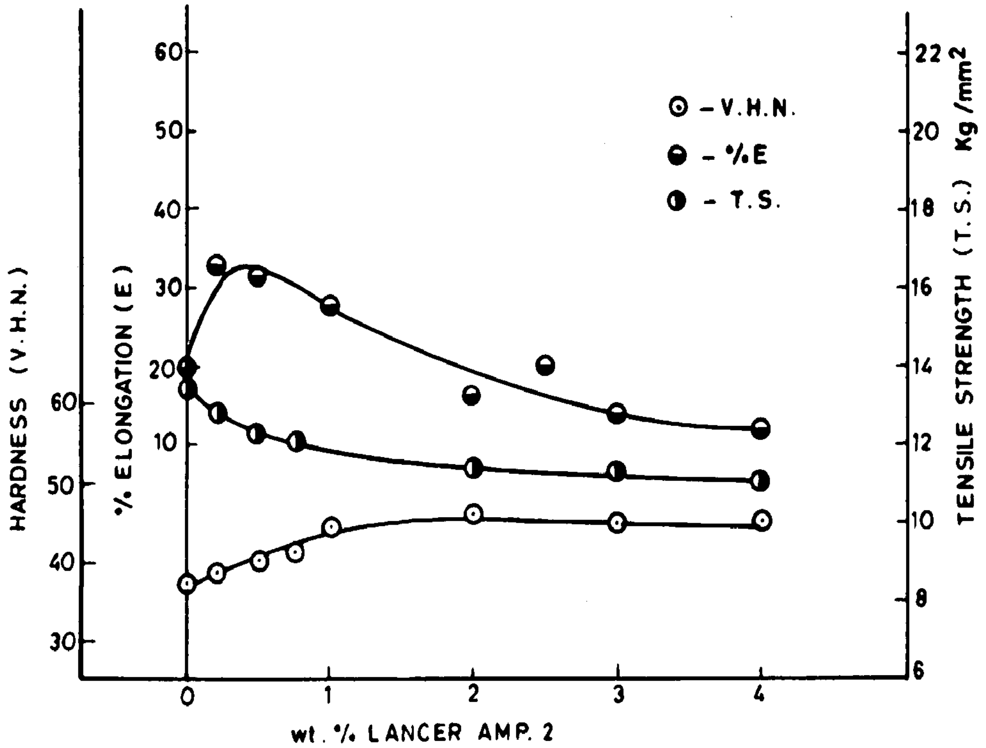


FIG.6.12_EFFECT OF LANCER AMP. 2 ON THE MECHANICAL PROPERTIES OF ALUMINIUM

TABLE 6.06

EFFECTS OF RARE EARTH FLUORIDES OF LANCER-AMP 2 ADDITIONS ON THE MECHANICAL PROPERTIES OF ALUMINIUM.

S. No.	% alloying element	Rare earth fluoride			Lancer amp. 2		
		T.S. kgm/mm^2	% elongation	V.H.N	T.S. kgm/mm^2	% elongation	V.H.N
1	0.00	13.32	13	27	13.3	26	27
2	0.20	-	-	-	12.8	33.0	39
3	0.50	12.2	15	45	12.3	31.0	40
4	0.75	-	-	-	12.2	-	41
5	1.00	13.0	18	46	-	28.0	44
6	1.50	12.5	27	49	-	-	-
7	2.00	13.2	26	49	11.0	16.0	46
8	2.50	-	-	-	11.3	20.0	-
9	3.00	12.20	25	44	11.3	14.0	45
10	4.00	11.6	26	46	11.1	12.0	45

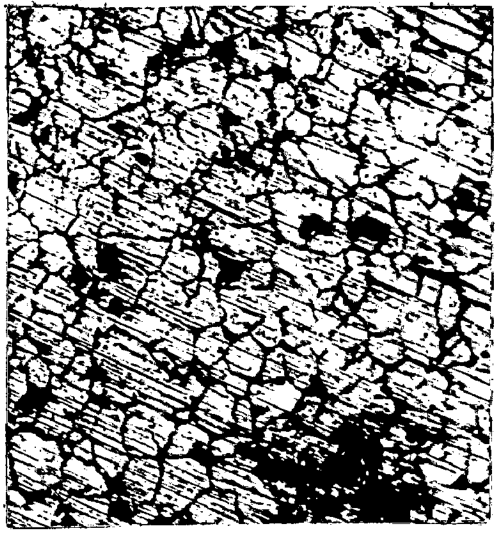


PLATE 6.26 106 X

Al + 0.2 wt. % LANCER AMP.2

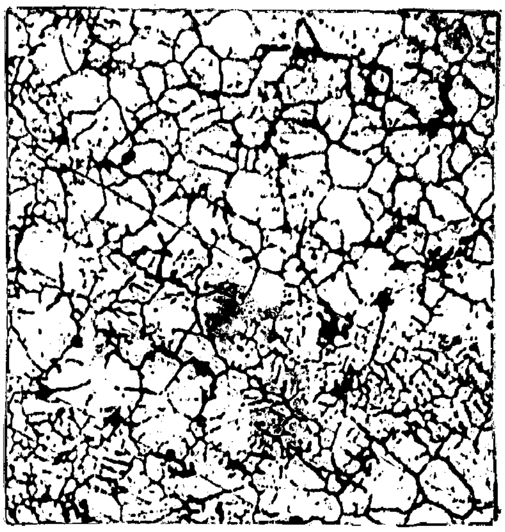


PLATE 6.27 106 X

Al + 0.5 wt. % LANCER AMP.2

PLATE 6.28 106 X

Al + 0.75 wt. % LANCER AMP.2

PLATE 6.29 106 X

Al + 1 wt. % LANCER AMP. 2

PLATE 6.30 106 X

Al + 2 wt. % LANCER AMP. 2

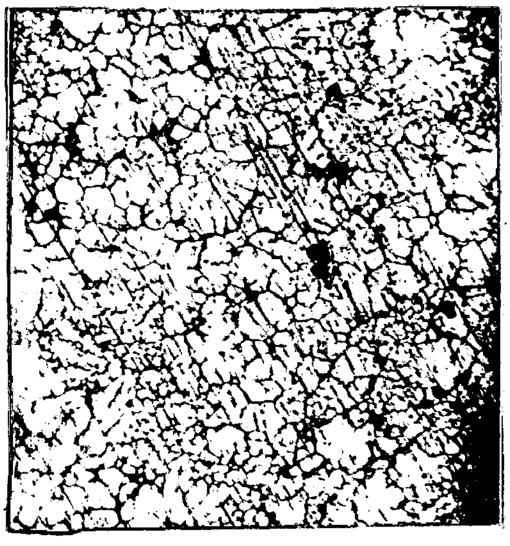


PLATE 6.31 106 X

Al + 3 wt. % LANCER AMP. 2

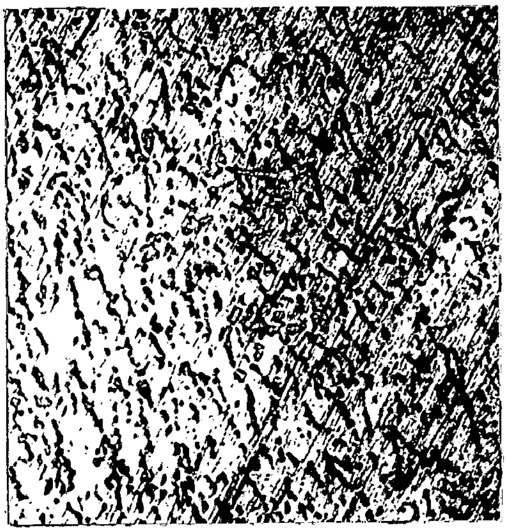


PLATE 6.32 106 X

Al + 1.0 wt.% R.E. FLUORIDE

PLATE 6.33 106 X

Al + 1.5 wt.% R.E. FLUORIDE

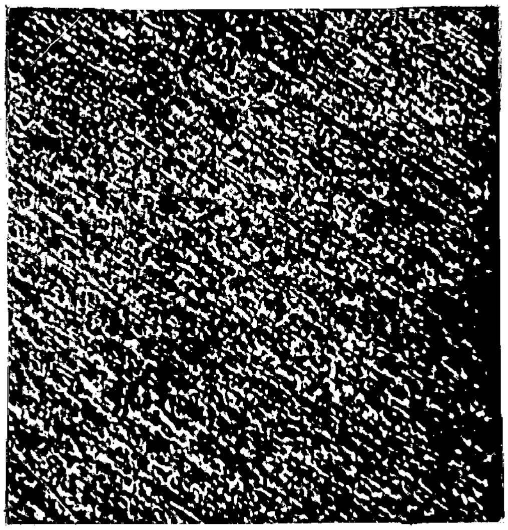


PLATE 6.34 106 X

Al + 2.0 wt.% R.E. FLUORIDE

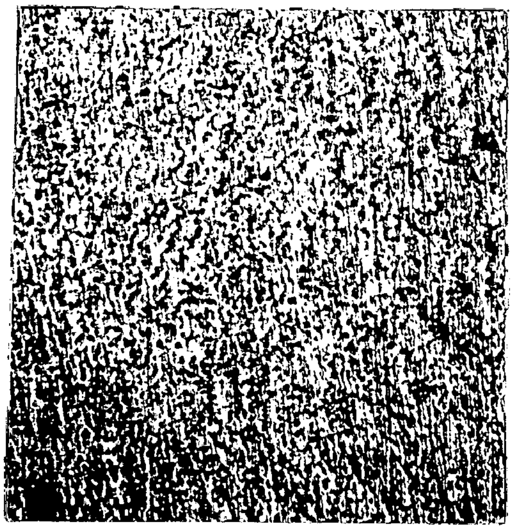


PLATE 6.35 106 X

Al + 4.0 wt.% R.E. FLUORIDE

6.4 DISCUSSION OF THE RESULTS .

6.4.1 Effects of tantalum additions :

The salient features of the results are given below :

i) On metallographic study it is observed that there is large refining even with minor addition of tantalum upto 0.01%. With further addition little detectable effect is observed on the grain size (plates 6.01 to 6.08).

ii) No beneficial effect is produced on the tensile strength. It decreases to a constant value at about .05% addition.

iii) The % elongation curve also follows the pattern of the tensile strength curve. It is found to decrease to a constant value at about .025% addition.

iv) The hardness increases slightly upto about 0.15% addition beyond which it decreases.

v) Both chip thickness ratio and shear plane angle Ø decrease slightly to constant value, at about 0.1% addition resulting in decrease of machinability.

vi) HP/cm^3/min. is found to increase upto 0.025% addition indicating the detrimental effect upto 0.025% addition. Beyond 0.025% addition the value drops down to a constant value at about 0.1% addition. Value is however, slightly more than that of original alloy indicating that there is no beneficial effect of tantalum addition upto 0.3% on the machinability from power considerations.

Tantalum is found to have no solid solubility in aluminium. Marignac (154) and Braner (155) have reported the existence of the compound $TaAl_3$ formed at 69.09 weight percentage. Thus with the addition of tantalum there is possibility of formation of very small amount of new phase, the amount of which should increase with the increased addition. The increase in hardness can easily be attributed to the formation of the fine particles of these intermetallic compounds. The decrease in hardness beyond 0.15% may be due to large coarsening of the precipitate of the intermediate phase. The refining action may also be understood, as due to particles of these compounds acting as nuclei for re-crystallization.

The increase in machinability beyond 0.025% addition can easily be explained on the basis of the following two major points:-

i) The refinement effect produced due to tantalum addition should bring about increase in machinability.

ii) The increasing addition of tantalum may bring about the increase in the hardness resulting in decrease of machinability.

It is interesting to note that beneficial effect due to refining predominates over the adverse effect due to the presence of hard $TaAl_3$ phase resulting in increase of machinability upto about 0;1% addition. With further addition the effects seem to balance each other and machinability becomes constant.

6.4.2 Effects of cerium additions :

The salient features of the results are given below :-

i) An important effect of cerium addition is the refinement of grain structure. Addition at 1.5% cerium gives a very fine grain size. (Plates 6.09 to 6.12).

ii) Tensile strength is found to decrease to a constant value at about 0.5% addition.

iii) The percentage elongation remains almost unchanged upto 0.5% addition beyond which it decreases to a constant value at about 1.5% addition.

iv) There is slight but regular increase in hardness value with cerium addition. The rate appears to be nearly linear.

v) Both chip thickness ratio and shear plane angle Ø increase to constant values at about 0.5% addition.

vi) HP/cm^3/min. first decreases upto 0.2%. Between 0.2 and 1% the value is almost constant. Beyond 1% addition it again goes on decreasing slightly.

The refining action and hardness increase are not surprising. Cerium is reported to dissolve upto 0.05% by weight in aluminium in the solid state. Four intermetallic phases are reported; Ce_3Al, CeAl, $CeAl_2$ and $CeAl_4$(84). So with the addition of cerium, very little of it goes into solid solution, most of the amount reacting with aluminium to form the intermetallic compounds the predominating being $CeAl_4$. The increase in hardness can easily be attributed to the formation of fine particles of these intermetallic compounds. Their presence, however, could not be detected in our metallographic study, possibly because of their submicroscopic size. The refining action may also be understood due to the particles of these compounds acting as nuclei during solidification.

Figure 8.04 gives the values of chip thickness ratio, shear plane angle Ø and HP unit for the aluminium treated with cerium. It can easily be seen that both chip

v) Both chip thickness ratio and shear plane angle Ø increase to constant values at about 0.5% addition.

vi) HP/cm^3/min. first decreases upto 0.2%. Between 0.2 and 1% the value is almost constant. Beyond 1% addition it again goes on decreasing slightly.

The refining action and hardness increase are not surprising. Cerium is reported to dissolve upto 0.05% by weight in aluminium in the solid state. Four intermetallic phases are reported; Ce_3Al, CeAl, $CeAl_2$ and $CeAl_4$(84). So with the addition of cerium, very little of it goes into solid solution, most of the amount reacting with aluminium to form the intermetallic compounds the predominating being $CeAl_4$. The increase in hardness can easily be attributed to the formation of fine particles of these intermetallic compounds. Their presence, however, could not be detected in our metallographic study, possibly because of their submicroscopic size. The refining action may also be understood due to the particles of these compounds acting as nuclei during solidification.

Figure 8.04 gives the values of chip thickness ratio, shear plane angle Ø and HP unit for the aluminium treated with cerium. It can easily be seen that both chip thickness ratio and shear plane angle Ø increase to constant values at about 0.5% addition showing beneficial effects of cerium addition on machinability from chip formation considerations. Beyond 0.5% addition the values become constant. The results of HP unit also indicate the beneficial effects of cerium addition upto 0.5% on the machinability. The improvement in the machinability upto 0.5% can easily be understood on the basis of the following two considerations :-

i) Large refining effect produced due to cerium addition results in increase of machinability.

ii) With the increasing cerium content in aluminium the amount of harder submicroscopic intermediate phases finally dispersed in the matrix should result in improvement of hardness but corresponding decrease in machinability.

It appears that for the lower additions upto about 0.5% the beneficial effect due to large refinement of grain predominates over the adverse effect due to intermediate phases, the amount of which will be smaller resulting in the improvement in machinability from both chip formation study and power considerations. For higher additions due to large increase in the amount of intermediate phases beyond 0.5 to 1%, the two effects seem to balance each other resulting in constant value of machinability.

6.4.3 Effects of lanthanum additions :

The salient features of the results are given below :

i) On metallographic study it is observed that there is large refining effect of lanthanum addition upto about 0.1%, beyond which there appears to be the presence of some additional phases in the microstructure which seems to increase with further addition (Plates 6.13 to 6.16)

ii) There is large decrease in tensile strength upto about 0.1% addition beyond which it becomes constant.

iii) The % elongation is found to decrease upto 0.05% addition beyond which it goes on increasing with further additions. The rate of increase is nearly linear.

iv) There is slight effect of lanthanum addition on the hardness value. The value slightly decreases upto 0.1% beyond which it increases slightly with further additions.

v) Both chip thickness ratio and shear plane angle Ø slightly increase upto 0.1% addition beyond which they become constant with further additions.

vi) H.P./cm3/min. is found to decrease slightly upto 0.05% addition beyond which there is gradual but regular increase in the horse power value with lanthanum addition.

Lanthanum is found to have no solid solubility in aluminium. Five intermediate phases are reported La_3Al, La_3Al_2, La Al, $LaAl_2$, and $LaAl_4$ the predominating phase being Al_4La. Thus with lanthanum addition there is possibility of the formation of these phases, the predominating being Al_4La, the amount of which should increase with increased addition. Presence of some additional phases has already been detected in our metallographic study, although the exact nature could not be detected and needs careful X-ray or electron study.

The slight increase in hardness may be due to the presence of these phases. The large refining effect produced may also be due to the fine particles of these phases, acting as nuclei, during solidification.

Fig. 6.06 indicates the beneficial effect of lanthanum addition upto 0.05% on the machinability. Both chip thickness ratio and shear plane angle Ø increase, while HP unit decreases with 0.05% lanthanum addition. As explained earlier the increase in machinability may be due to large refining effect, predominating over the adverse effect, due to the phases the amount of which should be very small. With larger additions the adverse effect due to increasing amount of phases predominates over the beneficial effect due to the refining, resulting in decrease of machinability.

6.4.4 Effects of neodymium additions :

The salient features of the results are given below :-

i) There is large refinement of grain structure with neodymium addition at about 0.2%. The effect is not so pronounced as in the case of cerium addition (Plates 6.17 to 6.19).

ii) The tensile strength is found to decrease to a constant value at about 0.2% addition.

iii) The percentage elongation increases upto 0.2% addition. Beyond 0.2% the value is found to decrease with increased addition.

iv) There is no effect of neodymium addition on the hardness value upto 0.1%. With further addition the value is found to decrease slightly to a constant value at about 0.3% addition.

v) Both chip thickness ratio and shear plane angle ϕ increase with neodymium addition upto 0.3% addition.

vi) There is large decrease in the value of H.P/cm^3/min. upto about 0.1% addition. Beyond 0.1% the value increases to a constant value at about 0.2% addition.

The maximum solid solubility of neodymium in aluminium is found to be 0.2%. Following phases have been reported in the system $NdAl$, $NdAl_2$ and $NdAl_4$. Savitskii and others have observed the strengthening effect of Nd addition on aluminium (97). They have reported an increase in hardness from 25 to 155 and decrease in plasticity by 5 to 10 %.

In the present investigation no detectable effect on the hardness value has been observed upto about 0.2% addition. The percentage elongation is found to increase upto about 0.2%, beyond which it decreases slightly, the maximum increase being by about 4%. The result is not in agreement with the results of Savitskii and others who have reported definite increase in hardness and decrease in plasticity with Nd addition.

Since the solubility of Nd in aluminium is 0.2%, there is no possibility of formation of any intermediate phase upto 0.2%. Even the amount of intermediate phases at 0.3% addition will be very small. As explained in the previous cases, the increase in hardness is due to the presence of fine dispersion of intermediate phases, it is expected that with Nd addition no change in hardness should take place. The results are in concurrence with the above argument. It is interesting to note that little refinement has occured upto 0.2% Nd addition. Effect is not so pronounced as in the case of cerium addition since the amount of intermediate phases acting as nuclei during solidification will be negligibly small in case of Al-Nd alloy.

It can easily be seen that large beneficial effect of Nd addition on the machinability has been observed both from chip formation and power considerations. The increase in machinability with Nd addition may be attributed to the Nd going into solid solution and slight refinement in the grain structure. A slight increase in power consumption beyond 0.2% may be due to the presence of small amounts of hard intermediate phases bringing about a slight adverse effect on machinability.

6.4.5 Effects of misch metal :

The salient features of the results are given below:-

i) An important effect of misch metal addition is the refinement of grain structure. There is little effect upto 1.5% addition. Addition between 2 to 3 % gives a very fine grain size. (Plates 6.20 to 6.25)

ii) There is very little effect on tensile-strength at room temperature. It decreases slightly upto about 1% and then remains almost unchanged.

iii) The % elongation increases upto 0.5% addition and then decreases to a constant value.

iv) There is a regular increase in the hardness of aluminium with the addition of misch metal.

v) Both chip thickness ratio and shear plane angle Ø increase with the increase in misch metal addition, very rapidly upto 0.5% addition, beyond which they increase gradually.

vi) $HP/cm^3/min$ decreases to a constant value at about 1% addition.

There are conflicting reports about the effects of addition of misch metal on the properties of aluminium. Schofield and Wyatt (21) and Danmison & Tull (22) have observed grain refinement in aluminium but Bowen and Bernstein (23) found little or no refining effect. Our investigations confirm the earlier results, although little grain refining effect was observed below 1.5% addition.

The refining effect produced by misch metal can easily be understood on the basis of effects of individual elements contained in the misch metal. 1.5% misch metal will contain nearly 0.75% Ce, 0.3% La and 0.2% Nd. Cerium is reported to dissolve upto 0.05% by weight in aluminium in the solid state, while lanthanum is reported to have no solid solubility in aluminium. Nd is reported to dissolve upto 0.2% by weight. So out of Ce-La-Nd content of misch metal upto 1.5% addition, very little Ce goes into solid solution; most of the Ce and whole amount of La will react with aluminium to form intermetallic compounds Al_4Ce, Al_4La etc., which contain 56.4% Ce and 56.3% La by weight respectively. The whole of Nd will go into solid solution. The slight refinement effect may be understood as due to the particles of these compo-unds acting as nuclei during solidification. With further addition of misch metal the amount of intermediate phases will increase.

Part of neodymium will also react with aluminium to form the intermetallic compounds. With the addition between 2 - 3 % there may be presence of large amount of fine particles of these phases Al_4Ce, Al_4La and little of Al_4Nd acting as nuclei during solidification leading to a large amount of grain refinement.

The fact, that the larger amount of grain refinement as observed on metallographic study is due to cumulative effects of Ce, La, Nd, is also confirmed by seeing separately the effects of these individual elements. It can be seen that additions of all these elements separately have resulted in slight refinement of grain-structure. Effect is not so pronounced as in the case of misch metal, where the large effect may be due to cumulative effects of all the three elements. The increase in hardness may also be attributed to the presence of fine particles of these intermetallic compounds. The % elongation is found to increase upto 0.5% beyond which it decreases to a constant value. The effect of misch metal on the % elongation can be explained on the basis of the following considerations.

i) With the additions of cerium upto 0.05% the % elongation remains unchanged beyond which it decreases to a constant value at 1.5% addition.

ii) With lanthanum addition beyond 0 05% addition, the % elongation goes on increasing linearly.

iii) With addition of neodymium upto 0.2% the %-elongation increases, beyond which it is found to decrease with further addition.

0.5% misch metal will contain nearly 0.25% Ce 0.10% La and 0.075% Nd. Thus it is clear from the above considerations that the beneficial effect of misch metal on % elongation is due to La and Nd contents. For higher additions between 0.5% to 1% the adverse effect due to (i) predominates over the slight beneficial effect due to (ii) and (iii) resulting in decrease of % elongation. It is interesting to note that with addition beyond 1%, the large beneficial effect produced by La and Nd contents become almost equal to that of detrimental effect, produced by cerium leading to constancy in the value.

Fig. 6.10 gives the values of chip thickness ratio, shear plane angle Ø and unit horse power. Both chip thickness ratio and shear plane angle Ø increase very rapidly upto 0.5% beyond which they increase gradually. The unit horse power decreases to a constant value at about 1.5% addition. The large improvement in machinability with misch metal addition is not surprising. By studying the effects of individual elements, it has been seen that both Ce and Nd bring about improvement in machinability by addition upto 0.5% addition. Since 1% misch metal contains about 0.5% Ce, 0.2% La and 0.15% Nd, it can easily be seen that both Nd and Ce have mainly participated in improving the machinability.

The improvement in machinability with misch metal addition can easily be understood onthe basis of the following considerations :-

i) With misch metal addition very little of Ce and large part of Nd will go into solid solution resulting in slight increase of machinability.

ii) Large refinement effect on misch metal addition will result in increase of machinability.

iii) The increasing amount of misch metal addition will bring about a large increase in additional hard intermediate phases resulting in decrease of machinability.

For additions upto 0.5% the effect due to (iii) will be negligibly small, with a net result of large improvement in machinability. For additions beyond 0.5% the increasing amount of additional phases will have an increased amount of adverse effect resulting in decrease of the rate of increase of machinability. With misch metal addition beyond 1% effects seem to balance each other, resulting in constant values of machinability as indicated by only a slight increase in chip thickness ratio and shear plane angle ϕ and almost constant values of unit horse power.

6.4.6 Effects of lancer-amp 2 and rare earth fluorides:

Salient features of the results are given below:-

i) Metallographic study indicates that there is little effect of lancer-amp 2 addition on the grain size of aluminium. Very slight refinement has been seen on metallographic study. Plates 6.2 to 6.35).

ii) There is little effect of lancer amp 2 on tensile strength. It decreases upto about 1 percent and remains almost unchanged.

iii) The % elongation increases upto about 0.5% lancer amp 2 addition beyond which it goes on decreasing with more additions.

iv) There is a slight and regular increase in the hardness of aluminium with addition of lancer-amp 2 upto about 2.5% beyond which it becomes constant.

v) An important effect of rare-earth fluoride addition is the refinement of the grain structure. There is a little effect upto 2% addition beyond which ,there is large refining effect. The effect is not so pronounced as in the case of misch metal (Plates 6.32 to 6,35),

vi) There is a little effect of rare-earth fluorides addition on the tensile strength. It slightly decreases with the addition.

vii) The % elongation increases to a constant value at 1.5% addition of rare earth fluoride.

viii) There is a large increase in hardness upto about 2% rare earth fluorides addition beyond which it slightly decreases.

The only difference between lancer-amp 2 and misch metal is that lancer-amp 2 contains 10% Zr and 5% Fe, in addition to misch metal It is, therefore, expected that effects of lancer-ampp 2 should be similar to that of misch metal on aluminium. From Fig. 6.21) it can easily be seen that effects of lancer-amp.2 are similar to that of misch metal on tensile strength and % elongation. On metallographic study it is seen that unlike misch metal there is only little refinement effect on aluminium, probably due to the presence of zirconium or iron which have acted in the opposite way to minimise the refining action of misch metal. As previously discussed the increase in hardness with lancer-amp 2 can be attributed to the presence of fine particles of intermetallic compounds of Al_4Ce, Al_4La etc. The decrease in % elongation beyond 0.5% may be due to large increase in the amount of brittle intermetallic phases.

Since fluorides are mixtures of rare-earth metals in the same proportion in which they exist in misch metal, they when added may decompose partly to their elemental forms and effect the properties in a similar way as that of misch metal. The fluorides so added may also act as flux in removing hydrogen. It is considered that the increase in % elongation may be partly due to the degassification that takes place due to fluoride addition. The refining effect may be due to fine particles of intermediate phases formed because of partial decomposition and acting as nuclei during solidification.

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CHAPTER VII

EFFECTS OF TANTALUM, RARE EARTH ELEMENTS AND MISCH METAL ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM-MAGNESIUM ALLOYS AT ORDINARY TEMPERATURE.

CHAPTER VII

EFFECTS OF ADDITIONS OF TANTALUM, RARE EARTH ELEMENTS AND MISCH METALS ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM MAGNESIUM ALLOYS AT ORDINARY TEMPERAUTRE.

7.1 GENERAL INTRODUCTION:

Alloys of aluminium with magnesium are characterised by their high resistance to corrosion, superior high temperature properties and good machinability. These are extensively used for structural members in aircrafts and marine vessels, brake shoes etc. The present investigation has been undertaken with a view to study systematically the effects of tantalum and other rare earths on the microstructure, machinability and mechanical properties of aluminum-magnesium alloy 218 with 8% magnesium. This alloy has large applications where exceptional mechanical properties, resistance to corrosion and finishing characteristics are required. The rare earths has been added in the form of mixtures as misch metal. The effects of individual metals Ce, La, and Nd contained in the misch metal have also been investigated. Literature survey has indicated that although little work is available on the effects of misch metal on microstructure and mechanical properties of Al-M_g alloy practically no work has been reported in the direction of machinability . No work has been done previously to study the effects of individual elements Ce, La and Nd on the machinability, microstructure and mechanical properties. Thus the above investigation will help in the development of new alloys from strength and machinability standpoint, as well as to understand more fundamentally the causes of the effects of misch metal on the alloy based on the effects of individual elements contained in it.

7.2 EXPERIMENTAL PROCEDURE.

7.2.1 Preparation of alloy 218 and addition of rare-earth and tantalum:

Aluminium was first melted in a graphite crucible in a temperature controlled vertical muffle furnace. The required quantity of magnesium wrapped in aluminium foil was added to the melt at 750°c. The alloy was chemically analysed for its magnesium content. The required amount of rare-earths wrapped in aluminium foil was dipped into the molten metal and the melt stirred.

In order to add tantalum, the molten alloy with magnesium was super heated to about 1000°C and tantalum wrapped in aluminum foil was then added to it.

Casting was done at about 700°C in cast iron moulds in all the cases. After analysis, the magnesium content in the alloy was found to be 7.96%.

7.3 EXPERIMENTAL - RESULTS.

7.3.1 Effects of tantalum additions:

Results are given in Figs. 7.01 & 7.02, table 7.01 and plates 7.01 to 7.06.

7.3.2 Effects of cerium additions:

Results are given in figs. 7.03 & 7.04, table 7.02 and plates 7.07 to 7.09.

7.3.3 Effects of lanthanum additions:

Results are given in Figs. 7.05 to 7.06 table 7.03 and plates 7.10 to 7.13 .

7.3.4 Effects of neodynium addition

Results are given in Figs. 7.07 & 7.08 table 7.04 and plates 7.14 to 7.15.

7.3.5 Effects of misch metal.

Results are given in Figs. 7.09 & 7.10, table 7.05 and plates 7.16 to 7.20.

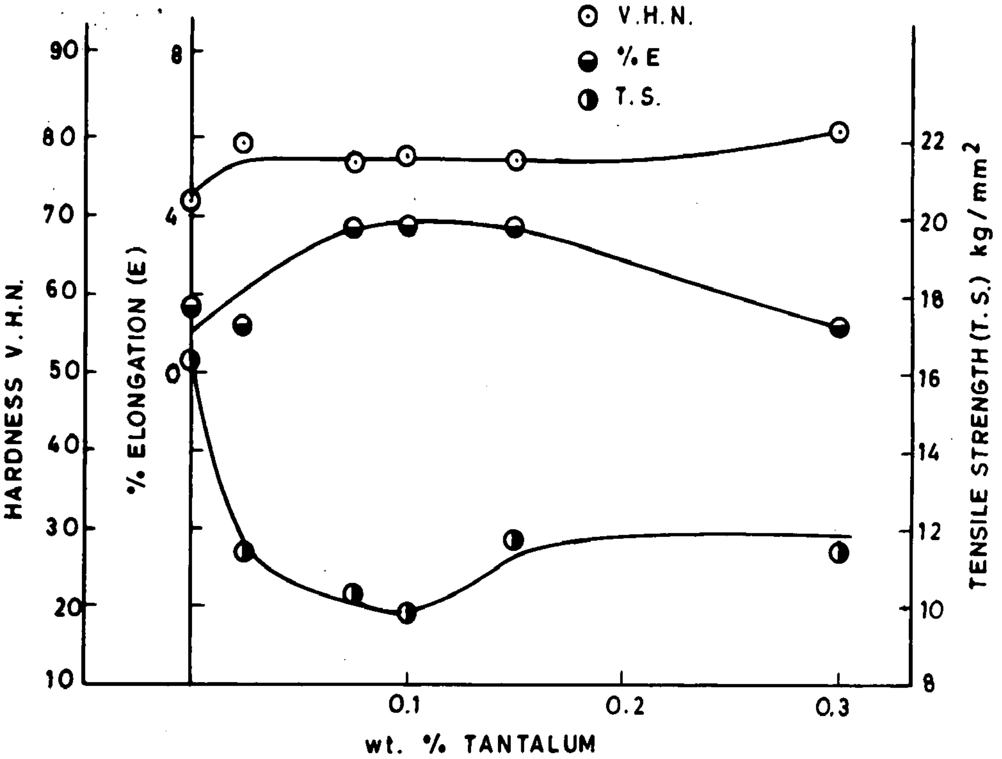

FIG. 7.01 – EFFECT OF TANTALUM ON THE MECHANICAL PROPERTIES OF Al - 8% Mg ALLOY

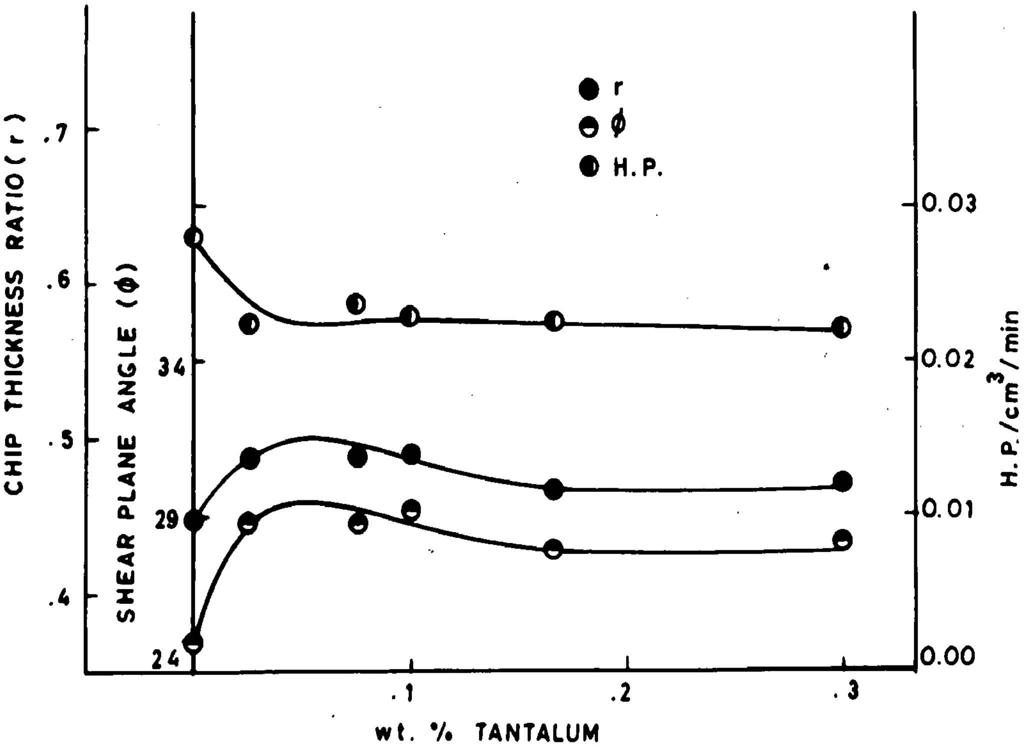

FIG. 7.02 – EFFECT OF TANTALUM ON THE MACHINABILITY OF Al - 8% Mg ALLOY.

TABLE 7.01

EFFECTS OF TANTALUM ADDITIONS ON THE MECHANICAL PROPERTIES, MACHINABILITY OF ALLOY 218, (8% Mg)

No.	% alloying element	Tensile strength kgm/mm^2	% elongation	V.H.N	Chip thickness ratio	Shear plane angle ϕ	HP/cm^3/min
1	0.000	16.3	1.75	72	0.424	$24^{0}50'$	0.0280
2	0.025	11.4	1.20	79	0.483	$28^{0}48'$	0.0225
3.	0.075	10.3	3.75	77	0.483	$28^{0}48'$	0.0238
4	0.10	9.8	3.75	78	0.489	$29^{0}13'$	0.0227
5	0.15	11.7	3.75	77	0.465	$27^{0}35'$	0.0224
6	0.30	11.4	1.20	81.	0.470	$29^{0}55'$	0.0221

TABLE 7.02

EFFECTS OF CERIUM ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 218 (8% Mg).

S. No.	% alloying element	Tensile strength kgm/cm^2	% elongation	V.H.N.	Chip thickness ratio	Shear plane angle ϕ	HP/cm^3/ min.
1	0.00	16.3	2.00	72	0.424	$24^{0}50'$	0.0280
2	0.20	13.9	2.25	75	0.467	$27^{0}43'$	0.0233
3	0.50	10.6	2.25	77	0.446	$26^{0}17'$	0.0204
4	1.00	12.8	1.20	77	0.470	$27^{0}55'$	0.0211
5	1.50	11.6	1.20	78	0.470	$27^{0}55'$	0.0204

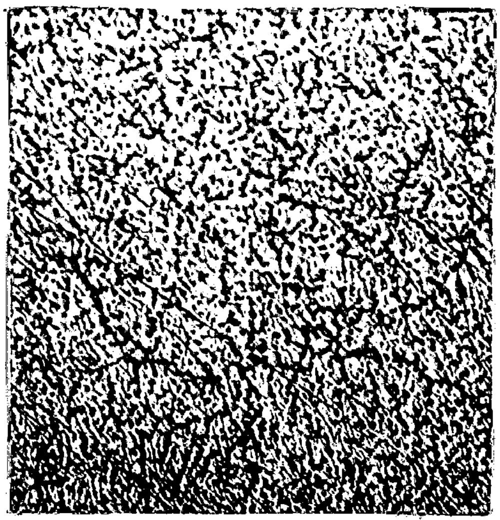

PLATE 7.01 106 X

Al + 8 % Mg

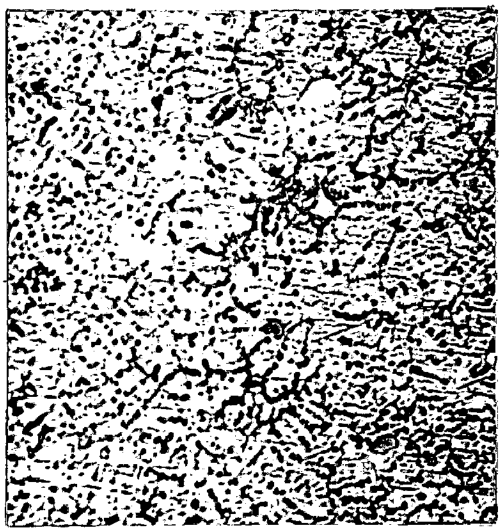

PLATE 7.02 106 X

Al+8 % Mg + 0.025 wt.% Ta

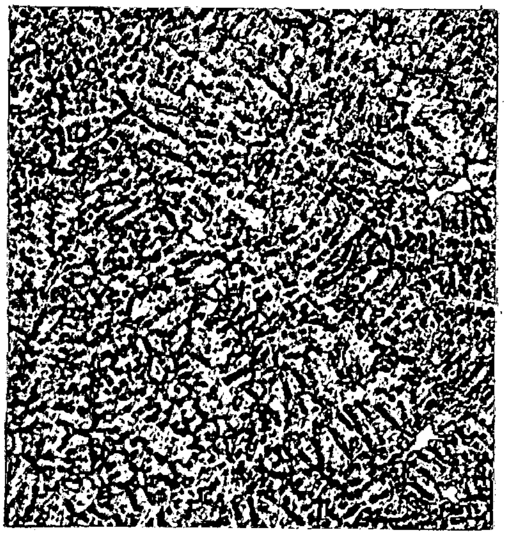

PLATE 7.03 106 X

Al + 8 % Mg + 0.075 wt.% Ta

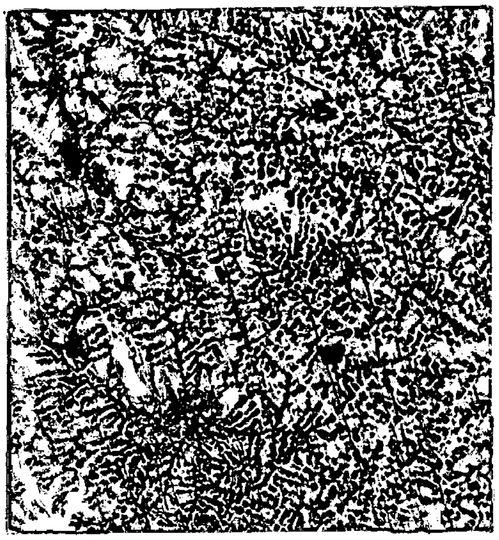

PLATE 7.04 106 X

Al + 8 % Mg + 0.10 wt. % Ta

PLATE 7.05 106 X

Al + 8 % Mg + 0.15 wt. % Ta

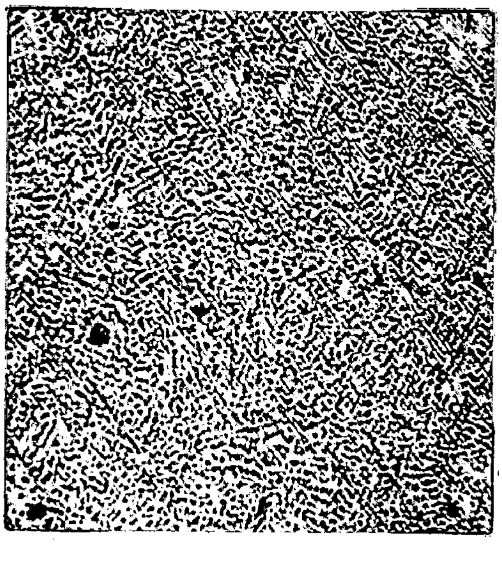

PLATE 7.06 106 X

Al + 8% Mg + 0.30 wt. % Ta

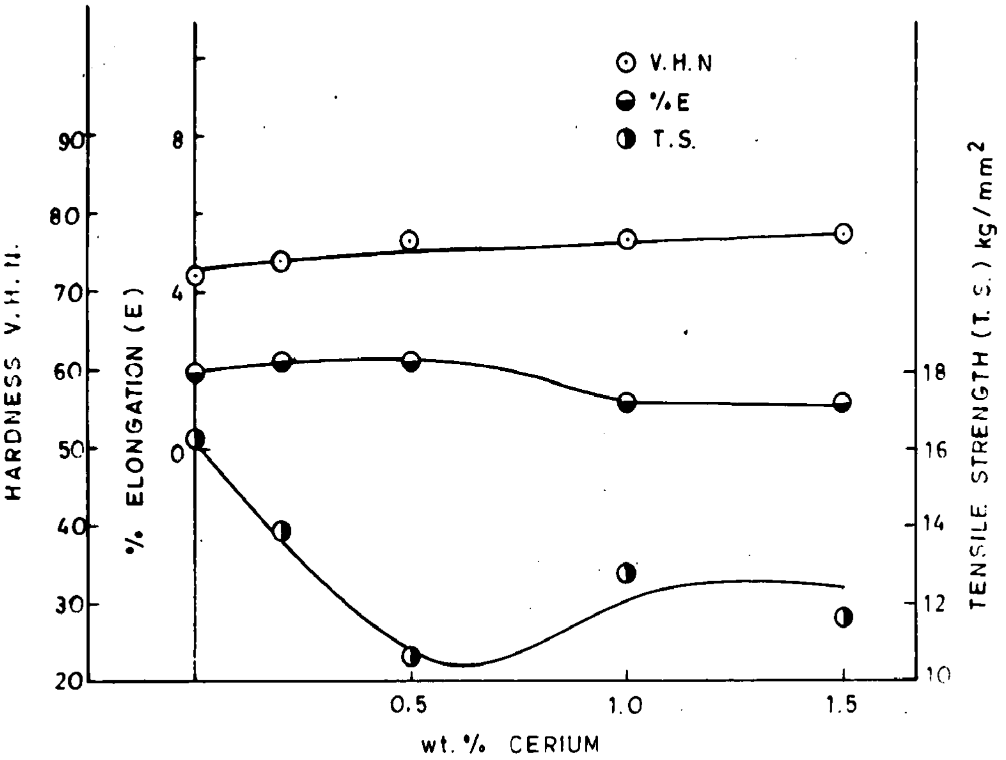

FIG. 7.03_EFFECT OF CERIUM ADDITION ON THE MECHANICAL PROPERTIES OF Al - 8% Mg ALLOY.

