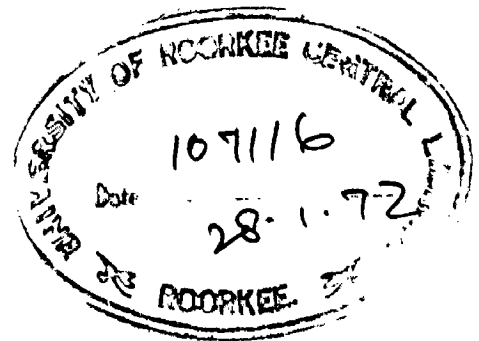# EFFECT OF TERNARY ADDITIONS ON THE AGEING CHARACTERISTICS OF SOME ALUMINIUM-ZINC ALLOYS

A Dissertation
submitted in partial fulfilment
of the requirements for the degree
of
MASTER OF ENGINEERING
in
METALLURGICAL ENGINEERING
[PHYSICAL METALLURGY]

BY

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C E R T I F I C A T E

Certified that the Dissertation entitled,
"EFFECT OF TERNARY ADDITIONS ON THE AGEING CHARAC-
TERISTICS OF SOME Al-Zn ALLOYS", which is being sub-
mitted by Mr. VINAY KUMAR GARG in partial fulfilment for
the award of the Degree of Master of Engineering in Meta-
llurgical Engineering [Physical Metallurgy] of the Uni-
versity of Roorkee, Roorkee(India), is a record of his own
work carried out by him under my supervision and guidance,
from 2nd January 1971 to October 11, 1971.

The matter embodied in this dissertation
has not been submitted for the award of any other degree.

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A B S T R A C T

The influence of homogenisation temperature, ageing temperature and solute concentration on the rate of pre-precipitation has been studied employing resis - tivity and hardness measurements. Whereas hardness mea- surement was restricted only to binary Al-Zn alloys, resistivity measurements were extended to two ternary additions of In also.

It has been found that both resistivity and hardness increase with ageing time and that the hardness curves are characterized by a well-defined incubation period while resistivity starts increasing almost instanta- neously. Time taken to reach the resistivity maxima and "flat hardness" are strongly dependent of the homogenisa- tion and ageing temperatures. Homogenisation temperature is found to have no influence on the resistivity peak and "flat hardness" while these are strongly dependent on the ageing temperature.

It is observed that addition of In to the binary alloys lowers the rate of pre-precipitation and higher the In content, slower is the ageing process.

A C K N O W L E D G E M E N T S

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CHAPTER - I

LITERATURE REVIEW

1.1 INTRODUCTION :

The phenomenon of age hardening, as is understood in its present form, is known to involve a precipitation reaction leading to the formation of a fully or partially coherent phase from a supersaturated solid solution and thus resulting in net increase in hardness. This technique has acquired a conspicuous position as one of the modern strengthening techniques because of the obvious advantage of possibility of heat-treating after the alloy has been worked to its final shape. Most important age hardenable alloys include Al-Zn, Al-Cu, Al-Mg and Cu-Be. The strength of such alloys can be increased by ageing to as much as five times.

Al-Zn system was chosen to be studied in the present investigation since it offers a wide range of solid solubility and hence constitutes one of the best alloy systems suited for age hardening.

1.2 HISTORICAL SURVEY :

Alfred Wilm[1,2] a German Scientist, in the years 1903-1911, discovered that Duralumin (an aluminium alloy) when quenched from high temperature below its melting point to low temperature and then allowed to stand at room temperature the hardness increased with time. It was later on found that the change in hardness could occur even at other temperatures but the course of hardening differed at different temperatures. This process has come to be known as "Ageing" and the change in hardness as Age Hardening.

Subsequently, some alloys other than duralumin were found to exhibit ageing and hence the phenomenon attracted the attention of many workers[3-7] due primarily to the fact that super high strength light alloys needed for many industrial applications could be developed. Since ageing is undoubtably followed by precipitation of a second phase from the parent matrix of supersaturated α, it came to be known as "Precipitation Hardening" also.

Age hardening consists of :

1. Homogenisation ; i.e., heating the alloy to a temperature where homogeneous solid solution is obtained,

2. quenching the alloy from solutionizing temperature to lower temperatures so that supersaturated solid solution is obtained, and

3. ageing ; holding the quenched specimen at a definite temperature for various lengths of time.

Prerequisits of age hardening :

Following are the prerequisits for an alloy system to be precipitation hardened :

1. The solubility of the solute must decrease with decreasing temperature in the solid solution or, in other words, the phase diagram of the alloy system should have a sloping solvus,

2. on quenching from solutionising temperature a super-saturated solid solution must form, and

3. on heating, this supersaturated solid solution must decompose into a non-equilibrium or inter-metallic phase which is fully or partially coherent with the matrix.