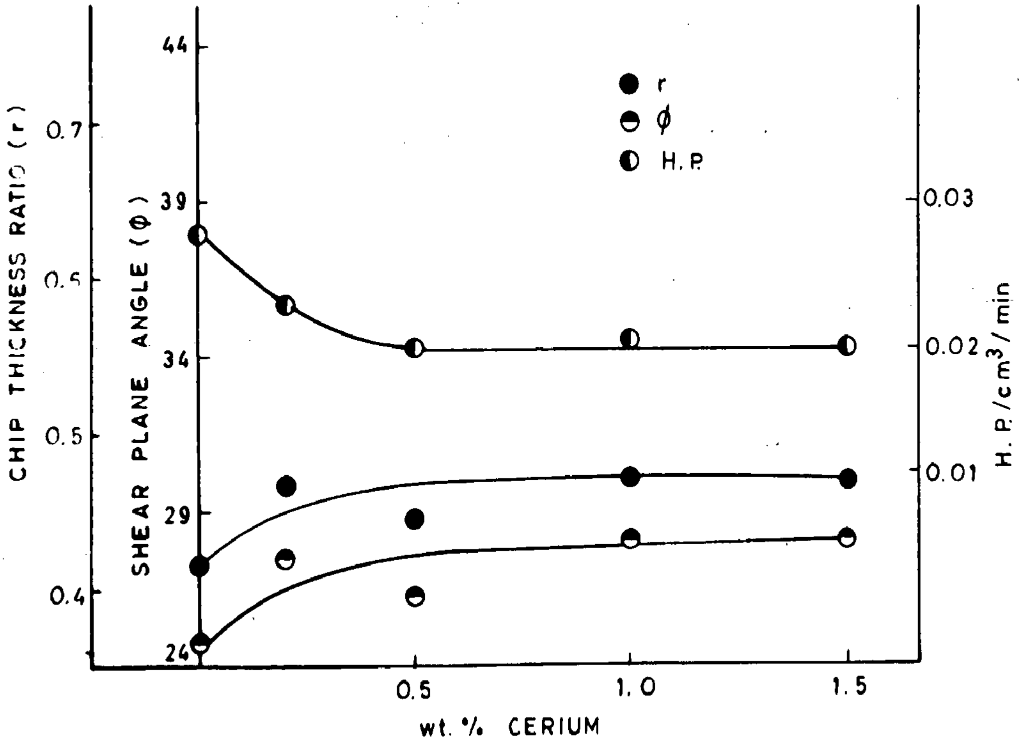

FIG. 7.04_ EFFECT OF CERIUM ADDITION ON THE MACHINABILITY OF Al - 8% Mg ALLOY

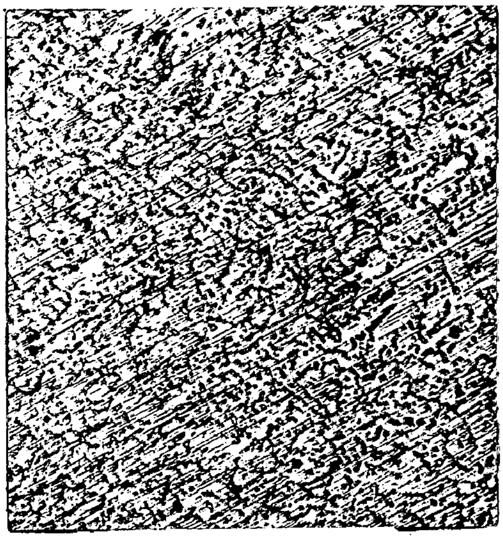

PLATE 7.07 106 X

Al + 8% Mg + 0.5 wt.% Ce

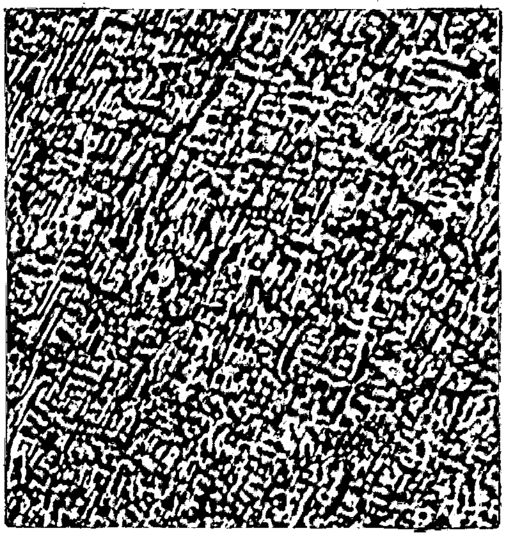

PLATE 7.08 106 X

Al+8% Mg + 1.0 wt. % Ce

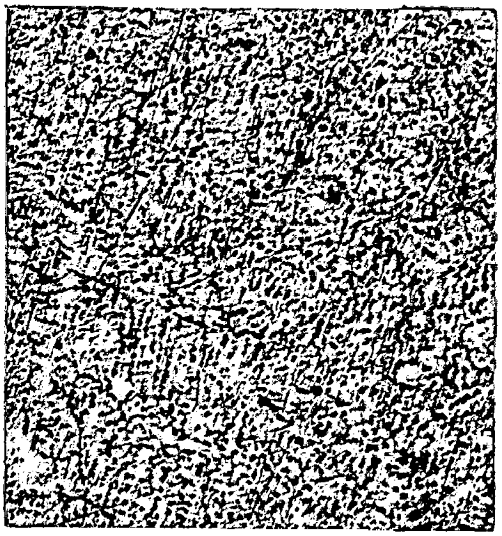

PLATE 7.09 106 X

Al + 8% Mg + 1.5 wt. % Ce

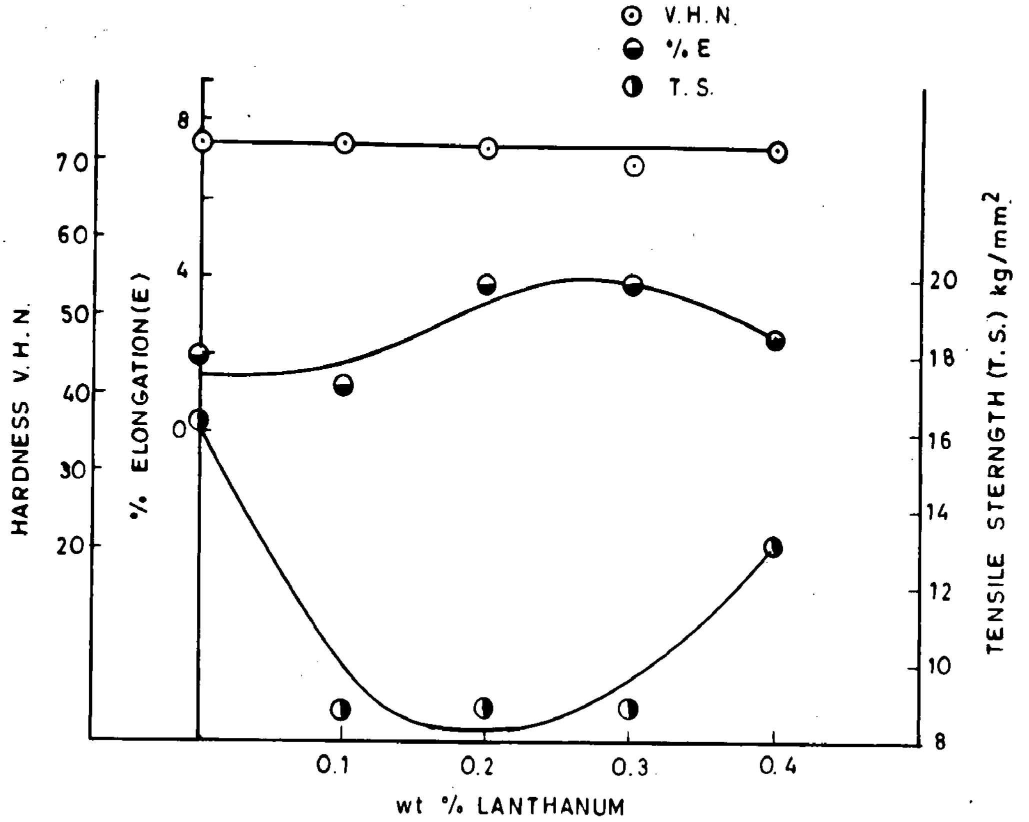

FIG. 7.05_ EFFECT OF LANTHANUM ADDITION ON THE MECHANICAL PROPERTIES OF Al - 8% Mg ALLOY

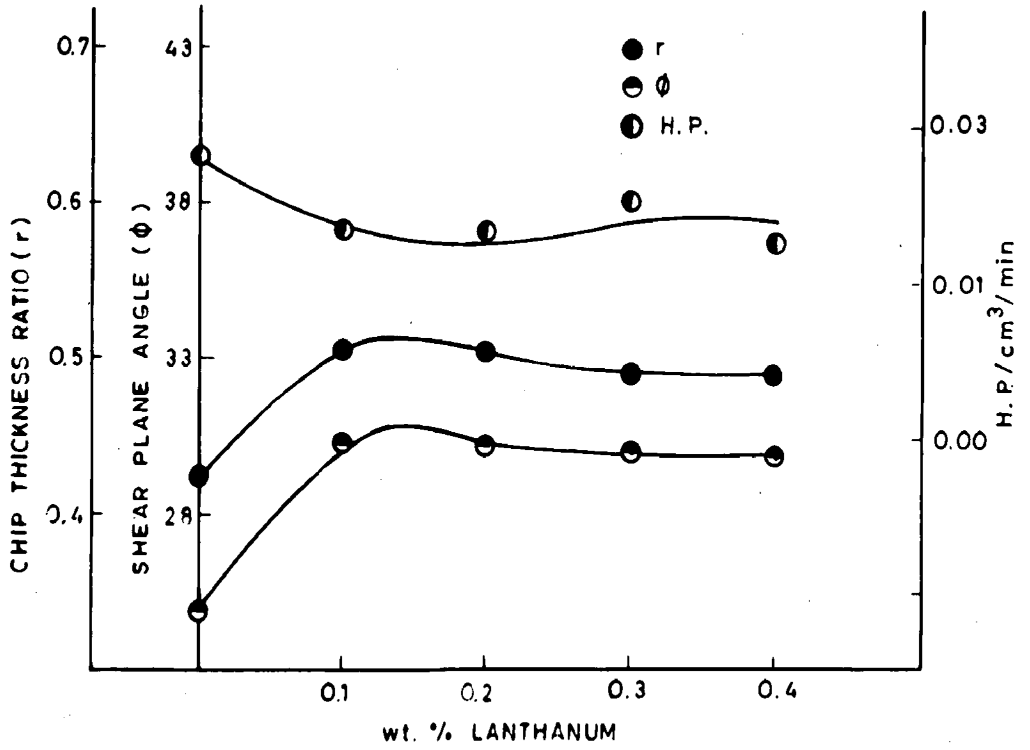

FIG. 7.06_EFFECT OF LANTHANUM ADDITION ON THE MACHINABILITY OF Al - 8% Mg ALLOY.

TABLE 7.03

EFFECT OF LANTHANUM ADDITION ON MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 218 (8% Mg)

S. No.	% alloying element	Tensile strength kgm/mm^2	% elongation	V.H.N.	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min.
1	0.00	16.3	2.00	72	0.424	24°50'	0.0280
2	0.10	8.8	1.20	72	0.505	30°14'	0.0232
3	0.20	8.9	3.78	72	0.505	30°14'	0.0232
4	0.30	8.9	3.75	70	0.491	30°01'	0.0253
5	0.40	13.1	2.50	72	0.491	30°01'	0.0226

TABLE 7.04

EFFECT OF NEODYMIUM ADDITION ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 218 (8% Mg)

S. No.	% alloying element	Tensile strength kgm/mm^2	%elongation	V.H.N	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min
1	0.00	16.3	2.00	72	0.424	24°50'	0.0280
2	0.10	16.4	7.50	69	0.506	30°21'	0.0198
3	0.20	10.4	1.20	72	0.503	30°10'	0.0223
4	0.30	11.6	2.25	89	0.527	31°48'	0.0233

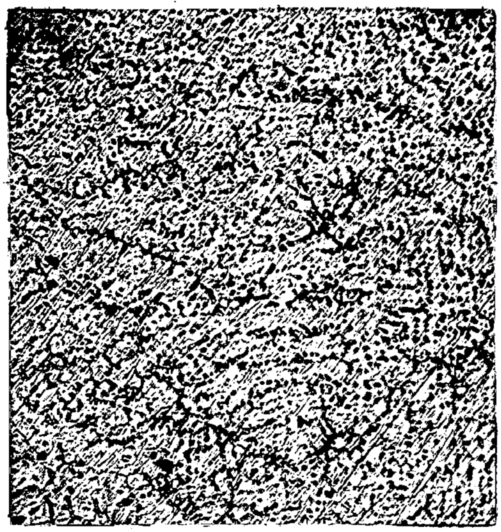

PLATE 7.10 106 X

Al + 8 % Mg + 0.1 wt.% La

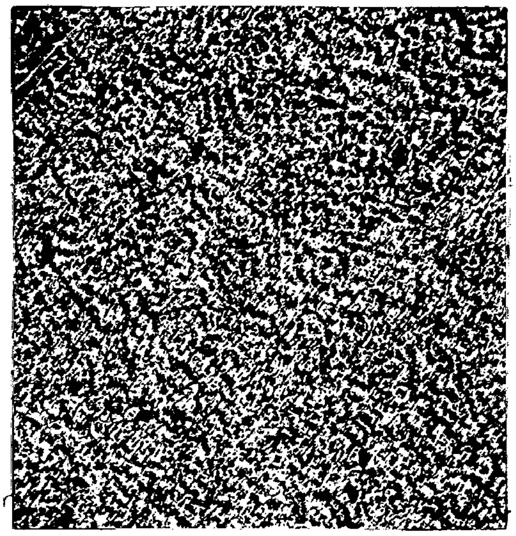

PLATE 7.11 106 X

Al + 8 % Mg + 0.2 wt.% La

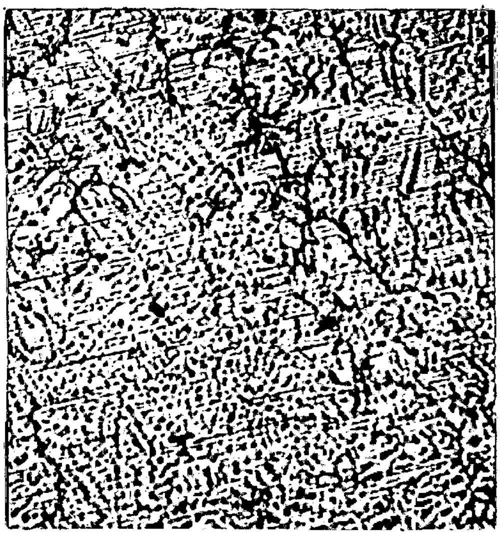

PLATE 7.12 106 X

Al + 8 % Mg + 0.3 wt.% La

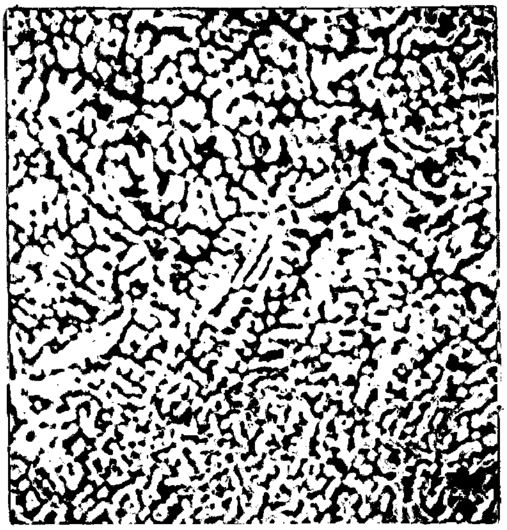

PLATE 7.13 106 X

Al + 8 % Mg + 0.4 wt.% La

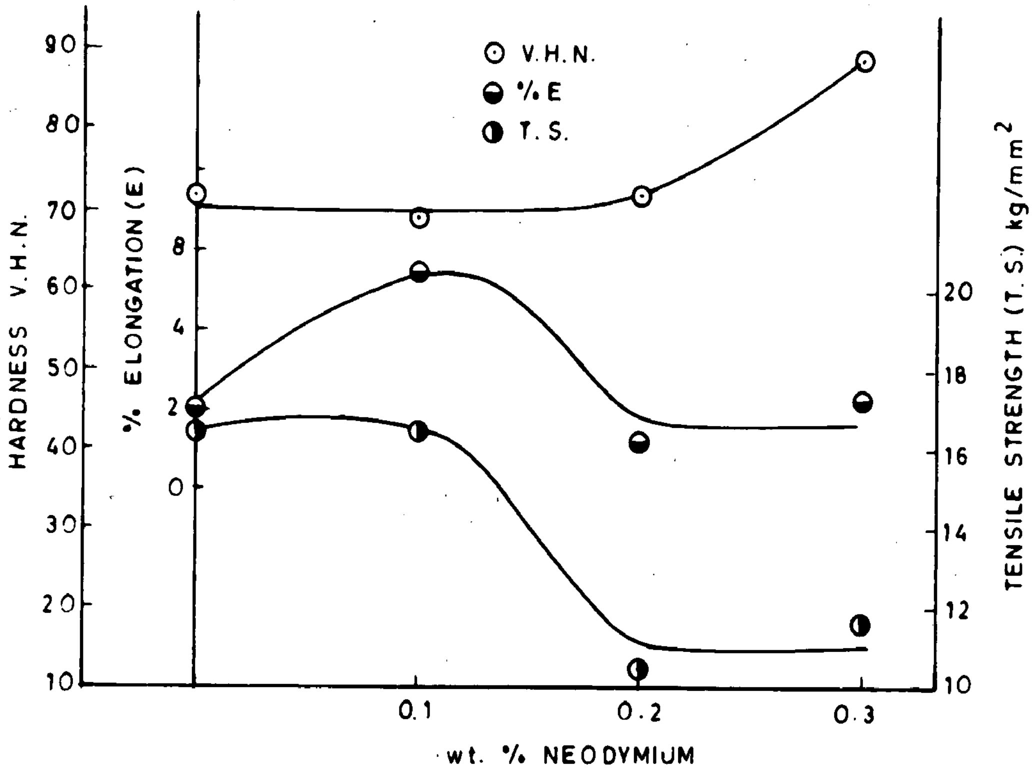

FIG.7.07_EFFECT OF NEODYMIUM ON THE MECHANICAL PROPERTIES OF Al - 8% Mg ALLOY

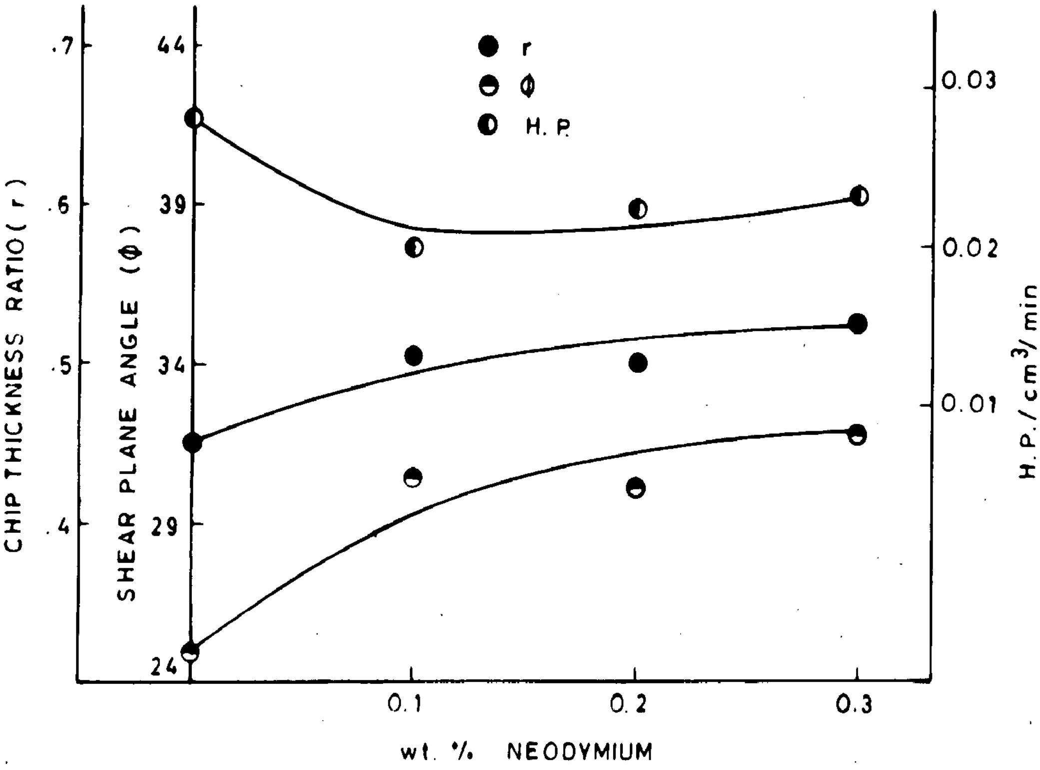

FIG. 7.08_ EFFECT OF NEODYMIUM ON THE MACHINABILITY OF Al - 8% Mg ALLOY.

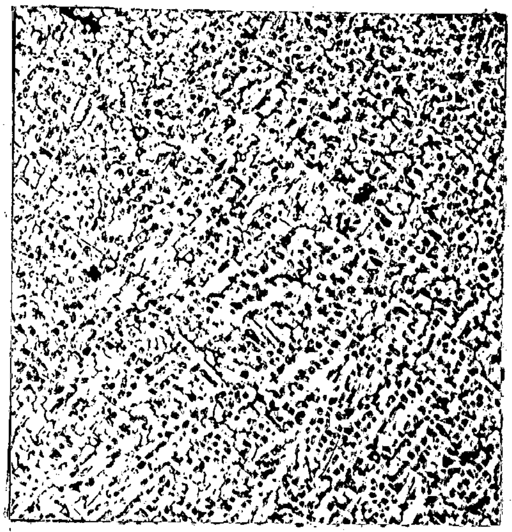

PLATE 7.14 106 X

Al+8 % Mg+0.1 wt. % Nd.

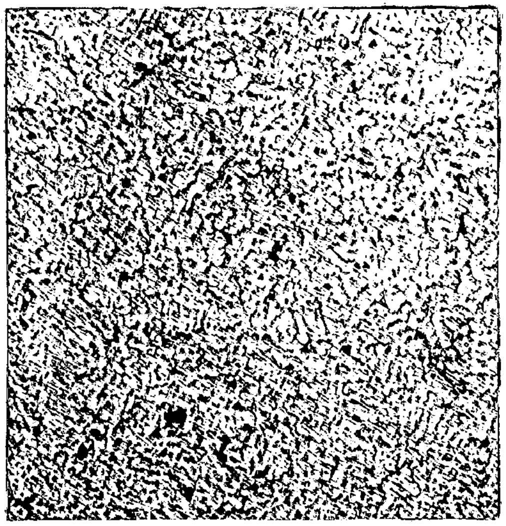

PLATE 7.15 106 X

Al + 8 % Mg+0.3 wt.% Nd

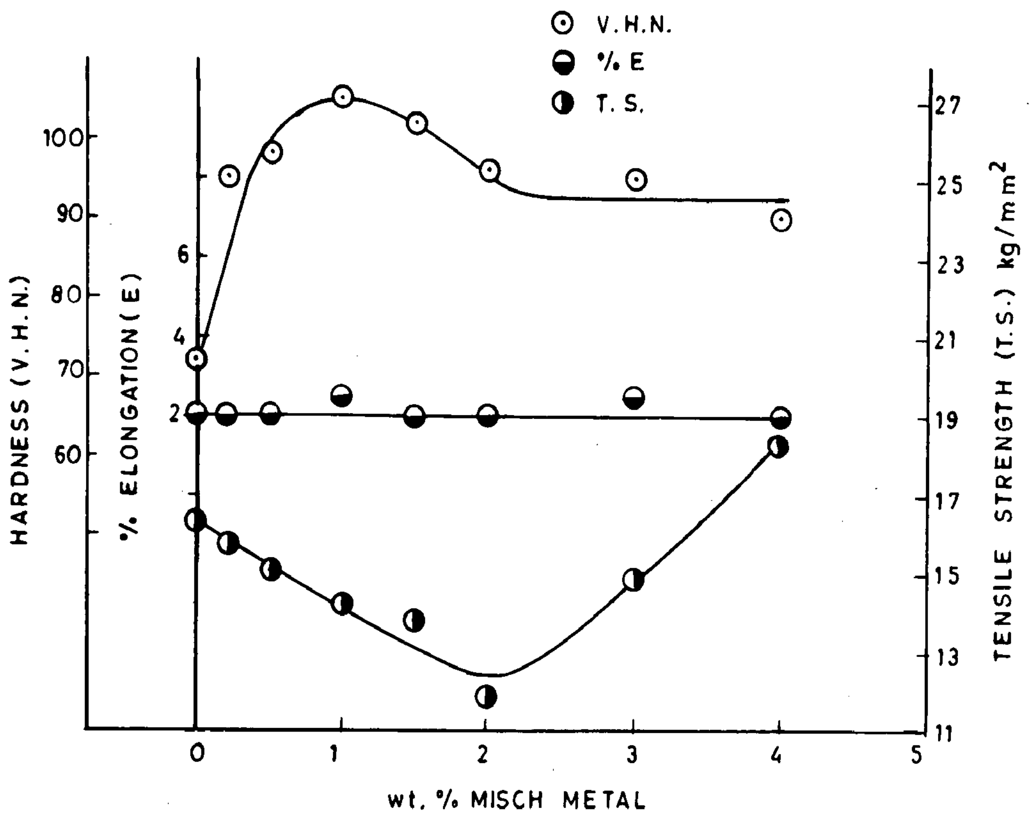

FIG. 7.09_ EFFECT OF MISCH METAL ON THE MECHANICAL PROPERTIES OF Al + 8% Mg ALLOY.

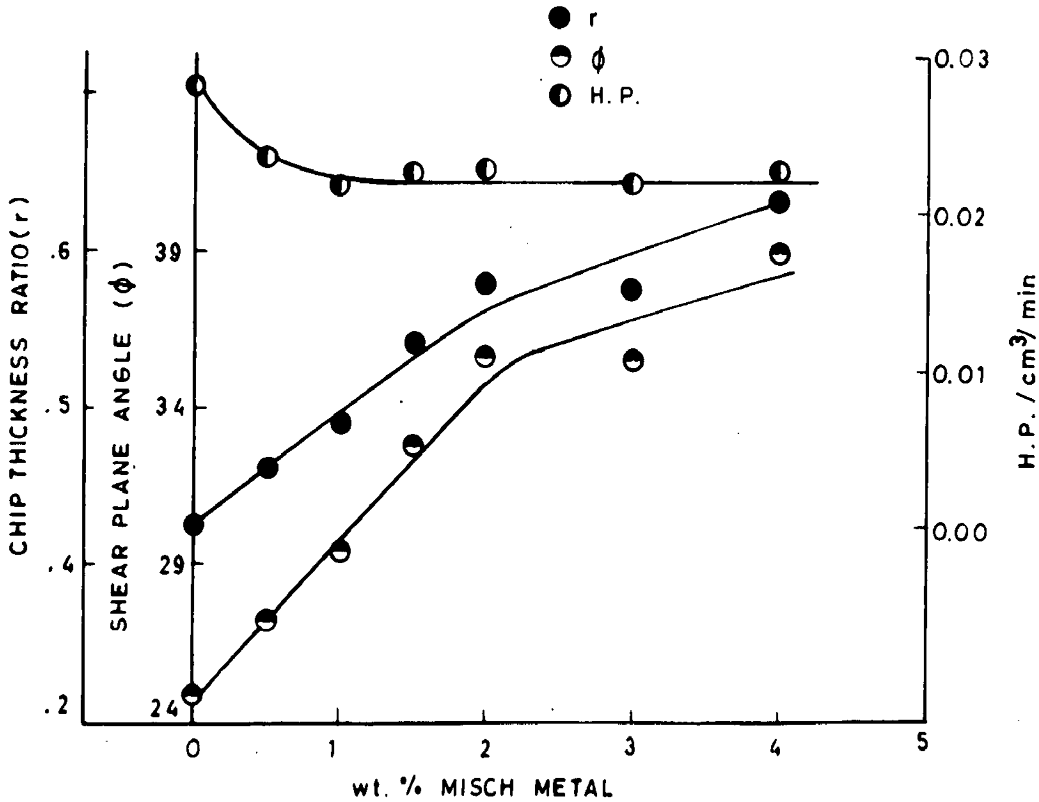

FIG. 7.10_ EFFECT OF MISCH METAL ON THE MACHINABILITY OF Al + 8% Mg ALLOY.

TABLE 7.05

EFFECTS OF MISCH METAL ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 218 (8% Mg)

S. No.	% alloying element	Tensile strength kgm/mm^2	%elongation	V.H.N.	Chip thickness ratio	Shear plane angle ϕ	HP/cm^3/min.
1	0.00	16.3	2.0	72	0.424	24°50'	0.0280
2	0.20	15.8	2.0	95	-	-	-
3	0.50	15.0	2.0	98	0.580	35°24'	0.0236
4	1.00	14.2	2.5	105	0.492	29°25'	0.0218
5	1.50	13.9	2.0	102	0.542	32°50'	0.0225
6	2.00	12.0	2.0	96	0.580	35°24'	0.0228
7	3.00	15.9	2.5	95	0.571	34°48'	0.0220
8	4.00	13.1	2.0	90	0.635	39°9'	0.0228

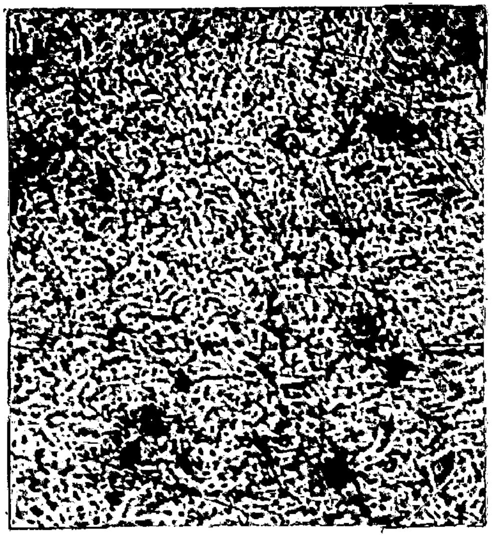

PLATE 7.16 106 X

Al + 8% Mg + 0.5 wt.% MISCH METAL

PLATE 7.17 106 X

Al+8%Mg + 1.0 wt.% MISCH METAL

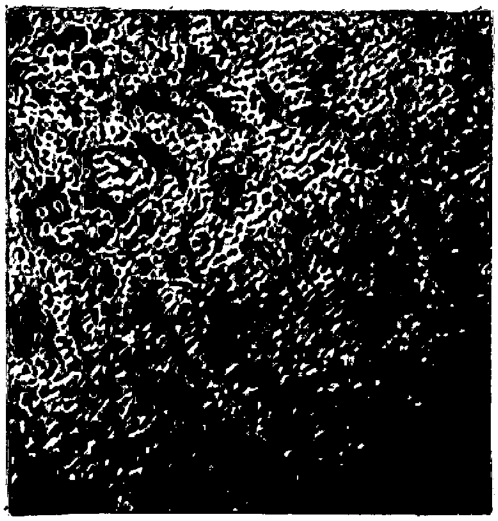

PLATE 7.18 106 X

Al + 8%Mg + 2.0 wt.% MISCH METAL

PLATE 7.19 106 X

Al + 8% Mg + 3 wt. % MISCH METAL

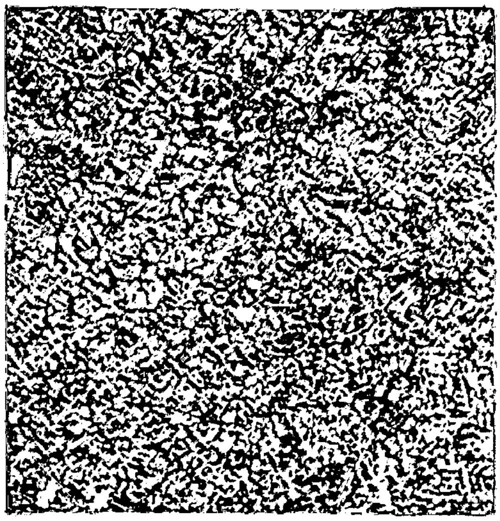

PLATE 7.20 106 X

Al + 8% Mg + 4 wt. % MISCH METAL

7.4 DISCUSSION OF THE RESULTS

7.4.1 Effects of tantalum additions

The salient features of the results are given below:

i) On metallographic study it is observed that only phases that are present in the original alloy are Al-Mg and Mg_2Si, the latter due to silcon present as an impurity in the metal in addition to α. With increase in tantalum addition there appears to be large refining effect of grains. The microstructure of the alloy treated with 0.15% tantalum indicates a very fine grain size. (Plates 7.01 to 7.06).

ii) No beneficial effect is produced on the tensile strength with the addition of tantalum. The tensile strength decreases upto 0.1% addition, beyond which it slightly increases to a constant value at about 0.15% addition.

iii) The % elongation increases upto 0.15% tantalum addition. With further addition the % elongation goes on decreasing.

iv) There is slight but regular increase in the hardness value with tantalum addition.

v) Large beneficial effect of tantalum addition is produced on machinability. Both chip thickness ratio and shear plane angle ϕ increase upto 0.075% addition beyond which they decrease to constant values at about 0.15% addition.

vi) The values of unit horse power are found in agreement with the values of chip thickness ratio and shear plane angle ϕ . The large decrease is observed in the unit horse power value upto 0.075% addition beyond which it increases very slightly to a constant value at about 0.1% addition, indicating a large beneficial effect of tantalum addtion on machinability.

Tantalum is found to have no solid solubility in aluminium and magnesium. The only phase formed is $TaAl_2$, The large refining effect is due to the presence of fine particles of intermetallic compounds acting as nuclei during solidification. The regular increase in hardness may be attributed to the increased amount of hard intermetallic compounds. The increase in % elongation upto 0.15% addition may be due to the refining effect. The decrease in % elongation beyond 0.15% addition may be due to large increase in the amount of brittle intermetallic phases.

The improvement in machinability with tantalum addition can be explained on the basis of the following considerations:

i) The large refining effect produced due to tantalum addition tends to increase the machinability.

ii) With increasing tantalum addition the amount of hard intermediate phases should increase tending to decrease the machinability.

It appears that for additions upto 0.075% the beneficial effect due to the refining predominates over the slight adverse effect due to the presence of intermediate phases. For higher additions however, increased amount of hard intermediate phases brings about a large adverse effect, predominating over the beneficial effect due to (i) with a net result of decrease in machinability.

The presence of the intermediate phases as explained above could not be detected on metallographic study probably due to their being of submicroscopic size and needs more careful X-ray or electron diffraction study.

7.4.2 Effects of cerium additions:

The salient features of the results are given below:

i) No detectable effect is observed on metallographic structure with the addition of cerium except that a very slight refinement occurs in the grain. (Plates 7.07 to 7.02)

ii) With the addition of cerium, the tensile strength is found to decrease upto about 0.5%, the rate being almost linear. With further addition the strength increases to a constant value at about 1% addition.

iii) There is very little effect of cerium addition on the % elongation of the alloy. It increases very slightly upto 0.5% beyond which it drops down to a more or less constant value at about 1% addition.

iv) There is slight but regular increase in hardness value.

v) Large beneficial effect is observed on the machinability of the alloy when treated with cerium from chip formation consideration. Both chip thickness ratio and shear plane angle ϕ increase to more or less constant value, at about 1% addition.

vi) H.P. / C m^3/min is found to decrease very rapidly to more or less a constant value at about 0.5% addition. The unit horse power curve is found to be in agreement with the curve of shear plane angle ϕ, the minima and maxima being obtained in unit horse power and shear angle ϕ respectively at about 0.5% addition. With further addition; both are found to remain unchanged.

The solid solubilities of cerium are reported to be 0.5 wt% and 0.85 wt % in aluminium and magnesium respectively. Thus with the addition of cerium upto 0.1% a part of it will go into solid solution in aluminium and remaining into magnesium phase. With further addition the cerium should react with aluminium and magnesium to form intermediate phases Al_4Ce, Al_2Ce, AlCe, $AlCe_3$ GeMg, $CeMg_2$, $CeMg_3$, CeMg the predominating phase being Al_4Ce and $CeMg_9$.

A slight refinement may be due to fine particles of intermediate phases acting as nuclei during solidification. A slight but regular increase in hardness with cerium addition may also be attributed to the presence of fine harder precipitates of these phases. The presence of these phases could not be detected on metallographic study probably due to their being of sub-microscopic size.

The slight increase in % elongation upto 0.2% may be due to cerium going into solid solution and refinement of primary α . The decrease in % elongation beyond 0.5% addition is attributed to the fine despersion of the precipitation of the intermediate phases being coherent with the lattice.

Fig. 7.04 indicates the effects of cerium addition on the chip thickness ratio, shear plane angle Ø and horse power consumed during drilling. From all the three criteria the machinability is found to improve by addition of cerium upto 0.5%. The improvement in the machinability with cerium addition can be understood on the basis of the following considerations:-

i) Slight amount of cerium going into solid solution in aluminium and magnesium will lead to improvement in machinability.

ii) Refinement effect observed with cerium addition should have beneficial effect on machinability.

iii) The decrease in tensile strength upto 0.5% addition should bring about an improvement in machinability as lesser power should be consumed for cutting.

iv) With increasing cerium addition beyond 0.1% the amount of hard brittle intermediate phases should increase, resulting in decrease of machinability.

It is interesting to note that for additions upto 0.5%, the beneficial effects due to (i), (ii) and (iii) predominate over the adverse effect due to (iv) resulting in improvement in machinability. For higher additions, the increased amount of intermediate phases and corresponding increase in hardness and tensile strength, will bring about an adverse effect, which seems to balance the beneficial effects due to cerium going in solid solution and slight refinement, with a net result of constant value of machinability.

7.4.3 Effects of lanthanum additions:

The salient feature of the results are given below:-

i) On metallographic study there appears to be large refining effect in α grains as well as in Al-Mg phase upto 0.2% addition, beyong which slight reversal takes place and grain becomes again coarser. (Plates 7.10 to 7.13)

ii) The tensile strength decreases very rapidly upto about 0.2% addition, beyond which it again goes on increasing with further additions.

iii) The % elongation is found to increase with lanthanum addition. There is very little effect upto 0.1%, beyond which it increases at a large rate, the optimum composition giving the maximum % elongation being 0.3% . Beyond 0.3% addition the % elongation again decreases.

iv) There is very little effect on the hardness value. It remains almost unchanged with lanthanum addition.

v) Both chip thickness ratio and shear plane angle $\emptyset$ increase upto 0.1% addition.

to constant values at about 0.2% addition indicating the beneficial effects of lanthanum addition on machinability upto about 0.1% addition from chip formation considerations.

vi) HP/Cm^3/min. decreases with increasing lanthanum addition slowly between 0.1 and 0.2% addition. Beyond 0.2% the value slightly increases to a constant value at about 0.3% addition. Though the values of H.P. for all the alloys treated upto 0.4% are lesser than that of the original alloy showing the beneficial effect on machinability for all additions.

Lanthanum is found to have no solid solubility in aluminium. The intermediate phases formed are La_3Al, La_3Al_2, $LaAl$, $LaAl_2$ and $LaAl_4$ the predominating phase being $LaAl_4$. The solid solubility of lanthanum in magnesium is about 0.4 %. Four intermediate phases exist $LaMg$, $LaMg_2$, $LaMg_3$ and $LaMg_9$. Thus with the addition of lanthanum to alloy 218 about 0.03% will go into solid solution in magnesium beyong which it may react with aluminium and magnesium to form intermediate phases La_3Al, La_3Al_2, $LaAl$, $LaAl_2$, $LaAl_4$ and $LaMg$, $LaMg_2$, $LaMg_3$, $LaMg_9$ respectively, the predominating phases being La Al_4 and $LaMg_9$. The large refining effect upto 0.2% addition may be due to the presence of the particles of these intermediate phases acting as nuclei during solidification.

The large increase in % elongation may be partly due to small amounts of lanthanum going into solid solution and partly due to refinement of α grains and entectic. The decrease beyond 0.3% may be due to large increase in the amount of brittle hard intermediate phases and recoarsening of the matrix as observed in microstructure. The increase in tensile strength beyond 0.3% may also be attributed to the large increase in the amount of intermediate phases.

The large improvement in machinability upto 0.1% may be attributed partly to the lanthanum, going into solid solution in magnesium and partly to refinement effect produced, predominating over the small adverse effect produced due to the presence of intermediate phases. For higher additions, however, the large increase in the amount of intermediate phases with a corresponding increase in tensile strength and recorsening of the matrix will bring about the decrease in machinability. The results obtained on the basis of both the criteria, horse power and chip formation, are in pefect agreement.

7.4.4 Effect of neodymium additions:

The salient features of the results are given below:

i) On metallographic study there appears to be slight refining effect of neodymium addition. (Plates 7.14 & 7.15).

ii) The tensile strength remains almost unchanged upto about 0.1% addition beyond which it decreases sharply to a constant value at about 0.2% addition.

iii) Large beneficial effect is produced on % elongation with neodymium addition, upto about 0.1%. With further additions the value drops to a constant value at 0.2 % addition.

iv) The hardness remains almost constant upto 0.15% addition, beyond which it goes on increasing with more and more neodymium addition.

v) Both chip thickness ratio and shear plane angle ϕ increases with increase in neodymium addition, rapidly upto about 0.1% beyond which the rate goes on decreasing, and the values tend towards more or less a constant value at about 0.3% addition.

vi) HP/Cm^3/min. decreases rapidly upto about 0.1% addition. Between 0.1 to 0.2% the value is almost constant. With further additions there appears to be very slight increase. The unit H.P. values of all the alloys treated,upto about 0.3% are much lower than that of the initial alloy indicating the beneficial effect on machinability in all the cases.

The maximum solid solubilities of neodymium in aluminium and magnesium are reported to be 0.2 and 4.0 weight percentages respectively. The solid solubility however decreases with dercease in temperature. Thus with the addition upto about 0.2% most of the neodymium will go into solid solution in aluminium and magnesium. With further additions it may react with aluminium and magnesium to form NdAl, $NdAl_2$, $NdAl_4$, NdMg, $NdMg_2$, $NdMg_3$, the amount of which should increase with increased additioms. The improvement in % elongation is attributed to neodymium going into solid solutio n, and slight refinement of matrix- The increase in hardness beyond 0.2% addition may also be due to increasing amount of fine dispersion of the above mentioned intermediate phases. The improvement in machinability with neodymium can also be explained on the basis of the above considerations. It can easily be seen from Fig. 7.08 that both chip thickness ratio and shear plane angle ∅ are found to increase with Nd addition upto 0.2%. The slope decreases beyond 0.1% . The same fact is confirmed from H.P. plot also, where large decrease in the horse power value was observed only upto 0.1%. The large increase in machinability upto 0.1% addition may be due to neodymium going into solid solution and slight refinement of the grains. For additions between 0.1% and 0.2% there may be possibility of formation of small amount of intermediate phases, bringing about a slight adverse effect on machinability resulting in decrease of slope. With further additions the adverse effect due to increased amount of intermediate phases seems to predominate over the beneficial effect produced, with a net result of slight decrease in machinability.

7.4.5 Effect of misch metal additions:

The salient features of the results are given below

i) Metallographic study indicates that with increased misch metal addition there is a gradual decrease in the AlMg phase and a corresponding increase in new phase. There appears to be little refining action with misch metal addition

ii) The tensile strength is found to decrease somewhat upto about 2.0% misch metal addition. Beyond 2.0% misch metal the tensile strength again increases with further and further additions, the minimum being observed between 2.0 and 3.0% .

iii) The % elongation was found unchanged with the addition.

iv) The hardness increases upto 1.0% misch metal addition, beyond which it decreases to a constant value at a-bout 2.0% misch metal.

v) Both chip thickness ratio and shear plane angle Ø increase with increase in misch metal addition appreciably upto 2.0%, beyond which the rate decreases slightly indicating beneficial effect of misch metal addition on machinability upto 4.0% misch metal.

vi) H.P./Cm^3/min. is also found to decrease with misch metal addition. There is an appreciable decrease upto about 1.5% addition, beyond which the value decreases very slightly upto 4.0% addition. The unit H.P. curve agrees well with the chip thickness ratio and shear plane angle Ø curves, confirming the large beneficial effect of misch metal additions on machinability.

Our results of mechanical properties are in general agreement with the results of Trehan et al on

Al-Mg alloys. The work has been extended further to study the effects of additions on machinability.

The effect of misch metal addition on the properties can easily be explained on the basis of three major individual elements cerium, lanthanum and neodymium contained in it. Cerium and neodymium are reported to dissolve upto 0.05 % and 0.2Wt% in aluminium respectively, while La has no solid solubility. The solid solubilities of cerium, lanthanum and neodymium in magnesium are reported to be 0.85, 0,4 and 4.0 weight percentage, respectively. Thus out of cerium, lanthanum and neodymium contents of misch metal ,cerium will go into solid solution upto about 0.1% partly into aluminium and partly into magnesium. About 0.03% lanthanum will go into solution in magnesium, while neodymium may go into solution upto about 0.2% in aluminium and 0.3% in magnesium. Thus with the addition of misch metal very little of cerium and lanthanum will go into solution, most of them may react with aluminium and magnesium to form intermediate phases Al_4Ce, Al_4La, $CeMg_9$, $CeMg_3$, $LaMg_9$ and $LaMg_3$ etc. ,the amount of which should increase with increased addition. The whole of Nd will go into solid solution upto 4.0% addition of misch metal. With the increase of misch metal content from 1.0 to 4.0% the amount of Al-Mg phase gradually decreases and there is an increase in some new phases Slight refining effect has been observed on metallographic study. Refining may be due to the fine particles of intermediate phases acting as nuclei during solidification.

It is interesting to note that the % elongation remains almost unchanged with misch metal addtion, when all the three elements cerium, lanthanum and neodymium have individually produced the beneficial effect on % elongation. This may probably be due to the coarsening of the precipitates in the case of misch metal as observed on metallographic study.

It may be noted that there is a very large increase in hardness upto about 1.0% misch metal, which will contain roughly about 0.5% Ce, 0.2% La and 0.15% Nd. There has been no effect of lanthanum on the hardness value. In the case of cerium addition hardness continuously increases with increased addition upto 1.5%. The hardness remains uneffected with Nd additions upto 0.15%, beyond which increases appreciably with increased additions. Thus the large increase in hardness as observed upto 0.5% misch metal addition may therefore be mainly contributed by cerium and partly by combined effects of neodynium and lanthanum. The increase in hardness may be attributed to the presence of intermetallic compounds of Ce and La. The decrease in hardness beyond 1.0% may be due to the coarsening of the matrix.

Fig. 7.09 gives the values of tensile strength with increasing misch metal additions. The tensile strength decreases upto 2.0% beyond which it again goes on increasing, the minimum being obtained at 2.0%. The tensile strength of the alloy treated with 4.0% misch metal is much greater than that of the untreated alloy. The effects of misch metal on the tensile strength can easily be understood on the basis of the following considerations:-

i) With increasing cerium and lanthanum additions the tensile strength decreases upto 0.5 and 0.2% respectively, beyond which it again increases. The curve follows the same pattern as in the case of misch metal **addition.**

ii) With increase in neodymium content the tensile strength increases upto 0.1% beyond which it decreases.

It is interesting to note that for additions upto 2.0 % misch metal, the detrimental effect due to (i) predominates over the beneficial effect due to (ii). For higher additions beyond 2.0%, however, the large beneficial effect due to cerium and lanthanum predominates over the effect produced due to Nd content with a net result of increase in tensile strength with misch metal addition. There is a large increase in machinability upto 1.5% both addition from chip formation as well as horse power consideration. The rate however decreases with further addition to more or less a constant value at higher additions. The large improvement in machinability upto 1.5% may be due to cerium and neodymium going into solid solution, slight refinement effect and decrease in tensile strength. With further additions the increase in amount of intermediate phases, coarseness and their non-uniform distribution and increase in tensile strength will bring about an increasing adverse effect, resulting in decrease of slope. The values become nearly constant at about 2.0% indicating that the adverse effect produced balances against the beneficial effect produced due to elements going into solid solution and slight refining effect. It is clear from figures 7.03, 7.06 and 7.08 that large beneficial effect is produced on addition of cerium, lanthanum and neodymium upto 0.5%, 0.15% and 0.1% respectively. 1.5% misch metal will contain roughly about 0.5% Ce, 0.2% La and 0.15% Nd. Thus for additions upto 1.0% misch metal all the three elements cerium, lanthanum and neodymium should contribute to the improvement in machinability.

-:o:-

CHAPTER VIII

EFFECTS OF TANTALUM, RARE EARTH ELEMENTS AND MISCH METAL ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM COPPER ALLOYS AT ORDINARY TEMPERATURE.

CHAPTER VIII

EFFECTS OF TANTALUM, RARE EARTH ELEMENTS AND MISCH METAL ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM - COPPER ALLOYS AT ORDINARY TEMPERATURE.

8.1 GENERAL INTRODUCTION:

In the present chapter a systematic study has been made of the effects of tantalum, rare earth elements and misch metal on the microstructure, machinability and mechanical properties of alloy 195 with 4.5% copper and alloy 122 with 10% copper and 0.2% magnesium. Alloy 195 has got applications where a combination of high tensile properties and good machinability are required. Flywheel housings, rear axle housings, bus wheels, aircraft wheels, fittings and crank cases are some typical applications. Alloy 122 has got the applications, where good high temperature strength, high hardness, resistance to wear and good machinability are required. Air cooled cylinder heads, automotive piston, bushings, are typical applications.

In the present investigation the rare earths have been used in the form of mixture viz. misch metal. Effects of individual elements cerium, lanthanum and neodymium contained in misch metal have also been investigated. Effects of tantalum additions have also been studied. As practically no work seems to have been reported on the above lines, it is hoped that the above investigation will be useful in the development of aluminium alloys with additions of rare earth metals ,misch metal & tantalum from machinability and mechanical strength standpoint.

8.2 EXPERIMENTAL PROCEDURE:

8.2.1 Preparation of alloys:

8.2.1.1 Preparation of alloy 195 and addition of rare earth metals and tantalum to the same:

The master alloy of aluminium and copper with about 50% copper and 50% aluminium was first prepared. The alloy was chemically analysed. The necessary quantities of aluminium and the master alloy were melted in a graphite crucible. When the temperature reached 800°C, the required amount of rare earth metal wrapped in aluminium foil was added. In order to add tantalum, the master alloy containing copper was super heated to 1000°C and tantalum wrapped in aluminium foil was added to the melt and stirred.

The alloys were cast in cylindrical metal moulds to prepare specimens for tensile testing at room temperature, hardness and metallographic study, at about 700°C in all the cases. Aluminium chloride was used as flux for degassification. The alloy 195 was chemically analysed for its copper content by iodometric method and the results of chemical analysis based on duplicate samples are given in table 8.01

8.2.1.2 Preparation of alloy 122 and addition of rare earth metals and tantalum to the same:

Necessary quantities of aluminium and the master alloy were melted in graphite crucible. When the temperature of the melt reached 800°C, the magnesium equivalent of 0.2% was wrapped in aluminium foil, and added to the melt with stirring. The alloy now containing 10% copper and 0.2% magnesium was kept in the furnace for 10 minutes for homogenization. Alloy was then cast and chemically analysed. Results of chemical analysis based on duplicate samples are given in table 8.01. The required amount of the alloy was melted. When the temperature reached 800°C the weighed amount of rare earth metal wrapped in aluminum foil was added to the melt. In order to add tantalum, molten alloy was super-heated to 1000°C and then the tantalum wrapped in aluminum foil was added. The alloys were kept in the furnace for about 10 minutes for complete homogenization. The alloys were then cast in cylindrical metal moulds at a temperature of about 700°C. The aluminum chloride was used as flux for degassification.