1.3 STRUCTURAL CHANGES DURING AGEING :

Complex structure changes occur in age harden-ing alloys during the process of ageing. A typical seq-uence of structural changes can be represented as :

Solid solution(at homogenisation temperature)

$\downarrow$

Quench : supersaturated solid solution

$\downarrow$

Excess vacancies $\longrightarrow$ Loops, Helics, Stacking faults

+

Solute atoms $\longrightarrow$ clusters $\longrightarrow$ zones

$\downarrow$

Intermediate precipitate $\longleftarrow$

$\downarrow$

Equilibrium precipitate
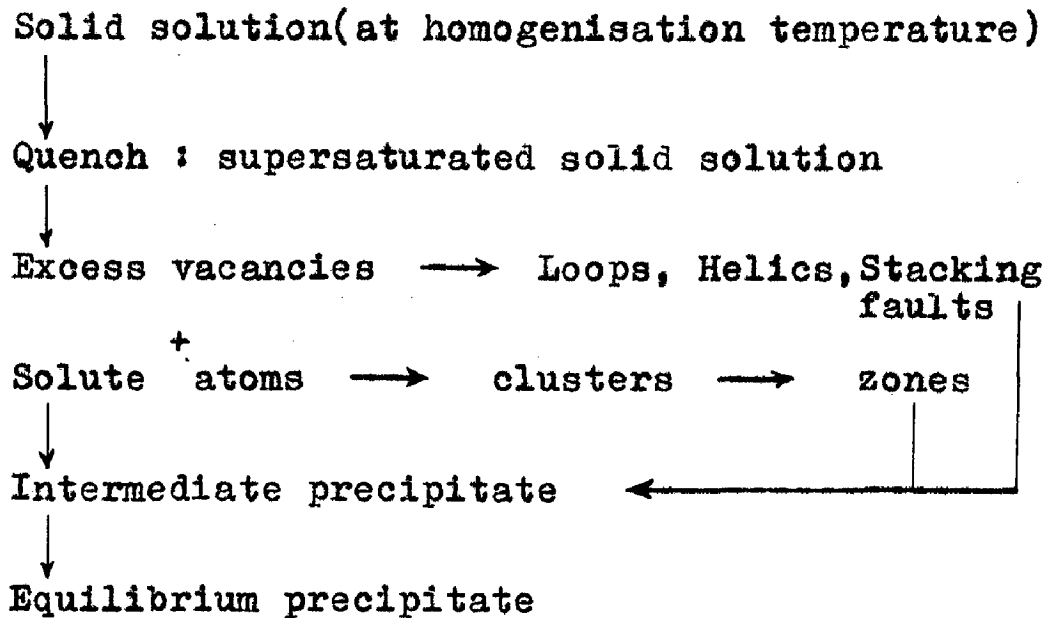
1.3.1 Defect structure :

At the homogenisation temperature, solute atoms and vacancies are randomly distributed in the substi-

tutional solid solution. On quenching, the alloy beco-

mes supersaturated with vacancies and solute atoms.

These excess vacancies give rise to various structural

defects which act as potential sites for nucleation of

metastable precipitate.

The defect structure introduced by quenching

has been studied in Al and Al-alloys by various authors.[8-10]

In aluminium, the microstructure consists of prismatic

dislocation loops about 500 A^o in diameter which are for-

med by the collapse and shear of discs of vacancies in

the manner predicted by Kuhlman-Wilsdorf.[11] In dilute

aluminium alloys excess vacancies are generally precipi-

tated as prismatic dislocation loops while in concentra-

ted alloys they form helical dislocations.

According to Kuhlman-Wilsdorf,[11] the excess

retained vacancies first form clusters or discs. Pris-

matic dislocation loops are then formed by collapse and

shear of these vacancy clusters or discs. The mechanism

of formation of helical dislocations has been suggested

by Cottrell[12] which is based on the absorption of vacancies

on a screw dislocation. Thomas and Whelan[13] have

investigated the mode of formation of helical disloca-

tions in quenched alloys and have suggested a mechanism

to explain the large number of helical dislocations found

in these alloys. According to them, a prismatic disloca-

tion loop may elongate along its glide cylinder to produce

two long screw dislocations. These screw dislocations

then transform to helical ones by the mechanism suggested

by Cottrell.[12]

Kelly and Nicholson[14] concluded that the

defect structure of a quenched dilute supersaturated alloy

is generally similar to that of pure metals although the

number of excess vacancies is higher due to the binding

energy between a solute atom and a vacancy. In concentra-

ted supersaturated alloys, formation of defects is diffi-

cult, and no defects other than helical dislocations and

a few isolated dislocation lines and prismatic loops have

yet been observed.

Rajan and Saxena[15] studied the effect of cold

working on precipitation hardening. They concluded that

the effect of cold working on precipitation, hardening is

a complex phenomenon involving interaction between vacan-
cies, dislocations, solute atoms, zones, intermediate
phases and the equilibrium precipitate.

1.3.2. Pre-precipitation :

During the early stage of ageing, solute atoms
segregate on certain preferential crystallographic planes
thus leading to the formation of solute rich clusters.
These clusters are mobile and grow at the expense of each
other forming G.P. zones. The clusters are completely
homogeneous with the matrix and so also the zones and
do not possess structures of their own. Zones, however,
have a definite shape depending upon the degree of mis-
match between the atomic diameters of solute and solvent.
Strictly speaking, if zones are left to itself, they have
their lattice parameter different from that of the parent
matrix. But since zones are homogeneous or, in other
words, have one to one correspondence with the matrix,
the overall zone is either compressed or expanded to
match its lattice parameter with that of the parent matrix.