Table 8.01

Chemical analysis of Alloy

S.No.	Alloy No.	Desired composition	Actual Composition (by chemical analysis)
1.	195	4.5% cu Rest aluminium.	4.48% cu
2.	122	10%, Cu,0.2% Mg Rest aluminium	9.92% Cu,0.19% Mg.

8.3 EXPERIMENTAL RESULTS:

8.3.1 Effects of additions of tantalum, rare earth elements and misch metal on the properties of alloy 195.

8.3.1.1 Effects of tantalum additions.

The results are given in figs. 8.01 & 8.02 plates 8.01 to 8.05 and table 8.02.

8.3.1.2 Effects of cerium additions.

The results are given in figs. 8.03 & 8.04, plates 8.06 to 8.08 and table 8.03.

8.3.1.3 Effects of tantalum additions

The results are given in figs. 8.05 & 8.06 plates 8.09 to 8.11, and table 8.04.

8.3.1.4 Effects of neodymium additions.

The results are given in figs. 8.07 & 8.08 plates 8.12 to 8.14 and table 8.05.

8.3.1.5 Effects of misch metal additions.

The results are given in figs. 8.09 & 8.10 plates 8.15 to 8.18 and table 8.06.

8.3.2 Effects of additions of tantalum, rare - earth elements and misch metal on the properties of alloy 122.

8.3.2.1 Effects of tantalum additions.

The results are given in figs.8.11 & 8.12 plates 8.19 to 8.23 and table 8.07.

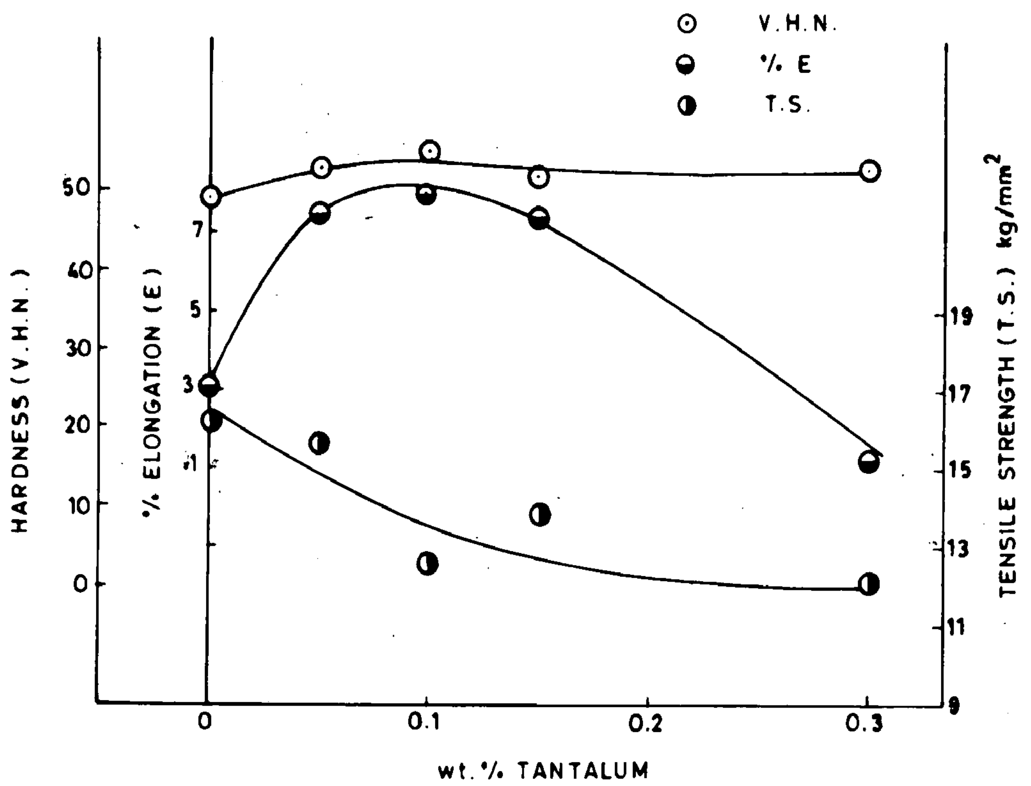

FIG. 8.01_EFFECT OF TANTALUM ON THE MECHANICAL PROPERTIES OF Al + 4.5 % Cu ALLOY

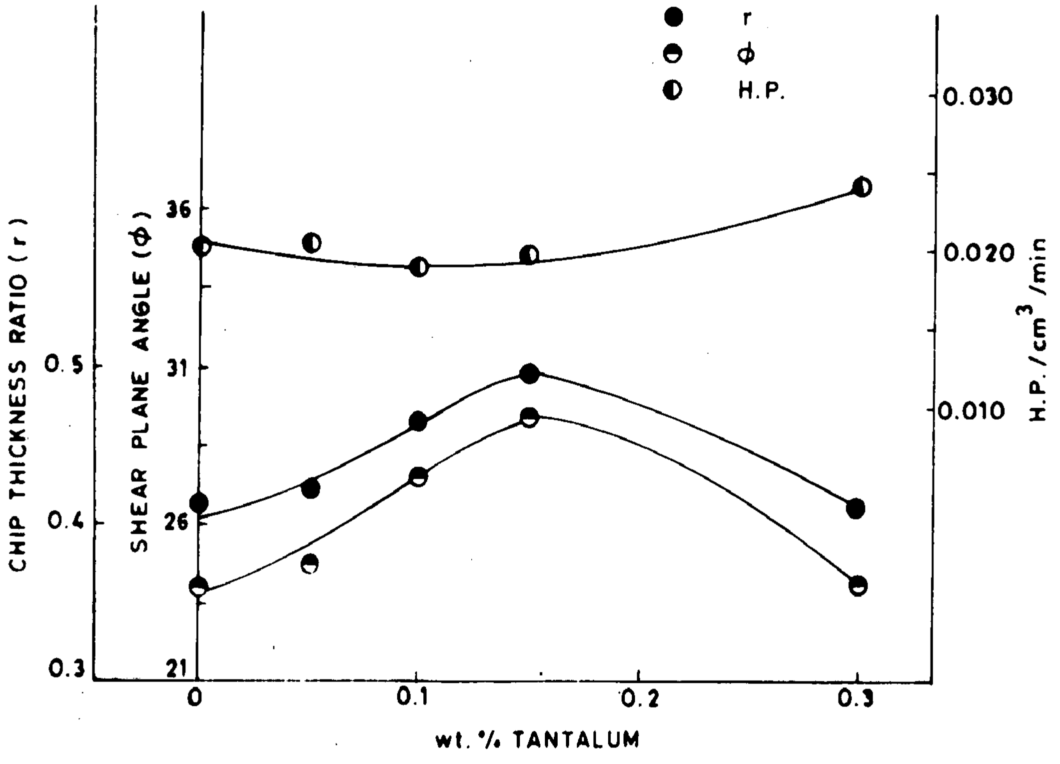

FIG. 8.02_EFFECT OF TANTALUM ON THE MACHINABILITY OF Al + 4.5 % Cu ALLOY

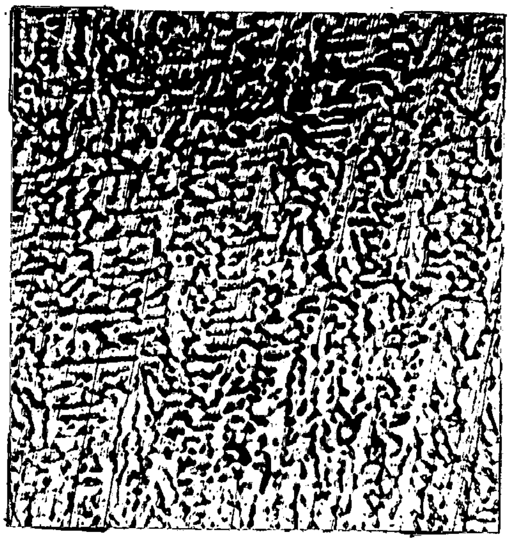

PLATE 8.01 106 X

Al + 4.5 % Cu

PLATE 8.02 106 X

Al + 4.5% Cu +0.05wt.% Ta

PLATE 8.03 106 X

Al +4.5% Cu+0.1wt.% Cu

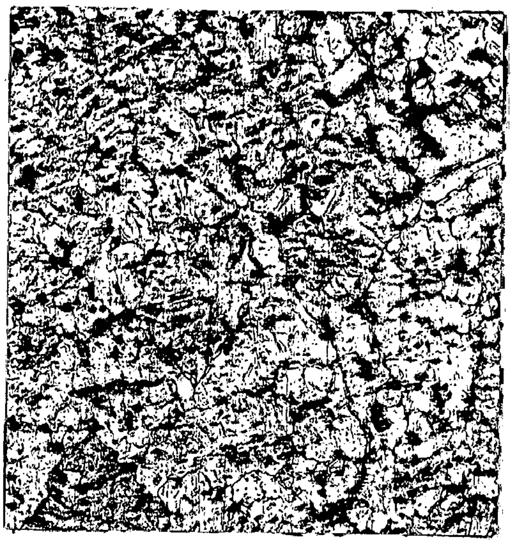

PLATE 8.04 106 X

Al + 4.5 % Cu + 0.15 wt.% Ta

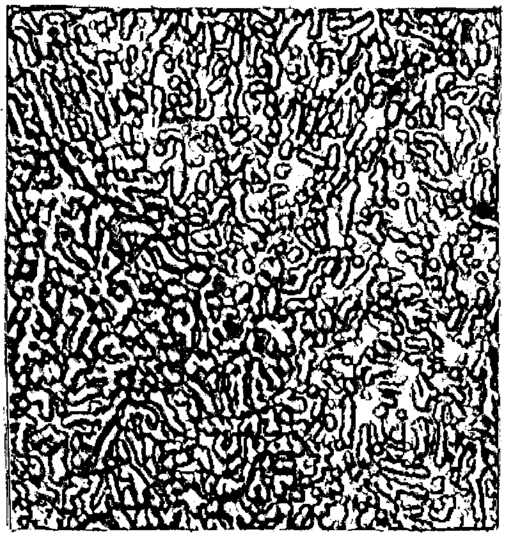

PLATE 8.05 106 X

Al + 4.5% Cu + 0.3 wt.% Ta

TABLE 8.02

EFFECTS OF TANTALUM ADDITION ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 195(4.5% Cu)

S. No.	% alloying element	Tensile strength kgm/mm^2	%elongation	Hardness V.H.N.	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min
1	0.00	16.2	3.1	49	0.412	23°58'	0.0201
2	0.05	15.3	7.5	53	0.422	24°42'	0.0203
3	0.10	12.6	8.0	55	0.464	27°30'	0.0188
4	0.15	13.8	7.5	52	0.493	29°29'	0.0197
5	0.30	12.1	1.2	53	0.414	24°95'	0.0241

TABLE 8.03

EFFECTS OF CERIUM ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 195(4.5% Cu).

S. No.	%alloying element	Tensile strength kgm/mm^2	% elongation	Hardness VHN.	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min
1	0.00	16.2	3.12	49	0.412	23°58'	0.0201
2	0.20	10.8	2.25	50	0.412	23°58'	0.0193
3	0.75	7.1	2.25	52	0.418	24°25'	0.0205
4	1.50	15.5	6.25	51	0.397	23°20'	0.0193

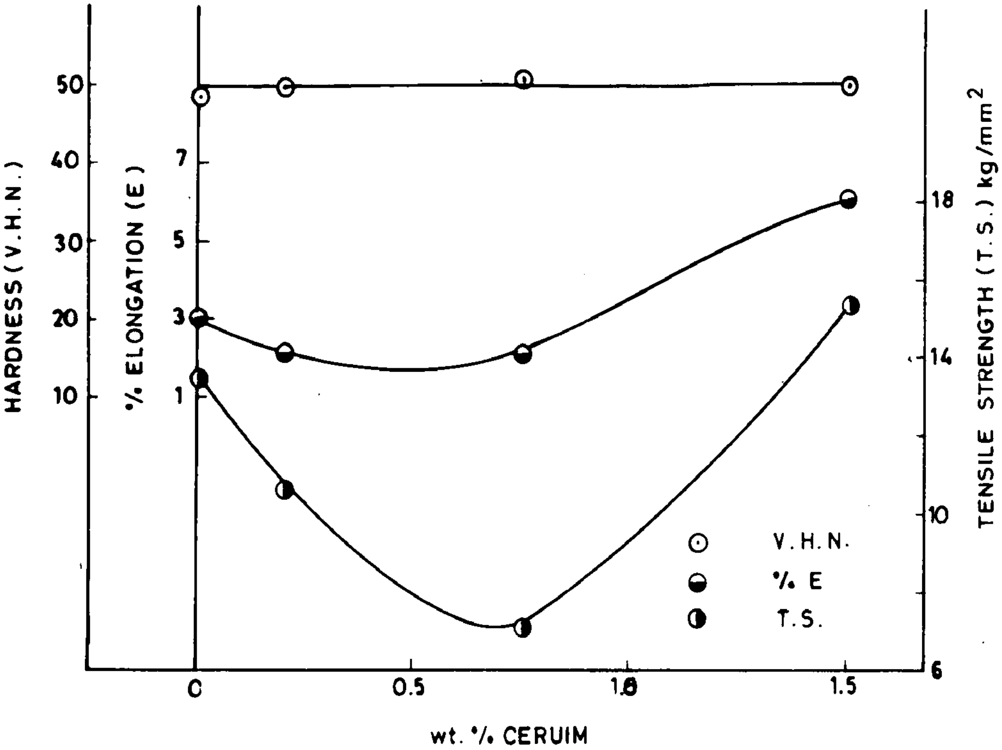

FIG. 8.03_EFFECT OF CERUIM ON THE MECHANICAL PROPERTIES OF Al+4.5 % Cu.ALLOY

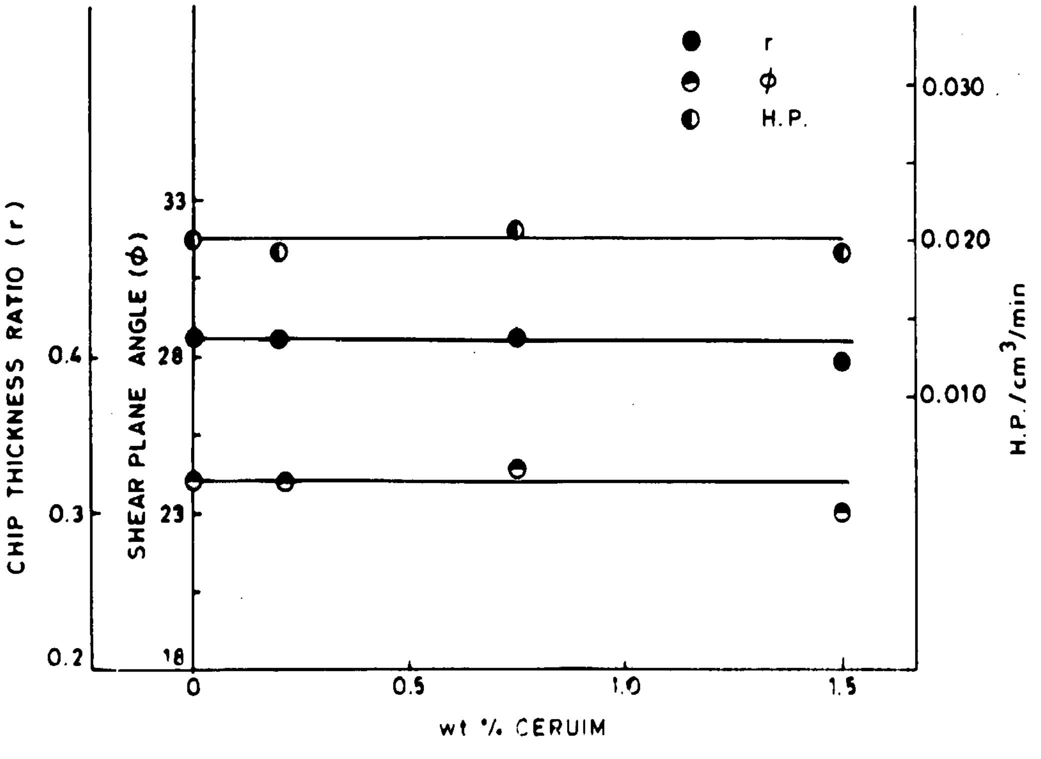

FIG. 8.04_EFFECT OF CERUIM ON THE MACHINABILITY OF Al + 4.5 % Cu ALLOY

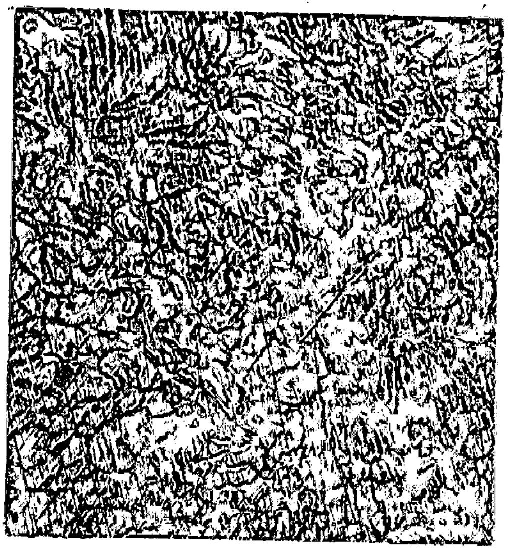

PLATE 8.06 106 X

Al+4.5% Cu+0.2 wt.% Ce

PLATE 8.07 106 X

Al + 4.5% Cu+ 0.75 wt.% Ce

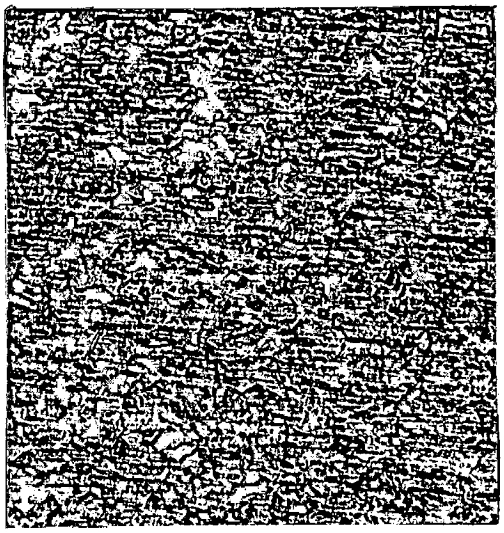

PLATE 8.08 106 X

Al+ 4.5 % Cu+ 1.5 wt.% Ce

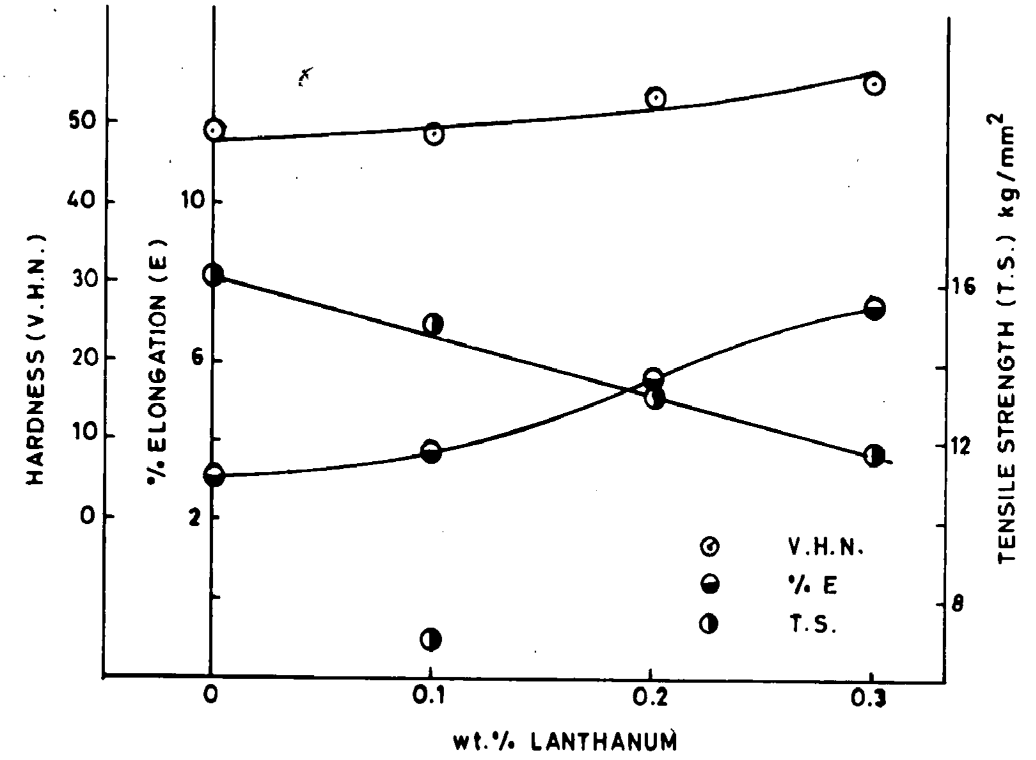

FIG.8.05_EFFECT OF LANTHANUM ON THE MECHANICAL PROPERTIES OF Al+4.5% Cu ALLOY

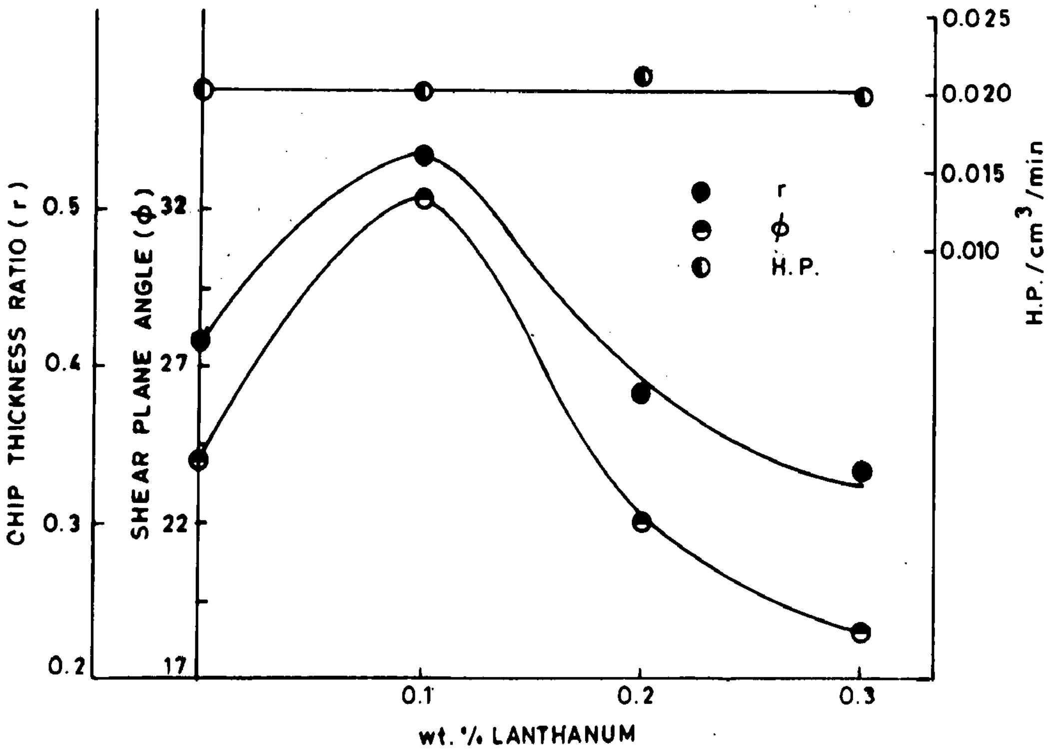

FIG.8.06_EFFECT OF LANTHANUM ON THE MACHINABILITY OF Al+4.5% Cu ALLOY

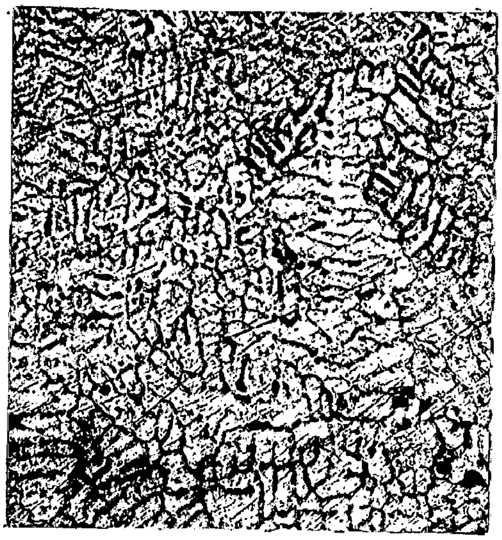

PLATE 8.09 106 X

Al + 4.5% Cu + 0.1wt.% La

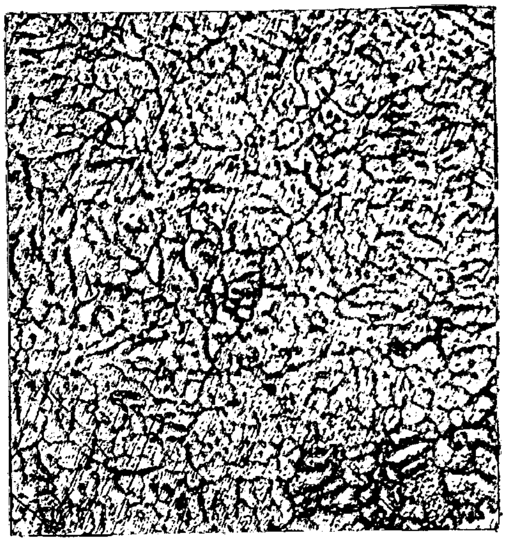

PLATE 8.10 106 X

Al + 4.5% Cu + 0.2 wt.% La

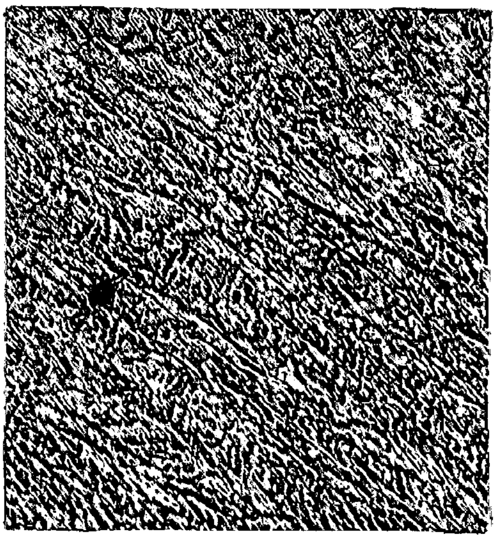

PLATE 8.11 106 X

Al + 4.5% Cu + 0.3wt.% La

TABLE - 8.04

EFFECTS OF LANTHANUM ADDITION ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 195 (4.5% Cu)

S. No.	% alloying element	Tensile strength kgm/mm^2	% Elongation	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min
1	0.0	16.2	3.13	49	0.412	23°38'	0.0201
2	0.1	15.0	3.75	49	0.535	32°21'	0.0201
3	0.2	13.3	5.65	54	0.382	22°02'	0.0210
4	0.3	11.8	7.50	56	0.328	18°34'	0.0197

TABLE 8.05

EFFECTS OF NEODYMIUM ADDITION ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 195 (4.5% Cu)

S. No.	% alloying element	Tensile strength kgm/mm^2	%elongation	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min
1	0.0	16.2	3.10	49	0.412	23°58'	0.0201
2	0.1	13.8	6.50	66	0.340	19°18'	0.0201
3	0.2	11.5	6.30	62	0.348	19°49'	0.0198
4	0.3	13.9	5.50	62	0.348	19°49'	0.0210

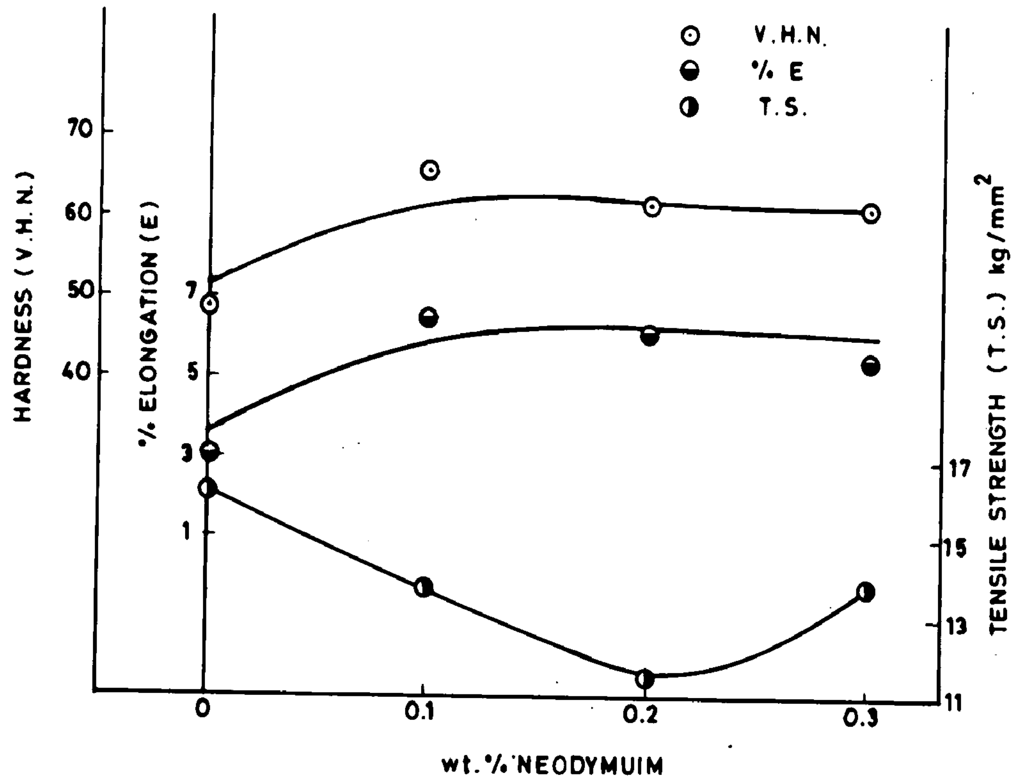

FIG.8.07_EFFECT OF NEODYMIUM ON THE MECHANICAL PROPERTIES OF Al + 4.5 % Cu ALLOY

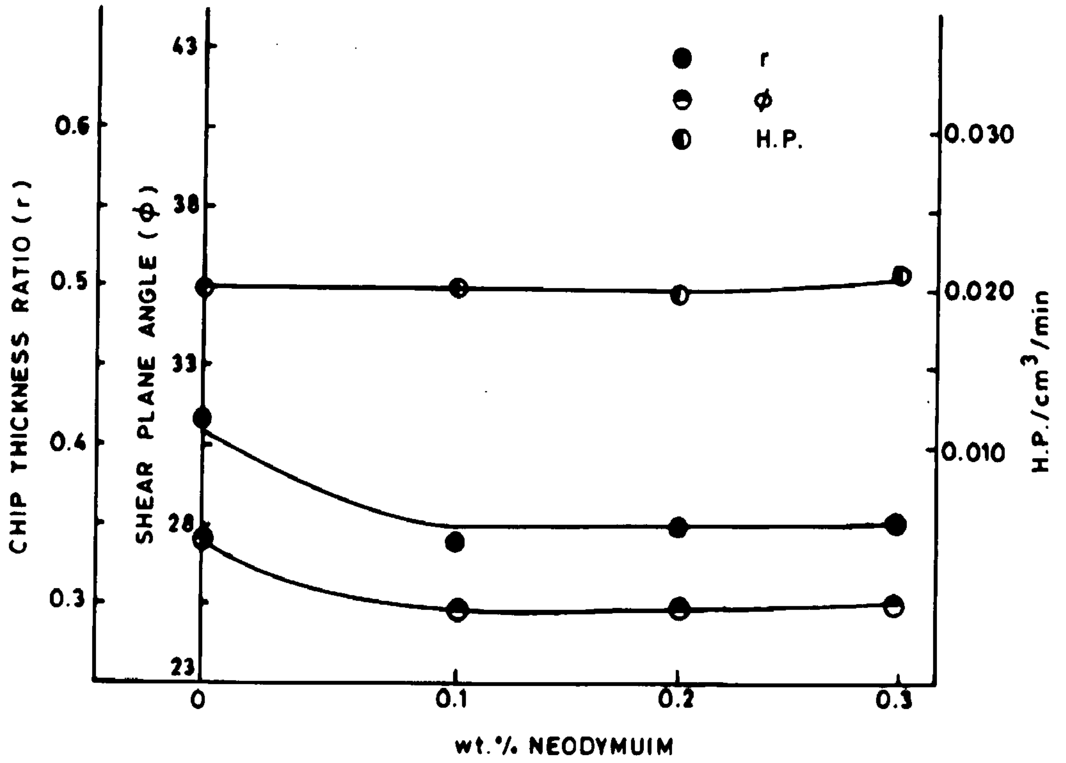

FIG.8.08_EFFECT OF NEODYMIUM ON THE MACHINABILITY OF Al + 4.5 % Cu ALLOY

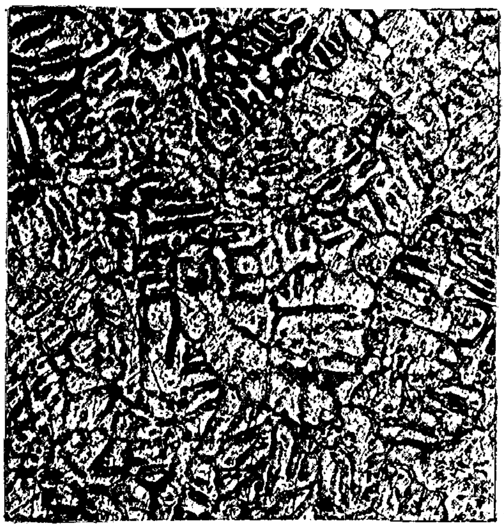

PLATE 8.12 106X

Al + 4.5% Cu + 0.1wt.% Nd

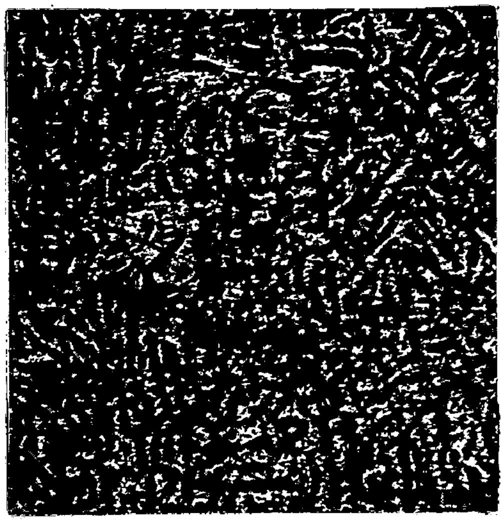

PLATE 8.13 106 X

Al + 4.5% Cu + 0.2 wt.% Nd

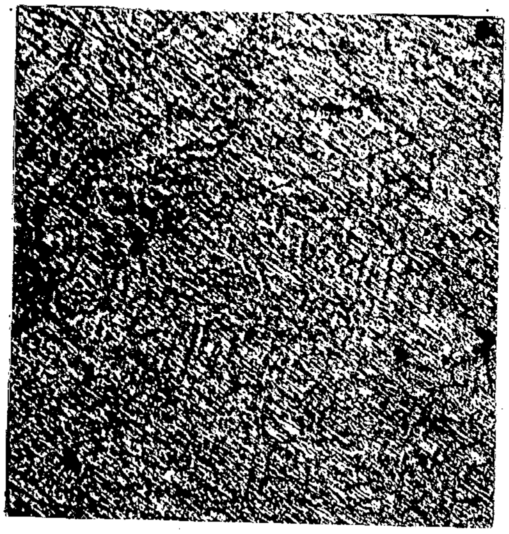

PLATE 8.14 106X

Al + 4.5% Cu + 0.3 wt.% Nd

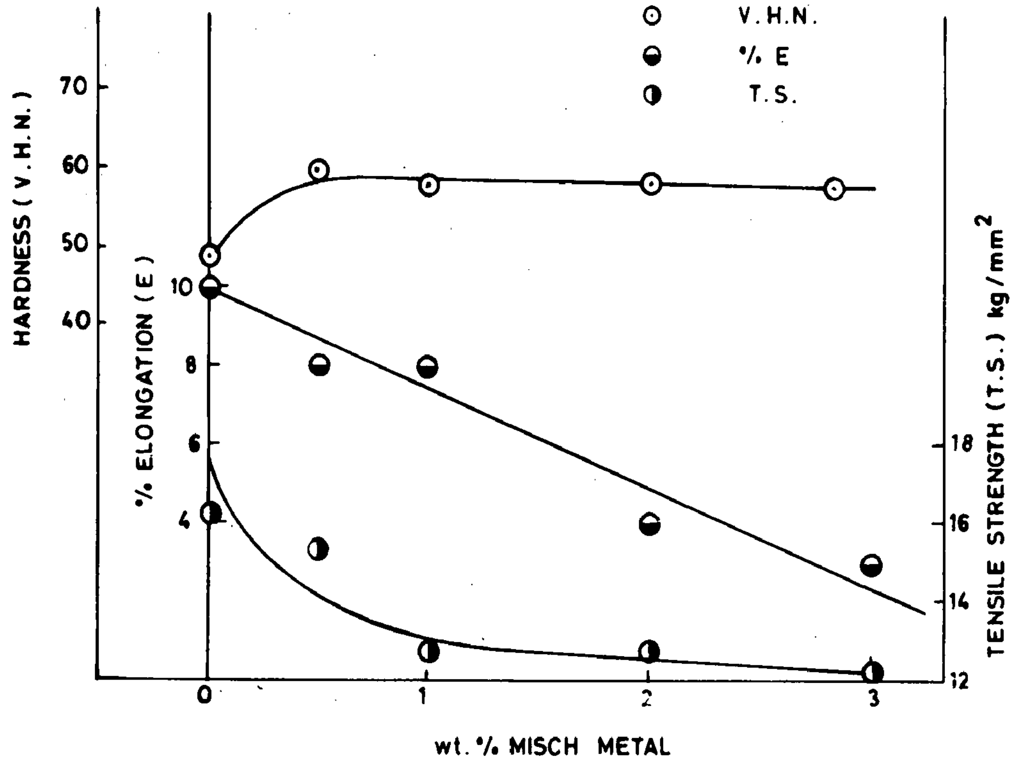

FIG.8.09_EFFECT OF MISCH METAL ON THE MECHANICAL PROPERTIES OF Al+4.5% Cu ALLOY

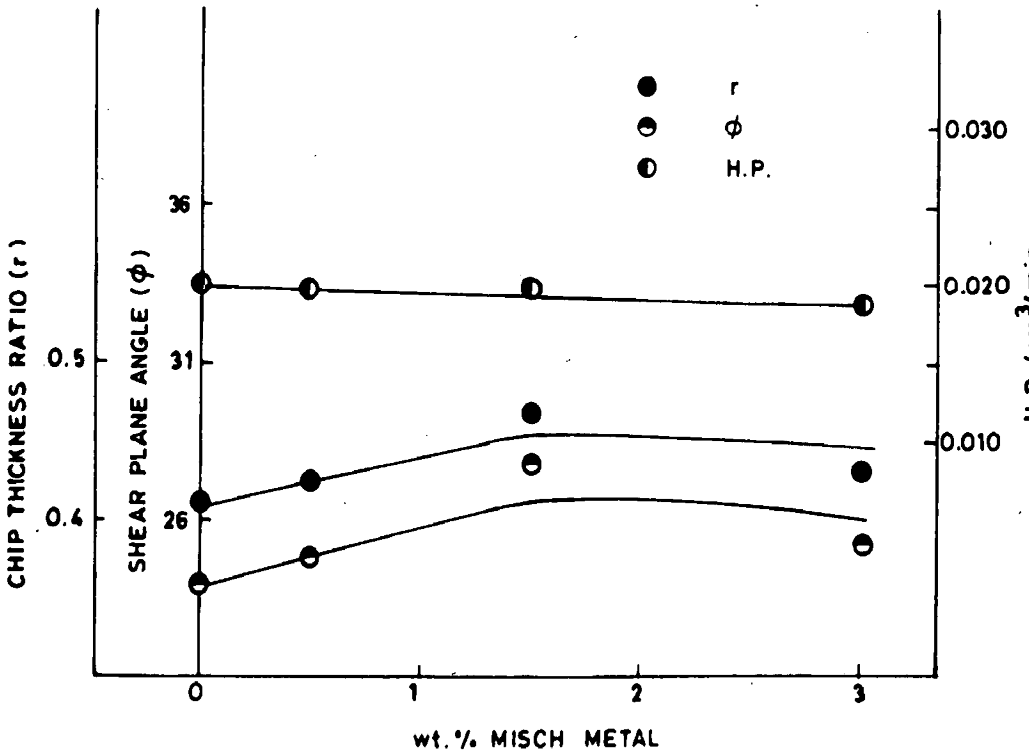

FIG.8.10_EFFECT OF MISCH METAL ON THE MACHINABILITY OF Al+4.5% Cu ALLOY

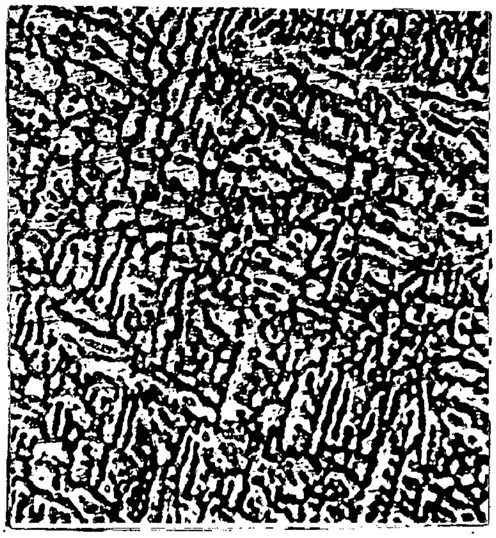

PLATE 8.15 106 X

Al+ 4.5%Cu+ 0.5 wt.% MISCH METAL

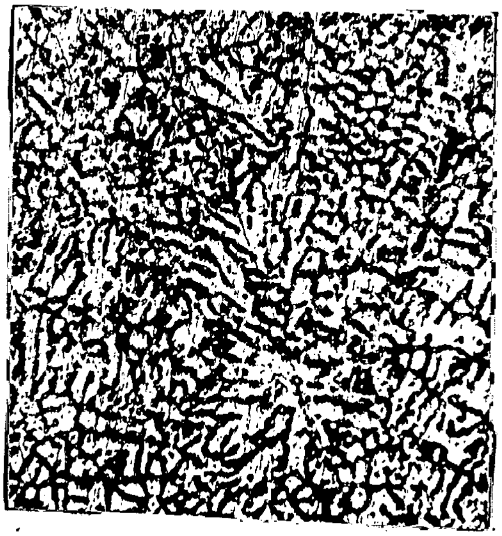

PLATE 8.16 106 X

Al+ 4.5% Cu+ 1.0 wt % MISCH METAL

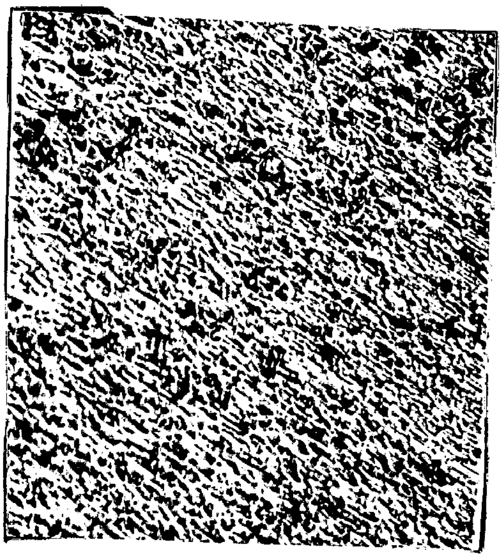

PLATE 8.17 106X

Al+ 4.5% Cu+ 2wt.% MISCH METAL

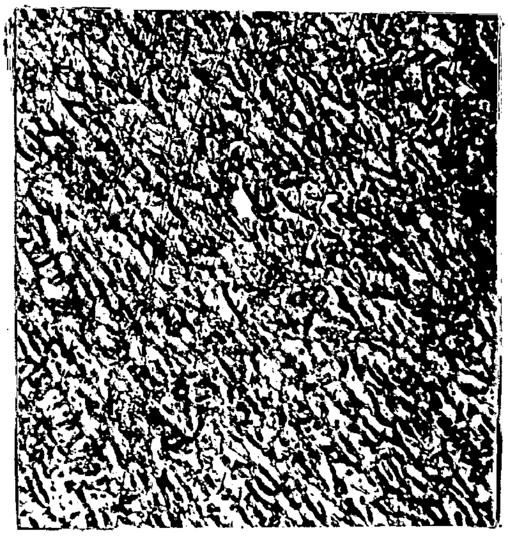

PLATE 8.18 106 X

Al+ 4.5% Cu + 3 wt.% MISCH METAL

TABLE 8.06

EFFECTS OF MISCH METAL ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 195 (4.5% Cu)

S. No.	% alloying element	Tensile strength kgm/mm2	% elongation	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm3/min
1	0.0	16.2	10.00	49	0.412	23°58′	0.0201
2	0.2	14.1	7.00				
3	0.5	15.3	8.00	60	0.422	24°42′	0.0197
4	1.0	12.7	8.00	58			
5	1.5	-	-		0.467	27°43′	0.0198
6	2.0	12.7	5.00	59			
7	3.0	12.3	4.00	58	0.429	25°7′	0.0190

TABLE 8.07

EFFECTS OF TANTALUM ADDITION ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 122 (0% Cu, 0.2% Mg)

S. No.	% alloying element	Tensile strength kgm/mm2	% elongation	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm3/min
1	0.00	12.5	4.38	109	0.589	36°20′	0.0247
2	0.05	25.7	1.20	99	0.611	37°31′	0.0252
3	0.10	22.5	3.00	95	0.611	37°31′	0.0235
4	0.15	23.6	2.25	107	0.580	35°24′	0.0233
5	0.30	22.1	1.20	102	0.580	35°24′	0.0236

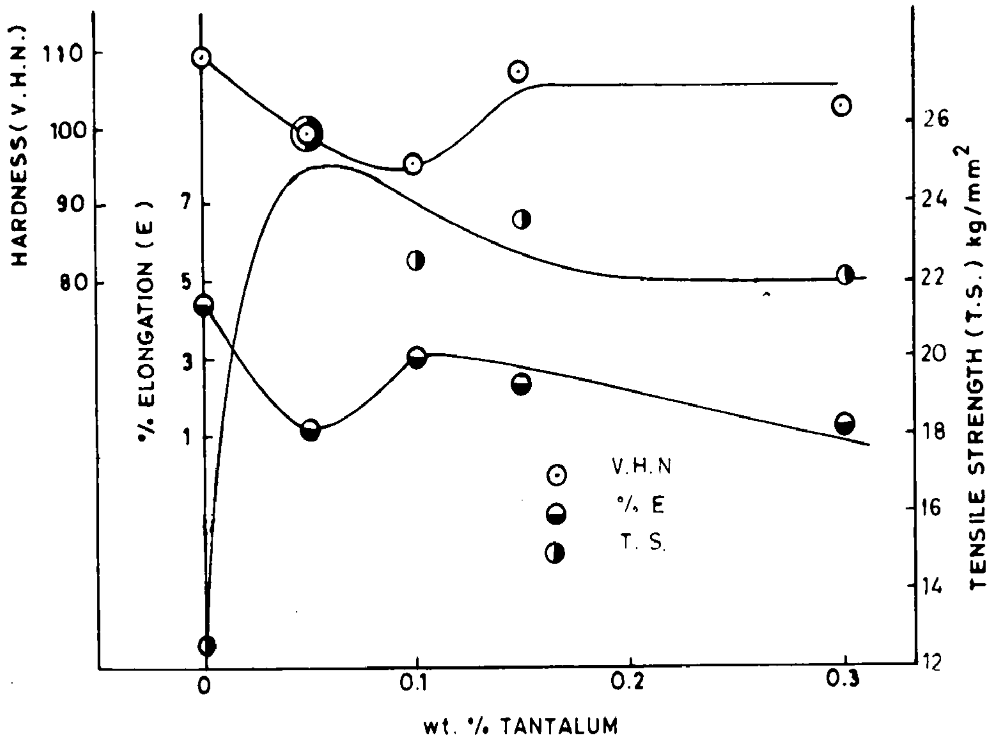

FIG.8.11_EFFECT OF TANTALUM ON THE MECHANICAL PROPERTIES OF ALLOY 122

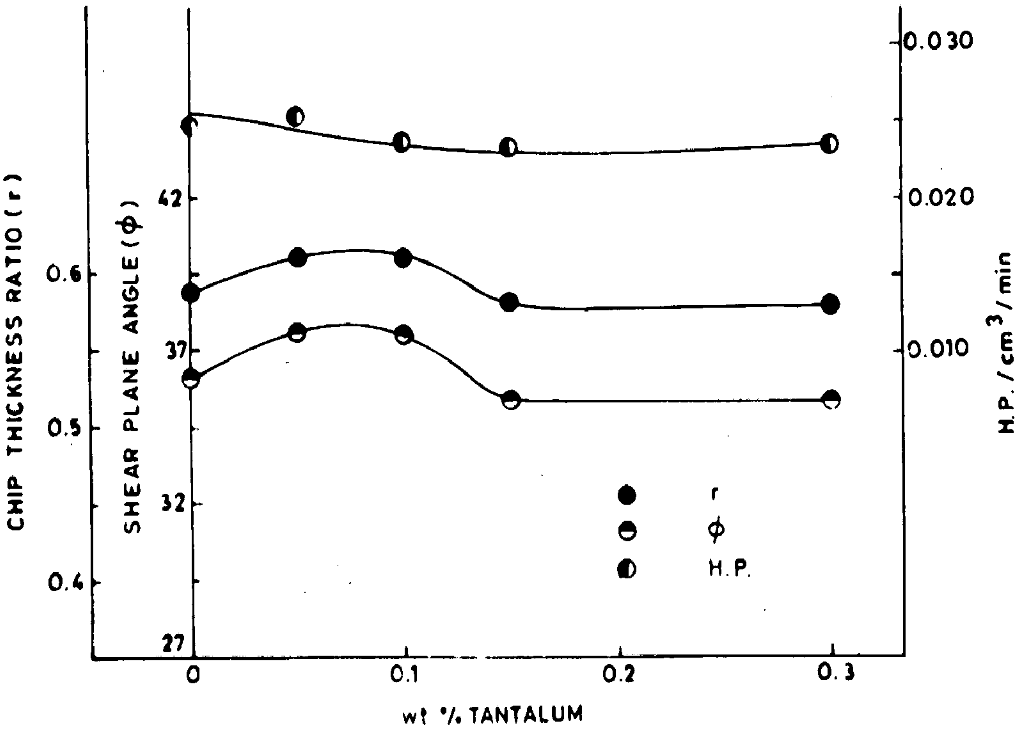

FIG.8.12_EFFECT OF TANTALUM ON THE MACHINABILITY OF ALLOY 122

PLATE 8.19 106 X

ALLOY 122

PLATE 8.20 106 X

ALLOY 122+0.05 wt.% Ta

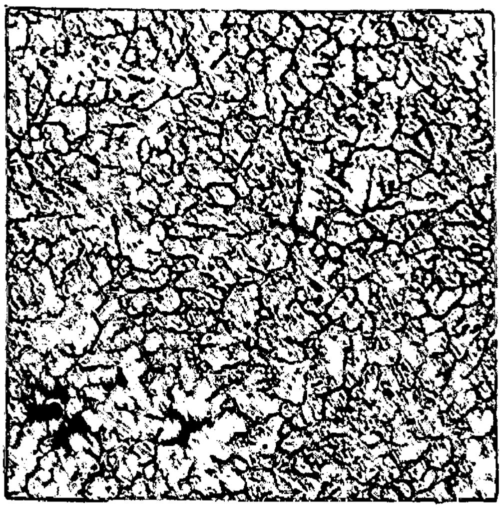

PLATE 8.21 106 X

ALLOY 122+0.1 wt.% Ta

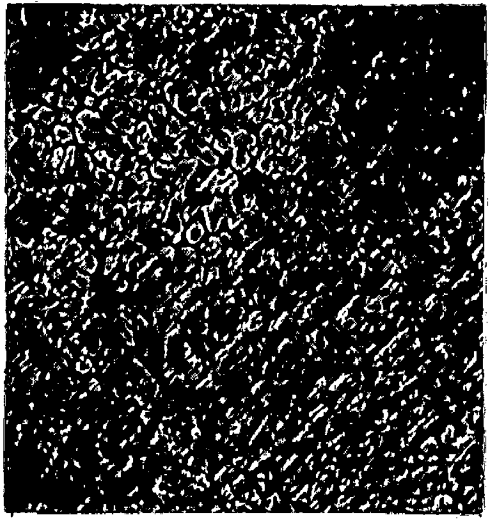

PLATE 8.22 106 X

ALLOY 122+0.15 wt. % Ta

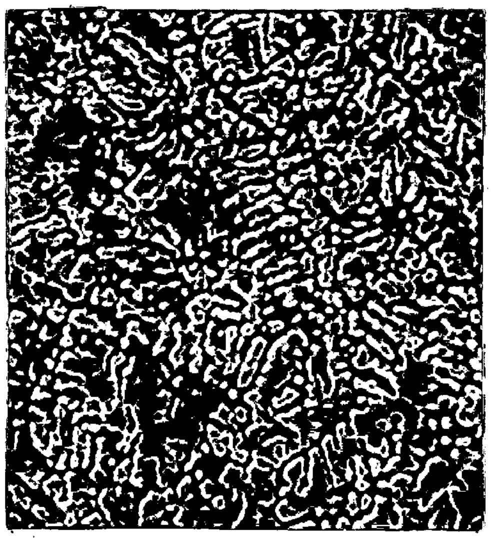

PLATE 8.23 106 X

ALLOY 122+ 0.3 wt. % Ta

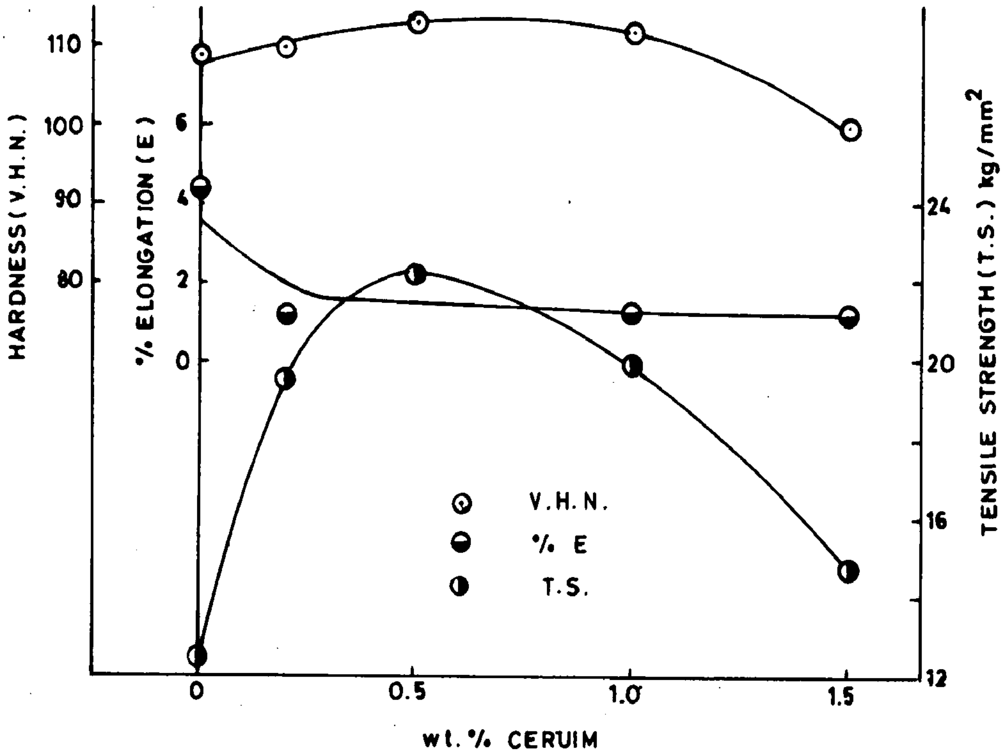

FIG.8.13_EFFECT OF CERUIM ON THE MECHANICAL PROPERTIES OF ALLOY 122

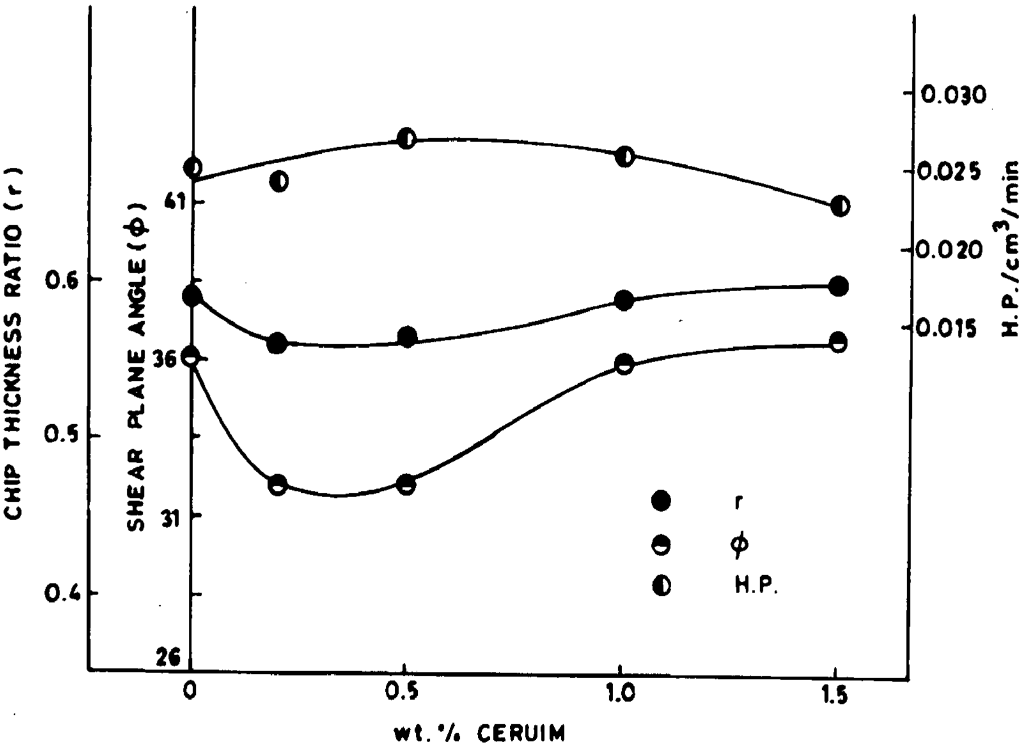

FIG. 8.14_EFFECT OF CERUIM ON THE MACHINABILITY OF ALLOY 122

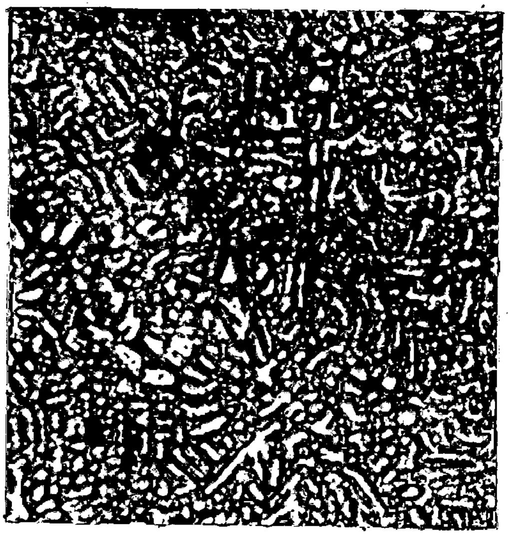

PLATE 8.24 106 X

ALLOY 122 + 0.2 wt. % Ce

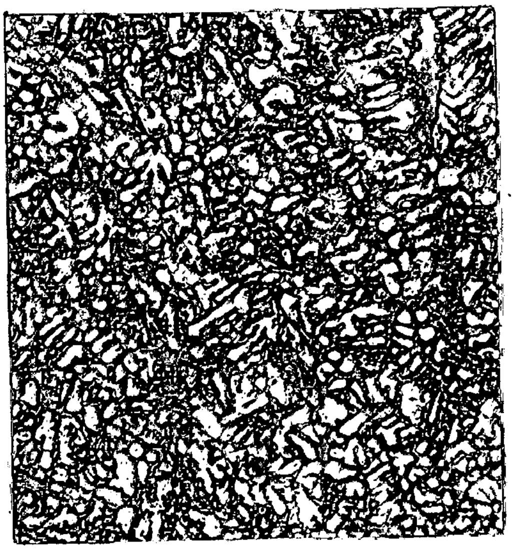

PLATE 8.25 106 X

ALLOY 122 + 0.5 wt. % Ce

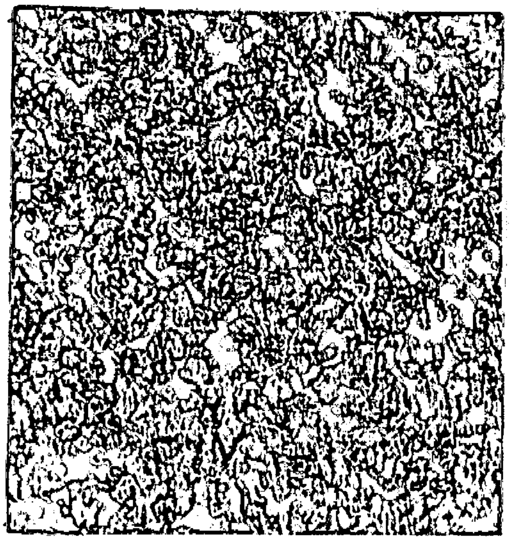

PLATE 8.26 106 X

ALLOY 122 + 1.5 wt. % Ce

TABLE 8.08

EFFECTS OF LANTHANUM ADDITION ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 122 (10% Cu, 0.2% Mg)

S. No.	% alloying element	Tensile strength kgm/mm2	% elongation	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm3/min
1	0.0	12.5	4.38	109	0.589	$36^{\circ}21'$	0.0247
2	0.1	23.1	3.75	108	0.527	$31^{\circ}48'$	0.0254
3	0.2	23.0	2.25	110	0.611	$37^{\circ}31'$	0.0254
4	0.3	23.6	2.00	109	0.600	$36^{\circ}46'$	0.0256

TABLE 8.09

EFFECTS OF CERIUM ADDITION ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF ALLOY 122 (10% Cu, 0.2% Mg)

S. No.	% alloying element	Tensile strength kgm/mm2	% elongation	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm3/min
1	0.0	12.5	4.38	109	0.589	$36^{\circ}20'$	0.0247
2	0.2	19.6	1.20	110	0.561	$32^{\circ}50'$	0.0242
3	0.5	22.2	2.25	113	0.561	$32^{\circ}50'$	0.0268
4	1.0	19.9	1.20	112	0.579	$36^{\circ}60'$	0.0258
5	1.5	14.7	1.20	100	0.600	$36^{\circ}46'$	0.0227

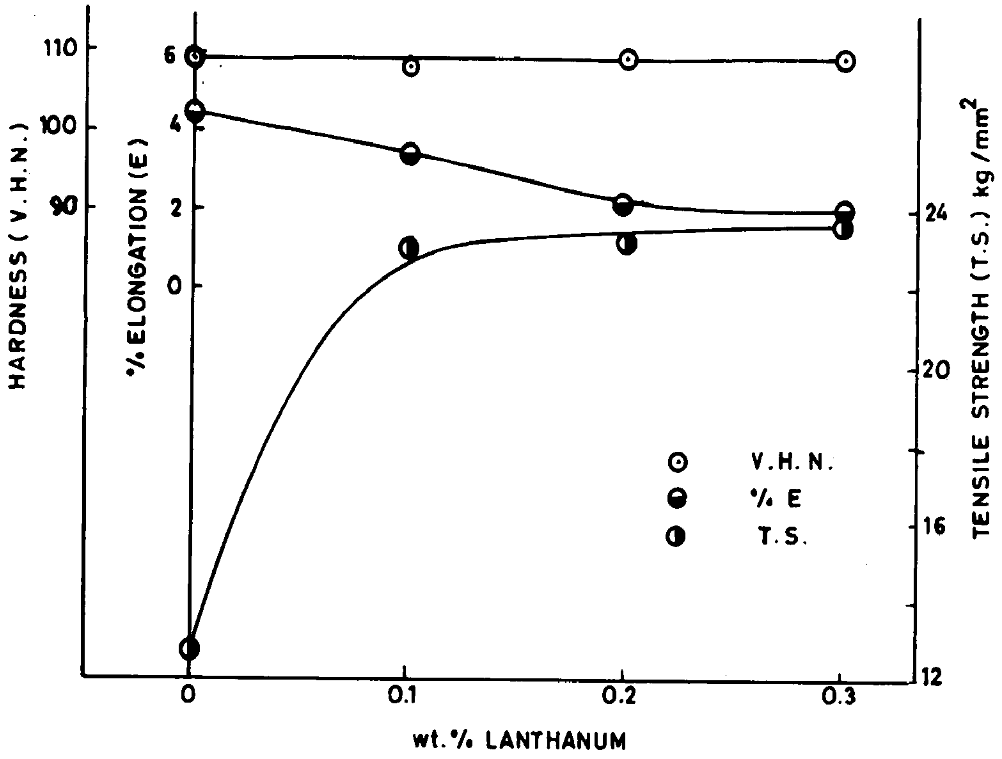

FIG. 8.15_EFFECT OF LANTHANUM ON THE MECHANICAL PROPERTIES OF ALLOY 122

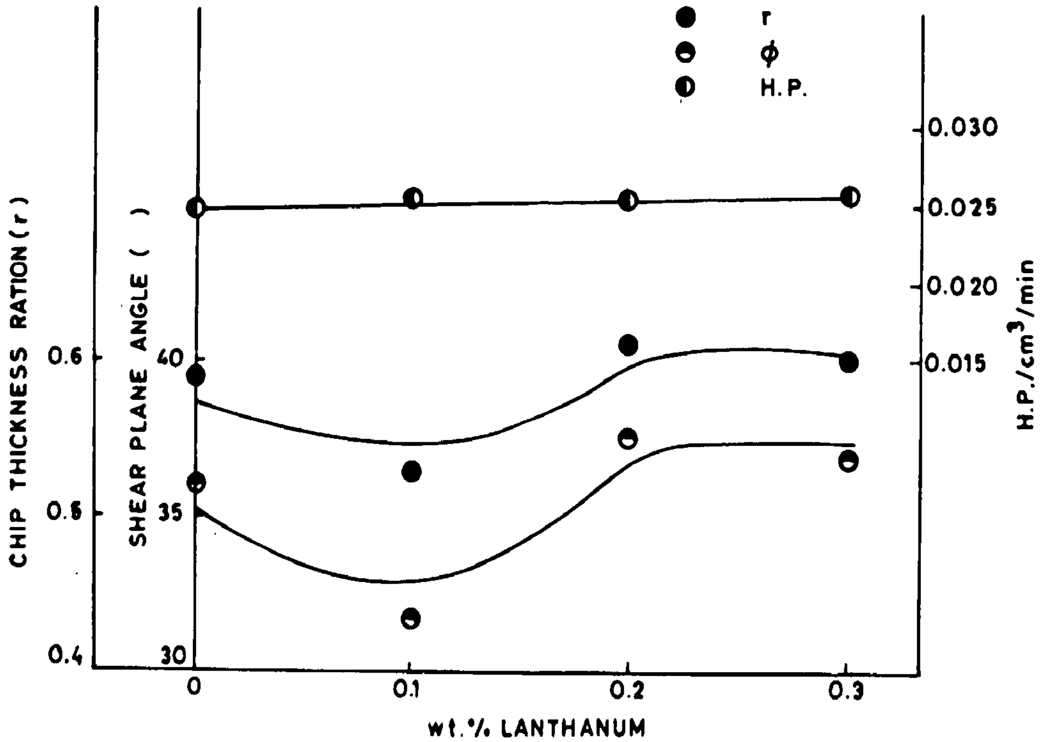

FIG.8.16_EFFECT OF LANTHANUM ON THE MACHINABILITY OF ALLOY 122

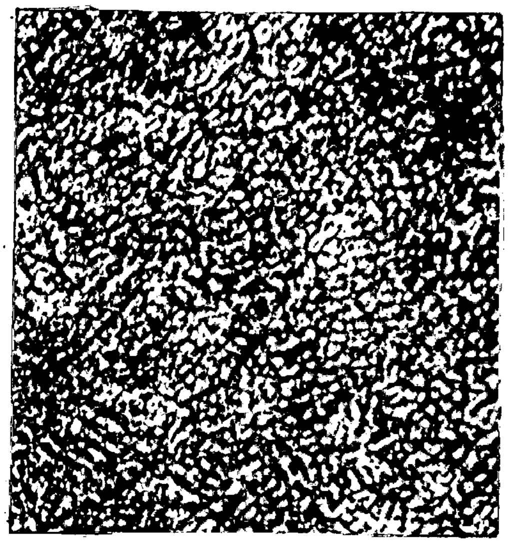

PLATE 8.27 106 X

ALLOY 122+0.1 wt. % La

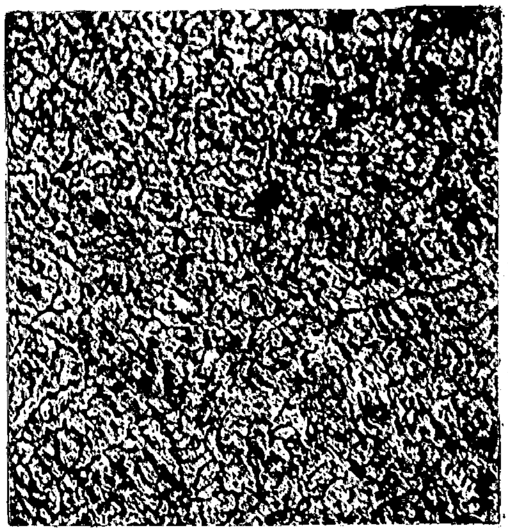

PLATE 8.28 106 X

ALLOY 122+0.2 wt. % La

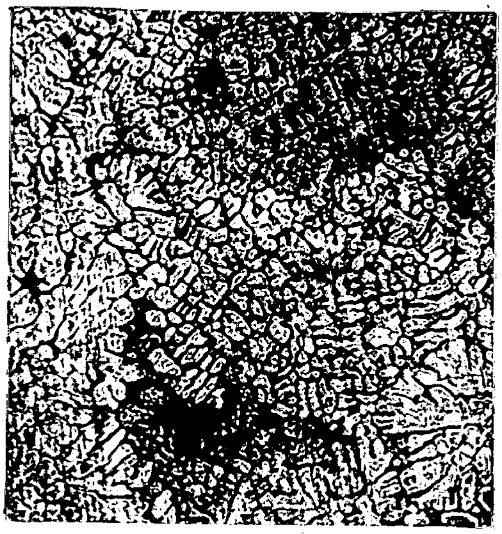

PLATE 8.29 106 X

ALLOY 122+0.3 wt. % La

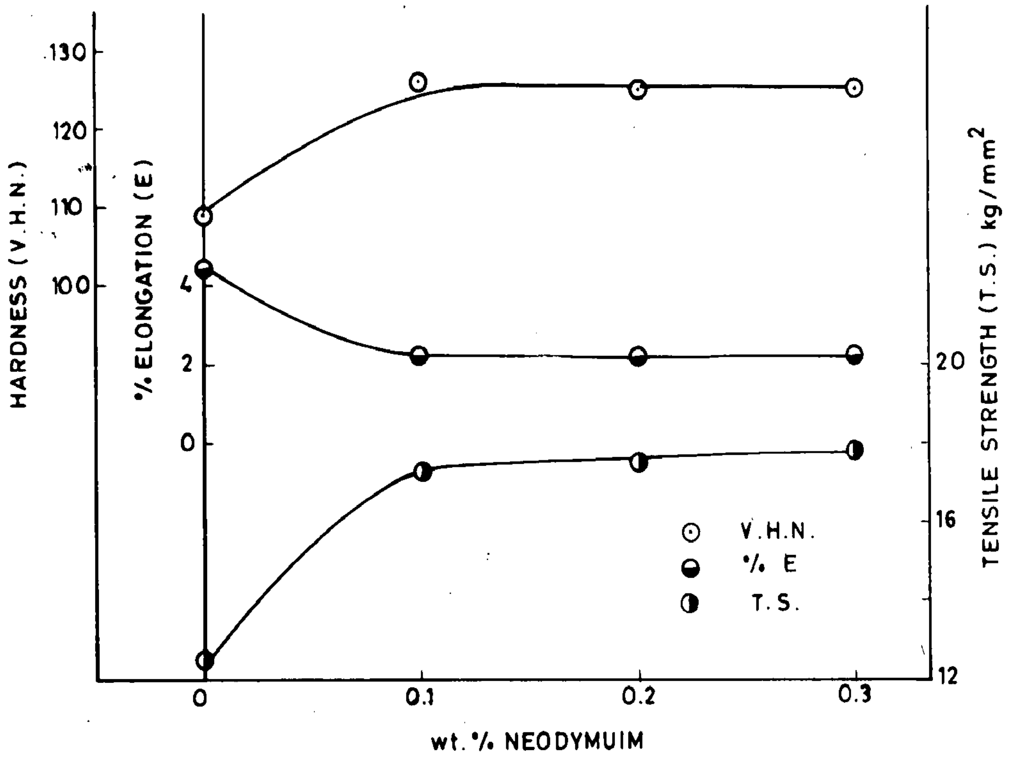

FIG.8.17_EFFECT OF NEODYMIUM ON THE MECHANICAL PROPERTIES OF ALLOY 122

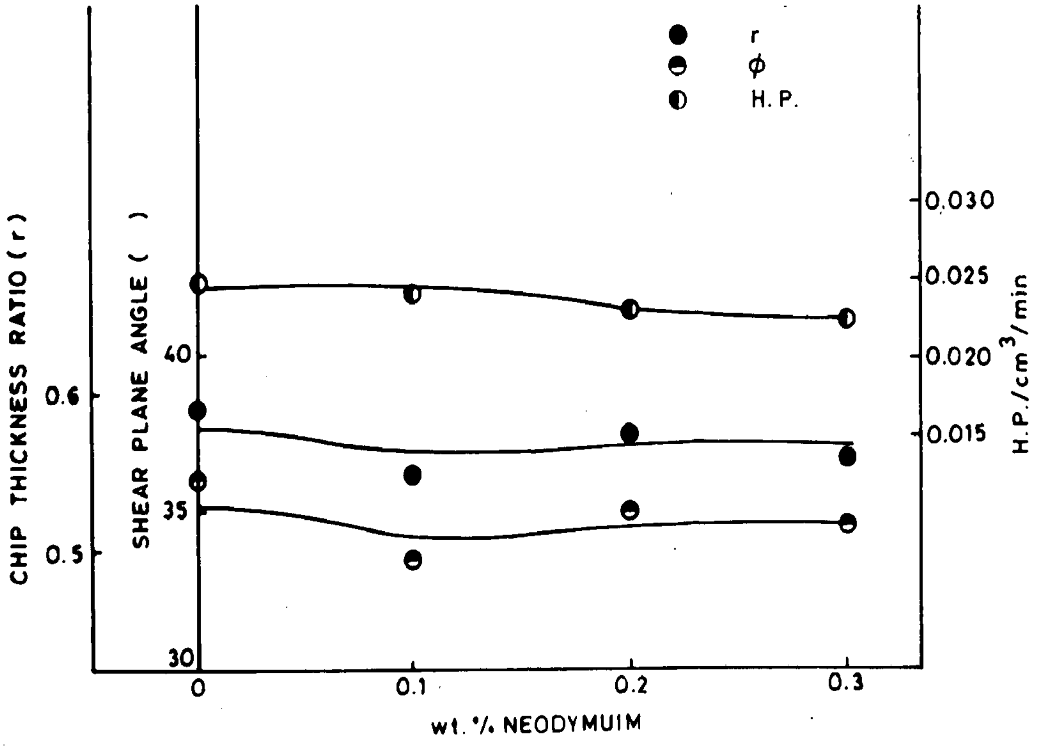

FIG.8.18_EFFECT OF NEODYMIUM ON THE MACHINABILITY OF ALLOY 122

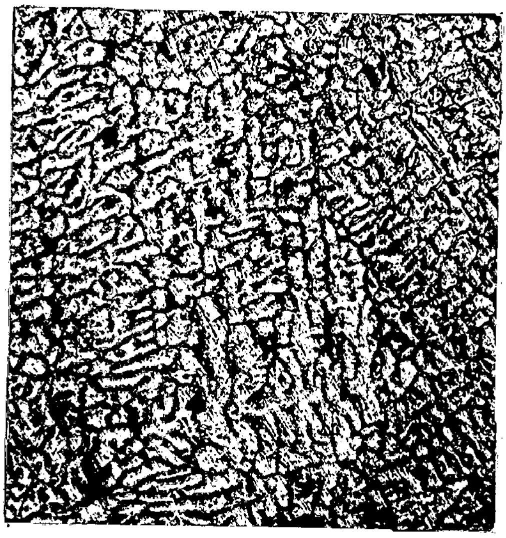

PLATE 8.30 106 X

ALLOY 122+ 0.1 wt.% Nd

PLATE 8.31 106 X

ALLOY 122 + 0.3 wt.%Nd

TABLE 8.10

EFFECTS OF NEODYMIUM ON THE MECHANICAL PROPERTIES & MACHINABILITY OF AL + 10% Cu + 0.2% Mg ALLOY.

S. No.	% alloying element.	Tensile strength kgm/cm^2	% elongation.	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min
1	0.0	12.5	4.38	109	0.589	36°02'	0,0247
2	0.1	17.3	2.25	126	0.551	33°27'	0.0239
3	0.2	17.5	2.25	125	0.575	35°05'	0.0229
4.	0.3	17.8	2.25	125	0.575	35°05'	0.0225

TABLE

EFFECT OF MISCH. METAL ON THE MECHANICAL PROPERTIES & MACHINABILITY OF AL + 10% CU + 0.2% Mg.

S. No	% alloying element	Tensile strengh kgm/cm^2	% elongation	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min
1	0.0	12.5	2.00	109	0.589	36°02'	0.0247
2	0.2	13.5	2.00	155	0.600	36°46'	0.0236
3.	0.5	17.3	2.00	176			
4	1.0	16.2	2.00	194	0.635	39°09'	0.0222
5	1.5	-	-	-	0.611	37°31	0.0248
6	2.0	20.4	4.50	240			
7	3.0	20.4	4.50	255			
8	4.0	18.8	2.10	235	0.600	36°46	0.0220
9	5.0	16.5	1.50	197			

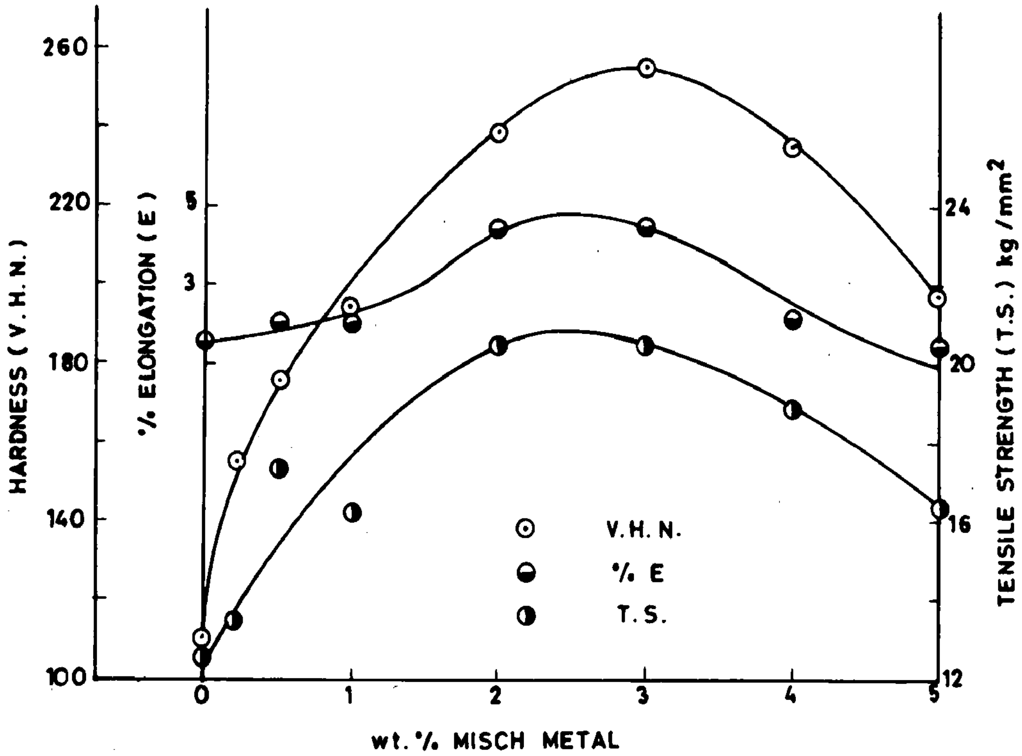

FIG. 8.19_EFFECT OF MISCH METAL ON THE MECHANICAL PROPERTIES OF ALLOY 122

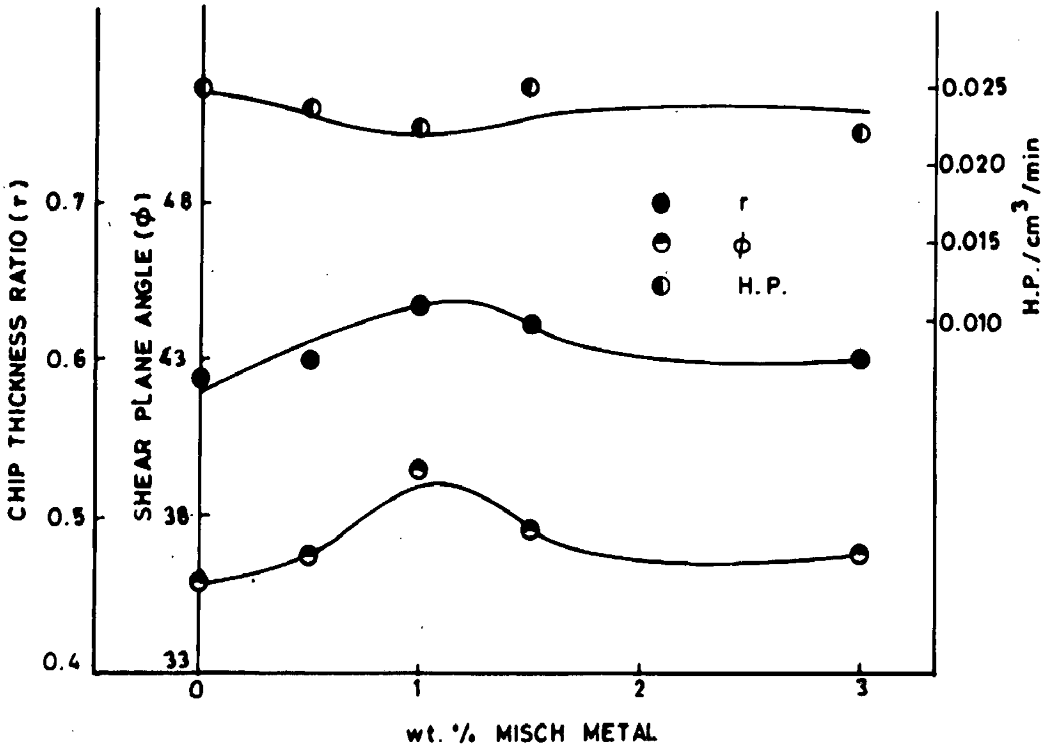

FIG. 8.20_EFFECT OF MISCH METAL ON THE MACHINABILITY OF ALLOY 122

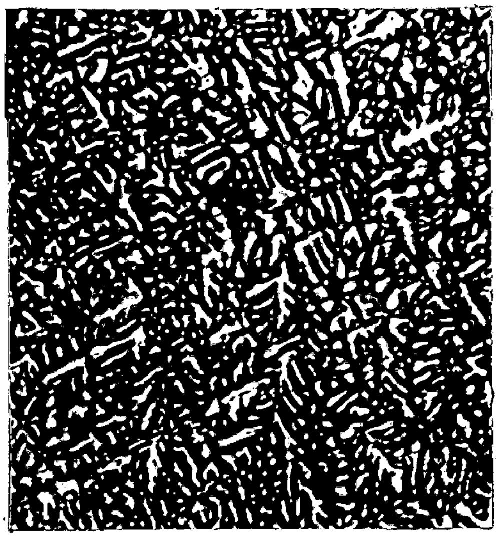

PLATE 8.32 106 X

ALLOY 122+0.5 wt.% MISCH METAL

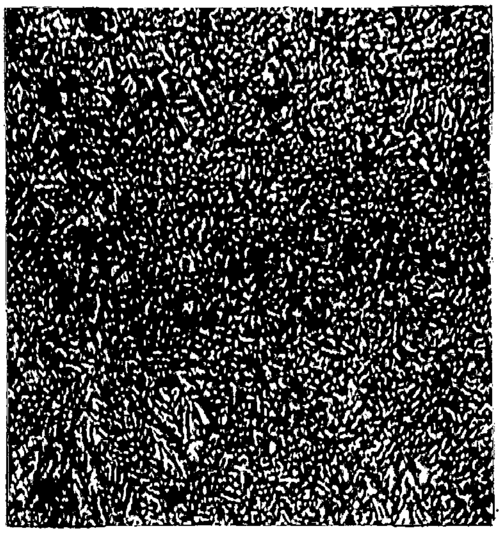

PLATE 8.33 106 X

ALLOY 122+1.0 wt.% MISCH METAL

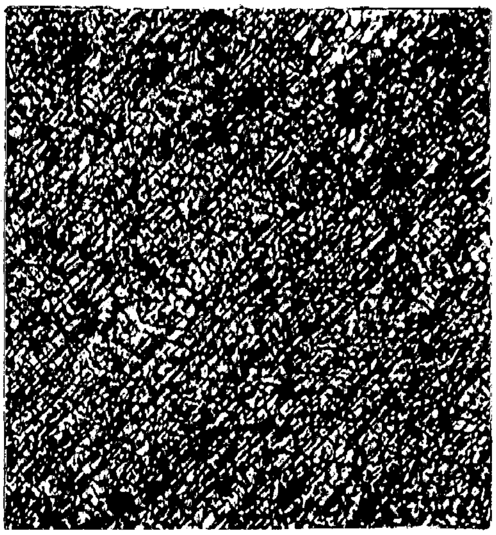

PLATE 8.34 106 X

ALLOY 122+2.0 wt.% MISCH METAL

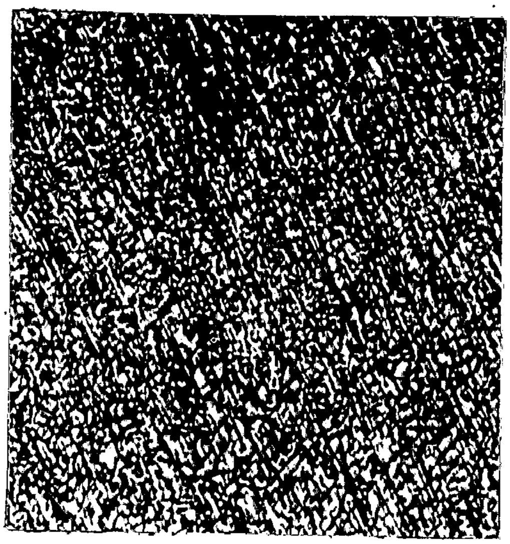

PLATE 8.35 106 X

ALLOY 122+3 wt. % MISCH METAL

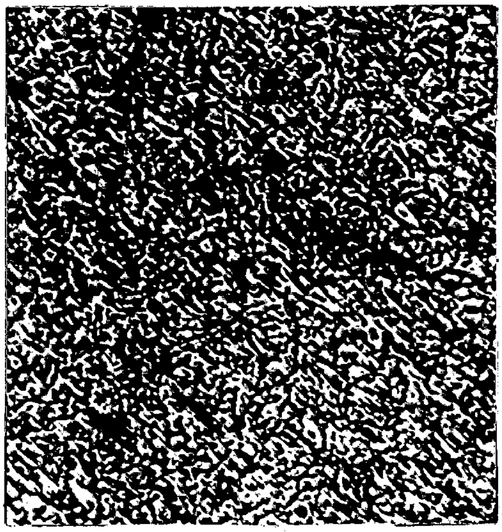

PLATE 8.36 106 X

ALLOY 122+4wt. % MISCH METAL

PLATE 8.37 106 X

ALLOY 122+5wt. % MISCH METAL

8.3.2.2 Effects of cerium addition :

The results are given in Figs. 8.13 & 8.14 plates 8.24 to 8.26 & table 8.08.

8.3.2.3 Effects of lanthanum addition :

The results are given in Figs. 8.15 & 8.16, plates 8.27 to 8.29, and table 8.09 .

8.3.2.4 Effects of neodymium additions :

The, results are given in Figs. 8.17 & 8.18, plates 8.30 to 8.31 and table 8.10 .

8.3.2.5 Effects of misch metal additions :

The results are given in Figs. 8.19 & 8.20, plates 8.32 to 8.37 ; table 8.11.

8.4 DISCUSSION OF THE RESULTS.

8.4.1 Effects of additions of tantalum, rare earth metals and misch metal on the properties of alloy 195 :

84.1.1 Effects of tantalum additions :-

The salient features of the results are given below :-

i) There appears to be very little effect of tantalum addition on the grain size of the alloys. The metallographic study indicates the presence of some additional phases in addition to the normal phases, the amount of which appears to increase with further addition.

ii) With increase in tantalum addition, the tensile strength decreases to nearly a constant value at about 0.3% addition.

iii) There is large improvement in % elongation upto 0.15% tantalum addition, beyond which it goes on decreasing with further and further additions.

iv) There is slight but regular increase in hardness value with tantalum addition.

v) Both chip thickness ratio and shear plane angle Ø increase upto 0.15% addition, beyond which the values go on decreasing with tantalum additions.

vi) H.P/Cm^3/min decreases upto 0.1 % tantalum addition With further additions the value goes on increasing. The curve of unit horse power vs % tantalum addition is found in agreement with chip thickness ratio 'r' and shear plane angle Ø curve.

Tantalum is found to have no solid solubility in aluminium. The only intermetallic compound formed as reported by Marignac and Braner (155) is $TaAl_3$ at 69.09 %. No intermediate phase is formed with copper. Thus with the addition of tantalum there is possibility of formation of very small amount of new phases, the amount of which should increase with increased addition. The increase in hardness and tensile strength can easily be attributed to the formation of fine particles of these intermetallic compounds.

The improvement in machinability may be due

For higher additions large increase in the amount of hard intermediate phases resulting in large increase in hardness and tensile strength, will bring about a large adverse effect on machinability, predominating over the slight beneficial effect due to refining, with a net result of decrease in machinability.

8.4.1.2 Effects of cerium additions:

The salient features of the results are given below:

i) Metallograahic study indicates that effect of cerium addition is to refine the grain up to 0.75% beyond which a little coarsening takes place. There appears to be the presense of some new phase at grain boundary at 0.75% cerium, in addition to normal phases primary alpha and Cu Al_2, the amount of which seems to increase with further additions (Plates 8.06 to 8.08)

ii) The tensile strength is found to decrease up to about 0.75% addition beyond which it goes on increasing with further and further additions.

iii) There is slight decrease in % elongation up to about 0.75% cerium addition beyond which it goes on increasing with further additions.

iv) The hardness remains almost unchanged with cerium addition.

v) There appears to be almost no effect of cerium addition on machinability from chip formation consideration. Both chip thickness ratio and shear plane angle ϕ remain almost unchanged with cerium addition.

vi) H.P./cm^3 / min also remains almost constant with cerium addition, confirming that there is no detectable effect of addition on machinability.

Since cerium dissolves upto 0.05 wt % in aluminium and has no solubility in copper, with the addition of cerium very little of it goes into solid solution. Most of the cerium so added reacts with aluminium and copper to form Ce_3Al_2, CeAl, $CeAl_2$, $CeAl_4$ and CeCu, $CeCu_2$, $CeCu_4$ and $CeCu_6$ the predominating phases being $CeAl_4$, $CeCu_6$, and $CeCu_4$. Presence of these phases has already been seen in the microstructure. Though the exact nature can not be predicted on metallographic study only and needs cereful X-ray or electron diffraction study.

The slight refining action may be due to the particles of these intermediate phases acting as nuclei during solidification. The increase in tensile strength beyond 0.75% may be due to increase in the intermediate phases quite coherent with the lattice.

No detectable effect of cerium addition on machinability has been obtained. It is interesting to note that slight beneficial effect due to refining seems to balance with the adverse effect produced due to increase in the intermediate phases, resulting in constant value of machinability with cerium addition.

8.4.1.3 Effects of lanthanum additions:

The salient features of the results are given below:

i) Metallographic study indicates that there is little refinement produced by lanthanum addition.

ii) The tensile strength goes on decreasing almost linearly with lanthanum additions.

iii) Large beneficial effect of lanthanum additions is produced on % elongation. It goes on increasing with increase in lanthanum additions.

iv) Large increase in hardness value is observed with cerium additions. There is little effect upto 0.1% addition, beyond which it goes on increasing appreciably with further and further additions.

v) Both chip thickness ratio and shear plane angle Ø increase upto 0.1% addition, beyond which both the values are found to decrease with further and further additions.

vi) H.P. remains almost unchanged with lanthanum additions.

Lanthanum is found to have no solid solubility in aluminium and copper. Following intermediate phases are reported: La_3Al, La_3Al_2, $LaAl$, $LaAl_2$ and $LaAl_4$; $LaCu$, $LaCu_2$, $LaCu_4$, $LaCu_6$ and $LaCu_5$. Thus with the addition of lanthanum there is possibility of the formation of these phases, predominating phases being $LaAl_4$, $LaCu_4$, $LaCu_6$ and $LaCu_5$, the amount of which should increase with increased addition. Presence of these phases could not be dectected on our metallographic study probably due to their submicroscopic size. Increase in hardness and tensile strength with lanthanum additions beyond 0.1% may be due to the increasing amount of these fine intermediate phases. The refining effect produced is due to the presence of fine particles of these intermediate phases acting as nuclei during solidification. The increase in % elongation may be attributed to the refinement of α in the microstructure.

The slight beneficial effect of lanthanum addition upto 0.1% on the machinability can easily be understood on the basis of the following considerations:

i) The slight refinement effect produced due to lanthanum addition should result in the improvement of machinability.

ii) The increasing amount of intermediate phases with increased lanthanum addition should bring about an adverse effect on machinability.

Lanthanum is found to have no solid solubility in aluminium and copper. Following intermediate phases are reported: La_3Al, La_3Al_2, LaAl, $LaAl_2$ and $LaAl_4$; LaCu, $LaCu_2$, $LaCu_4$, $LaCu_6$ and $LaCu_5$. Thus with the addition of lanthanum there is possibility of the formation of these phases, predominating phases being $LaAl_4$, $LaCu_4$, $LaCu_6$ and $LaCu_5$, the amount of which should increase with increased addition. Presence of these phases could not be dectected on our metallographic study probably due to their submicroscopic size. Increase in hardness and tensile strength with lanthanum additions beyond 0.1% may be due to the increasing amount of these fine intermediate phases. The refining effect produced is due to the presence of fine particles of these intermediate phases acting as nuclei during solidification. The increase in % elongation may be attributed to the refinement of α in the microstructure.