Elastic strains are thus set up and there is an elastic interaction of the field due to dislocations and elastic strain set up due to the formation of zones. This offers resistance to the movement of dislocations and hence the hardness increases. The change in hardness eventually depends upon the difference in atomic diameter of solute and solvent. There is no nucleation barrier for clustering and zone formation. Guinier termed this phenomenon as pre-precipitation, to distinguish it from the true precipitation.

Kinetics of pre-precipitation :

Jagodzinski and Laves[16] first pointed out that Guinier's model[6] of G.P. zones in Al - Cu alloys required a very high diffusion coefficient, at least 10^7 times greater than normal diffusion coefficient obtained by extrapolation of high temperature measurements to room temperature. Two attempts have been made to explain this anomalously high solute diffusion coefficient in quenched alloys.

The first explanation proposed by Turnbull[17] assumed the diffusion of solute atoms to take place via dislocations. A large number of dislocations may act as channels for easier diffusion, but this theory failed to explain why the diffusion coefficient was so sensitive to quenching rate, homogenisation temperature etc.

Another plausible explanation proposed by Zener[18] and later developed by Federighi[19] and De Sorbo et al.[20], attributes this anomaly to quenched-in vacancies. The theory, came to be known as 'Excess Vacancy Theory', is widely accepted and assumes that high rate of diffusion is due to higher concentration of quenched-in vacancies.

The diffusion coefficient of Cu in Al at an ageing temperature T_A is given by:

$$D_{Cu} = A \exp \left[- \frac{E_F + E_M}{k\ T_A} \right] \qquad .. [1.1]$$

where E_F and E_M are the activation energies for the formation and migration of vacancies respectively, and A is a constant. At $20^{\circ}C$, assuming $E_F + E_M = 1.4$ eV, D_{Cu} was calculated to be 10^{-24} cm^2/sec. But De Sorbo et al.[20] calculated $D_{Cu} \sim 10^{-16}$ cm^2/sec by assuming

that Cu atoms move three or four atom diameters during clustering. Thus there is a discrepancy of $\sim 10^8$ between experimental and theoretical values. The excess vacancy theory attributes this discrepancy to the fact that diffusion is governed by a concentration of vacancies which would be in equilibrium at the homogenisation temperature T_H rather than at ageing temperature T_A. Thus :

$$D_{Cu} = A \exp \left[- \frac{E_F}{k \, T_H} \right] \exp \left[- \frac{E_M}{k \, T_A} \right]$$

.. [1.2]

Assuming $E_F = 0.76$ eV, $E_M = 0.58$ eV, $T_H = 550^\circ C$ and $T_A = 20^\circ C$, the value of D_{Cu} was calculated to be 5×10^{-16} cm^2/sec. which is in good agreement with the value estimated from experimental data. If the quenching is slow, some vacancies have time to migrate to sinks during quenching itself and thus the retained vacancy concentration will not be associated with T_H but with some lower temperature T_H'. The equation [1.2] can thus be modified as :

$$D_{Cu} = A \exp \left[- \frac{E_F}{k \, T_H'} \right] \exp \left[- \frac{E_M}{k \, T_A} \right]$$

.. [1.3]

where T_H' is the effective homogenisation temperature.

The excess vacancy theory is confirmed by the following experimental observations employing resistivity measurement :

1. With increase in T_H, the initial resistivity and also the initial rise in resistivity increases.

2. Ageing temperature remaining constant, time taken to reach peak is lesser for higher T_H.

1.3.3   Precipitation :   Intermediate and Equilibrium :

On continued ageing the G.P. zones are gradually replaced by a metastable phase having definite crystal structure and lattice parameter.   This phase termed as intermediate precipitate is partially or fully coherent with the matrix.

As the intermediate precipitate grows, the strain fields enlarge till they overlap.   Now on further growth of precipitate, the strains become too great to be accommodated merely by bending the lattice planes.   At this stage coherency breaks down and the strains at the interface are now accommodated by the formation of structural dislocations.

This is the stage where intermediate precipitate transforms to the equilibrium precipitate and since it is followed by disappearance of the coherency strain, softening results. Hence we get a peak after the hardness maxima in the hardness vs. ageing time curve.

1.4   Al-Zn   SYSTEM :

Zn has an extensive solid solubility in Al which ranges from 16 at. pct. Zn at 275°C to 1.5 at. pct. at 100°C. Thus, Al-Zn alloys are very favourable for the study of precipitation reaction over a wide range of compositions. The Al-Zn alloy system has attracted considerable attention in recent years as precipitation sequence has been found to be more complicated than thought of earlier. Existence of spinodal transformation has also been reported in some alloys.[21-24] The Al-Zn phase diagram is shown in Fig.1.1. A short account of the ageing characteristics of Al-Zn alloys is being given in the following sections :
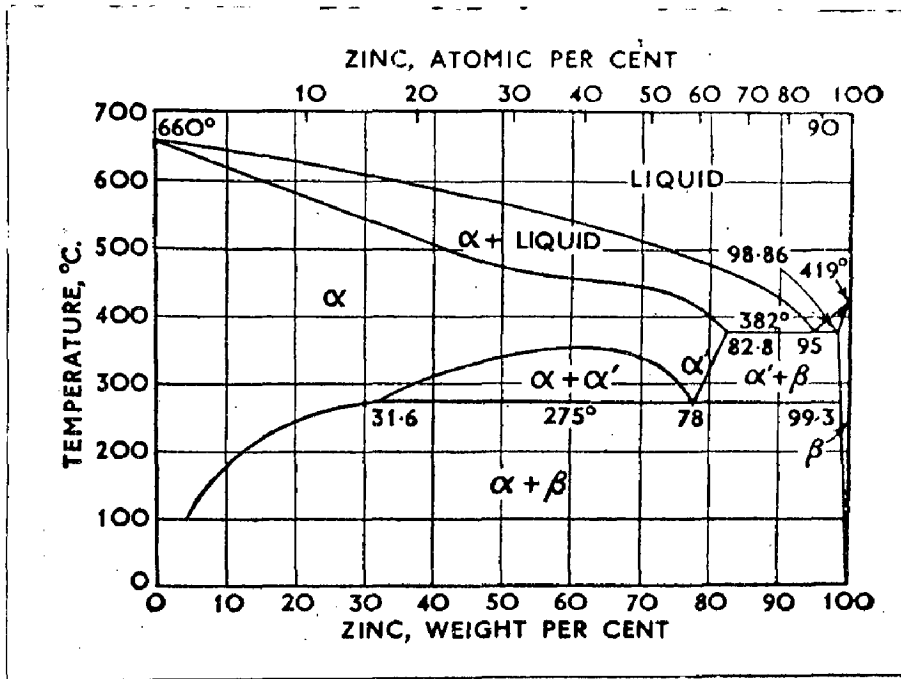
ZINC, ATOMIC PER CENT
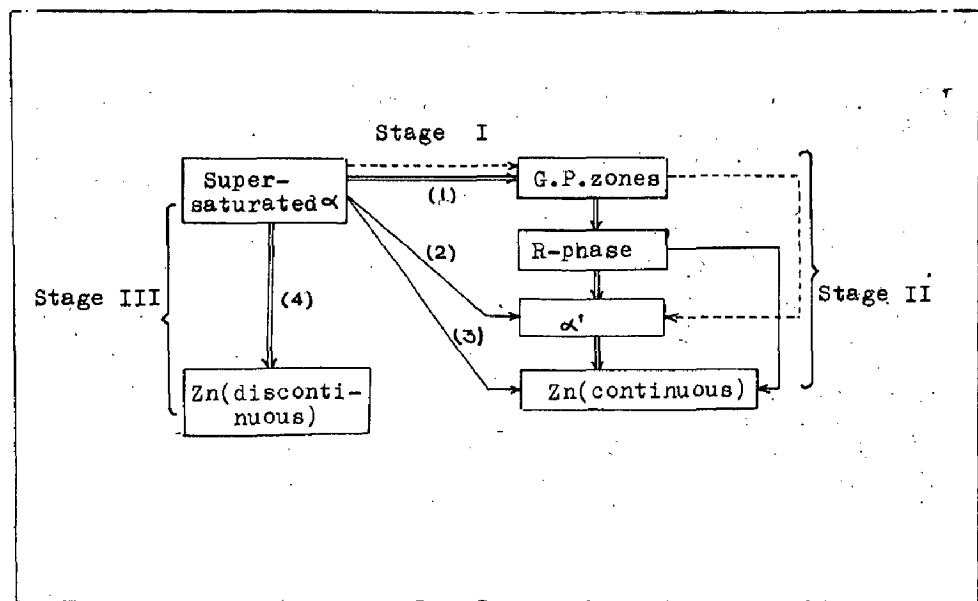
FIG. 1.1   Al - Zn phase diagram[41]

FIG. 1.2   Mode of precipitation during ageing
in Al -   Zn alloys[30]

1.4.1   Sequence of precipitation :

Till recently the precipitation sequence in

Al-Zn alloys was accepted to be quite simple :

Supersaturated solid solution ⟶ spherical

G.P. zones ⟶ equilibrium precipitate.

Garwood et al[25] for the first time detected

a face-centred cubic transition precipitate in an Al-12.5

at. per cent Zn alloy aged at 200°C. Since then, inves-

tigations by various workers[22,26-29] have established

the existence of another phase (called R phase) in the

precipitation sequence of Al-Zn alloys.

The effect of solute concentration and ageing

temperature on the sequence of precipitation is, however,

not still clearly known. Reviewing various experimental

data, Krishna Rao et al[30] have summarised the mode of

precipitation in Al-Zn alloys as indicated in Fig. 1.2.

According to them,

1)   zones form by normal solute clustering or by spinodal

decomposition. These zones are homogeneously distri-

buted. Zone composition is controlled by metastable

miscibility gap in two phase region, but its size

and thus volume increases which introduces a large

amount of strain, which leads to the formation of
R-phase and then following sequence takes place :

$$R\text{-phase} \longrightarrow \alpha' \longrightarrow Zn \text{ (continuous)},$$

2) heterogeneous formation of α' takes place in small
amounts. It probably occurs within the grains at
dislocations, normally present or arising due to
vacancy aggregates,

3) continuous Zn precipitation by heterogeneous
nucleation on structural defects is not very signi-
ficant as it will require a great deal of solute to
be localised and again surmounting of a large free
energy barrier for nucleation,

4) discontinuous precipitation of Zn is the common
mechanism for high Zn alloys. It becomes more
prominent with decreasing ageing temperature and
increasing solute concentration.