The slight beneficial effect of lanthanum addition upto 0.1% on the machinability can easily be understood on the basis of the following considerations:

i) The slight refinement effect produced due to lanthanum addition should result in the improvement of machinability.

ii) The increasing amount of intermediate phases with increased lanthanum addition should bring about an adverse effect on machinability.

Since the amount of intermediate phase should be very small, for lanthanum addition upto 0.1% the beneficial effect due to (i) may predominate over the adverse effect due to the presence of intermediate phases with a net result of improvement in machinability. For higher additions, however, the increased amount of hard intermediate phases should bring about the large adverse effect, predominating over the slight beneficial effect due to slight refinement, resulting in decrease of machinability.

8.4.1.4 Effects of neodymium additions:

The salient features of the results are given below:

i) On metallographic study it is observed that there is appreciable refining effect in grain with neodymium addition beyond 0.2% upto which very little refining takes place.

ii) The tensile strength decreases almost linearly upto 0.2% addition beyond which it again rises.

iii) The % elongation increases to more or less a constant value at about 0.1 Wt % addition.

iv) The hardness also increases to a constant value at about 0.1 % neodymium addition.

v) Both chip thickness ratio and shear plane angle ϕ decrease to constant values at 0.1 Wt% addition.

vi) H.P./Cm^3/min. is found to remain almost constant with neodymium additions.

Neodymium is reported to have solubility of 0.2% in aluminium and no solubility in copper. The only phases that are formed are NdAl, $NdAl_2$ and $NdCu_5$. Thus with the addition of Nd content upto about 0.2% of it will go into solid solution, beyond which it may react with aluminium and copper to form the above intermediate phases. These phases could not be detected on metallographic study probably due to their being of submicroscopic size. The refining action produced beyong o.2% may be due to the presence of these fine intermediate phases acting as nuclii during solidification. The increase in tensile strength and hardness may be due to fine dispersion of the intermediate phases. The increase in % elongation upto 0.1% may be due to neodymium going into solid solution.

The machinability has been found out to remain almost unchanged with neodymium addition though detectable refining effect has been produced at about 0.2% addition.

It appears that little beneficial effect produced due to refining on machinability is counter-balanced by the adverse effect due to the presence of intermediate phases resulting in almost constant value of machinability.

8.4.1.5 Effect of misch metal additions:

The salient features of the results are given below:

i) There appears to be appreciable refinement of the α grains with misch metal addition. The normal phases that exist are α grains, and $CuAl_2$ phase at the grain boundary. With increasing misch metal, there appears to be the presence of some additional phases in addition to the normal phases, the amount of which seems to increase with further addition. (Plates 8.15 to 8.18).

ii) The tensile strength decreases to a cons-tant value at about 3% addition. It decreases at an appreciable rate upto 1% beyond which the rate decreases.

iii) The % elongation goes on decreasing with increasing misch metal addition.

iv) Hardness increases to more or less a constant value at about 1% addition.

v) Both chip thickness ratio and shear plane angle ϕ increase, with misch metal addition, appreciably upto 1.5%, and at a slow rate between 1.5% to 3%, indicating the beneficial effect on machinability upto 3% addition from chip formation consideration.

vi) HP/Cm^3/min. is found to decrease gradually with misch metal addition upto 3% the result being in agreement with chip thickness ratio and shear plane angle Ø plot.

The above effects so observed with misch metal addition can easily be explained on the basis of the effects of individual elements viz. cerium, lanthanum, and neodymium contained in misch metal. Cerium and neodymium dissolve upto 0.05% and 0.2% by weigh t in aluminium respectively. Lanthanum has no solid solubility. The above three elements are found to have no solid solubility in copper. Thus out of cerium , lanthanum & neodymium content of misch metal very little goes into solid solution. Most of the cerium and lanthanum will react with aluminium and copper to form the intermediate phases: Al_4Ce, Al_4La, $CeCu_4$, $CeCu_6$, $LaCu_4$, $LaCu_5$ and $LaCu_6$ etc. The presence of the above phases has been actually observed on metallographic study though the exact nature could not be detected on metallographic study only. The refining effect produced may be due to the presence of fine particles of these intermediate phases acting as nuclei during solidification. It can easily be seen from the effect of individual elements, that all the three elements cerium, lanthanum and neodymium have contributed in refining the grains. The large refining effect produced may be due to large cumulative effects produced due to these elements.

The decrease in tensile strength upto 1% addition of misch metal is not surprising. 1% misch metal will contain approximately 0.5% Ce, 0.2% La and 0.15% Nd. It can easily be seen from the effects of individual elements that detrimental effect produced by misch metal upto 1% is contributed by all the three elements. Both cerium and neodymium have detrimental effect, on tensile strength upto 0.75% and 0.2% addition respectively, beyond which they have beneficial effects, while lanthanum has detrimental effect produced due to lanthanum addition is counterbalanced by the slight beneficial effect produced due to cerium and neodymium.

Fig. 8-05 indicates the effect, of misch metal addition on % elongation. The % elongation is found to decrease almost linearly with misch metal addtion. It is interesting to note that where all the three elements cerium, lanthanum and neodymium have beneficial effects on % elongation, the effect of misch metal addition can not be explained on the basis of the effects of individual elements. The decrease in % elongation may be probably due to large increase in the amount and coarsening of the precipitates.

Large beneficial effect of misch metal addition on machinability has been produced upto 1.5% addition both from chip formation and power consumption considerations. Beyond 1.5% addition the machinability tends towards constancy from chip formation consideration although from power consideration it steadily decreases with misch metal additions. The improvement in machinability with misch metal addition can easily be understood on the basis of the following considerations :

i) With misch metal addition refinement of grain takes place, tending to improve the machinability.

ii) With increasing misch metal addition there is an increase in the amount of the brittle intermediate phases resulting in improvement in machinability.

For lower additions since the amount of intermediate phase is very small, there appears to be large increase in machinability due to refinement. For higher additions however a part of the beneficial effect is counterbalanced by the adverse effect produced due to increase in the amount of intermediate phases resulting in decrease in the rate of increase of machinability with misch metal addition.

8.4.2 Effects of addition of tantalum, rare-earth metals and misch metal on the properties of alloy 122

8.4.2.1 Effects of tantalum additions;

The salient features of the results are given below;

i) There is pronouneed refining action of tantalum addition upto about 0.15% addition. No significant effect is observed on further additions.

ii) The tensile strength increases upto 0.025% addition beyond which it decreases to a constant value at about 0.1% addition.

iii) The % elongation decreases upto 0.05% addition. Between 0.05% to 0.1% addition there is a slight increase. With further additions it again decreases almost linearly.

iv) The hardness first decreases upto about 0.1% addition, beyond which it increases to a constant value at about 0.15% addition.

v) Both chip thickness ratio and shear plane angle Ø increase upto 0.1% addition, beyond which each decreases to a constant value at about 0.15% addition.

vi) H.P./Cm^3/min. is found to decrease upto 0.15% addition beyond which it again increases.

As discussed earlier since tantalum has no solubility in aluminium, with the addition of tantalum the only intermediate phases formed is $TaAl_3$, the aamount of which will increase with increased additions. The large refining effect may be due to the presence of fine particles of the intermetallic compound. The increase in tensile strength may be due to the large aamount of refinement of primary alpha and eutetic.

The improvement in machinability for smaller additions upto 0.15% addition, may be attributed to the large refinement of primary alpha and eutectic , since the adverse effect due to intermediate phase will be very small. For larger additions the large adverse effect due to large increase in the amount of intermediate phase, will predominate over the beneficial effects due to refining with a net result of decrease in machinability.

8.4.2.2 Effects of cerium addition.

The salient features of the results are given below:

i) There is a pronounced refining action of cerium addition upto a bout 1.5% cerium. At about 1.5% cerium, some new phase in addition to the alpha solid solution and the eutectic makes its appearance and seems to increase with fruther addition.

ii) The tensile strength is found to increase appreciably with cerium addition, gradually upto 0.5% beyond which it decreases.

iii) The % elongation decreases to a constant value at about 0.5% addition..

iv) The hardness is found to increase upto about 0.5% addition beyond which it decreases.

v) Both chip thickness ratio and shear plane angle ϕ decrease upto about 0.5% cerium addition beyond which the values go on increasing almost linearly with cerium additions.

vi) HP/Cm^3/min. increases upto about 0.5% cerium addition beyond which it goes on decreasing with further additions.

Since cerium dissolves only upto 0.05% in aluminium and has no solubility for copper, with the addition of cerium very little of it goes into solid solution, most of it reacting with aluminium and copper to form the intermediate phases Al_4Ce, Al_2Ce, CeAl, Ce_3Al_2, $CeCu_2$, $CeCu_4$ and $CeCu_6$; the predominating phases being Al_4Ce, $CeCu_4$ and $CeCu_6$. The refining effect produced as already explained may be due to the presence of these fine particles acting as unclei during solidification. The large increase in tensile strength and hardness upto 0.5% addition may be attributed to the refinement of alpha, and increasing amount of intermediate phases finely dispersed in the matrix. The decrease in tensile strength and hardness beyond 0.5% may be due to recoarsening of the precipitates.

Detrimental effect of cerium addition on machinability has been observed upto 0.5% beyond which machinability increases. The machinability of the alloy treated with 1.5% cerium is slightly better than the untreated original alloy. The detrimental effect of misch metal addition upto 0.5% on machinability may be due to increase in hardness and tensile strength, thereby offering larger resistance to cutting. The increase in machinability beyond 0.5% may be due to decrease in hardness and tensile strength due to coarsening of the precipitates.

8.4.2.3 Effects of lanthanum additions :

i) Metallographic study indicates that there is a large refining effect of lanthanum addition upto about 0.2%. On further additions no detectable effect is observed.

ii) The tensile strength is found to increase with lanthanum additions appreciably upto about 0.1% and slightly between 0.1 to 0.3%.

iii) The % elongation decreases to a constant value at about 0.2% lanthanum addition.

iv) The hardness remains almost unchanged with lanthanum addition.

v) Both chip thickness ratio and shear plane angle ∅ decrease slightly upto about 0.1%. Between 0.1 and 0.2% both the values increase, and with further additions the values become almost unchanged. The values for the alloy treated with 0.3% are almost equal to that of the original untreated alloy,

vi) HP/cm^3/min remains almost constant with lanthanum additions.

Lanthanum is found to have no solid solubility in aluminium and copper. Thus with the addition of lanthanum there will be possibility of formation of large amount of intermediate phases: La_3Al, La_3Al_2, $LaAl$, $LaAl_2$, $LaAl_4$, $LaCu$,

$LaCu_2$, $LaCu_4$; $LaCu_5$, $LaCu_6$ resulting in increase in tensile strength and corresponding decrease in % elongation The above phases could not be detected on metallographic study probably due to their being of submicroscopic size. From Fig. 8.16, it can be seen that almost no beneficial effect is obtained on machinability both from chip formation and power considerations. It appears that slight beneficial effect due to refinement is counterbalanced by the adverse effect produced due to presence of brittle intermediate phases, resulting in almost constant value of machinability.

8.4.2.4 Effects of neodymium additions :-

i) There appears to be a little refining effect of neodymium addition on the microstructure.

ii) The tensile strength increases with neodymium addition appreciably upto 0.1% and slightly between 0.1 and 0.3% addition.

iii) The % elongation decreases to a constant value at about 0.1% addition.

iv) The hardness increases to a more or less constant value at about 0.1% addition.

v) Both chip thickness ratio and shear plane angle Ø slightly decrease upto 0.1% addition, indicating the slight detrimental effect with further additions they increase to more or less constant values at 0.2% addition.

vi) HP/Cm^3/min. remains almost unchanged upto about

0.1% addition beyond which it decreases upto 0.2%. With further additions, value becomes almost constant.

With the addition of neodymium, it will go into solid solution upto about 0.2% in aluminium. Thus the amount of intermediate phases $NdAl_2$, NdAl etc. formed due to reaction of neodymium with aluminium should be very small. If refining effect observed is due to the presence of fine intermediate phases, as discussed in earlier cases, it is expected that in case of neodymium addition, very little refining effect should be observed. The results are in confirmity with these facts.

The increase in tensile strength may be attributed partly to the neodymium going into solid solution distorting the lattice, enhancing the strength and partly to fine dispersion of intermediate phases particularly at higher additions.

Fig. 8.18 indicates that almost no beneficial effect is obtained on machinability. Machinability remains almost unchanged with neodymium addition. It slightly decreases upto about 0.1% addition beyond which it increases almost to its original value. The slight decrease in machinability upto 0.1% may be attributed to the increase in hardness and tensile strength. Since tensile strength and hardness remain almost constant beyond 0.1% a slight refinement effect will bring about slight increase in machinability. The above fact is not so much apparent from chip formation consideration, the H.P. values are in complete confirmity with the above explanation.

8.4.2.5 Effect of misch metal additions:

i) There is a pronounced refining action of misch metal upto about 1% addition. No significant change is observed on further additions. Beyond 3% addition some new phase in addition to the alpha solid solution phase and the eutectic, makes its appearance and seems to increase with fruther additions.

ii) Beneficial effects due to the addition of misch metal have been observed on tensile strength, % elongation and hardness at room temperature upto about 3% addition. With further additions the values are found to decrease.

iii) Both chip thickness ratio and shear plane angle Ø increase upto 1% addition beyond which they decrease to constant values at about 1.5% addition.

iv) HP/Cm^3/min. is found to decrease upto about 1% addition beyond which it increases to a constant value at about 1.5% addition. Values are found in agreement with the chip thickness ratio and shear plane angle Ø values indicating the beneficial effect of misch metal on machinability upto about 1% from both the criteria.

The striking increase in tensile strength, hardness and % elongation at room temperature in the alloy on adding misch metal may be attributed as in the earlier cases to the refining of the structure, and also the appearance of submicroscopic particles of new phases. Most of the rare earth elements contained in misch metal do not go into solid solution and may be expected to react with aluminium and copper to form intermetallic compounds which will increase with increase in misch metal content. The refining effect produced with misch metal addition is not surprising. All the three elements cerium, lanthanum and neodymium contained in misch metal have produced refining effect on the alloy individually. The large refining effect of misch metal may be the cumulative effects of all the three elements. The decrease in mechanical properties beyond 3% may be due to relatively increased amounts of intermetallic compounds, their non.uniform distribution as well as recoarsening of the matrix. The nature of the insoluble phases could not be confirmed by metallographic study only.

The increase in tensile strength and hardness with misch metal addition can easily be explaihed on the basis of the effects of individual elements cerium, lanthanum contained in misch metal. It is interesting to note that beneficial effects have been produced due to all the three elements when studied seperately. Thus the large beneficial effects of misch metal may be the cumulative effects of all the three elements.

The increase in % elongation with misch metal addition can not be explained on the basis of the effects of individual elements, since all the three elements have individually produced detrimental effects on % elongation and the effect of misch metal is to increase the % elongation. This is probably due to the fact that very large refining of alpha grains, is obtained with misch metal addition,predominating over the adverse effect due to brittle intermediate phases.

It is interesting to note that the alloy treated with misch metal upto 1% is superior from strength as well as machinability consideration. Alloy treated with further addition upto 3% is better only from strength consideration.

The improvement in machinability upto 1% misch metal addition may be attributed to the large refining effect produced, predominating over the slight adverse effect due to intermediate phases. For further additions upto 1.5% the adverse effect due to the large increase in the amount of intermediate phases and increase in strength predominates over the beneficial effect produced due to refining, resulting in slight decrease of machinability.

CHAPTER IX

EFFECTS OF TANTALUM AND RARE EARTH ELEMENTS AND MIXTURE OF RARE EARTH ELEMENTS ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM-SILICON ALLOYS AT ORDINARY TEMPERATURE.

CHAPTER IX

EFFECTS OF TANTALUM , RARE EARTH ELEMENTS AND MIXTURE OF RARE EARTH ELEMENTS ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM-SILICON ALLOYS AT ORDINARY TEMPERATURE.

9.1 GENERAL INTRODUCTION.

The commercial importance of aluminium-silicon alloys is borne by the fact that they now account for about **half** of the world's light alloys production. The alloys have very wide applicability in the manufacture of objects like pistons for internal combustion engines and cylinder blocks requiring good resistance to corrosion and wear. They have very good casting properties and a wide strength/weight ratio.

Aluminium dissolves very little or silicon in the solid state. The percentage of silicon found in the solid solution in aluminium at eutectic temperature is 1.65% by weight. The solubility decreases with decrease of temperature and comes down to 0.65% at 200° C. Ordinarily silicon is coarsely dispersed in the form of flakes which affect the mechanical properties of the alloys adversely. The distribution and size of silicon particles can favourably be modified by certain additions like sodium, alkali metals and alkali fluorides. Modification process refers to a process in which eutectic temperature, structure and composition of alloys are apparently altered by addition of small amounts of another constituent.

The present investigation reported has been undertaken with a view to explore the possibility of rare-earth metals principally as modifier and secondarily as a beneficial alloying addition to aluminium silicon alloys. The micro-structure as well as mechanical properties of hyper- and hypo alloys, alloyed with different amounts of misch metal, lancer amp 2 and rare earth fluorides in the molten condition, have been studied. The cooling rate during solidification affects the structure and mechanical properties drastically. Faster rate of cooling makes the structure finer and improves the mechanical properties. In order to investigate the effects of rate of cooling during solidification alloys have been cast in permanent as well as in sand moulds. The mixture of metals in the form of misch metal and lancer-amp 2 was preferred to one or more of the pure metals because of its low cost and easy availability in India. The cheaper mixture of rare earth fluorides has also been used with the expectation that fluorides will have an additional advantage of hydrogen degassification in aluminium alloys. Influence of minor additions of pure metals has also been investigated to understand more fundamentally the causes of behaviour of the mixtures based on the effects of individual elements, viz., cerium, lanthanum and neodymium contained in the mixture of rare earths.

As no systematic work seems to have been carried out on these lines, it is hoped that these investigations will prove useful in the development of new lighter alloys with better machinability, mechanical and structural properties.

9.2 PREVIOUS WORK AND THEORIES ACCOUNTING FOR MODIFICATION:

Uptil 1900, the addition of silicon to aluminium was considered to be detrimental. In 1891, however, Minet (156) recognized that an abnormal mode of crystallization could occur in the alloys of aluminiumand silicon. In 1920 Pacz(157) discovered that the mechanical properties of aluminium silicon alloys could be materially improved by the addition of a small quantity of an alkaline fluoride. This process was given the name "Modification". In 1922 and 1923, Edwards, Frary and Churchil (158) introduced the use of alkali metals while Guyer and Philips (159) introduced the use of alkaline earth metals and their oxides and hydroxides for modification. The function of the modifying agent is not known and several theories to account for modification have been put forward.

9.2.1 Guillet's True flux theory :

The theory was proposed by Guillet (160) in 1722. According to him the modifying reagent reacts chemically in such a way as to remove the impurities, alumina and silica by reduction. The improvement in the mechanical properties was believed to be due to increased purity. The theory failed to explain any of the facts associated with modification.

9.2.2 Allotropic modification theory : (161)

Jefferes proposed, at about the same time as true flux theory, that two systems of aluminium silicon alloys exist, one of which is metastable. This theory was quickly abondoned when it was shown by X-ray diffraction methods that the same allotropic modification existed in normal and modified alloys.

9.2.3 Ternary phase theory: (162,163,164) :

Curran (162) and Otani (163) suggested independently the ternary Al-Si-Na system as an explanation of the part, sodium played in the production of modified structure.

Ransley and Newfield (164) attributed the formula Al Na Si 1.25 or Al Na Si 1.33 to the ternary phase. They found a ternary eutectic at about 0.017 wt. percentage sodium.

The ternary alloy theory attempted not only to account for the improvement in the properties but also to explain the depression of the eutectic solidification temperature. This theory fails , however, to account for the change in the eutectic composition during modification and for the fact that the modified eutectic melts at the same temperature as the normal eutectic. The theory does not also account for the presence of both primary phases in modified hyper eutectic alloys.

9.2.4 Colloidal Theory :

The colloidal theory was proposed by Guyer and Phillips (165). Above the melting point, the state is of complete atomic dispersion. But on solidification the state is crystalline. This theory states that in transition from the dispersion state to crystalline state substance passes through a colloidal state. The particle size is 10^{-5} to 10^{-7} cm . in diameter. The crystal normally grows because the colloids usually coalesce. But in presence of certain "colloid protectors" this coalescence is retarded and so the crystal growth is inhibited. These colloid protectors are known as modifying agents.

The theory failed to explain the following(166):

i) According to this theory coalescence of smaller particles results in eutectic, which is not true.

ii) In order to account for the effect of sodium, the theory depends on the limited solubility of sodium in aluminium and silicon and offers no explanation for the fact that Li and to a small extent Ca and Mg, which are soluble in aluminium-silicon alloys in all proportion, can effect modification.

iii) Further there are metals such as lead and bismuth, which like sodium, are partially soluble in aluminium silicon alloys and yet do not cause modification.

iv) C hange brought about by rapid chilling of sodium free alloys.

v) Presence of both primary phases in modified hper-eutectic alloys.

9.2.5 Thall and Chalmer's theory (167)

Examinations of the early literature revealed numerous inconsistencies in the result of thermal analyses, the temperature of solidification of modified eutectic for example being reported over the range of 576° C - 560° C.

The beginning of primary solidification in modified alloys is variously to be above and below the normal temperature. None of the theories was able to account for modification. Both modification and thermal experiments were carried out. Because of the high volatility of sodium and its ease of oxidation, one of the uncertainties in all the earlier work on modifications has been the actual sodium content of the alloys. Thall and Chalmer adopted radioactive method to estimate Na content. From the results of the radioactive experiments it may be concluded that sodium is present in all the alloys to which it was added, the percentage depending on the amount added to the melt and on the rate of solidification. Remoulding the alloys causes a loss of some but not all the sodium and superheating increases the loss.

Thermal experiments using the more modern equipment and purer metals were repeated. The object of these experiments was to determine the temperature of eutectic solidification and of begining of primary solidification for alloys of various composition with and without sodium addition and with different cooling rates.

Following conclusions were derived from the experiments:-

i) The presence of sodium depresses the temperature of eutectic solidification.

ii) The presence of sodium does not affect the temperature of primary solidification.

Previous investigators had concluded that the sodium burned off in remelting the modified alloys, since heating curves of the modified alloys indicated that the modified eutectic melts at the same temperature as the normal eutectic. The results of the radioactive experiments showed that sodium was still present after remelting.

Thall & Chalmers suggested that the presence of a concentration of sodium in the aluminium surrounding the silicon crystals decreased the surface energy and the interfacial tension between the solid aluminium and the solid solution. This will result in the suppression of the silicon crystals, and will evidently have no effect on the melting point of the eutectic.

9.2.6 Mitchel's theory (168) :

Mitchel and others have proposed that sodium removes or poisons nuclei such as hydrogen (or hydrides) and nonmetallic, inclusions and alters the nucleation pattern for the eutectic crystallization of silicon. The suggestion that sodium poisons or removes nonmetallic nuclei such as hydrogen or hydrides do not seem reasonable because their treatments such as chlorine fluxing, which removes hydrogen, do not cause modification.

9.2.7 Plumb and Lewis theory (169) :

Plumb and Lewis reported that they had not observed selective adsorption of sodium by the molten phase through the use of autoradiography. Instead they found, the sodium uniformly distributed with in 10μ resolution limit of their technique. Also using chemical analysis methods they found no evidence of selective segregation of sodium in the silicon phase with increasing sodium concentration. From the results of an electron microscope study they concluded that fine modified eutectic produced by chil casting and that produced by sodium treatment were sufficiently alike microstructurally to cast doubt on any contention that different mechanisms were involved in these two ways of obtaining modification.

A theory similar to a hypothesis advanced by Edward (158) was proposed. According to it the addition of modifying agent retards nucleation of eutectic silicon at the 'unmodified' eutectic temperature and causes solidification to occur at a steady state temperature considerably below the freezing point where diffusion is slower and individual silicon crystals cannot attain so large a size. Two ways in which sodium could present the formation of stable nuclei at the normal eutectic temperature are (i) by adsorption on the nuclei so making it difficult for them to grow to a stable size or (ii) by adsorption on the solid sur-faces (most probably on aluminium) where the silicon crystals form.

9.2.8 Kum and Heine's hypothesis (170) :

The literature survey revealed a lack of data supporting the theories of modification. Specifically, items not covered in the literature are as follows :-

i) The mechanism or path or solidification of the normal and modified alloys, including the temperature of initial nucleation of the silicon phase, growth-temperature range, and completion of solidification.

ii) The influence of nucleation and growth temperature on the morphology of the silicon phase.

iii) The effect of variable cooling rates.

iv) The influence of silicon content or alloy composition on items (i) and (ii) above.

v) The difference existing between the true modified eutectic and the pseudo modified eutectic produced by rapid cooling, if any.

vi) Supporting evidence concerning contact angles, interfacial energies, and concentration gradients.

vii) Common properties of elements causing eutectic modification

viii) A critical nucleation and growth temperature experiment.

ix) A hypothesis applicable to other systems involving modification.

Some of these gaps were filled in by this theory, particularly in relation to items (i) to (v) and (vii) to (ix).

The phenomenological aspects of the processes of solidification in modified and normal aluminium silicon alloys were described for the composition range 7.0 - 17.0 % silicon. The techniques of interrupted slow cooling followed by quenching and interrupted isothermal quenching were used for this purpose. Elements that produce modification were studied. and the requirements of a modifying element were formulated from the results. The shape of the silicon particles, which grow during solidification has proved to be temperature dependent. The specific temperature ranges of nucleation and growth in the 7 - 17% silicon range studied were as follows :-

i) Polyhedral silicon: nucleation and growth at temperature down to 583 °C.

ii) Coarse silicon plates: nucleation and growth at and slightly below 583° C.

iii) Fine silicon plates growth below 583° C with increasing cooling rate.

iv) Globular silicon: nucleation and growth at 580°C and below. This produces the modified eutectic in sodium treated alloys and pseudo-modified eutectic in chill cast alloys.

v) Idiomorphic silicon (primary silicon in hyper eutectic alloys) from the liquidus temperature down to temperatures of plate growth.

Polyhedral silicon and coarse silicon plates in normal alloys is produced by the habit characteristics of the higher temperature of formation. In modified alloys the growth is favoured on crystallographic planes of the lattice, which produces a globular shape. According to them, the modifying element depresses the nucleation temperature, lowers the growth temperature thereby changing the growth shape from coarse plates to globular silicon.

Following conditions for modification were given:-

i) The binary system should be of proper type. There should be a substantial difference in melting point between the two components. Further the precipitating or high melting point phase should have low solubility for the solvent phase. In addition the eutectic should be rich in the solvent or low melting phase component.

ii) The modifying element should have a tendency to form compounds with the precipitating phase at a temperature below the normal eutectic temperature. Further the modifying element should have a low compound forming tendency, low solubility and possibly a miscibility gap with the solvent phase. The modifying element probably should have a low melting point (and boiling point) relative to low melting binary phase component.

9.3 EXPERIMENTAL PROCEDURE.

9.3.1 Preparation of aluminium-silicon alloys and addition of tantalum and rare earths to the same :-

The required amounts of aluminium and silicon including an additional amount of silicon to account for the loss in preparation were charged into a graphite crucible heated in an oil fired furnace. Precautions were taken to avoid gas absorption. Four heats were prepared. Alloys were cast into blocks. Each heat was chemically analysed for silicon content. The chemical analysis report is given in Table 9.01.

TABLE 9.01

No. of heat		Desired composition	Actual composition
1	Hyper	12% Si	11.9 % Si
2		12% Si	13.0 % Si
3	Hypo	8% Si	7.5 % Si
4		8% Si	9.0% Si

Pieces weighing about 400 gms, cut from the block were taken in a temperature controlled electric furnace. When the temperature of the melt reached 800 °C, the crucible was taken out and weighed amount of rare earths wrapped in aluminium foil was plunged into the molten metal and agitated mildly with a graphite rod. In order to add tantalum melt was super heated to 1000 °C.

After the additions, the alloys were allowed to remain in the furnace for about 20 minutes for complete homogenization. The crucible was taken out and when the temperature of the melt reached 700 °C, alloys were cast in sand or cast iron moulds as desired. Pouring temperature and other conditions during addition and casting were kept exactly similar in all the cases. Heats No. 1 was used for additions of tantalum, cerium lanthanum and neodymium and heat No.2 was used for misch metal addition. Heats No. 3 and 4 were used for additions of misch metal, lancer-amp^2 and rare earth fluorides respectively.

9.4 EXPERIMENTAL RESULTS :

9.4.1 Effects of tantalum additions to hyper-eutectic Al Si alloy (11.9% Si)

The results are given in Figs. 9.01 and 9.02, Table 9.02 and plates 9.01 to 9.06.

9.4.2 Effects of cerium additions to hyper-eutectic Al-Si alloy (11.9% Si)

The results are given in Fig. 9.03 and 9.04, Table 9.03 and plates 9.07 to 9.11.

9.4.3 Effects of lanthanum additions to hyper-eutectic Al-Si alloy (11.9% Si)

The results are given in Figs. 9.05 & 9.06, table 9.04 and plates 9.12 to 9.16.

9.4.4 Effects of neodymium additions to hyper-eutectic Al-Si alloy (11.9% Si)

The results are given in Figs. 9.07 & 9.08, table 9.05 and plates 9.17 to 9.19.

9.4.5 Effects of misch metal addition to hyper-eutectic Al-Si alloy (13% Si) :-

The results are given in Figs. 9.09 to 9.11, tables 9.06 & 9.07 and plates 9.20 to 9.32.

9.4.6 Effects of lancer-amp^2 additions to hyper-eutectic aluminium silicon alloys (13% Si) :-

The results are given in Fig. 9.12, table 9.08 and plates 9.33 to 9.38.

9.4.7 Effects of rare-earth fluorides addition to hyper-eutectic aluminium silicon alloys (13% Si):

The results are given in Figs. 9.13 & 9.14, table 9.09 and plates 9.39 to 9.48.

9.4.8 Effects of rare-earth fluorides additions to hypo-eutectic aluminium-silicon alloys (7.5 % Si) :

The results are given in Figs. 9.15 , table 9.10, and plates 9.49 to 9.52.

9.4.9 Effects of lancer-amp^2 and misch metal additions to hypo-eutectic aluminium silicon alloys (9% Si) :-

The results are given in Figs. 9.16 & 9.17, table 9.11 and plates 9.53 to 9.62.

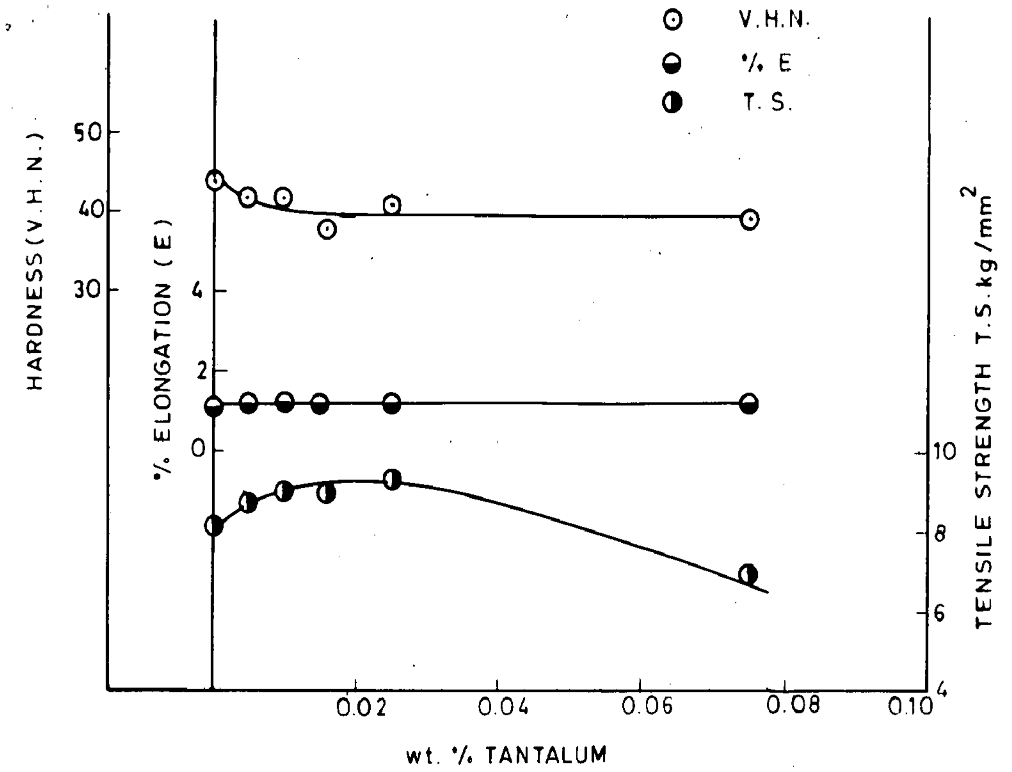

FIG.9.01_EFFECT OF TANTALUM ON THE MECHANICAL PROPERTIES OF Al+11.9% Si (METAL MOULD CAST)

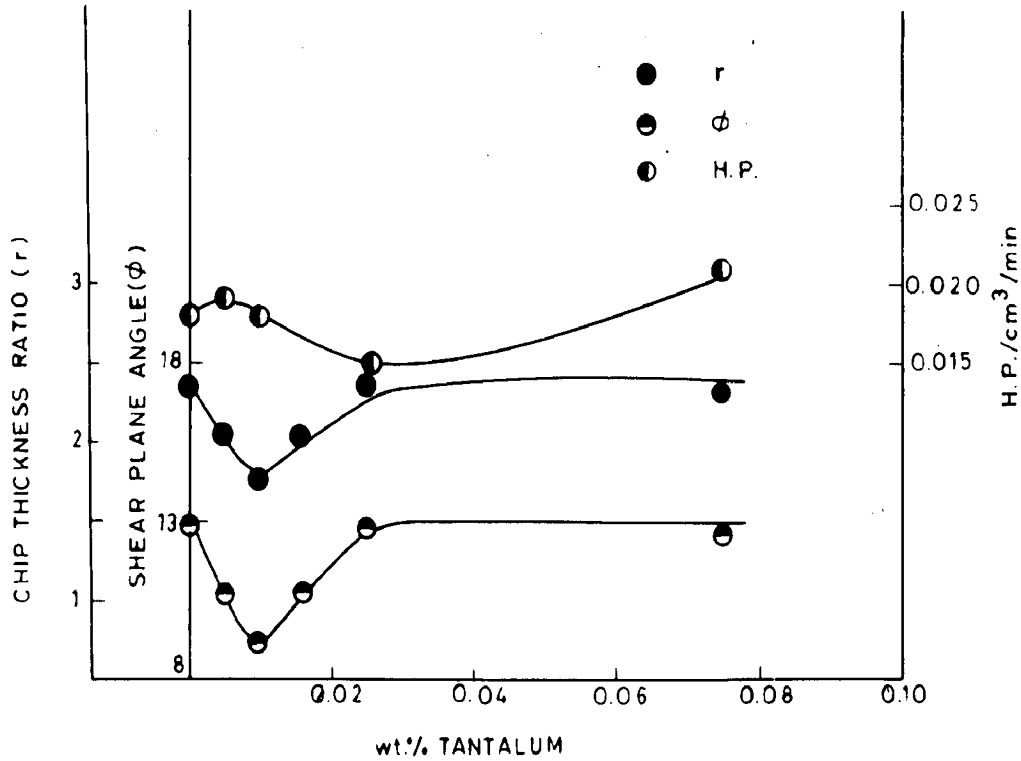

FIG.9.02_EFFECT OF TANTALUM ON THE MACHINABILITY OF Al+11.9% Si (METAL MOULD CAST)

TABLE 9.02

EFFECTS OF TANTALUM ADDITIONS ON THE MECHANICAL PROPERTIES & MACHINABILITY OF HYPEREUTECTIC ALUMINIUM-SILCON ALLOY (11.9% Si), METAL MOULD CAST.

S. No.	% alloying element.	Tensile strength kgm/cm^2	% elongation.	Hardness VHN	Chip thickness ratio	Shear plane angle ϕ	HP/cm^3 min.
1	Pure	6.1	2.50	44	0.238	$12^{\circ}56'$	0.0181
2	0.005	8.7	1.20	42	0.204	$10^{\circ}40'$	0.0189
3	0.010	8.9	1.20	42	0.174	$9^{\circ}11'$	0.0177
4	0.017	8.9	1.20	38	0.202	$10^{\circ}47'$	-
5	0.025	9.3	1.20	41	0.235	$12^{\circ}45'$	0.0150
6	0.075	7.0	1.20	40	0.229	$12^{\circ}24$	0.021

TABLE 9.03

EFFECT OF CERIUM ADDITION ON THE MECHANICAL PROPERTIES & MACHINABILITY OF HYPEREUTECTIC ALUMINIUM-SILCON ALLOY (11.9% Si) METAL MOULD CAST.

S. No.	% alloying element.	Tensile strength kgm/cm^2	% elongation.	Hardness VHN	Chip thickness ratio	Shear plane angle ϕ	HP/cm^3 min.
1	0.00	6.1	2.50	44	0.238	$12^{\circ}56'$	0.0181
2	0.10	-	-	-	0.490	$29^{\circ}17'$	0.0155
3	0.30	7.2	2.25	65	0.348	$19^{\circ}49'$	0.0230
4	0.50	6.9	1.25	68	0.405	$23^{\circ}34'$	0.0243
5	0.75	7.0	2.25	66	0.355	$20^{\circ}17'$	0.0250
6	1.00	7.3	1.90	74	0.293	$16^{\circ}19'$	0.0235

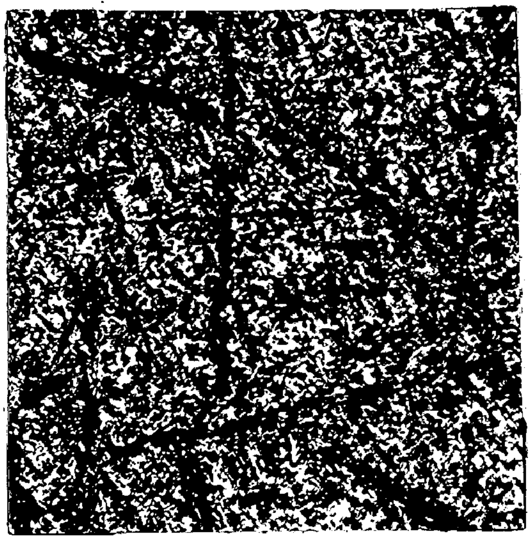

PLATE 9.01 106 X

Al + 11.9 % Si

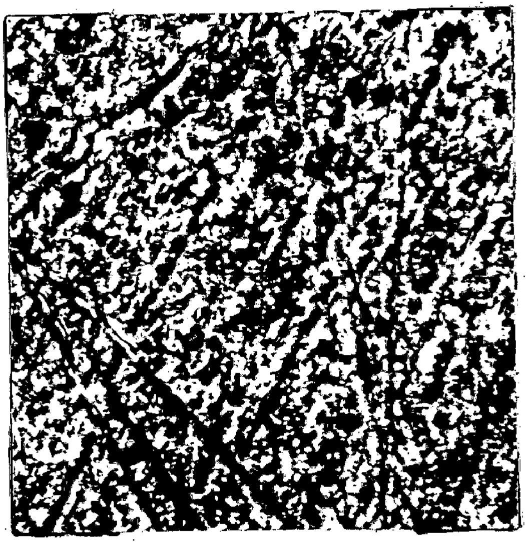

PLATE 9.02 106 X

Al+11.9 % Si+ 0.005 wt. % Ta

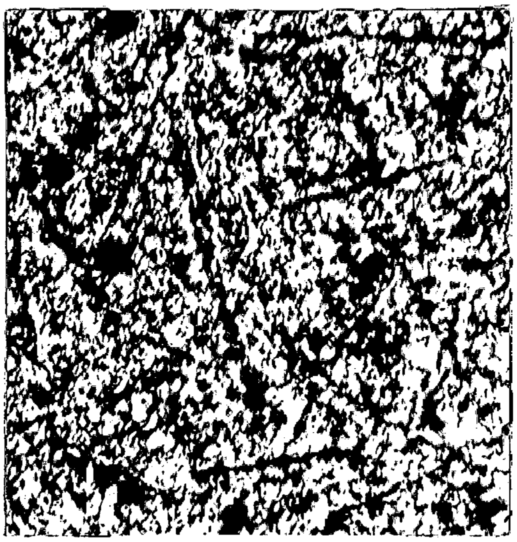

PLATE 9.03 106 X

Al + 11.9 % Si + 0.010 wt. % Ta

PLATE 9.04 **106 X**

Al + 11.9 % Si + 0.016 wt. % Ta

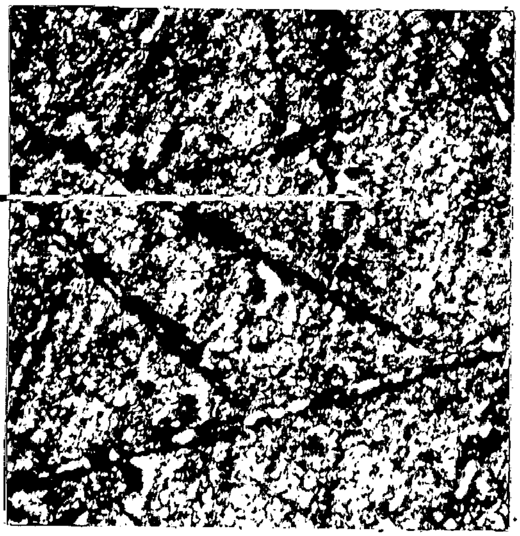

PLATE 9.05 **106 X**

Al + 11.9 % Si + 0.025 wt. % Ta

PLATE 9.06 **106 X**

Al + 11.9 % Si + 0.075 wt. % Ta

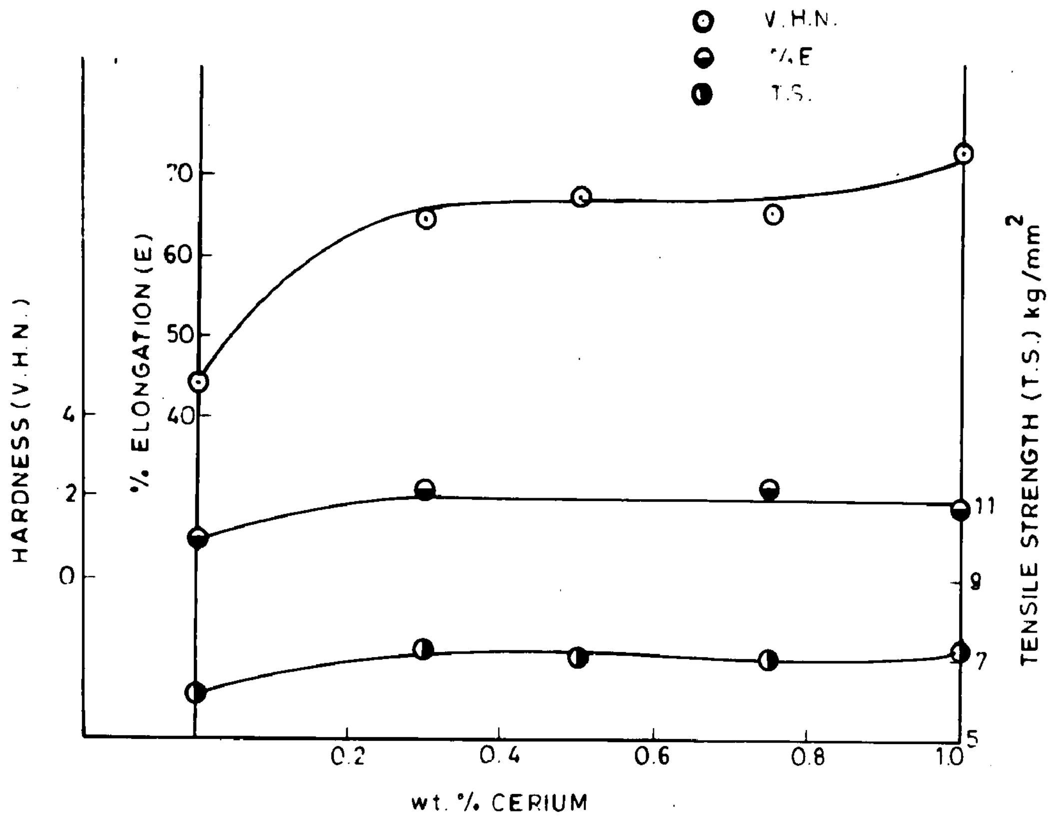

FIG. 9.03_EFFECT OF CERIUM ON THE MECHANICAL PROPERTIES OF Al+11.9 % Si (METAL MOULD CAST)

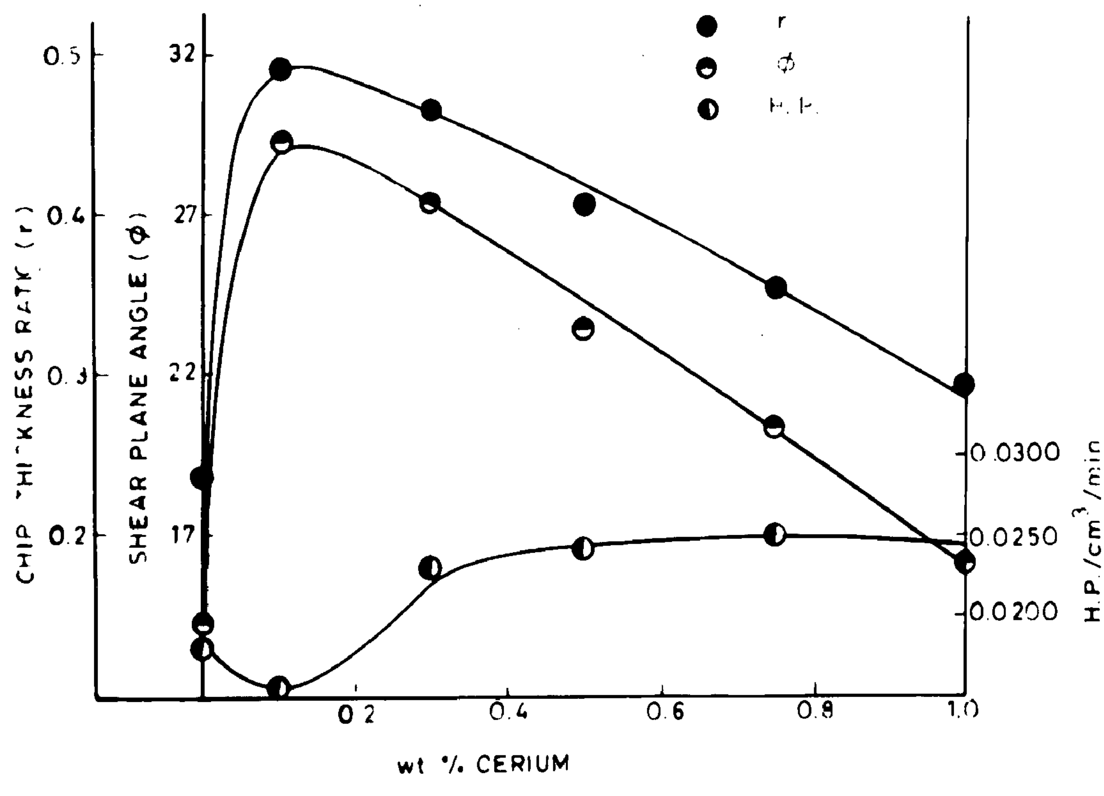

FIG. 9.04_EFFECT OF CERIUM ON THE MACHINABILITY OF Al+11.9 % Si (METAL MOULD CAST)

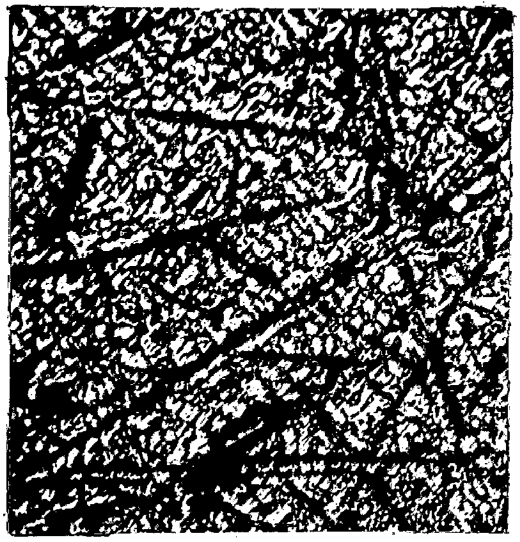

PLATE 9.07 106 X

Al + 11.9 % Si + 0.1 wt. % Ce

PLATE 9.08 106 X

Al + 11.9 % Si + 0.3 wt.% Ce

PLATE 9.09 106X

Al + 11.9 % Si + 0.5 wt. % Ce

PLATE 9.10 106 X

Al+11.9 % Si + 0.75 wt. % Ce

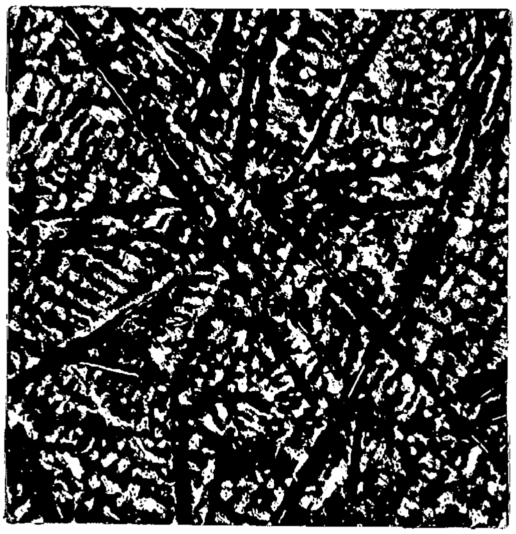

PLATE 9.11 106 X

Al +11.9 % Si + 1.0 wt. % Ce

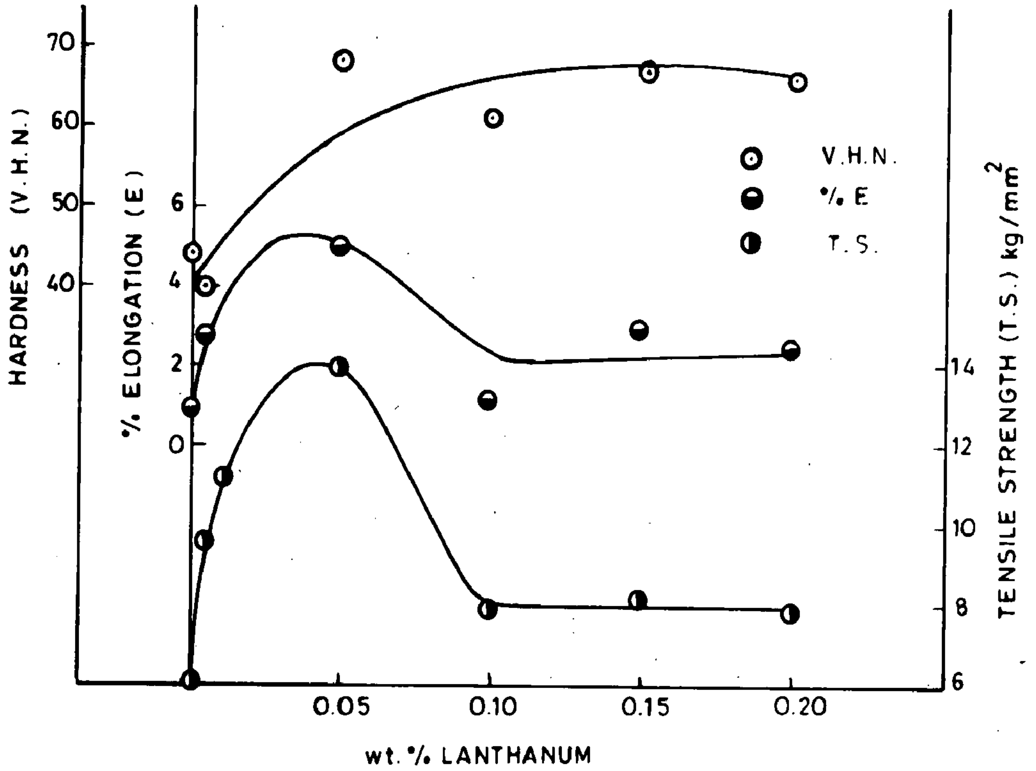

FIG.9.05_EFFECT OF LANTHANUM ON THE MECHANICAL PROPERTIES OF Al+11.9 % Si (METAL MOULD CAST)

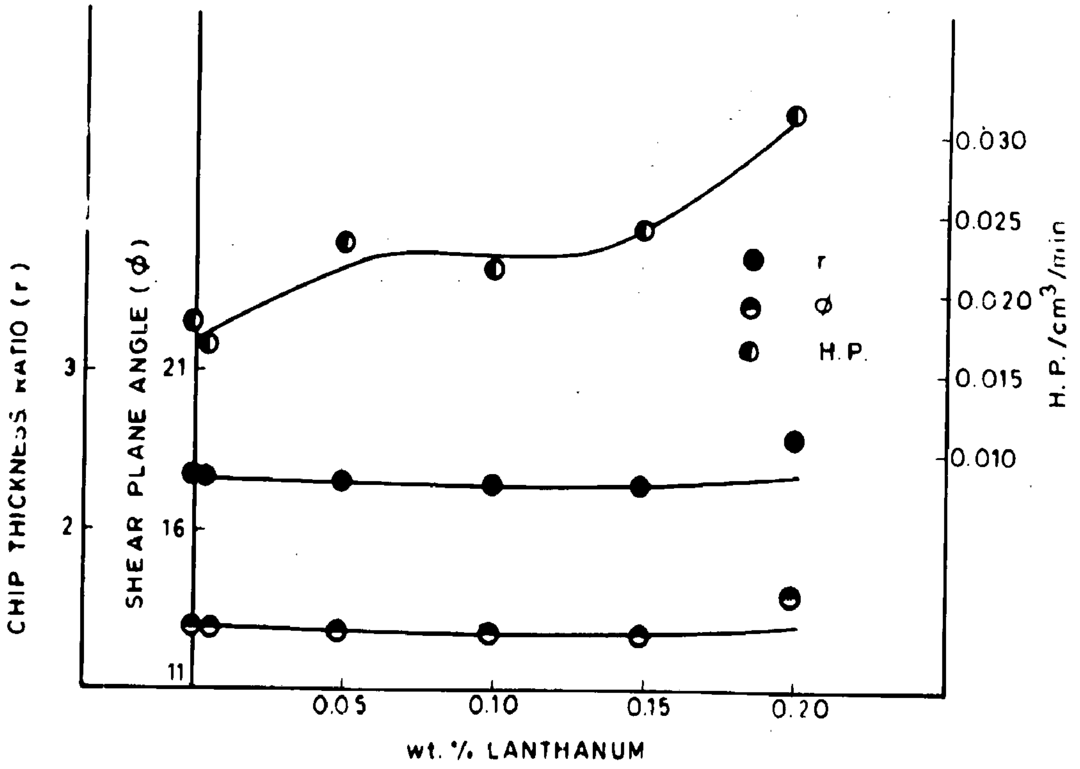

FIG.9.06_EFFECT OF LANTHANUM ON THE MACHINABILITY OF Al + 11.9 % Si (METAL MOULD CAST)

TABLE 9.04

EFFECTS OF LANTHANUM ADDITIONS ON THE MECHANICAL PROPERTIES & MACHINABILITY OF HYPEREUTECTIC ALUMINIUM-SILICON ALLOY (11.9% Si), METAL MOULD CAST.

S. No.	% alloying element	Tensile strength kgm/cm^2	% elongation.	Hardness VHN	Chip thickness ratio.	Shear plane angle ∅	HP/cm^3 min.
1	0.00	6.1	2.50	44	0.238	$12^{\circ}56'$	0.0181
2	0.005	9.6	2.80	40	0.238	$12^{\circ}56'$	0.0165
3	0.050	14.0	5.07	68	0.233	$12^{\circ}38'$	0.0231
4	0.100	7.9	1.20	60	0.233	$12^{\circ}38'$	0.0214
5	0.150	8.2	3.00	67	0.232	$12^{\circ}34'$	0.0241
6	0.200	7.9	2.50	65	0.260	$14^{\circ}16'$	0.0241
6							

TABLE 9.04

EFFECTS OF NEODYMIUM ADDITIONS ON THE MECHANICAL PROPERTIES & MACHINABILITY OF HYPEREUTECTIC ALUMINIUM-SILICON ALLOY (11.9% Si), METAL MOULD CAST.

S. No.	% alloying element	Tensile strength kgm/cm^2	% elongation.	Hardness VHN	Chip thickness ratio	Shear plane angle ∅	HP/cm^3/ min.
1	0	6.1	2.50	44	.238	$12^{\circ}56'$	0.0181
2	0.1	4.8	2.50	54	.232	$12^{\circ}34'$	0.0234
3	0.2	6.2	2.50	67	.218	$11^{\circ}44'$	0.0214
4	0.3	7.3	2.30	66	.262	$14^{\circ}24'$	0.0217

PLATE 9.12 106 X

Al + 11.9 % Si + 0.005 wt. % La

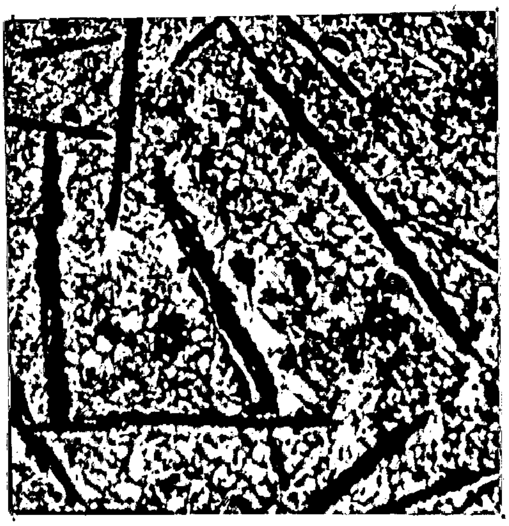

PLATE 9.13 106 X

Al + 11.9 % Si + 0.05 wt. % La

PLATE 9.14 106 X

Al + 11.9 % Si + 0.1 wt. % La

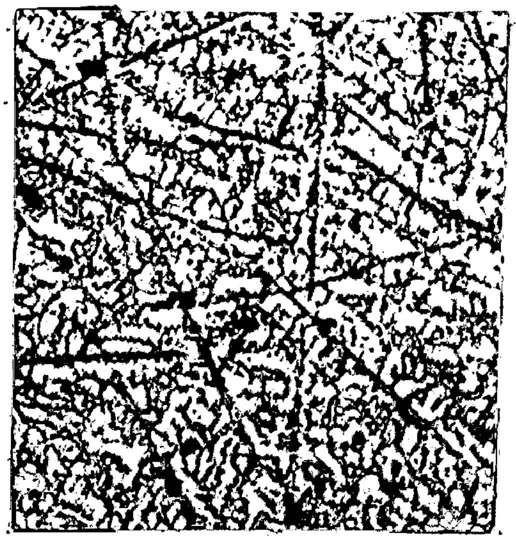

PLATE 9.15 106 X

Al + 11.9% Si + 0.15 wt.% La

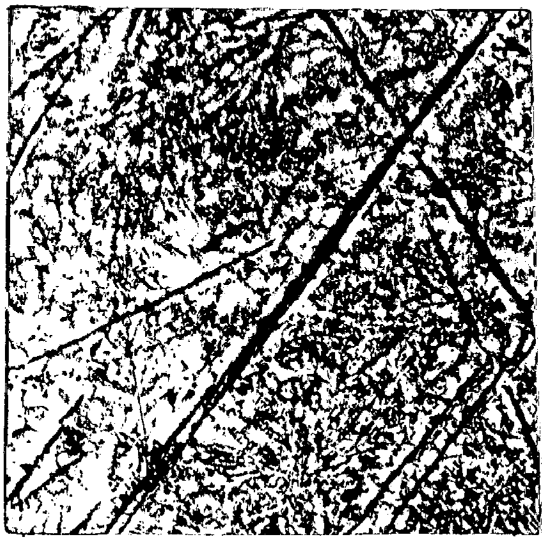

PLATE 9.16 106 X

Al + 11.9% Si + 0.2 wt. % La

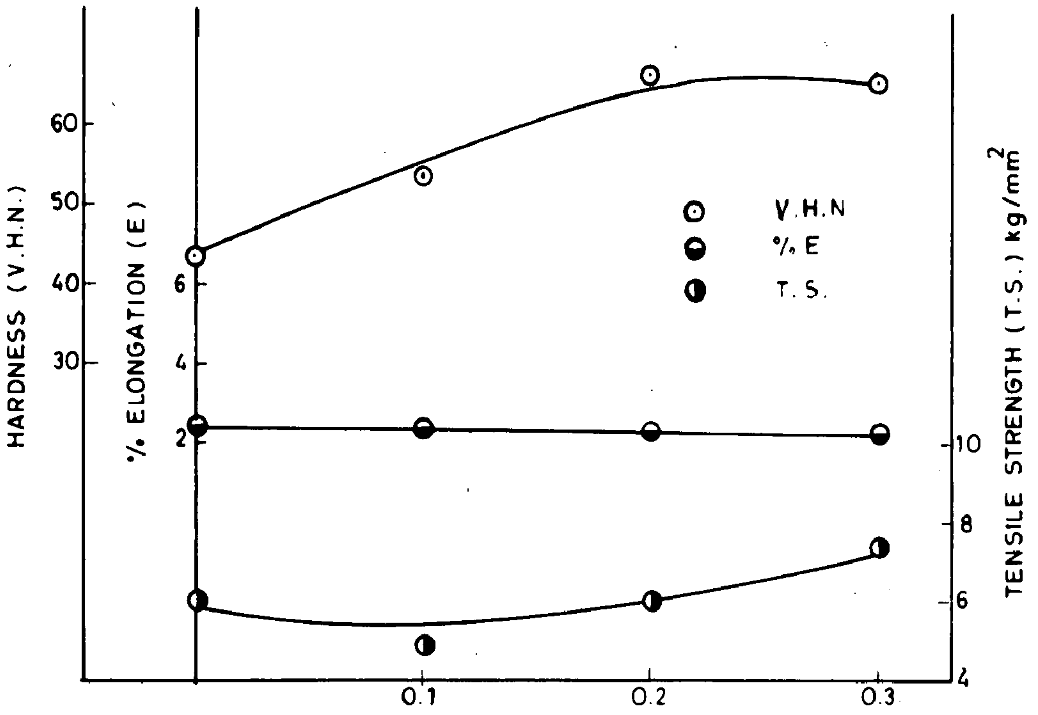

FIG. 9.07_EFFECT OF NEODYMIUM ON THE MECHANICAL PROPERTIES OF Al + 11.9 % Si (METAL MOULD CAST)

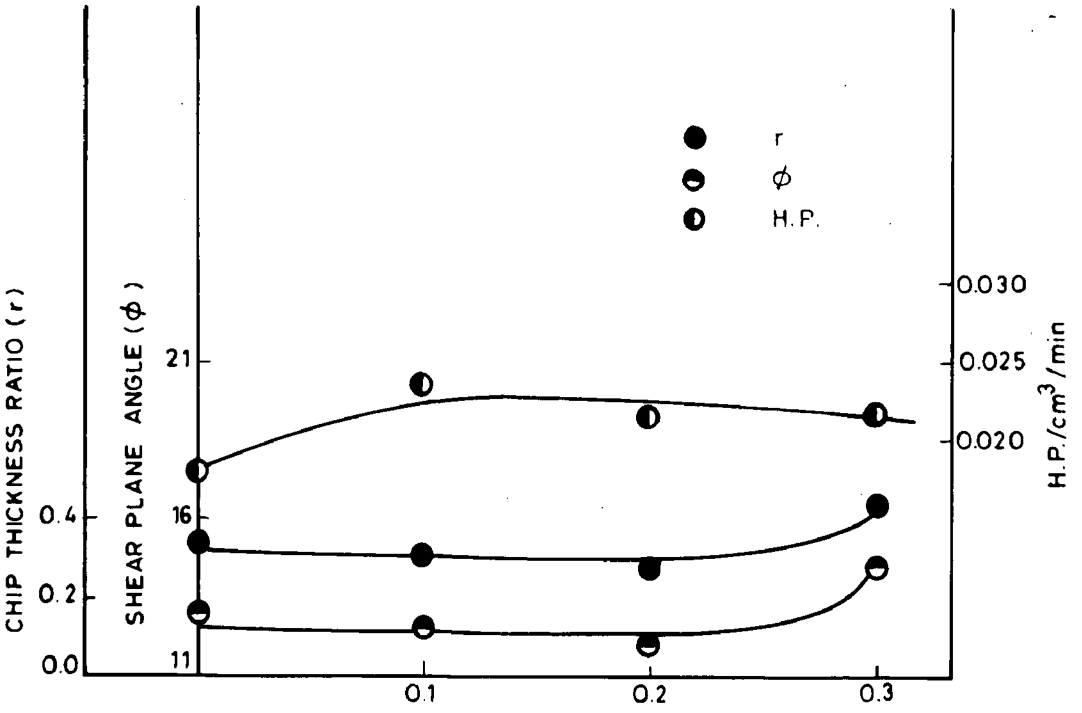

FIG. 9.08_ EFFECT OF NEODYMIUM ON THE MACHINABILITY OF Al + 11.9 % Si (METAL MOULD CAST)

PLATE 9.17 106 X

Al+11.9 % Si + 0.1 wt.% Nd

PLATE 9.18 106 X

Al + 11.9 % Si + 0.2 wt.% Nd

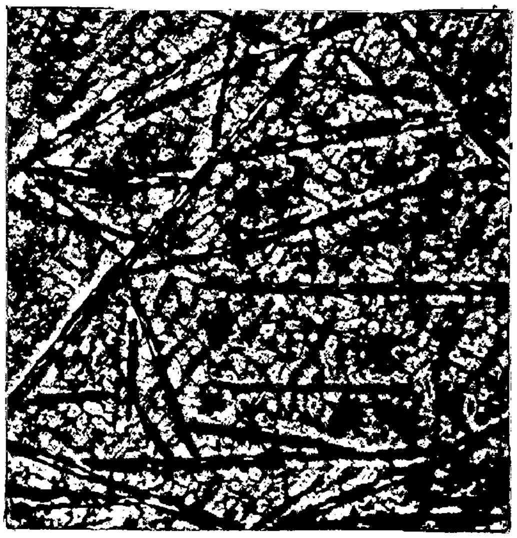

PLATE 9.19 106 X

Al+11.9 % Si+ 0.3 wt. % Nd

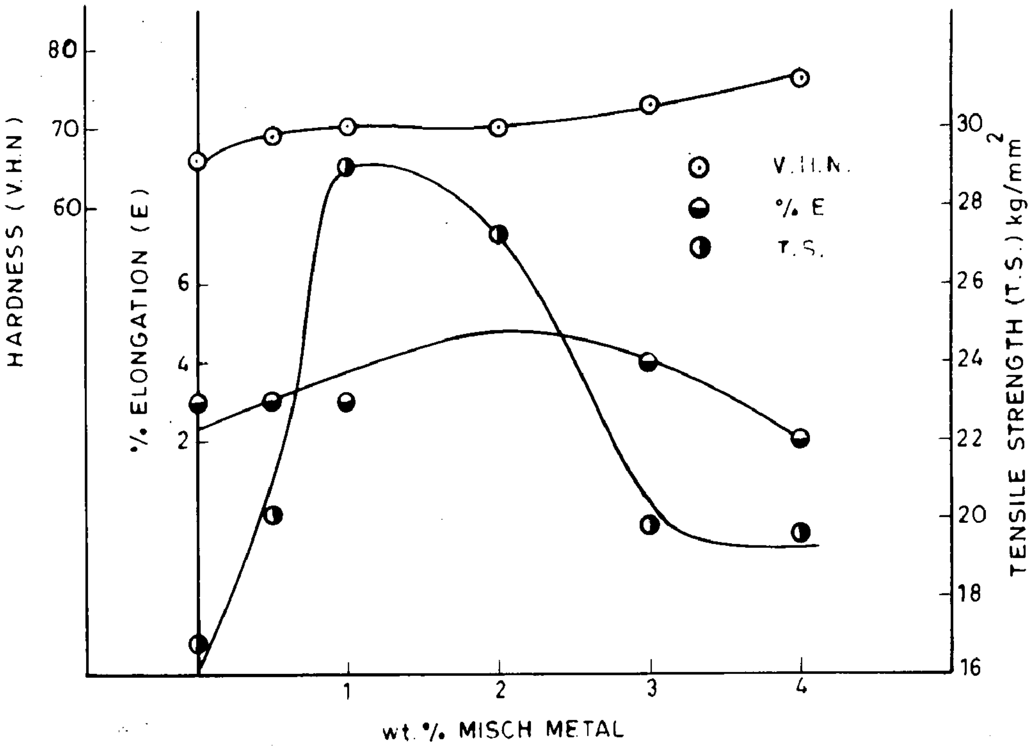

FIG.9.09_EFEECT OF MISCH METAL ON THE MECHANICAL PROPERTIES OF Al + 13 % Si

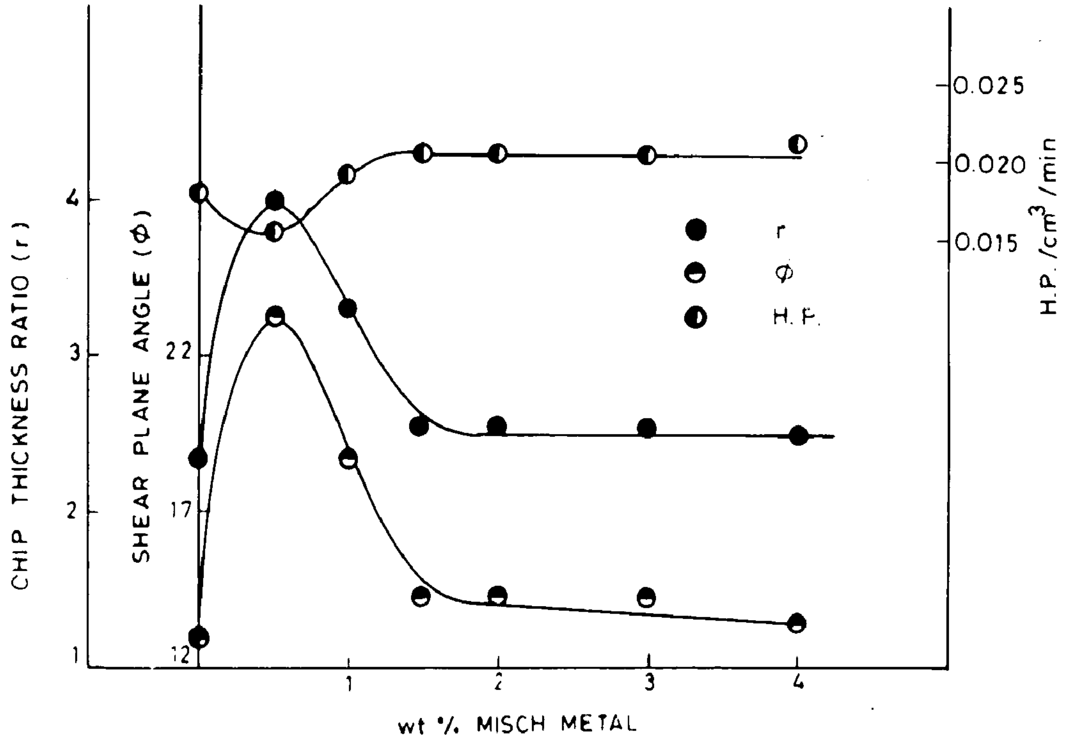

FIG. 9.10_EFFECT OF MISCH METAL ON THE MACHINABILITY OF Al + 13 % Si

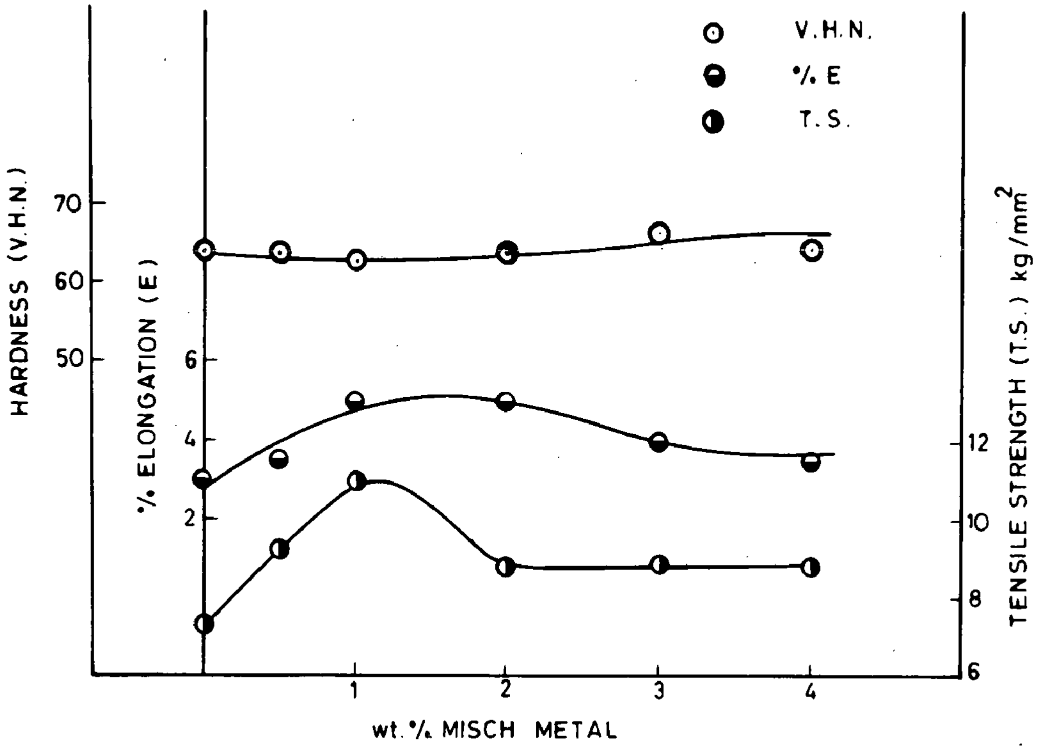

FIG.9.11_EFFECT OF MISCH METAL ON THE MECHANICAL PROPERTIES OF Al+13% Si (SAND CAST)

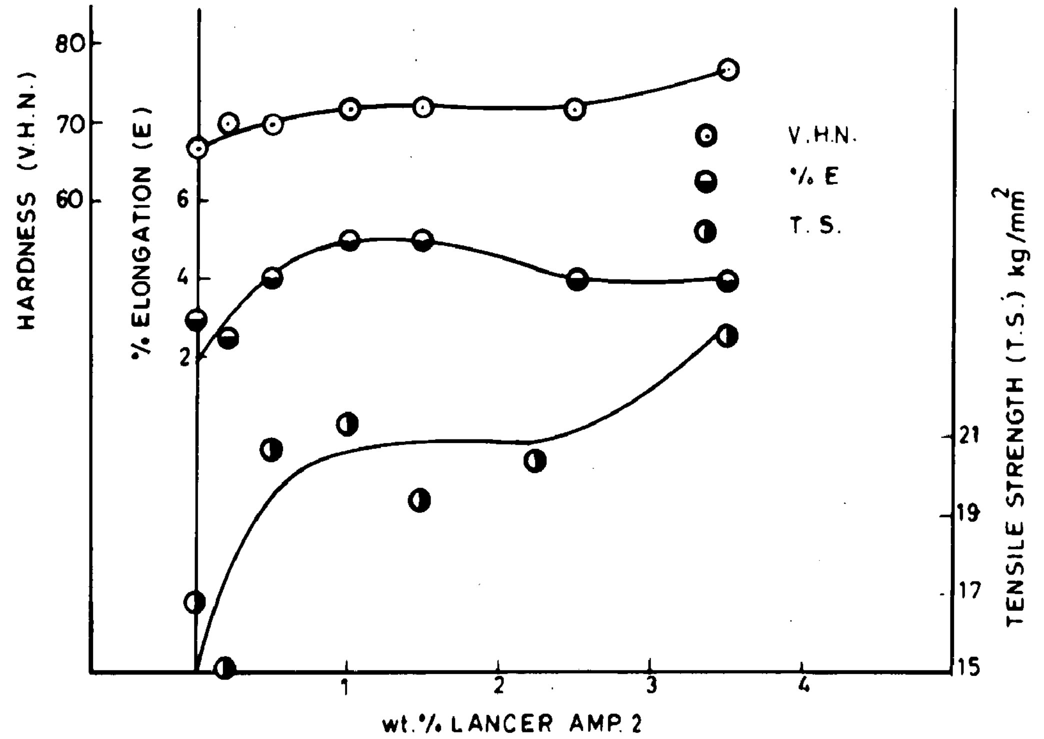

FIG.9.12_EFFECT OF LANCER AMP. 2 ON THE MECHANICAL PROPERTIES OF Al+13% Si (METAL MOULD CAST)

TABLE 9.06

EFFECTS OF MISCH METAL ADDITIONS ON THE MECHANICAL PROPERTIES AND MACHINABILITY OF HYPEREUTECTIC ALUMINIUM-SILICON ALLOY (13% Si) METAL MOULD COST.

S. No.	% alloying element.	Tensile strength kgm/cm^2	% elongation.	Hardness VHN	Chip thickness ratio	Shear plane angle Ø	HP/cm^3/min.
1	0.0	16.8	3.0	66	.238	12°56'	0.0181
2	0.2	20.1	3.0	69	.398	23° 6'	0.0155
3	1.0	29.1	3.0	70	.332	18°47'	0.0190
4	1.5	-	-	-	.258	14°9'	0.0204
5	2.0	27.3	4.0	70	.262	14°23'	0.0204
6	3.0	19.9	4.0	73	.262	14°23'	0.0204
7	4.0	19.6	2.0	76	.250	13°39'	0.0212
8	5.0	15.7	1.5	79			

TABLE 9.07

EFFECTS OF MISCH METAL ADDITION ON THE MECHANICAL PROPERTIES OF HYPEREUTECTIC ALUMINIUM SILICON-ALLOY (13% Si), SAND CAST.

S. No.	% alloying element	Tensile strength kgm/cm^2	% elongation.	Hardness VHN
1	0.0	7.5	3.0	64
2	0.5	9.2	3.5	64
3	1.0	10.9	5.0	63
4	2.0	8.8	5.0	64
5	3.0	8.8	4.0	67
6	4.0	8.9	3.5	65

PLATE 9.20 106 X

Al + 13 % Si
(SAND CAST)

PLATE 9.21 106 X

Al + 13 % Si + 0.5 wt. % MISCH METAL
(SAND CAST)

PLATE 9.22 106 X

Al + 13 % Si + 1.0 wt. % MISCH METAL
(SAND CAST)

PLATE 9.23 106 X

Al + 13 % Si + 2 wt. % MISCH METAL
(SAND CAST)

PLATE 9.24 106 X

Al + 13 % Si + 3 wt. % MISCH METAL
(SAND CAST)

PLATE 9.25 106 X

Al + 13 % Si + 4 wt. % MISCH METAL
(SAND CAST)

PLATE 9.26 106 X

Al + 13% Si
(METAL MOULD CAST)

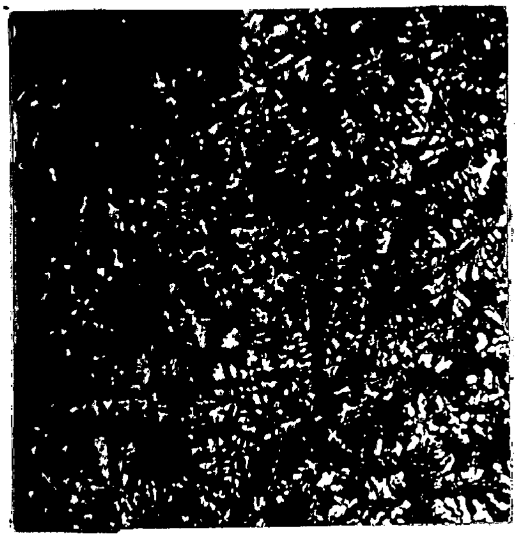

PLATE 9.27 106 X

Al + 13% Si + 0.2 wt.% MISCH METAL
(METAL MOULD CAST)

PLATE 9.28 106 X

Al + 13% Si + 1.0 wt.% MISCH METAL
(METAL MOULD CAST)

PLATE 9.29 106 X

Al + 13% Si + 2.0 wt.% MISCH METAL
(METAL MOULD CAST)

PLATE 9.30 106 X

Al + 13% Si + 3 wt.% MISCH METAL
(METAL MOULD CAST)

PLATE 9.31 106 X

Al + 13% Si + 4 wt. % MISCH METAL
(METAL MOULD CAST)

PLATE 9.32 106 X

Al + 13% Si + 5 wt. % MISCH METAL
(METAL MOULD CAST)

TABLE 9.08

EFFECTS OF LANCER AMP-2 ADDITION ON THE MECHANICAL PROPERTIES OF HYPEREUTECTIC ALUMINIUM-SILICON ALLOY (13% Si) METAL MOULD COST.

S. No.	% Alloying element.	Tensile strength kgm/cm^2.	% elongation	Hardness VHN
1	0.0	16.8	3.0	67
2	0.2	15.1	2.5	70
3	0.5	20.7	4.0	70
4	1.0	21.8	5.0	72
5	1.5	19.4	5.0	72
6	2.5	20.4	4.0	72
7	3.5	23.5	4.0	77

TABLE 9.09

EFFECTS OF RARE EATH FOUORDS ON THE MECHANICAL PROPERTIES OF HYPEREULECTIC ALUMINIUM SILCON ALLOY (13% Si) METAL MOULD CAST.

	METAL MOULD CAST				SAND CAST			
S. No.	% alloying element.	Tensile strength kgm/cm^2	% elongation.	Hardness VHN	% alloying element	Tensile strength kgm/cm^2	%elongation	Hardness VHN
1	0.0	16.9	3.0	66	0.00	7.3	3.0	64
2	0.2	21.3	10.0	73	0.10	8.3	2.0	68
3	0.5	-	-	68	0.20	9.8	2.0	68
4	1.0	19.4	2.5	72	0.50	7.6	1.5	65
5	1.5	16.5	3.0	70	0.75	7.4	2.0	65
6	2.5	-	-	81	1.00	6.8	1.5	64

PLATE 9.33 106 X

Al + 13 % Si + 0.2 wt. % LANCER AMP. 2
(METAL MOULD CAST)

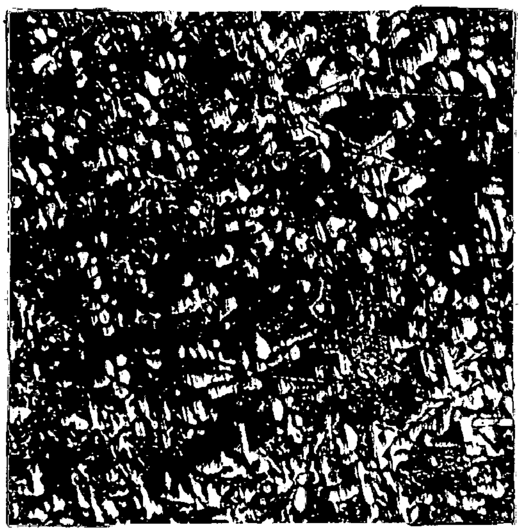

PLATE 9.34 106 X

Al + 13 % Si + 0.5 wt. % LANCER AMP. 2
(METAL MOULD CAST)

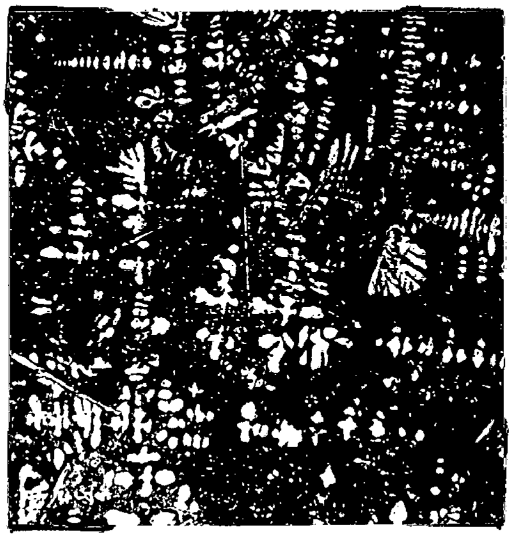

PLATE 9.35 106 X

Al + 13 % Si + 1 wt. % LANCER AMP. 2
(METAL MOULD CAST)

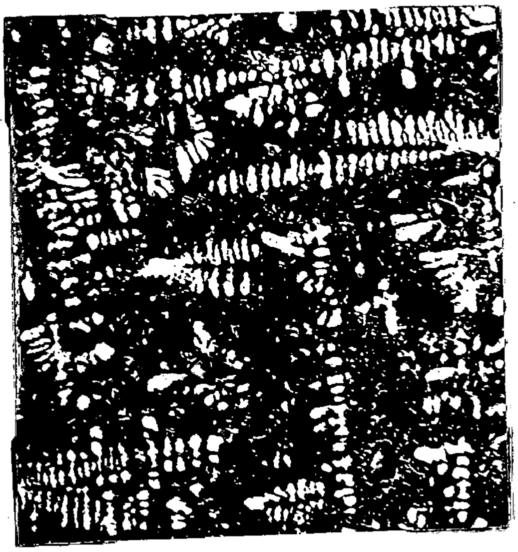

PLATE 9.36 106 X

Al+13% Si+1.5 wt.% LANCER AMP.2
(METAL MOULD CAST)

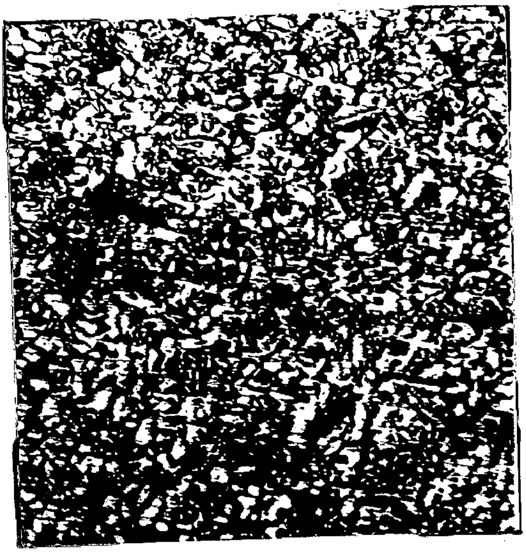

PLATE 9.37 106 X

Al+13% Si+2.5wt.% LANCER AMP.2
(METAL MOULD CAST)

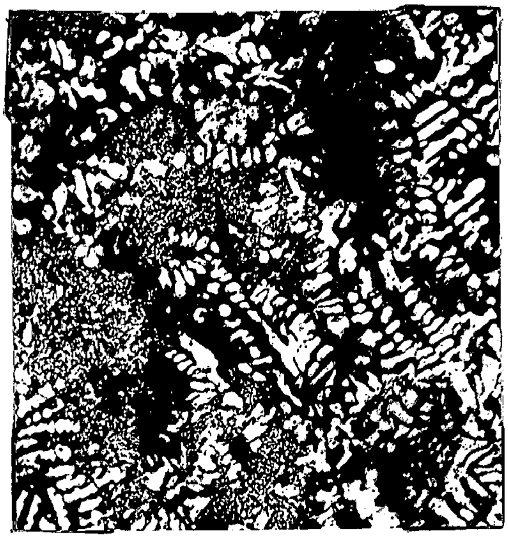

PLATE 9.38 106 X

Al+13%Si+3.5wt.% LANCER AMP.2
(METAL MOULD CAST)

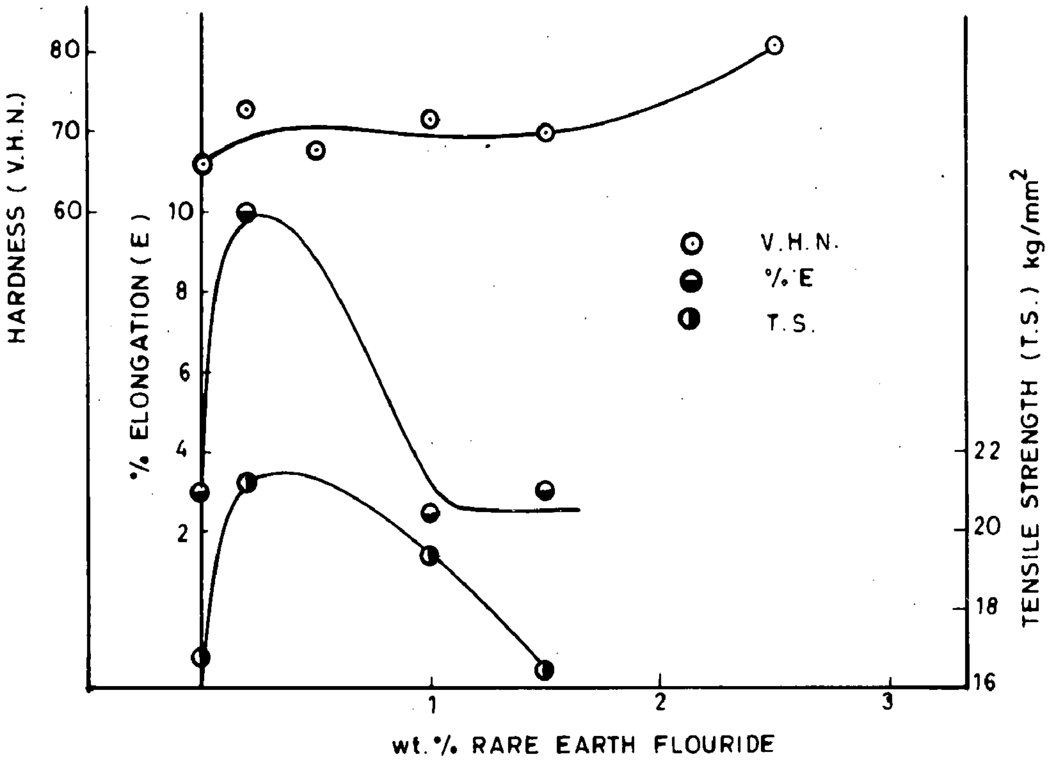

FIG.9.13_EFFECT OF RARE EARTH FLOURIDE ON THE MECHANICAL PROPERTIES OF Al+13 % Si (METAL MOULD CAST)

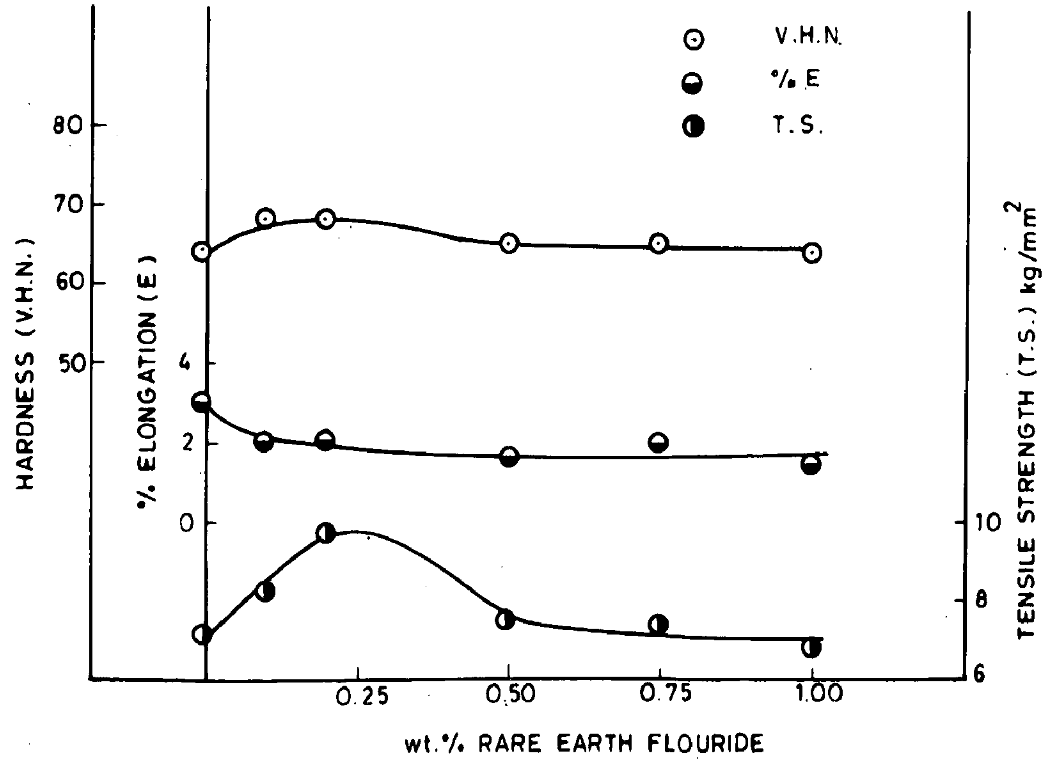

FIG.9.14_EFFECT OF RARE EARTH FLOURIDE ON THE MECHANICAL PROPERTIES OF Al+13 % Si (SAND CAST)

PLATE 9.39 106 X

Al+13% Si+0.2 wt.% R.E.FLUORIDE
(METAL MOULD CAST)

PLATE 9.40 106 X

Al+13% Si+0.5 wt.% R.E.FLUORIDE
(METAL MOULD CAST)

PLATE 9.41 106 X

Al+13% Si+1.0 wt.% R.E.FLUORIDE
(METAL MOULD CAST)

PLATE 9.42 106 X

Al+13% Si+1.5 wt.% R.E.FLUORIDE
(METAL MOULD CAST)

PLATE 9.43 106 X

Al+13% Si+0.05wt%R.E.FLUORIDE
(SAND CAST)

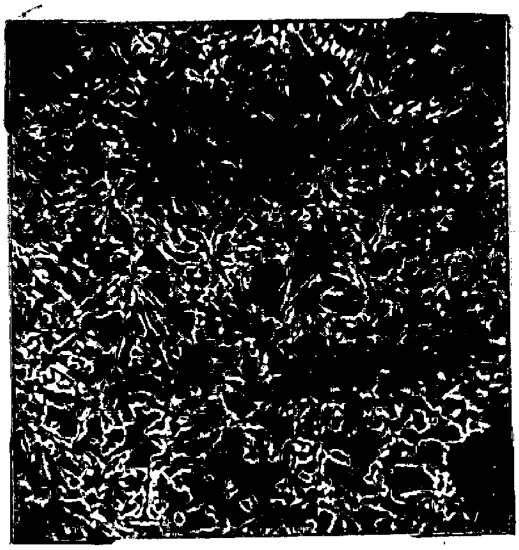

PLATE 9.44 106 X

Al+13% Si+0.1wt% R.E.FLUORIDE
(SAND CAST)

PLATE 9.45 106 X

Al+13% Si+0.2wt% R.E.FLUORIDE
(SAND CAST)

PLATE 9.46 106 X

Al+13 % Si+0.5wt % R.E. FLUORIDE
(SAND CAST)

PLATE 9.47 106 X

Al+13 % Si+0.75wt % R.E. FLUORIDE
(SAND CAST)

PLATE 9.48 106 X

Al+13 % Si+1.0 wt % R.E. FLUORIDE
(SAND CAST)

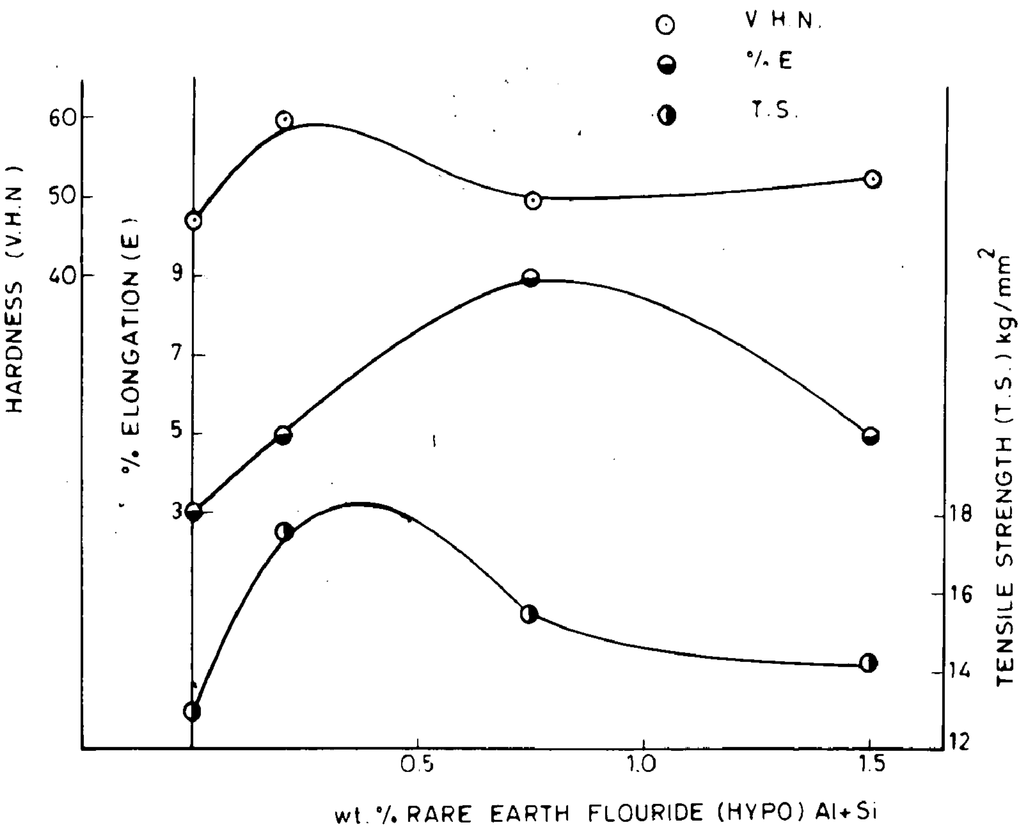

FIG.9.15_EFFECT OF RARE EARTH FLOURIDE ON THE MECHANICAL PROPERTIES OF Al+7.5 % Si

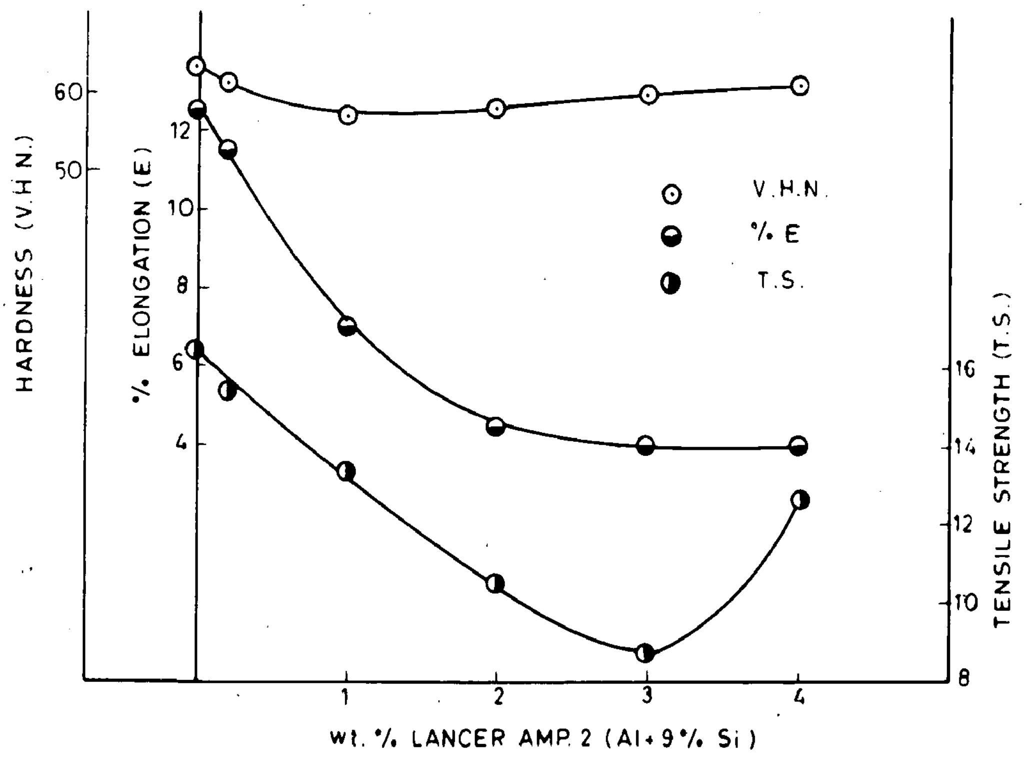

FIG.9.16_EFFECT OF LANCER AMP. 2 ON THE MECHANICAL PROPERTIES OF Al+9 % Si (METAL MOULD CAST)

TABLE 9.10

EFFECTS OF RARE EARTH FLUORIDES ON THE MECHANICAL PROPERTIES OF HYPOEUTECTIC ALUMINIUM-SILICON ALLOY (7.5% Si), METAL MOULD CAST.