1.4.2   Effect of ternary addition :

Since binding energy between a Zn atom and
vacancy is small, Al-Zn system happens to be most suit-
able one for studying effect of trace additions of a
third element.[31,32]

As has already been pointed out in this section, the clustering rate in Al-Zn system is very fast. This clustering rate, however, can be hampered appreciably by trace additions of a third element. Noting this decreased clustering rate, the binding energy of the third element with vacancies can be evaluated.

Silcock[33] and Kimura and Hasiguti[34] have made an alternative explanation for this decreased clustering rate that the trace element atoms might get paired-off with the excess vacancies and thereby reduce the rate of clustering and zone formation. This explanation seems to be quantitatively plausible since the concentrations of vacancies and trace element atoms are similar. Kimura and Hasiguti conducted resistivity measurements on Al-1.7 pct. Cu - 0.006 pct. Sn alloy and found that binding energy of a Sn atom is 0.2 eV greater than between a Cu atom and a vacancy. Hence majority of vacancies are bound to Sn atoms and the rate of diffusion of Cu atoms to form G.P. zones is rather slow.

1.5 SCOPE OF THE PRESENT INVESTIGATION :

The present work was undertaken to study the effect of Indium additions on the kinetics of pre-precipitation of Al-Zn binary alloys and thereby calculate the energies of formation and migration between a Zn atom and a vacancy and that between an In atom and a vacancy. It was therefore decided to take three compositions of the binary alloy and subsequently two additions of In were added to the same binary alloys. Values of E_F and E_M were calculated for the binary and ternary alloys using electrical resistivity measurements.

Efforts have been made to explain the change in kinetics of pre-precipitation brought about by the additions of In, in terms of the positive binding energy between an In atom and a vacancy.

Hardness measurement was also employed for the binary alloys to make the resistivity results more authentic.

CHAPTER - II

EXPERIMENTAL PART

2.1 ALLOY PREPARATION :

The binary and ternary alloys were prepared from high purity aluminium (99.99 pct. purity) supplied by Riedel-De Haen, Germany, zinc of 99.9 pct. purity and indium of 99.999 pct. purity.

Weighed amount of aluminium was melted in a clean graphite crucible in a muffle furnace and zinc in required quantity was added to the molten aluminium. To account for the loss of zinc during melting, 15 pct. excess zinc was taken. The melt was thoroughly stirred to ensure complete homogenisation of the melt and hexa-chloroethane was used as degasant before pouring the melt into cylindrical mild steel moulds and allowed to cool. The alloy ingots thus obtained were 7/8 in. in diameter and 6 in. long. The ingots were then hot forged at $450^{\circ}C$ to break up the cast structure.

Ternary alloys were made by remelting the scrap of the binary ones and adding to that the required quantity of pure indium. Two ternary compositions corresponding to each binary alloys were made.

Binary alloys were chemically analysed for zinc content. Aluminium content of the alloys was determined by difference. The table 2.1 gives the final composition of the various alloys .

2.2   SAMPLE PREPARATION :

Wire samples of 0.63 mm diameter for the binary and 0.67 mm diameter for the ternary alloys were made use of in the resistivity measure-ment. Wires, three meters in length, for the binary and one meter for that of ternary alloys, were wound like a lamp coil filament.

The samples employed for hardness measurement were cylindrical type, 20 mm in diameter and 8 mm in height. They were cut from the cast and hot forged cylindrical ingot (7/8" x 6"). The specimens were ground and polished, and used for hardness measurement.

TABLE 2.1 : Compositions of the various binary
and ternary alloys.

Alloy	Zn wt. pct.	In wt. pct.	Al wt. pct.
Binary			
D	8.1	-	Rest
E	9.5	-	..
F	12.1	-	..
Ternary - 1			
D1	8.1	0.05	..
E1	9.5	0.05	..
F1	12.1	0.05	..
Ternary - 2			
D2	8.1	0.15	..
E2	9.5	0.15	..
F2	12.1	0.15	..

2.3   HEAT   TREATMENT :

Initial homogenisation treatment was given to all the samples so as to give a single phase supersaturated α at room temperature on quenching from the homogenisation temperature. Wire samples were homogenised for 8 hours at 480°C in a muffle furnace and quenched down to room temperature. Hardness specimens were homogenised for 60 hours at 500°C in a muffle furnace. The specimens were then quenched down to room temperature and were ensured to be homogeneous as revealed by optical microscopy as well as by hardness measurement.

2.4   RESISTIVITY   MEASUREMENT :

Resistivity of the wire samples was measured potentiometrically with a least count of 0.01 mV. Specimens were homogenised in a muffle furnace, the temperature of which was controlled within ± 2°C . Quenching operations were carried out by manual extraction of samples from the furnace and by rapid immersion in the quenching media (ice water at 0°C) ; then samples were immediately transferred to the ageing bath.

Resistivity measurements carried out for Binary,

Ternary-1 and Ternary-2 alloys were made in two ways :

1. Varying T_H from 280°C to 500°C in steps of 20

degrees and keeping T_A constant (0°C).

2. Varying T_A from 0°C to 55°C in steps of 5 degrees

and keeping T_H constant (300°C).

The variations in resistivity were computed by

the relation :

$$\Delta \rho \quad = \quad \Delta R \cdot \frac{A}{L}$$

where $\Delta R = R - R_0$ is the variation in resistance

of the sample, with reference to R_0, when it has reached -

by ageing - the value R, A is the area of cross section

of the wire sample and L is the length of the sample

employed.