S.No.	% Addition	Tensile Strength Kgm/mm^2	% Elongation	Hardness V.H.N.
1	0.00	12.9	3.0	47
2	0.20	17.6	5.0	60
3	0.75	15.5	9.0	50
4	1.50	14.2	5.0	53

TABLE 9.11

EFFECTS OF LANCER-AMP-2 & MISCH METAL ADDITIONS ON THE MECHANICAL PROPERTIES OF HYPOEUTECTIC ALUMINIUM SILICON ALLOY (9% Si), METAL MOULD CAST.

S. No.	Lancer -amp2				Misch Metal		
	% Addition.	Tensile Strength Kgm/mm^2	% elongation.	Hardness	Tensile strength Kgm/mm^2	% elongation	Hardness
1.	0.0	16.4	12.5	63	16.4	12.5	63
2	0.2	15.3	11.5	61	15.0	10.0	59
3	1.0	13.3	7.0	57	14.8	8.0	55
4	2.0	10.5	4.5	58	12.8	7.0	58
5	3.0	8.7	4.0	65	11.4	7.0	67
6	4.0	12.7	4.0	66	8.4	4.0	66

PLATE 9.49 106 X

Al+7.5% Si+ 0 wt%R.E.FLUORIDE
(METAL MOULD CAST)

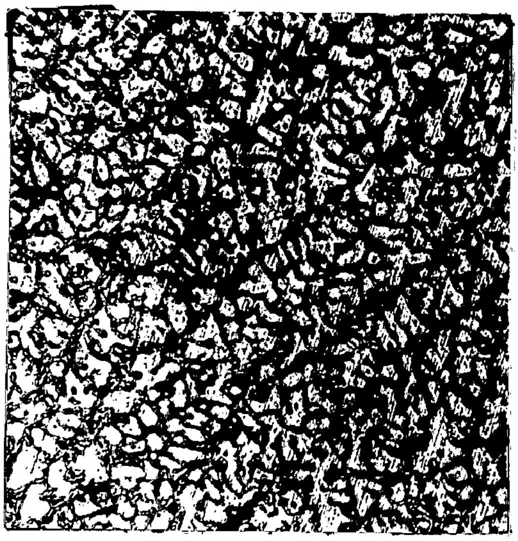

PLATE 9.50 106 X

Al+7.5% Si+0.2 wt%R.E.FLUORIDE
(METAL MOULD CAST)

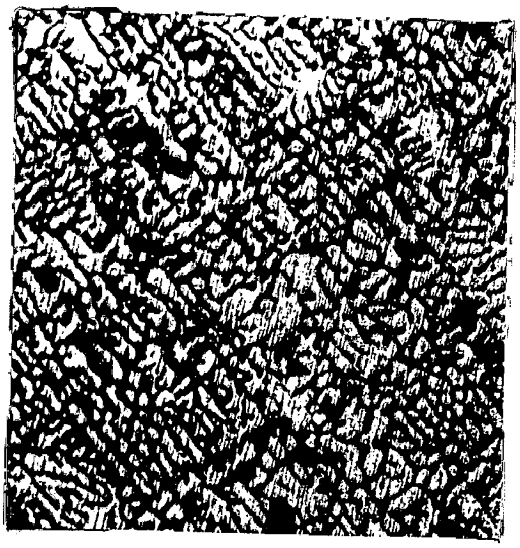

PLATE 9.51 106 X

Al+7.5% Si+0.75wt%R.E.FLUORIDE
(METAL MOULD CAST)

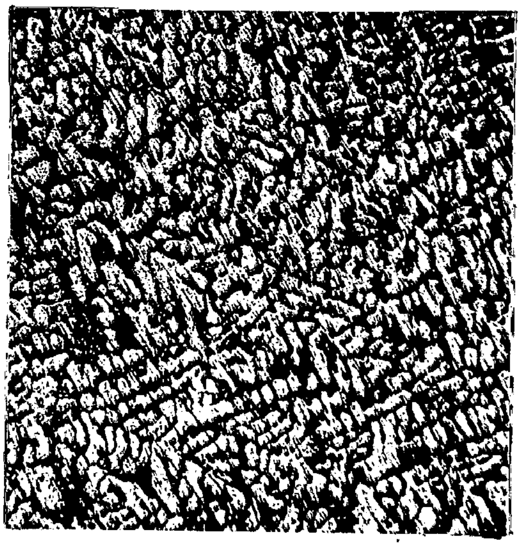

PLATE 9.52 106 X

Al+7.5%Si+1.5 wt %R.E.FLUORIDE
(METAL MOULD CAST)

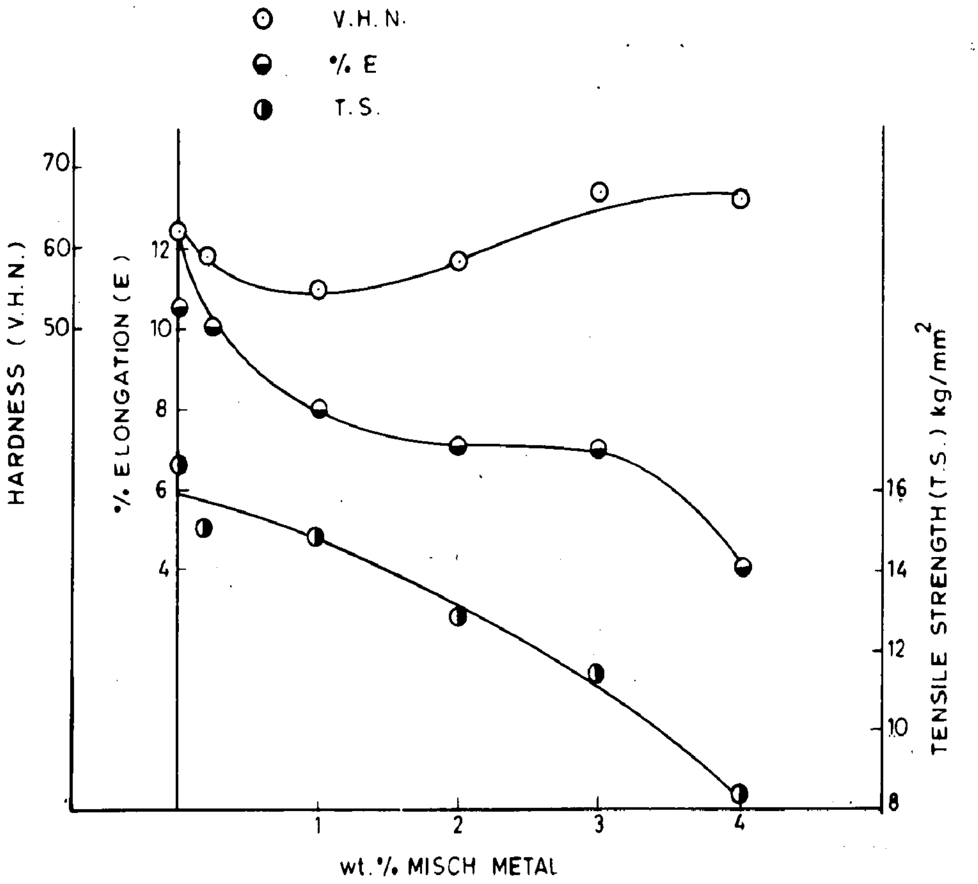

FIG.9.17_EFFECT OF MISCH METAL ON THE MECHANICAL PROPERTIES OF Al+9 %Si (METAL MOULD CAST)

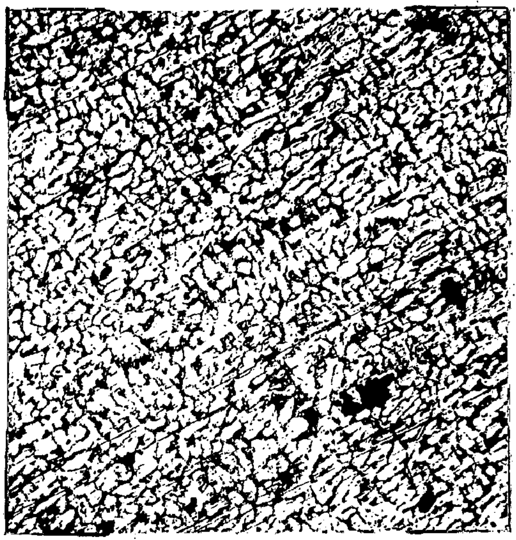

PLATE 9.56 106 X

Al + 9% Si + 3 wt% MISCH METAL
(METAL MOULD CAST)

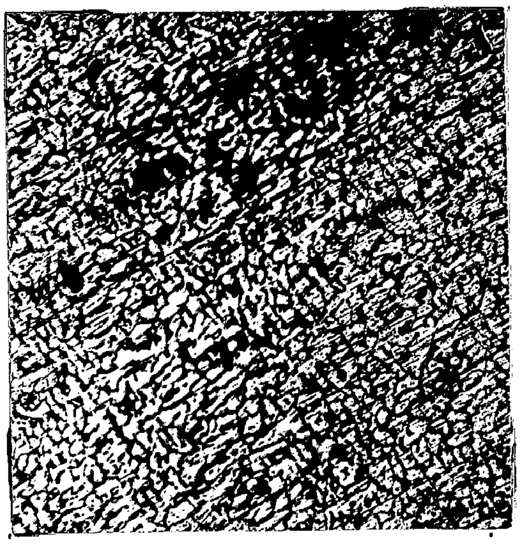

PLATE 9.57 106 X

Al + 9 % Si + 4 wt % MISCH METAL
(METAL MOULD CAST)

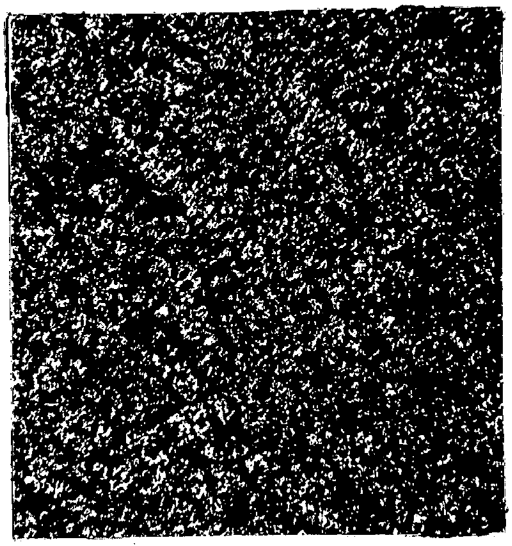

PLATE 9.58 106X

Al+9%Si
(METAL MOULD CAST)

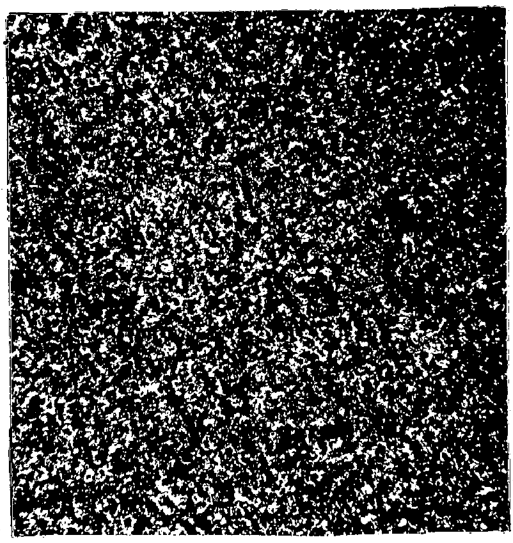

PLATE 9.59 106X

Al+9% Si+1.0 wt % LANCER AMP. 2
(METAL MOULD CAST)

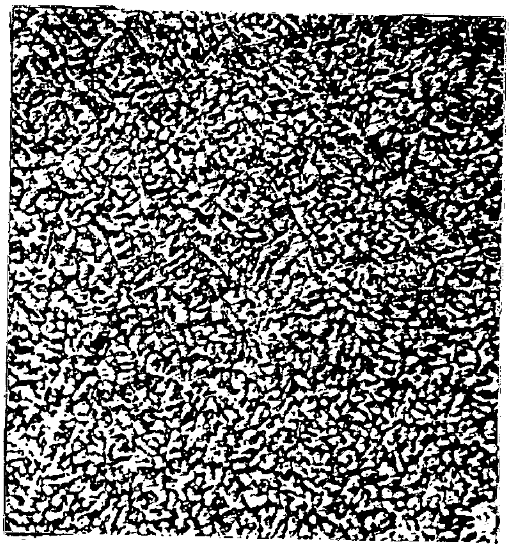

PLATE 9.60 106X

Al+9 % Si+2.0wt % LANCER AMP. 2
(METAL MOULD CAST)

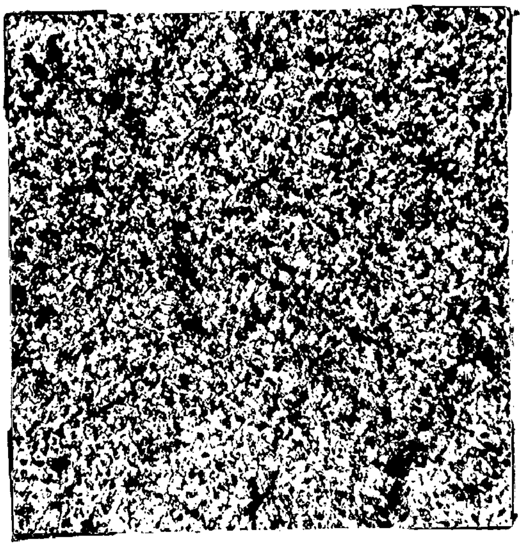

PLATE 9.61 106X

Al+9% Si+ 3wt % LANCER AMP 2
(METAL MOULD CAST)

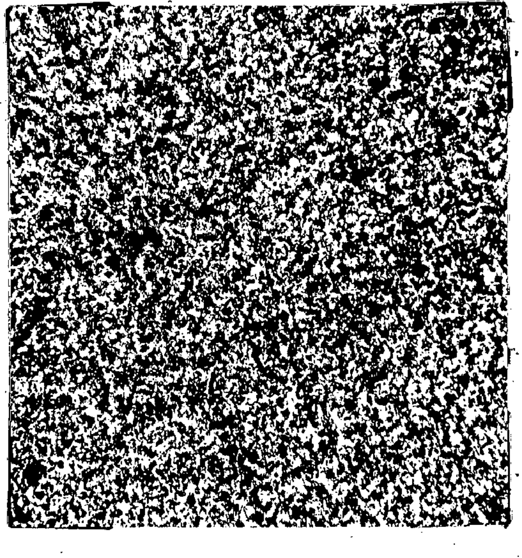

PLATE 9.62 106X

Al+9% Si+4 wt % LANCER AMP 2
(METAL MOULD CAST)

9.5 DISCUSSION OF THE RESULTS :

9.5.1 Effects of tantalum additions to hypereutectic aluminium-silicon alloys (11.9% Si) Metal mould cast:

Salient features of the results are given below:

i) Plate 9.01 indicates the microstructure of Al + 11.9 Wt.% Si metal mould cast. It consists of coarse eutectic and primary silicon plates. With increase in tantalum addition there appears to be the refinement of the silicon plates and the eutectic. The amount of α goes on increasing. At about 0.015% tantalum addition the structure appears to get modified. With further additions the amount of silicon containing eutectic phase decreases and a new phase starts appearing, the amount of which appears to increase with further additions.

ii) The tensile strength increases upto 0.025% addition beyond which it decreases.

iii) The %-elongation remains almost constant with tantalum additions.

iv) The hardness slightly decreases to a constant value at about .015% addition.

v) Both chip thickness ratio and shear plane angle ϕ decreases upto 0.01% addition indicating a slight detrimental effect on machinability. Beyond 0.01% addition it increases to a constant value at about 0.025% addition. The values of r and Ø of the alloy treated with 0.025% addition are almost equal to that of the untreated alloy.

vi) Unit HP/Cm^3/min. slightly increases upto 0.005% addition. With further additions it again decreases upto 0.025% addition. The unit HP curve is in agreement with the chip thickness ratio and shear plane angle Ø plot. The optimum amount of tantalum addition for best machinability from power consideration is found out to be 0.025% .

Tantalum is found to have no solid solubility in aluminium and silicon. Thus with the additions of tantalum it will react with aluminium to form $TaAl_3$ (154,155) and silicon to form $TaSi_2$ and Ta_5Si_3 (171). The increase in tensile strength may be partially attributed to modification, and partially to the fine dispersion of intermetallic compounds.

The increase in machinability beyond 0.005% upto 0.025% may be due to modification resulting in refinement of silicon plates and eutectic; predominating over the slight adverse effect due to the presence of intermediate phases. For additions beyond 0.025% addition the two effects seem to balance each other resulting in constant value of machinability.

The modification effect produced by tantalum addition can easily be explained on the basis of recent theory on modification proposed by Kim and Heine (170). According to this theory, phase shapes such as polyhedral coarse or fine plates in the normal eutectic or globular as in modified eutectic are growth-temperature dependent. The polyhedral silicon or silicon plates in normal alloy are produced by the habit characteristics of the higher temperature of formation. The modifying element does not cause the globular shape in modified eutectic instead it lowers the nucleation temperature to less than 580 °C, where the globular shape characteristically gorws. They have formulated the requirements of a modifying element that it should have a tendency to form compounds with the precipitating phase at temperature below the normal eutectic temperature, should have a low compound forming tendency, low solubility and possibly a miscibility gap with the solvent phase.

As explained above, it can be seen that tantalum satisfies the requirements of a modifying element except that it has also the tendency to react with the solvent phase preventing the added element to react with the precipitating phase. Because of the formation of intermetallic compounds with silicon the nucleation temperature of silicon is depressed to the temperature of the globular growth resulting in partially modified structure.

9.5.2 Effects of cerium additions to hyper-eutectic aluminium alloy (11.9% Si)

The salient features of the results are given below:-

i) The effect of cerium additions is to modify the structure. The eutectic gets refined, the amount of primary α increases. Complete modification is however, not brought about upto about 0.5 % cerium. The microstructure of the alloy treated with 0.5% cerium consists of fine eutectic, primary α and some silicon plates. Complete modification is brought about , however, at about 0.75 % cerium. There is almost complete disappearance of silicon plates in the microstructure. The microstructure of the alloy treated with 0.75% cerium contains fine eutectic in the matrix of primary alpha. With further additions there appears to be reversal and the silicon plates are again visible in the microstructure.

ii) The tensile strength increases to a constant value at about 0.5% addition.

iii) The % elongation also slightly increases upto about 0.3% addition, beyond which it becomes constant with further addition.

iv) There is large increase in hardness upto about 0.2% addition beyond which it increases to a constant value at about 0.3% addition. Beyond 0.75% addition it again goes on increasing with further additions.

v) Both chip thickness ratio and shear plane angle Ø increase upto 0.1% addition. With further additions they are found to decrease at a linear rate.

vi) HP/cm^3/min decreases upto 0.1% addition beyond which it increases to almost a constant value at about 0.5% addition.

Cerium is reported to dissolve upto 0.05% by weight in aluminium and has no solubility in the precipitating phase silicon. So most of the cerium reacts with aluminium to form intermetallic compounds Al_4Ce, Al_2Ce, AlCe and with the precipitating phase silicon to form compounds Ce_3Si, Ce_2Si and CeSi. The increase in hardness and tensile strength may partially be attributed to the modification and partially to the strengthening effect produced by the fine dispersion of the intermetallic compounds. Slight increase in % elongation may also be attributed to the modification effect produced on cerium addition.

Fig. 9.04 indicates the effect of cerium addition on the machinability of AlSi alloy. From both chip formation and power considerations, it can be seen that beneficial effect of cerium addition upto 0.1% has been produced on machinability. With further additions the machinability decreases.

The effect of cerium addition on the machinability can easily be understood on the basis of the following considerations:-

i) With cerium addition the refinement of the silicon plates and eutectic will bring about an improvement in machinability upto 0.75% addition.

ii) With increasing addition of cerium beyond 0.05% the amount of hard intermetallic phases will increase resulting in decrease of machinability.

For lower additions the beneficial effect due to (i) predominates over the little detrimental effect produced due to (ii) because cerium goes into solid solution upto 0.05% and the amount of intermediate phase will be very small. For higher additions , however, due to large increase in the amount of intermediate phases, detrimental effect produced predominates over the beneficial effect due to (i) resulting in decrease of machinability.

It can be seen here that though some modification effect has been produced by cerium it is not a very good modifier, as can be seen that with 0.75% addition even in metal mould cast the complete modification has not been brought about. The microstructure consists of some silicon plates in addition to primary alpha and fine eutectic. The modifying effect produced can easily be understood on the basis of the recent theory on modification proposed by Kim and Heine (170). Cerium satisfies the requirements of a modifying element as given by them except that it has also the tendency to react with the solvent phase to form intermetallic compounds, and also little solubility in the solvent phase. Thus the nucleation temperature of silicon is depressed to the temperature of globular growth only with larger additions of cerium when sufficient amount of modifying element is present to react with the silicon phase. The effect is not so pronounced and complete modification is not brought about even in metal mould cast due to above reasons, as the conditions are only partially satisfied.

9.5.3 Effects of lanthanum additions to hyper-eutectic aluminium-silicon alloy (11.9% Si):

The salient features of the results are given below :-

i) An important effect of lanthanum addition is to refine the eutectic. There appears to be very little change upto 0.01% addition. The structure appears to be partially modified at about 0.05% addition. The microstructure consists of fine eutectic in the matrix of α dendrites. Few silicon plates are also visible. With further additions there appears to be the presence of some new phase in addition to eutectic and few silicon plates.

ii) The tensile strength increases upto about 0.05% addition, beyond which it decreases to a constant value at about 0.1% addition, the value, however, being more than that of the untreated alloy, showing the beneficial effect of lanthanum addition on the tensile strength for all compositions.

iii) The % elongation also increases upto 0.05% addition beyond which it decreases to a constant value at about 0.1% addition.

iv) The hardness is found to increase with lanthanum addition. At about 0.15% addition the value becomes almost constant.

v) Both chip thickness ratio and shear plane angle Ø slightly decrease to constant values at about 0.05% addition.

vi) HP/cm^3/min also increases, with lanthanum additions indicating the detrimental effect of lanthanum on machinability.

As discussed previously lanthanum has no solid solubility in aluminium and silicon. Thus the whole amount of lanthanum added reacts with aluminium to form intermetallic compounds, Al_4La etc., and with the precipitating phase silicon to form LaSi. The increase in tensile strength and hardness may be partially attributed to the partial modification resulting in refinement of eutectic increase in the amount of primary alpha, and partially to the strengthening effect produced by the fine dispersion of the intermetallic compounds. The increase in % elongation may, however, be attributed to the refinement produced. The decrease in tensile strength and %-elongation beyond 0.1% may be due to relatively increased amount of the intermetallic compounds, increase in their size, non-uniform distribution as well as recoarsening of the matrix.

It can be seen that slight detrimental effect of lanthanum addition has been produced on machinability from both chip formation and power consideration. It appears that detrimental effect produced due to large amount of intermediate phases predominates over the little beneficial effect due to slight refinement with a net result in decrease of machinability with lanthanum additions.

For additions beyond 0.1% due to decrease in detrimental effect because of decrease in strength, the two effects seem to balance each other, resulting in constant value of machinability as obtained from both chip formation and power consumption considerations.

The modifying effect produced by lanthanum is not surprising. Lanthanum satisfies the requirements of a modifying element except that it has also the tendency to form intermetallic compounds with aluminium preventing the added element to react with the precipitating phase . Due to the above fact, it has not acted as good modifier and only partial modification takes place with lanthanum addition. Since there is no solubility of lanthanum in aluminium, immediately with the addition, there is possibility of the formation of LaSi to suppress the nucleation temperature and not as in the case of cerium which dissolves upto 0.05 %. This explains why in the case of lanthanum only small addition has been effective in modifying the structure when in case of cerium the effective addition has been found out to be about 0.75% .

9.5.4 Effects of neodymium additions to the hyper eutectic aluminium-silicon alloy (11.9% Si) :

The salient features of the results are given below:-

i) An important effect of neodymium addition is to refine the eutectic. There appears to be disappearance of silicon plates, decrease in the silicon containing phase and increase in the amount of alpha upto 0.2% neodymium. The structure of the alloy treated with 0.2% neodymium appears to be partially modified. Beyond 0.2% addition there is again reappearance of silicon plates in the microstructure.

ii) The tensile strength remains constant upto 0.1% addition. Beyond 0.1% it is found to increase with neodymium addition upto 0.3% .

iii) The % elongation remains almost unchanged with neodymium additions.

iv) The hardness increases appreciably to a constant value at about 0.2% addition.

v) Both chip thickness ratio and shear plane angle ø remain almost constant upto 0.2% addition. Beyond 0.2; **percent** they are found to increase with further additions.

vi) HP/cm3/min increases upto 0.1%, beyond which the value is found to decrease to a constant value at about 0.2% addition.

Neodymium has got the solubility of about 0.2% in aluminium. Thus with the addition of neodymium upto 0.2% most of the neodymium goes into solid solution and very small amount reacts with the precipitating phase to form $NdSi_2$. For additions beyond 0.2%, Nd should react with aluminium also to form $NdAl_2$, $NdAl_4$ etc. The increase in hardness and tensile strength may be attributed partially to the modification and partially to the strengthening effect produced due to the fine dispersion of intermetallic compounds especially at higher additions.

Improvement in machinability between 0.1% and 0.2% may be due to partial modification resulting in refinement of silicon plates and eutectic and increase in primary alpha.

The effect of neodymium to modify the structure partially can easily be understood on the basis of same considerations as discussed in previous cases. Neodymium does not fulfil all the requirements of a modifying element. Though it has no solubility for the precipitating phase, it has definite solubility upto 0.2% in aluminium. Because Nd has tendency to go into solid solution preventing the added element to react with the precipitating phase, with the addition of neodymium very little of $NdSi_2$ phase is available and is not sufficient to suppress the nucleation temperature completely to a temperature of globular growth. Thus large addition of about 0.2% Nd has been effective in modifying the structure and not small additions, as in the case of lanthanum. With the addition of neodymium upto 0.2% very little of $NdSi_2$ phase is available and is not sufficient to **su**press the nucleation temperature of all silicon to a temperature of globular growth and therefore, some silicon plates are formed and structure is only partially modified. For additions beyond 0.2% the tendency to form $NdSi_2$ is further hindered due to the formation of intermetallic compounds of neodymium with aluminium.

9.5.5 Effects of misch metal additions to hyper-eutectic aluminium silicon alloy (13% Si)

The salient features of the results are given below:-

i) An important effect of misch metal addition is to **modify the** structure. There is little effect upto 0.5% addition. Complete modification is, however, obtained at about 1% addition in both sand and metal mould cast alloys. With further additions recoarsening of the eutectic takes place. Beyond 2% addition some new phase makes its appearance and seems to increase with further additions.

ii) The tensile strength increases upto about 1% addition and then decreases in both sand and metal mould cast alloys (Fig. 9.09 and 9.11) The value has been found out to be higher than Na treated alloy, cast under the same conditiond.

iii) The % elongation increases upto about 2% addition and then decreases in both sand and metal mould cast alloys.

iv) There is gradual and slight increase in the hardness values of sand cast aluminium silicon alloys treated with misch metal additions. However, regular and larger increase in the hardness values with misch metal addition has been found in chill cast.

v) Both chip thickness ratio and shear plane angle Ø increase upto 0.5% addition beyond which they decrease to constant values at about 1.5% addition.

vi) HP/cm^3/min decreases upto 0.5% addition. With further additions it increases to a constant value at about 1.5% addition. The curve is found to be in agreement with chip thickness ratio and shear plane angle Ø curves. Optimum addition on the basis of both the criteria for best machinability has been found out to be 0.5% .

Cerium and neodymium are reported to dissolve upto 0.05% and 0.2% by weight in aluminium, while lanthanum is reported to have no solid solubility. So most of the amount of cerium, lanthanum content reacts with aluminium to form intermetallic compounds Al_4 Ce and Al_4La . The cerium and lanthanum content and little of neodymium may also react with the precipitating phase silicon to form compounds Ce_3Si, Ce_2 Si, CeSi, LaSi and $NdSi_2$. The increase in hardness can easily be attributed to the formation of fine particles of harder intermetallic compounds. The increase in tensile strength may be partially attributed to the modification and partially to the strengthening effect produced by the fine dispersion of the intermetallic compounds and not as in sodium modified alloy, where the increase may only be attributed to the modification of the structure. The increase in tensile strength upto 1% can also be understood on the basis of the effects of individual elements. 1% misch metal will contain about 0.5% cerium, 0.2% lanthanum and about 0.15% neodymium. It may be noted that all the three elements have produced beneficial effects on tensile strength. The large increase in hardness and tensile strength may be the cumulative effects of all the three elements. The decrease in tensile strength and % elongation beyond 1% addition may be due to relatively increased amount of intermetallic compounds, increase in their size and non uniform distribution as well as recoarsening of the matrix.

The improvement in machinability upto about 0.5% addition may be due to partial modification resulting in refinement of the eutectic and silicon plates. For higher additions the adverse effect due to increase in the amount of intermediate phases, thereby increasing the hardness and tensile strength predominates over the beneficial effect due to refinement of the structure with net result of decrease in machinability at an appreciable rate upto 1%. Beyond 1% since the tensile strength drops down, detrimental effect of misch metal due to harder intermediate phases seems to balance with the beneficial effect produced due to decrease in tensile strength, resulting in constant value of machinability.

The large modification effect produced by misch metal can easily be understood on the basis of individual elements cerium, lanthanum and neodymium contained in misch metal and the recent theory of Kim and Heine (170) on modification. Our experiments indicate that in the unmodified alloy microstructure consists of polyhedral silicon and coarse silicon plates. With increasing misch metal addition coarse silicon plates are converted into fine plates and finally into globular with about 1% addition. The microstructure of the alloy treated with 1% misch metal consists of only primary alpha solid solution and globular fine eutectic. If the temperature dependance of the silicon phase shape is confirmed it can be easily seen that misch metal depresses the nucleation temperature of silicon. Small addition does not seem to be effective and depression to temperature of globular growth is only possible with about 1% addition. It is interesting to note that the elements cerium, lanthanum and neodymium contained in misch metal satisfy the requirement of modifying elements except that these elements have also the tendency to react with the solvent phase to form intermetallic compounds, and cerium and neodymium have solubilities of 0.05 % and 0.2% respectively for solvent phase, preventing the added elements to react with the precipitating phase. Thus the nucleation temperature of silicon is depressed to the temperature of globular growth only with larger additions of misch metal when sufficient amount of modifying element is present to react with the silicon phase.

The modification effect of misch metal can also be explained on the basis of the effects of individual elements. It can be easily seen that since all the three elements cerium, lanthanum and neodymium do not satisfy all the requirements of a modifying element, they are, therefore, only partial modifiers, the optimum additions of Ce, La and Nd being 0.75%, 0.05% and 0.2% additions respectively. Although as discussed above the individual elements contained in misch metal are only partial modifiers the optimum addition of 1% of misch metal has been found out to be very effective in modifying the structure completely in both sand and metal mould cast alloys. It appears that this may be due to the combined effects of all the three elements cerium, lanthanum and neodymium which could individually otherwise only partially modify the structure.

9.5.6 Effects of Lancer-amp 2 additions to hyper-eutectic aluminium silicon alloy (13% Si) :

The salient features of the results are given below:-

i) Metallographic study indicates that with increasing addition of lancer-amp2 more and more modification of the structure takes place upto about 1.5% addition. Beyond 2% addition, a new phase having eutectic structure makes its appearance, the amount of which increases considerably with further additions. The distribution of the phase appears to be uniform.

ii) The tensile strength increases upto about 1.0% addition. Between 1 to 2% addition, it is practically constant. Beyond 2% addition it again goes on increasing.

iii) The % elongation increases upto about 1.5% addition and then decreases to a constant value at 2.5% addition.

iv) There is **a** slight but regular increase in hardness values with lancer-amp2 additions.

Metallographic study indicates that striking influence of lancer-amp2 addition is to refine the structure. The complete modification is, however, obtained at about 1.5% addition. The modifying effect of lancer-amp 2 can be easily attributed to the presence of cerium, lanthanum and neodymium in lancer-amp 2 , and can be explained exactly in the same way as in misch metal. Beyond 2.5% addition there appears to be the formation of some new phase having eutectic appearance. The new phase may be the eutectic of $Al-Al_4La$ or $Al-Al_4Ce$. The increase in hardness and tensile strength may be attributed to the modification as well as to the strengthening effect produced by the fine dispersion of the intermetallic compounds. Unlike misch metal, beyond 1.5% addition no coarsening of the phase takes place. The distribution of the phase is fine and uniform resulting in increase of tensile strength. Upto 3.5% addition the refining action may be because of the presence of zirconium in lancer-amp 2.

9.5.7 Effects of rare-earth fluorides additions to hyper-eutectic aluminium silicon alloy (13% Si) :

The salient features of the results are given below.

i) Metallographic study indicates that there is definite modification of the eutectic with 0.2 % addition in metal mould cast alloys. Complete modification is however, not brought about in sand cast alloys, although there is definite refining of both silicon plates and polyhedral silicon upto 0.2% addition.

ii) Both tensile strength and % elongation increase upto 0.2% addition in sand and metal mould cast alloys.

iii) Very slight but regular increase in hardness values was observed in metal mould cast alloys, however, in sand cast alloys there is increase only upto 0.2% addition beyond which it decreases to a constant value.

Metallographic study indicates that there is definite modification of the structure with the addition of 0.2% fluorides in chill cast alloys. With further additions alpha dendrites become coarser and their distribution becomes non-uniform. In the case of sand cast alloys complete modification is not brought about. There is definite refining of both polyhedral silicon plates upto 0.2% additions. Rare earth fluorides when added to molten alloy will decompose to their elemental form, out of which very little will go into solid solution, most of which will react with the precipitating silicon and the solvent phase to form intermetallic compounds. Only a small amount decomposes to the metallic state till the temperature of polyhedral silicon growth is reached, and may not be sufficient to suppress the nucleation temperature of all silicon to the temperature of globular growth. This explains why polyhedral silicon has not vanished though there is definite refining due to suppression of nucleation temperature of silicon. It appears that complete suppression of nucleation temperature to the temperature of globular growth can however be obtained in chill cast alloys by combined effects of chilling and modifier.

The increase in tensile strength and hardness may be due to refining of the structure and due to the presense of sub microscopic precipitates of intermetallic compounds. The increase in % elongation and tensile strength may also be partially due to the hydrogen degassification of the molten metal by fluorides.

9.5.8 Effects of rare-earth fluorides addition to hypoeutectic aluminium-silicon alloys, (7.5% Si)

The salient features of the results are given below:

i) Metallographic study indicates that with rare-earth fluorides addition, both the eutectic and primary alpha get refined upto 0.2% addition. The amount of alpha dendrites does not appear to change with the addition.

ii) The tensile strength increases upto about 0.2% addition and then decreases to almost constant value at about 1.5% addition.

iii) The % elongation increases upto 0.75% addition and then decreases.

iv) The hardness increases upto 0.2% addition beyond which it decreases to a constant value at abount 0.75% addition.

Metallographic study indicates that there is definite refining of primary alpha and eutectic upto 0.2% addition. Beyond 0.2% addition recoarsening of the matrix takes place. The increase in hardness, tensile strength and % elongation upto 0.2% addition may be due to the refining of the structure. The modification effect produced can be explained on the same basis as discussed in the case of hypereutectic aluminium-silicon alloys.

9.5.9 Effects of misch metal and lancer-amp2 additions to hypoeutectic aluminium -silicon alloy (9.0% Si)

The salient features of the results are given below:-

i) Metallographic study indicates that with increase in misch metal and lancer-amp2 addition there is gradual decrease of the amount of eutectic phase and increase in the size and mount of primary alpha dendrites. Between 3 to 3 % lancher-amp2 addition there appears to be the formation of some new phase having eutectic appearance, the amount of which increases considerably with further additions.

ii) It can be seen that the tensile strength decreases in both the cases. The decrease may be due to the increase in the coarseness and amount of alpha dendrites with increased addition.

iii) The % elongation is found to decrease in both the cases of lancer-amp2 and misch metal. With lancer-amp2, however, it decreases to almost constant value at about 2% addition.

iv) Hardness vs. composition curves follows the same path in both the cases of misch metal and lancer-amp2 addition. The hardness decreases upto 1.5% addition. There is increase between 1.5 to 3%.

With about 3% lancer-amp2 addition there appears to be the formation of some new phase having eutectic appearance and uniform distribution in the matrix resulting in an increase in tensile strength unlike misch metal in which regular decrease is observed.

CHAPTER X

EFFECTS OF MISCH METAL ADDITIONS ON THE HIGH TEMPERATURE MECHANICAL PRORERTIES OF ALUMINIUM AND ITS ALLOYS.

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EFFECTS OF MISCH METAL ADDITIONS ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF ALUMINIUM AND ITS ALLOYS

10.1 GENERAL INTRODUCTION .

Review of the literature has indicated that there are conflicting reports of the effects of misch metal on the high temperature mechanical properties of aluminium. Practically no work seems to have been reported on the effects of misch metal addition on the high temperature mechanical properties of aluminium-silicon, aluminium-copper and aluminium-magnesium alloys.

The present investigation reported has been undertaken with a view to explore the possibilities of beneficial additions of misch metal on the high temperature mechanical properties of pure aluminium, alloy 122 (10% Cu and 0.2% Mg), alloy 218 (or G 8A) with 8% Mg, and aluminium silicon alloy with 13% silicon. The properties studied include tensile strength and percentage elongation at 200 and 400 degrees centigrade.

10.2 EXPERIMENTAL PROCEDURE.

10.2.1 Tensile tests at elevated temperature:

Avery tensile testing machine was used for the determination of tensile strength at elevated temperature. The machine is provided with an electric

The leads are of compensating material so that the temperature shown on the millivoltmeter corresponds to that of the cold junction position.

By means of controller the temperature can be maintained constant. The feeler of the temperature controller is introduced into the heating chamber through another hole provided in furnace and the controller connected to the regulating unit. The regulating unit is supplied in a carrying case containing the transformer, resistances, main switch with thermal release, and contactor switch with valves relay for automatic temperature control. The automatic temperature regulator has an accuracy of ± 2 °C.

Samples of appropriate specifications with gauge length of 10 cms were machined from cylindrical cast ingots. Three samples were taken for each composition for performing the tests.

10.3 EXPERIMENTAL RESULTS.

10.3.1 Effects of misch metal additions to aluminium.

The results are given in table 10.01 and Fig. 10.01.

10.3.2 Effects of misch metal addition to aluminium-silicon alloy (13% Si) :

The results are given in Table 10.02 and Fig. 10.02 .

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10.1 GENERAL INTRODUCTION .

Review of the literature has indicated that there are conflicting reports of the effects of misch metal on the high temperature mechanical properties of aluminium. Practically no work seems to have been reported on the effects of misch metal addition on the high temperature mechanical properties of aluminium-silicon, aluminium-copper and aluminium-magnesium alloys.

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10.2 EXPERIMENTAL PROCEDURE.

10.2.1 Tensile tests at elevated temperature:

Avery tensile testing machine was used for the determination of tensile strength at elevated temperature. The machine is provided with an electric furnace which is suspended on two hinges fixed to one of the columns or to the frame of the testing machine. The furnace is provided with three superimposed heating coils each of which can be regulated independently. This arrangement permits distributing the heat uniformly along **the** full length of the specimen. The heating coils of the furnace are completely covered with ceremic sections so that they cannot be possibly touched or damaged by the specimen or metal parts. The construction of the furnace is such that by correct adjustment of the heating coils a uniform temperature over the gauge length of the specimen can be attained. Thermo-couples , the junctions of which are in immediate proximity to the specimen surface or slightly touch the specimen, are used for measuring the temperature of the specimen. Provisions are made for three thermocouples which can be introduced through a series of radial holes in the furnace . In this way the temperature can be measured at the two ends and in the middle of the specimen.

The thermo-electric voltage is measured by a millvolt meter provided with two scales of which one indicates millivolts and the other permits of direct readings in degrees centigrade when using thermocouples of the chromel alumel type.

For taking the temperature each of three thermocouples is successively connected to the millivoltmeter by means of a measuring position selector switch.

The leads are of compensating material so that the temperature shown on the millivoltmeter corresponds to that of the cold junction position.

By means of controller the temperature can be maintained constant. The feeler of the temperature controller is introduced into the heating chamber through another hole provided in furnace and the controller connected to the regulating unit. The regulating unit is supplied in a carrying case containing the transformer, resistances, main switch with thermal release, and contactor switch with valves relay for automatic temperature control. The automatic temperature regulator has an accuracy of ± 2 °C.

Samples of appropriate specifications with gauge length of 10 cms were machined from cylindrical cast ingots. Three samples were taken for each composition for performing the tests.

10.3 EXPERIMENTAL RESULTS.

10.3.1 Effects of misch metal additions to aluminium.

The results are given in table 10.01 and Fig. 10.01.

10.3.2 Effects of misch metal addition to aluminium-silicon alloy (13% Si) :

The results are given in Table 10.02 and Fig. 10.02 .