2.5 HARDNESS MEASUREMENT :

It was confined only to the Binary alloys.

Homogenisation was done in a vertical tube furnace fitted

with a Mahindra Electroflo temperature controller having

a sensitivity of $\pm$ 2°C. The samples were rapidly quen-

ched in the cooling medium (ice water at 0°C) and then

immediately transferred to the ageing bath. The temperature of the ageing bath was controlled by means of a thermostat within an accuracy of $\pm$ 0.2°C. All the hardness measurements were done on a Vicker's hardness tester employing a load of 5 kgs.

Similar to resistivity, hardness measurements were done in two ways :

1. Varying T_H from 280°C to 500°C in steps of 20 degrees and keeping T_A constant (50°C).

2. Varying T_A from 30°C to 85°C in steps of 5 degrees and keeping T_H constant (300°C).

2.6 CALCULATION OF E_F AND E_M :

Arrhenius equation governing the rate of diffusion of Zn atoms in Al is given by :

$$D_{Zn} = D_0 \exp \left[- \frac{E_F}{k \, T_H} \right] \exp \left[- \frac{E_M}{k \, T_A} \right] \quad .. \quad [2.1]$$

where, D_{Zn} is the diffusion coefficient of Zn, D_0 is a constant, E_F and E_M are the activation energies of formation and migration of vacancies and T_H and T_A are the homogenisation and ageing temperatures respectively.

At constant ageing temperature :

$$\ln D_{zn} = C_1 - \frac{E_F}{k\, T_H} \qquad \qquad .. [2.2]$$

Now, since D_{Zn} is proportional to the diffusion rate

which inturn is inversely proportional to the ageing time,

t, the equation [2.2] may be written as :

$$\ln \frac{1}{t} = C_2 - \frac{E_F}{k\, T_H} \qquad \qquad .. [2.3]$$

Thus, for a temperature T_{H_1} and time t_1,

$$\ln \frac{1}{t_1} = C_2 - \frac{E_F}{k\, T_{H_1}} \qquad \qquad .. [2.4]$$

and at a temperature T_{H_2} and time t_2,

$$\ln \frac{1}{t_2} = C_2 - \frac{E_F}{k\, T_{H_2}} \qquad \qquad .. [2.5]$$

Subetracting equations [2.4] and [2.5] we get

$$\ln \frac{t_2}{t_1} = \frac{E_F}{k} \left[\frac{1}{T_{H_2}} - \frac{1}{T_{H_1}} \right] \qquad .. [2.6]$$

or

$$\log \frac{t_2}{t_1} = \frac{E_F}{2.303\, k} \left[\frac{1}{T_{H_2}} - \frac{1}{T_{H_1}} \right] \qquad .. [2.7]$$

Thus, slope of the line plotted between log t

and $\frac{1}{T_H}$ and equating it to $\frac{E_F}{2.303\, k}$, the value of E_F

can be evaluated.

Similarly, at constant T_H, values of E_M can be evaluated by equating the slope of the line drawn between $\log t$ and $\dfrac{1}{T_A}$ to $\dfrac{E_M}{2.303\ k}$.

In the present investigation, values of E_F and E_M have been calculated for binary and ternary alloys by means of various Arrhenius plots. Slope of the lines was calculated by Least Square Method. All the calculations were done on an IBM 1620 computer. Listing of the computer programs calculating various variables are given in the appendix.

CHAPTER - III

RESULTS AND DISCUSSION

3.1 RESULTS :

The variation in resistivity and hardness as a function of ageing time for binary andternary alloys at different T_H and T_A have been shown from fig.3.1 to 3.72. Figs. 3.1 to 3.27 indicate resistivity change with ageing time for the Binary, Ternary-1 and Ternary-2 alloys with varying T_H while those from fig.3.28 to 3.54 for the corresponding alloys with T_A varying. Hardness plots for the Binary alloys with T_H and T_A varying are shown in Figs. 3.55 to 3.72.