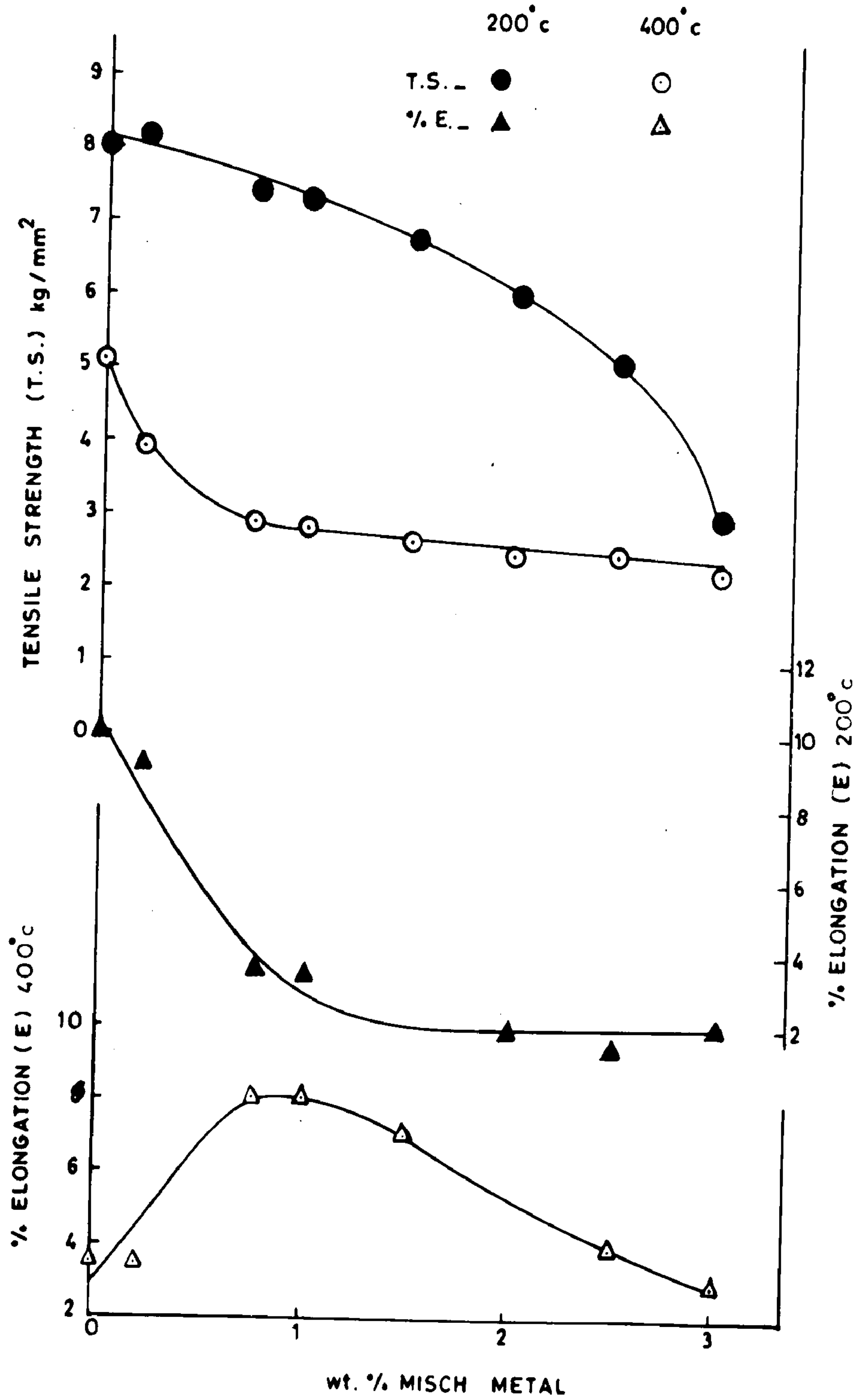

FIG.10.01_EFFECT OF MISCH METAL ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF Al.

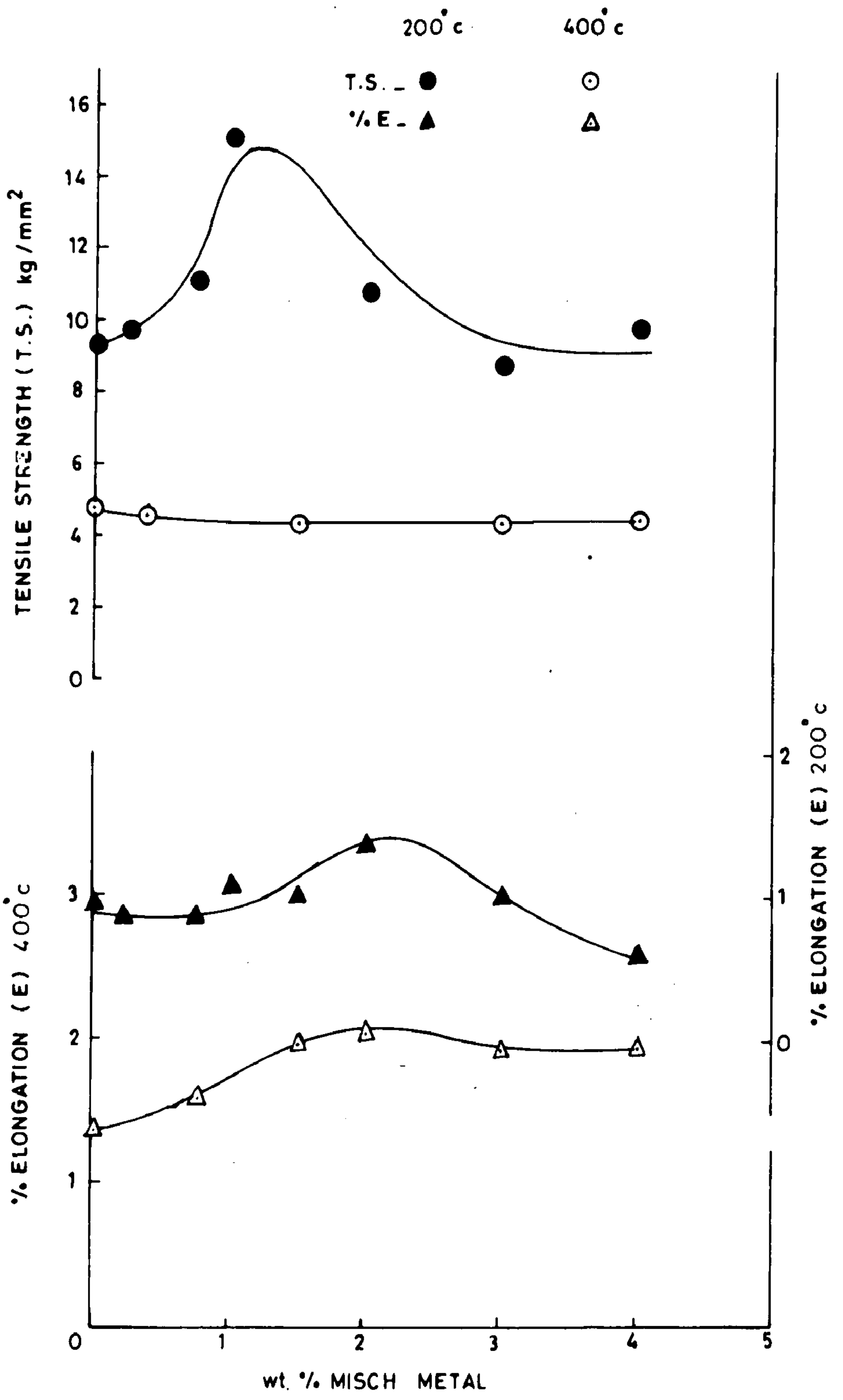

FIG.10.02_EFFECT OF MISCH METAL ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF Al - Si ALLOY

TABLE - 10.01

EFFECTS OF MISCH METAL ADDITIONS ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF ALUMINIUM

S. No.	% addition	200 °C		400 °C	
		Tensile strength kgm/mm2	% elongation	Tensile strength kgm/mm2	% elongation
1	0.00	8.05	10.0	5.09	3.5
2	0.20	8.14	9.1	3.88	3.5
3	0.50	-	-	-	-
4	0.75	7.42	3.5	2.91	8.1
5	1.00	7.28	3.5	2.82	8.1
6	1.50	-	-	2.64	7.1
7	2.00	-	2.0	2.47	-
8	2.50	5.12	1.5	2.47	4.0
9	3.00	2.98	2.0	2.24	2.5

TABLE - 10.02

EFFECTS OF MISCH METAL ADDITIONS ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF Al + 13% Si ALLOY (CHILL CAST)

S. No.	% addition	200° C		400° C	
		Tensile strength kgm/mm2	% elongation	Tensile strength kgm/mm2	% Elongation
1	0.00	9.38	0.95	4.78	1.00
2	0.20	9.74	0.85	-	-
3	0.75	11.09	0.80	4.59	1.60
4	1.00	15.08	1.05		
5	1.50	7.66	1.00	4.39	1.95
6	2.00	10.89	1.37		
7	3.00	8.77	1.00	4.38	1.95
8	4.00	9.93	0.60	4.57	1.95

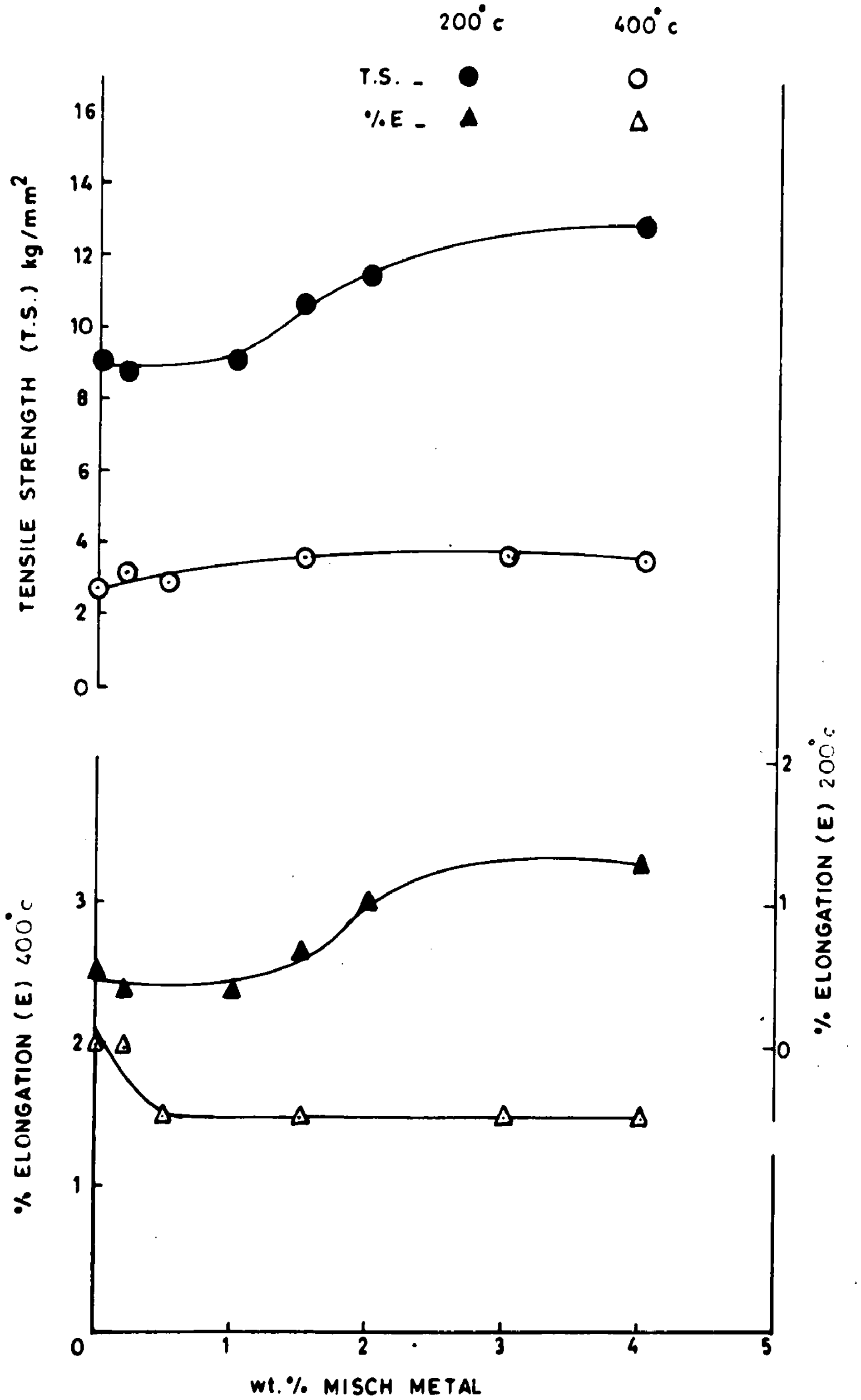

FIG.10.03_EFFECT OF MISCH METAL ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF Al + 8 % Mg ALLOY

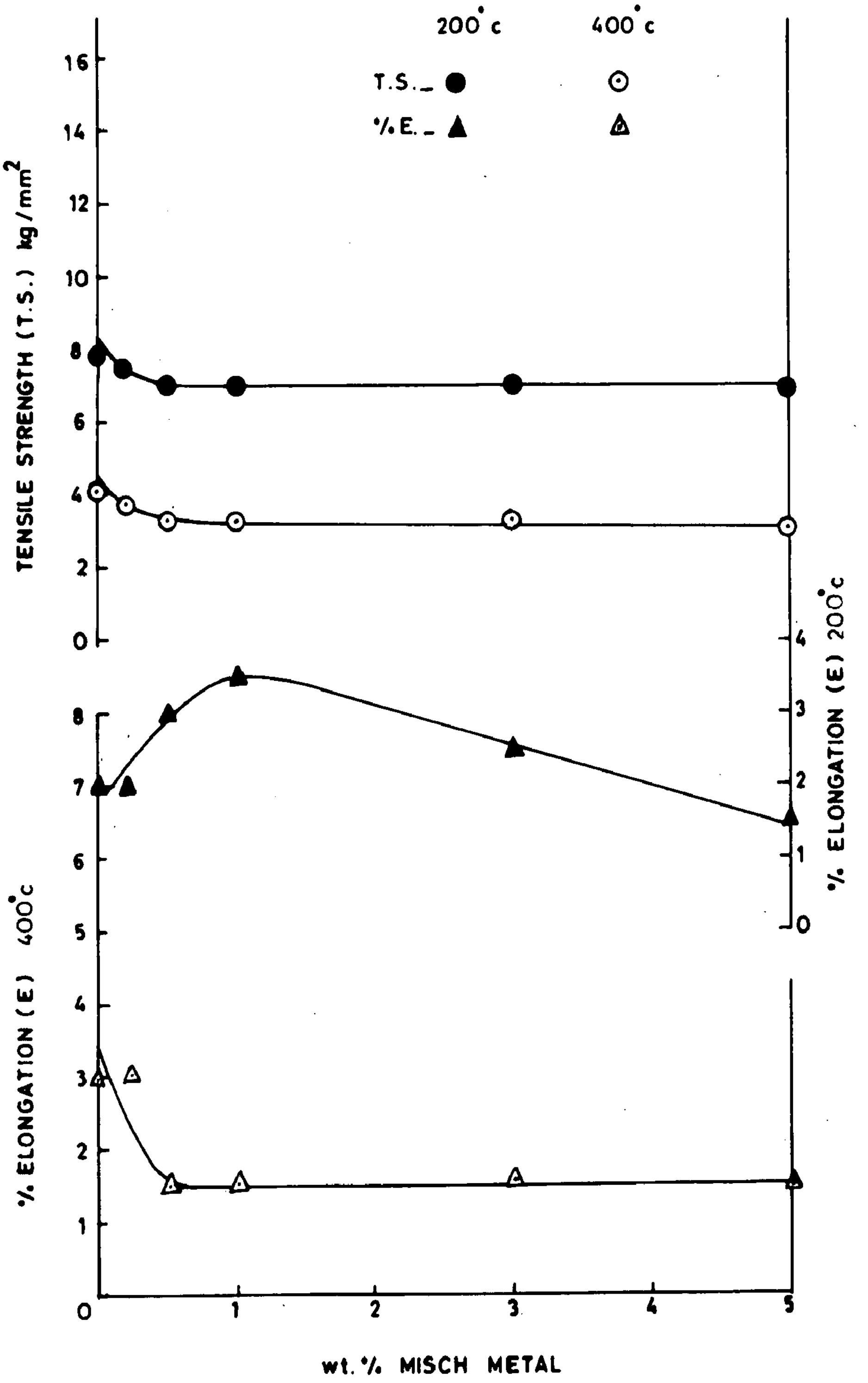

FIG.10.04_EFFECT OF MISCH METAL ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF ALLOY 122

TABLE - 10.03

EFFECTS OF MISCH METAL ADDITIONS ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF ALLOYS 218(8% Mg)

S. No.	% addition	200° C		400 ° C	
		Tensile strength kgm/mm²	% elongation	Tensile strength kgm/mm²	% elongation
1	0.0	9.11	0.51	2.71	2.00
2	0.2	8.80	0.41	3.06	1.98
3	0.5	-	-	2.85	1.51
4	1.0	9.14	0.38	-	-
5	1.5	10.72	0.64	3.51	1.51
6	2.0	11.48	1.00	-	-
7	3.0	-	-	3.74	1.51
8	4.0	12.89	1.27	3.60	1.51

TABLE - 10.04

EFFECTS OF MISCH METAL ADDITIONS ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF ALLOY 122(10%Cu & 0.2% Mg)

S. No.	% alloying element	200° C		400° C	
		Tensile strength kgm/mm2	% elongation	Tensile strength kgm/mm2	% elongation
1	0.0	7.90	2.0	4.16	3.0
2	0.2	7.60	2.0	3.82	3.0
3	0.5	7.00	3.0	3.30	1.3
4	1.0	7.00	3.5	3.26	1.5
5	2.0	7.00	2.5	3.26	1.6
6	4.0	-	-	-	-
7	5.0	6.90	1.5	3.07	1.5

10.3.3 Effects of misch metal addition to alloy 218 :

The results are given in table 10.03 and Fig. 10.03 .

10.3.4 Effects of Misch metal addition to alloy 122 :

The results are given in table 10.04 and Fig. 10.04.

10.4 DISCUSSION OF THE RESULTS :

10.4.1 Effects of misch metal addition to aluminium :

The salient features of the results are given below :

1) The tensile strength decreases at 200° and 400°C with the addition of misch metal. The rate of decrease is rather drastic at 200 °C. The % - elongation at 200 °C decreases to a constant value at about 1% addition of misch metal while at 400°C it increases upto 1% beyond which it falls down.

Hodge & Eastwood(24) did not find any significant effect of cerium addition on the elevated temperature properties of cast aluminium alloy. Our results, however, show a definite and significant decrease in tensile strengths at 200 ° and 400 °C with misch metal addition on aluminium, possibly due to the coagulation and growth of particles of the intermetallic compounds formed by the reaction of cerium, lanthanum and neodymium contained in misch metal with aluminium as discussed in previous chapters. Although no beneficial effect was obtained at 200° C on % elongation, it was found to increase upto 1% addition at 400 °C. No definite explanation can be given for the above effects. Probably this may be due to grain refinement resulting in the increase of grain boundary volume offering less resistance to plastic deformation as compared to grains at 400° C.

10.4.2 Effects of misch metal additions to aluminium-silicon alloys :-

The salient features of the results are given below:

(i) As at room temperature the tensile strength of Al-Si alloy at 200° C increases appreciably upto about 1 % addition of misch metal and thus decreases to a constant value. At 400° C the tensile strength remains almost unchanged with misch metal addition.

(ii) The % elongation increases at 200° as well as at 400° C upto about 2% addition. With further addition, it decreases appreciably at 200° while at 400° C it decreases slightly to a constant value at about 3.0% addition.

10.4.3 Effects of misch metal addition to alloy 218:

The salient features of the results are given below:

i) There is an appreciable increase in tensile strength of alloy 218 at elevated temperatures with misch metal additions. The increase was striking at 200° C upto about 3% addition.

ii) The % elongation at 200° C remains almost constant upto about 1% addition, beyond which it increases to a constant value at about 4% addition. The % elongation at 400° C, however, decreases to a constant value at about 0.5% addition.

As explained earlier with the addition of misch metal little of cerium and lanthanum will go into solid solution, most of them may react with aluminium and magnesium to form intermetallic compounds Al_4Ce, Al_4La, $CeMg_9$, $CeMg_3$, and $LaMg_3$, the amount of which should increase with increased addition. The above phases could be observed on metallographic study although the exact nature can be identified only by careful X-ray or electron diffraction studies. The large increase in strength at elevated temperature may be due to fine dispersion of some of these stable phases. Trehan and others (29) have already reported improvement in hot workability of the alloy with misch metal addition. Our investigations indicate larger improfement in machinability upto about 1.5% addition. Thus the alloy 218 treated with 1.5% misch metal is superior to that of the untreated alloy, from machinability, hot workability and high temperature strength considerations.

10.4.4 Effects of misch metal addition to alloy 122:

The salient features of the results are given below:-

i) The tensile strength at 200° C as well as at

400° C of alloy 122 remains almost unchanged with misch metal addition. A very slight decrease was observed upto about 1% addition, the tensile strength remaining almost constant thereafter, (Fig. 10.03).

ii) The % elongation at 200° C increases upto about 1% addition of misch metal beyond which it decreases, while at 400° C it decreases to a constant value at about 0.5% addition.

It is interesting to note that although large beneficial effects of misch metal have been obtained on the mechanical properties of alloy 122 at ordinary temperature, the tensile strength remains practically unchanged with misch metal addition at elevated temperatures. It appears that slight detrimental effect produced due to refining, balances the beneficial effects produced due to strengthening by the pressure of fine intermediate phases with a net result of constant value of strength at elevated temperatures.

-:o:-

$$a = 4.04711 + 0.140783 \times 10^{-3} T - 0.339197 \times 10^{-7} T^{2}$$

2. The values of lattice parameter at various temperatures as determined from equation are in close agreement with the experimental values.
3. Lattice spacing at 0 °C was found out to be 4.0471 Å and is in agreement with the results of Wilson(71) within ± 0.0001 Å
4. There is gradual decrease in density upto about 250 °C beyond which there is change in slope and density decreases at an appreciable rate.
5. Density determined at 20 °C was found out to be 2.6980 gm/cm³ and is in agreement with the previously reported values of Perryman and Taylor(142) within ± 0.0002 gm/cm³
6. Coefficient of thermal expansion ($\frac{1}{ao} \frac{da}{dt}$) as determined by equation decreases slightly with increase in temperature upto 548 °C. The relation is found out to be linear.

11.1.2 Aluminium-copper alloys :-

1. Lattice parameters of aluminium-copper solid solutions with varying compositions in equilibrium at different temperatures have been precisely determined using Nelson Riley's extrapolation method and are given in table 4.47.

2. The equation representing the relation between lattice parameter and temperature for each composition was determined with the help of computer. The values of lattice parameter as determined from equations are in close agreement with the experimentally determined values.

3. The lattice parameter is found to increase with increasing temperature for all the alloys. However, the curve shifts vertically downwards with increasing copper concentrations from 0 to 1.9 % and the values are lower than pure Al for all temperatures for 1.9% Cu alloy. The slope of the curve decreases for the lower temperatures and becomes parallel to the pure aluminium curve beyond 400 °C.

4. With increase in copper concentrations the same curve is obtained for lower temperatures. However, for higher temperatures the slope is found to decrease more appreciably in the case of alloy containing higher copper contents.

5. The density values of aluminium calculated at 525° and 548° C and the slopes of the curves increase with the increase in copper concentrations.

400^0 C of alloy 122 remains almost unchanged with misch metal addition. A very slight decrease was observed upto about 1% addition, the tensile strength remaining almost constant thereafter, (Fig. 10.03).

ii) The % elongation at 200^0 C increases upto about 1% addition of misch metal beyond which it decreases, while at 400^0 C it decreases to a constant value at about 0.5% addition.

It is interesting to note that although large beneficial effects of misch metal have been obtained on the mechanical properties of alloy 122 at ordinary temperature, the tensile strength remains practically unchanged with misch metal addition at elevated temperatures. It appears that slight detrimental effect produced due to refining, balances the beneficial effects produced due to strengthening by the pressure of fine intermediate phases with a net result of constant value of strength at elevated temperatures.

-:O:-

400° C of alloy 122 remains almost unchanged with misch metal addition. A very slight decrease was observed upto about 1% addition, the tensile strength remaining almost constant thereafter, (Fig. 10.03).

ii) The % elongation at 200° C increases upto about 1% addition of misch metal beyond which it decreases, while at 400° C it decreases to a constant value at about 0.5% addition.

It is interesting to note that although large beneficial effects of misch metal have been obtained on the mechanical properties of alloy 122 at ordinary temperature, the tensile strength remains practically unchanged with misch metal addition at elevated temperatures. It appears that slight detrimental effect produced due to refining,balances the beneficial effects produced due to strengthening by the pressure of fine intermediate phases with a net result of constant value of strength at elevated temperatures.

-:o:-

CHAPTER XI

CONCLUSIONS

11.1 HIGH TEMPERATURE X-RAY DIFFRACTION STUDY OF ALUMINIUM ALLOYS

11.1.1 Pure Aluminium :-

1. The lattice parameter is found to increase linearly with temperature. Following equation as determined by computer represents the relation between lattice-parameter and temperature :-

   $$a = 4.04711 + 0.140783 \times 10^{-3} T - 0.339197 \times 10^{-7} T^2$$

2. The values of lattice parameter at various temperatures as determined from equation are in close agreement with the experimental values.

3. Lattice spacing at 0 °C was found out to be 4.0471 Å and is in agreement with the results of Wilson (71) within ± 0.0001 Å

4. There is gradual decrease in density upto about 250 °C beyond which there is change in slope and density decreases at an appreciable rate.

5. Density determined at 20 °C was found out to be 2.6980 gm/cm^3 and is in agreement with the previously reported values of Perryman and Taylor (142) within ± 0.0002 gm/cm^3

6. Coefficient of thermal expansion ($\frac{1}{a_o} \frac{da}{dt}$) as determined by equation decreases slightly with increase in temperature upto 548 °C. The relation is found out to be linear.

11.1.2 Aluminium-copper alloys :-

1. Lattice parameters of aluminium-copper solid solutions with varying compositions in equilibrium at different temperatures have been precisely determined using Nelson Riley's extrapolation method and are given in table 4.47.
2. The equation representing the relation between lattice parameter and temperature for each composition was determined with the help of computer. The values of lattice parameter as determined from equations are in close agreement with the experimentally determined values.
3. The lattice parameter is found to increase with increasing temperature for all the alloys. However, the curve shifts vertically downwards with increasing copper concentrations from 0 to 1.9 % and the values ? are lower than pure Al for all temperatures for 1,9% Cu alloy. The slope of the curve decreases for the lower temperatures and becomes parallel to the pure aluminium curve beyond 400 °C.
4. With increase in copper concentrations the same curve is obtained for lower temperatures. However, for higher temperatures the slope is found to decrease more appreciably in the case of alloy containing higher copper contents.
5. The density values of aluminium calculated at 525° and 548° C and the slopes of the curves increase with the increase in copper concentrations.

6. Coefficients of thermal expansion at 525° and 548° C were calculated from equations with reference to lattice parameter at that temperature. The values are found to remain almost unchanged upto 4% copper, beyond which they decrease at an appreciable rate, and are in slight disagreement with the results of Ellwood and Silcock for lower concentrations, who have determined regular decrease in the values with varying copper concentrations. However, for higher concentrations ourfindings confirm their results.

7. **No detectable effect on coefficient of thermal** expansion is observed due to increase in temperature from 525° to 548° C.

8. **The values of lattice spacings in equilibrium at** different temperatures are found to decrease linearly with increase in copper concentrations in single phase region and become constant in the two phase region.

9. The solubility of copper in aluminium at a particular temperature was found out by determining the point of intersection of the inclined and horizontal portions of the lattice spacing vs. composition curve at that temperature. The solubility line determined between 475° and 548° C by the high temperature X-ray method is in agreement with the previous results determined by other methods.

10. The maximum solubility at the eutectic temperature has been found out by the two methods ;

    i) By finding out lattice parameter at 548°C with the help of equations in single and double phase regions and plotting the curve between lattice parameter and composition, then finding out the point beyond which the lattice parameter becomes unchanged.

ii) By plotting log solubility in atomic percentage against 1/Absolute temperature and extrapolating to that of the eutectic temperature.

The values determined by both the methods have been found out to be 5.7 wt. percentage confirming the results due to Raynor. The value is lower than the value reported by Ellwood and Silcock who have found out the maximum solubility to be 5.85 wt. percentage by applying method (ii) at the eutectic temperature.

11.1.3 Aluminium-silver alloys :-

1. The lattice parameter increases with increase in temperature for both aluminium + 7.5% silver and aluminium + 19.9 % silver alloys except that the values of the alloy containing 19.9% silver are lower than that of the alloy containing 7.5% silver for all temperatures.
2. The equations representing lattice parameter vs. temperature for each composition as determined by computer are given in table 4.45. The results of lattice spacings determined from equations are in agreement with the experimentally determined values.
3. With increase in silver concentrations the lattice parameter is found to decrease for all temperatures in single phase regions. The rate , however, de-

creases with the increase in temperature.

4. Our results are in agreement with the results of lattice spacings at room temperature of Hull and Axon (64), Sharan and others (144) and are not in agreement with the results of Gulizaev and Trusova (44) and Dorn (45) who have reported increase in lattice spacings with increase in silver concentrations at ordinary temperature.

5. Density increases and decreases almost linearly with increasing silver concentration and temperature respectively.

6. The curve density vs. composition shifts vertically downwards with increase in temperature, while the curve density vs. temperature shifts vertically upwards with increase in composition.

7. The coefficients of thermal expansion for the alloy containing 7.5% Ag is lower and for the alloy containing 19.9% Ag is higher than that of pure aluminium at a given temperature determined in single phase region. The value for 7.5% Ag alloy and 19.9% Ag alloy decreases and increases respectively with rise in temperature.

11.2 EFFECTS OF TANTALUM, RARE EARTH ELEMENTS AND MIXTURES OF RARE EARTH ELEMENTS ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM AT ORDINARY TEMPERATURE:

11.2.1 Effects of tantalum additions :

1. There is large refining effect of tantalum addition on the grain size upto 0.01% addition.

2. No beneficial effect is produced on the tensile strength and % elongation. Both are found to

why

decrease to constant values at about 0.05 and 0.025% respectively. Hardness , however, increases slightly by maximum of 21% at about .015% addition.

3. There is detrimental effect of tantalum addition on machinability from both chip formation and horsepower considerations.

11.2.2 Effects of cerium additions :-

1. The important effect of cerium addition is refinement of grain-structure. The optimum addition to give maximum refining effect is found out to be 1.5%.
2. No beneficial effect of additions is produced on the mechanical properties. Both tensile strength and % elongation decrease to constant values at about 0.5 and 1.5% additions respectively, while hardness remains unchanged for all additions.
3. Beneficial effect is produced on machinability. The optimum addition has been found out to be 0.5% to give the maximum increase by 30% in the shear angle Ø value and decrease in power consumption by about 8%.

11.2.3 Effects of lanthanum additions :

1. There is large refining effect upto 0.1% lanthanum addition. For higher additions there appears to to be the presence of some additional phase in the microstructure.

11.2.5 Effects of misch metal additions :-

1. The effect of misch metal is the refi[illegible]
   grain size beyond 1.5% [illegible] % elongation beyond 0.1%
   2-3% gives [illegible]
   [illegible]ition. There is increase in % elongation by about 20% of its initial value with the addition of 0.3% lanthanum.
3. Beneficial effect is produced by addition upto 0.05% on machinability as evaluated from both chip formation and power consumption criteria. There is maximum increase in shear plane angle Ø by 11% and decrease in horse power value by 7%.

11.2.4 Effects of neodymium additions :

1. There is little refining effect at about 0.2% addition. The effect is not so pronounced as in the case of cerium additions.
2. Detrimental effect is produced on tensile strength and hardness by additions upto 0.2 and 0.3% respectively. The % elongation, however, increases by maximum of 27% of its initial value with the addition of 0.2% neodymium.
3. Beneficial effect is produced on machinability as evaluated from both chip formation and power consumption criteria. There is maximum increase in Ø by 21% and maximum decrease in power consumption by about 27% at 0.3% and 0.1% additions respectively.

decrease to constant values at about 0.05 and 0.025% respectively. Hardness , however, increases slightly by maximum of 21% at about .015% addition.

3. There is detrimental effect of tantalum addition on machinability from both chip formation and horsepower considerations.

11.2.2 Effects of cerium additions :-

1. The important effect of cerium addition is refinement of grain-structure. The optimum addition to give maximum refining effect is found out to be 1.5%.
2. No beneficial effect of additions is produced on the mechanical properties. Both tensile strength and % elongation decrease to constant values at about 0.5 and 1.5% additions respectively, while hardness remains unchanged for all additions.
3. Beneficial effect is produced on machinability. The optimum addition has been found out to be 0.5% to give the maximum increase by 30% in the shear angle Ø value and decrease in power consumption by about 8%.

11.2.3 Effects of lanthanum additions :

1. There is large refining effect upto 0.1% lanthanum addition. For higher additions there appears to to be the presence of some additional phase in the microstructure.

2. The detrimental effect is produced on tensile strength upto 0.1% addition while there is only slight effect on the hardness. The beneficial effect is produced on % elongation beyond 0.1% addition. There is increase in % elongation by about 20% of its initial value with the addition of 0.3% lanthanum.
3. Beneficial effect is produced by addition upto 0.05% on machinability as evaluated from both chip formation and power consumption criteria. There is maximum increase in shear plane angle Ø by 11% and decrease in horse power value by 7%.

11.2.4 Effects of neodymium additions :

1. There is little refining effect at about 0.2% addition. The effect is not so pronounced as in the case of cerium additions.
2. Detrimental effect is produced on tensile strength and hardness by additions upto 0.2 and 0.3% respectively. The % elongation, however, increases by maximum of 27% of its initial value with the addition of 0.2% neodymium.
3. Beneficial effect is produced on machinability as evaluated from both chip formation and power consumption criteria. There is maximum increase in Ø by 21% and maximum decrease in power consumption by about 27% at 0.3% and 0.1% additions respectively.

11.2.5 Effects of misch metal additions :-

1. The effect of misch metal is the refinement of grain size beyond 1.5% addition. Addition between 2-3% gives a very fine grain size.
2. Slight detrimental effect is produced on tensile strength upto 1.0% addition. Beneficial effect is produced on the % elongation and hardness. There is maximum increase in % elongation by about 35% of its initial value at about 0.5% addition, while the hardness increases by about 74% at 3% addition.
3. There is large beneficial effect of misch metal addition on machinability, as indicated by about 2½ times increase in shear plane angle ϕ value , and decrease in power consumption by 27% at about 3% and 1% additions respectively.

11.2.6 Effects of lancer-amp 2 and rare earth fluorides additions:-

1. There is very little effect of lancer-amp 2 addition on the grain size, when large refining effect is produced by fluoride additions beyond 2%.
2. There is detrimental effect of both lancer-amp 2 and rare earth fluorides on tensile strength while large beneficial effect is produced on % elongation and hardness. The % elongation increases by maximum of about 100% and 65% with the additions of 1.5% rare earth fluorides and 0.5% lancer-amp^2 respectively.

The hardness, however, increases by maximum of 82% with the addition of 1.5% rare earth fluorides and by maximum of 65% with the addition of about 2% lanc~~h~~eramp 2.

11.3 EFFECTS OF ADDITIONS OF TANTALUM, RARE EARTH ELEMENTS AND MISCH METAL ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM MAGNESIUM ALLOYS AT ORDINARY TEMPERATURE :

11.3.1 Effects of tantalum additions :-

1. There appears to be large refining of grains with additions upto 0.15%.
2. Detrimental effect is produced on the tensile strength by additions upto 0.1%, while large beneficial effect has been obtained on % elongation which increases 4 times by optimum addition of 0.15%. The hardness however, slightly increases by maximum of about 12% with 0.3% addition.
3. Large beneficial effect has been produced on machinability. The value of ϕ increases by 18% and horse power decreases by about 18% with optimum addition of 0.075%.

11.3.2 Effects of cerium additions :

1. Very slight refinement occurs in the grain with the addition of cerium.
2. Slight detrimental effect is produced on tensile strength while little beneficial effect is established on % elongation and hardness which increase by about 12% and 8% with the optimum additions of

0.5% and 1.5% respectively.

3. Large beneficial effect is produced on machinability. The maximum increase in Ø by about 10% and decrease in horse power by about 27% with an optimum addition 0.5 %, have been obtained.

11.3.3 Effects of lanthanum additions :

1. There appears to be large refining of α grains upto 0.2% addition.
2. Detrimental effect is produced on tensile strength by addition upto 0.2% while hardness remains almost unchanged. There is large increase in % elongation by about 87% with optimum addition of 0.3%.
3. Beneficial effect on machinability has been produced by an optimum addition of 0.1%. There is increase in Ø by maximum of 20% and decrease in horse power by maximum of 17%.

11.3.4 Effects of neodymium additions :

1. There is slight refining effect of neodymium addition on the microstructure.
2. Detrimental effect is observed on tensile strength by additions upto 0.2% and the large beneficial effect on % elongation which increases by 3½ times the original value by an optimum addition of 0.1%. The hardness, however, remains unaffected upto 0.2% but increases by about 24% with an addition of 0.3% .

3. Beneficial effect of neodymium addition is established on machinability. There is an increase in the value of Ø by 24% with optimum addition of 0.2% and decrease in horse power by about 30% with an addition of 0.1%.

11.3.5 Effects of misch metal additions :-

1. With increase in misch metal addition there is gradual decrease in Al-Mg phase and corresponding increase in the new phase. There appears to be little refining action.
2. There is decrease in tensile strength upto 2% addition beyond which the beneficial effect is produced. Tensile strength increases by about 11% with the addition of 4% misch metal. The toughness remains constant while increase in hardness is observed by about 45% with an optimum addition of 1%.
3. Large beneficial effect is established on machinability. There is very large increase in shear plane angle Ø value by about 57% and decrease in horse power by about 21% with the optimum addition of 4% misch metal.

11.4 EFFECTS OF TANTALUM, RARE EARTH ELEMENTS AND MISCH METAL ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM COPPER ALLOYS AT ORDINARY TEMPERATURE:

11.4.1 Additions to alloy 195 (4.5% cu):-

11.4.1.1 Effects of tantalum additions :-

1. There appears to be very little effect on grain size.

2. Detrimental effect is obtained on tensile strength upto 0.3% addition, while there is slight but regular increase in hardness value. The % elongation increases to 2½ times its initial value, with an optimum addition of 0.1% .
3. Beneficial effect is produced upto about 0.1% addition. The shear plane angle ∅ increases by about 23% and horse power decreases by 6% with the same addition.

11.4.1.2 Effects of cerium additions :-

1. The effect of cerium addition is to refine the grain upto 0.75% beyond which some new phase starts forming.
2. The detrimental effect is produced on tensile strength. The hardness remains unchanged, The large beneficial effect is produced on % elongation beyond 0.75% addition, which increases to about double of its value at about 1.5% addition.
3. There is no effect of cerium addition on machinability as evaluated from chip formation and power consumption criteria.

11.4.1.3 Effects of lanthanum additions :-

1. There is little refining effect produced by lanthanum addition.
2. Detrimental effect is produced on tensile strength. The % elongation increases to 2½ times its initial value and hardness increases by 14% with 0.3% addition.

3. The machinability is found to improve with the addition of lanthanum from chip formation criteria. The value of Ø increases by about 35% with optimum addition of 0.1%. The horse power , however, remains unchanged.

11.4.1.4 Effects of neodymium additions :

1. For additions beyond 0.2% large refining effect in α grains is observed.
2. Detrimental effect is produced on tensile strength. The % elongation, however, increases to double of its initial value and hardness increases by about 30% with an optimum addition of 0.1%.
3. Slight detrimental effect is produced on machinability.

11.4.1.5 Effects of misch metal additions :-

1. There appears to be an appreciable refinement of α grains. For higher additions there appears to be the formation of some new phase in additiom to the normal phases.
2. Detrimental effect is produced on tensile strength and % elongation upto about 3% addition. The hardness , however, increases by about 23% with an optimum addition of 1% beyond which it remains constant.
3. Beneficial effect is produced on machinability. Ø increases by about 16% and horse-power decreases by about 5.5 % with the addition of about 0.3% misch metal.

11.4.2 Additions to alloy 122 :-

11.4.2.1 Effects of tantalum additions :-

1. There is pronounced refining action of tantalum addition upto 0.15% .
2. Large beneficial effect is produced on tensile strength which increases to about double of its value with an optimum addition of 0.05%. Detrimental effect is produced on % elongation and hardness.
3. There is slight improvement in machinability with tantalum additions. The shear plane angle Ø increases by 4% and horse power decreases by 25% with an addition between 0.1% and 0.2% .

11.4.2.2 Effects of cerium additions :-

1. There is pronounced refining action of cerium addition. For higher additions some new phase appears in addition to the normal phases.
2. Large beneficial effect of cerium addition on tensile strength is established with an optimum addition of 0.5%. The hardness increases very slightly upto 0.1% while % elongation decreases to constant value at about 0.5% addition.
3. Although very little effect on machinability is observed from chip formation criterian, there is definite decrease in power consumption by about 8% with an addition of 1.5% cerium.

11.4.2.3 Effects of lanthanum additions :-

1. There is large refining effect of lanthanum addition upto 0.2%.
2. There is large improvement in tensile strength by about 80% with an optimum addition of 0.1%. The detrimental effect is produced on % elongation upto 0.2% while hardness remains unchanged.
3. There is no detectable effect of lanthanum addition on machinability from both chip formation and power consumption considerations.

11.4.2.4 Effects of neodymium additions :

1. There appears to be little refining effect In the microstructure.
2. There is large increase in tensile strength by about 90% with the addition of 0.3%. Detrimental effect is produced on % elongation while hardness remains almost unchanged.
3. There appears to be no detectable effect of neodymium addition on machinability.

11.4.2.5 Effects of misch metal additions :

1. There is a pronounced refining action of addition of misch metal upto about 1.0%. For additions beyond 3% some new phase makes its appearance.
2. Beneficial effects have been observed on the mechanical properties. The tensile strength

increases by about 64% while the hardness and % elongation increase to double of their initial values with an optimum addition of 3.0% misch metal.

3. The improvement in machinability with misch metal addition is established. The shear plane angle Ø increases by about 9% and horse power decreases by about 10% with an optimum addition of 1% misch metal.

11.5 EFFECTS OF TANTALUM, RARE EARTH ELEMENTS AND MIXTURES OF RARE EARTH ELEMENTS ON THE MICROSTRUCTURE, MACHINABILITY AND MECHANICAL PROPERTIES OF ALUMINIUM - SILICON ALLOYS AT ORDINARY TEMPERATURE :

11.5.1 Effects of tantalum additions to hypereutectic aluminium-silicon alloy (11.9 % Si) -:-

1. The structure appears to be partially modified at 0.015% addition, With further additions the amount of silicon containing phase decreases and the new phase starts appearing.
2. Beneficial effect on tensile strength is produced by tantalum addition. There is increase in tensile strength by about 46% with an optimum addition of 0.025 %. There is little effect on % elongation and hardness values.
3. No beneficial effect is produced on machinability as evaluated from chip formation criterian although very little beneficial effect is observed from horsepower consideration, the minimum being obtained at about 0.025% addition.

11.5.2 Effects of cerium additions to hyper-eutectic aluminium-silicon alloys (11.9% Si) :-

1. The effect of cerium addition is to modify the structure. Maximum modification is observed at about 0.75% addition.

2. There is large increase in tensile strength by about 21% with an optimum addition of 0.75% cerium.

   The hardness increases by about 68% at an addition of 1%. No beneficial effect is produced on % elongation.

3. Large beneficial effect is produced on machinability. The shear plane angle Ø increases to about double of its initial value and horse power decreases by about 14% with an optimum addition of 0.1% cerium.

11.5.3 Effects of lanthanum additions to hyper-eutectic aluminium-silicon alloy (11.9% Si) :-

1. The structure appears to be partially modified at about 0.05% addition. For higher additions there appears to be the presence of some new phase in addition to eutectic and silicon plates.

2. The large beneficial effect is produced on mechanical properties. The tensile strength increases to about 2½ times its initial value and % elongation increases by about 50% with an optimum addition of 0.05%. The hardness , however, increases by about 54% with the addition of 0.15% .

3. Slight detrimental effect is produced on machinability with lanthanum additions.

11.5.4 Effects of neodymium additions to hyper-eutectic aluminium -silicon alloy (11.9% Si) :-

1. Partial modification takes place in microstructure with the addition of 0.2% neodymium.
2. Beneficial effect is produced on tensile strength beyond 0.1% addition. There is maximum increase by about 20 %at 0.3% addition. The hardness increases by about 52% with 0.2% addition. The % elongation remains almost unchanged.
3. No beneficial detectable effect is produced on machinability.

11.5.5 Effects of misch-metal additions on the hyper-eutectic aluminium-silicon alloy (13.0% Si):-

1. Complete modification is obtained at about 1% addition in both sand and metal mould cast alloys. Beyond 2% addition some new phase makes its appearance and seems to increase with further additions .
2. Beneficial effects of additions are produced on the mechanical properties. Tensile strength increases by about 77% and 50% with an optimum addition of 1% in both metal mould and sand cast alloys respectively. % elongation , however, increases by 60% with the same addition in the both the cases. Although little effect on hardness is produced in sand cast, it increases by 20% in chill cast with an addition of 5%.
3. There is large improvement in machinability with misch metal addition. The value of ϕ increases by 85% and horse power decreases by 14% with an optimum addition of 0.5%

11.5.6 Effects of lancer -amp^2 additions to hypereutectic aluminium-silicon alloy (13.0% Si) :-

1. Modification of the structure takes place at about 1.5% addition. For higher additions there appears to be presence of some new phase having uniform distribution.
2. There is an increase in tensile strength by about 40%, % elongation by 66% of the initial value and hardness by 15% with an optimum addition of 3.5% 2% and 3% respectively.

11.5.7 Effects of rare earth fluorides additions to hypereutectic aluminium silicon alloy (13.0% Si):-

1. Complete modification is brought about with 0.2% addition in chill cast, while only partial modification takes place in sand cast with the same addition.
2. There is increase in tensile strength by about 34% and 26% with the addition of 0.2% in sand cast and metal mould cast alloys respectively. Large beneficial effect on % elongation and hardness is produced in metal mould cast alloys with the addition of 0.2% and 2.5% additions respectively, while little effect is produced in sand cast.

11.5.8 Effects of rare earth fluorides additions to hypereutectic aluminium-silicon alloys(7.5% Si):-

1. With the addition of rare earth fluorides both eutectic and primary α get refined upto 0.2%.
2. There is increase in tensile strength by 36%, % elongation to three times its original value and

hardness by 27% with the additions of 0.2% 0.75% and 0.2% fluorides respectively.

11.5.9 Effects of misch metal and lancer-amp^2 additions to hypoeutectic aluminium-silicon alloy (9% Si) :-

1. There is gradual decrease in the amount of eutectic phase and increase in the size and amount of primary α.
2. Detrimental effect is produced on tensile strength and % elongation with lancer-amp^2 and misch metal addition. Very slight beneficial effect is produced on hardness beyond 1.5% upto 3%.

11.6 EFFECTS OF MISCH METAL ADDITIONS ON THE HIGH TEMPERATURE MECHANICAL PROPERTIES OF ALUMINIUM AND ITS ALLOYS :-

11.6.1 Effects of misch metal additions to aluminium:

1. Definite and significant decrease in tensile strength at 200° C and 400° C is obtained with misch metal additions.
2. Although the toughness at 200° C decreases, % elongation at 400° C increases to about 1½ times its initial value with the optimum addition of 0.75% misch metal.

11.6.2 Effects of misch metal addition to alloy 218 :

1. The tensile strength at 200° and 400°C increase by 40% and 36% respectively with 4% misch metal addition.

2. The % elongation at 200 °C increases to about 2½ times its initial value, while detrimental effect is produced at 400 °C.

11.6.3 Effects of misch metal addition to alloy 122 :

1. The tensile strength remains almost unchanged with misch metal additions at 200 °C and 400 °C.
2. Although there is slight increase in % elongation at 200 °C upto 1%, it decreases slightly at 400 °C with misch metal additions.

11.6.4 Effects of misch metal additions to aluminium silicon alloys :-

1. Tensile strength at 200 °C increases appreciably by about 67% with 1% misch metal addition, while it remains almost unchanged at 400 °C.
2. % elongation , however, increases by 44% and 39% at 200 °C and 400 °C respectively with misch metal addition.

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APPENDIX

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