On the basis of these graphs, Arrhenius plots as shown in figs. 3.73 to 3.108 have been derived. Values of E_F and E_M for the binary and ternary alloys have been calculated and also the values of binding energy between a Zn atom and a vacancy and that between an In atom and a vacancy arrived at. All these results are discussed later in this chapter.
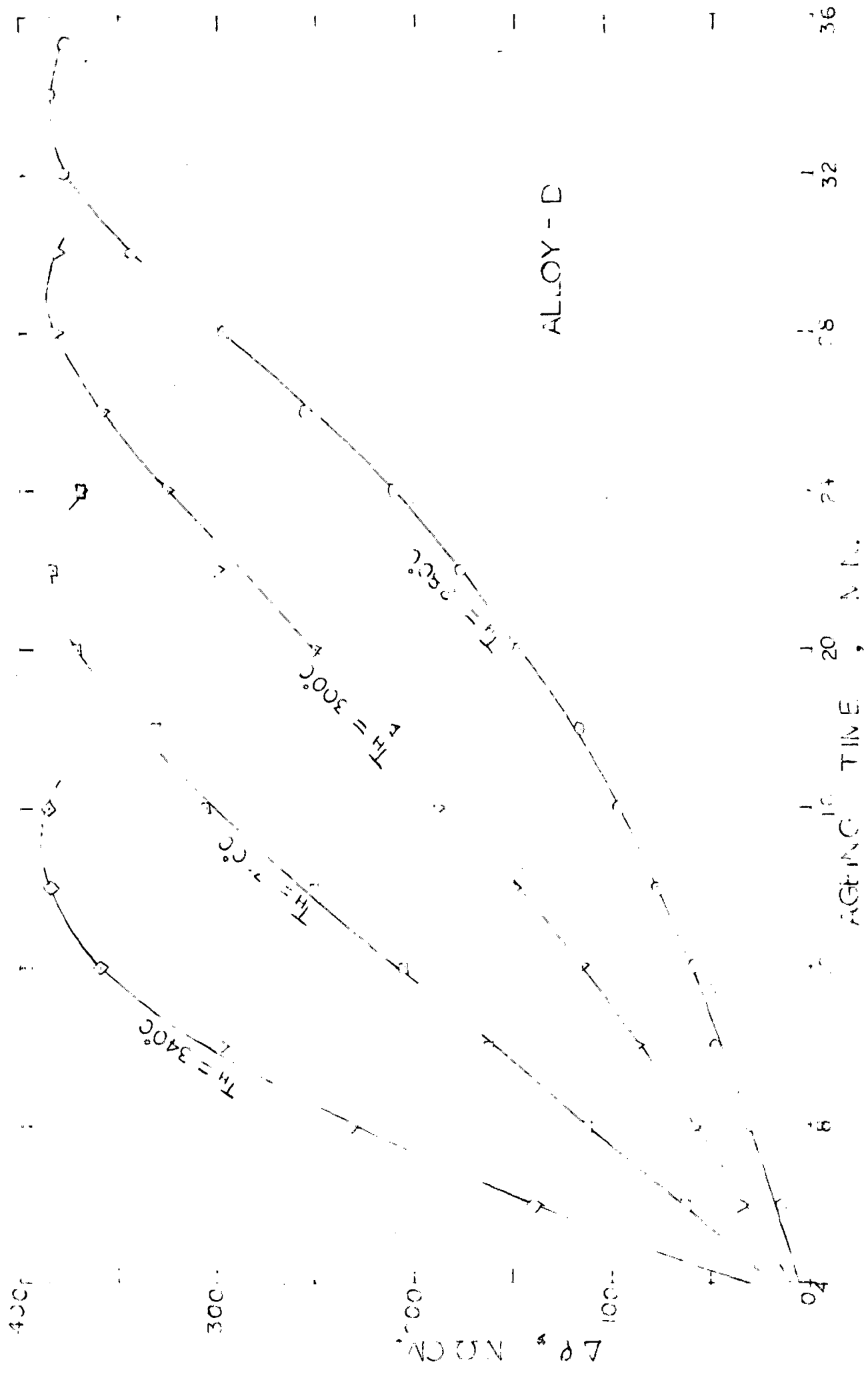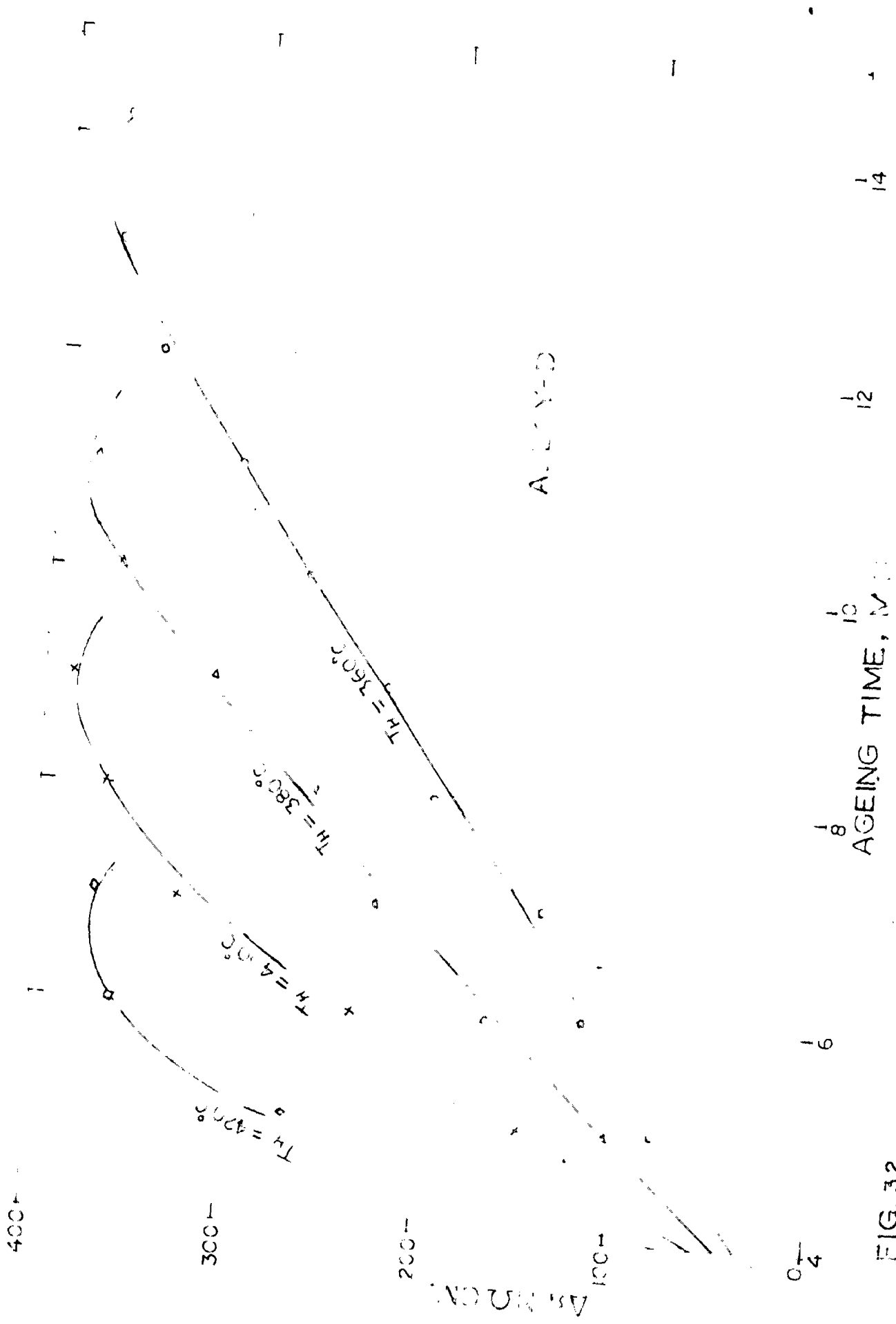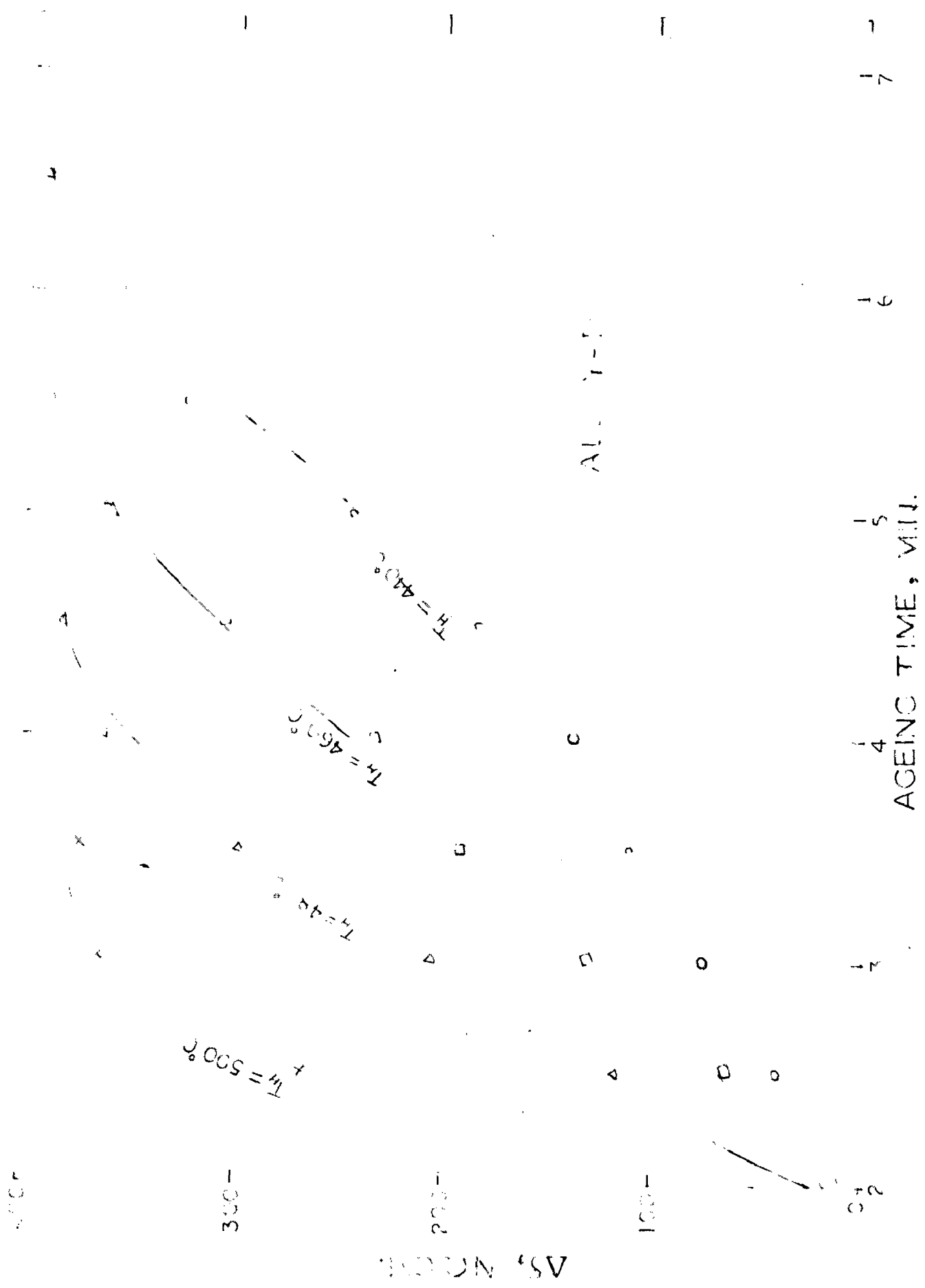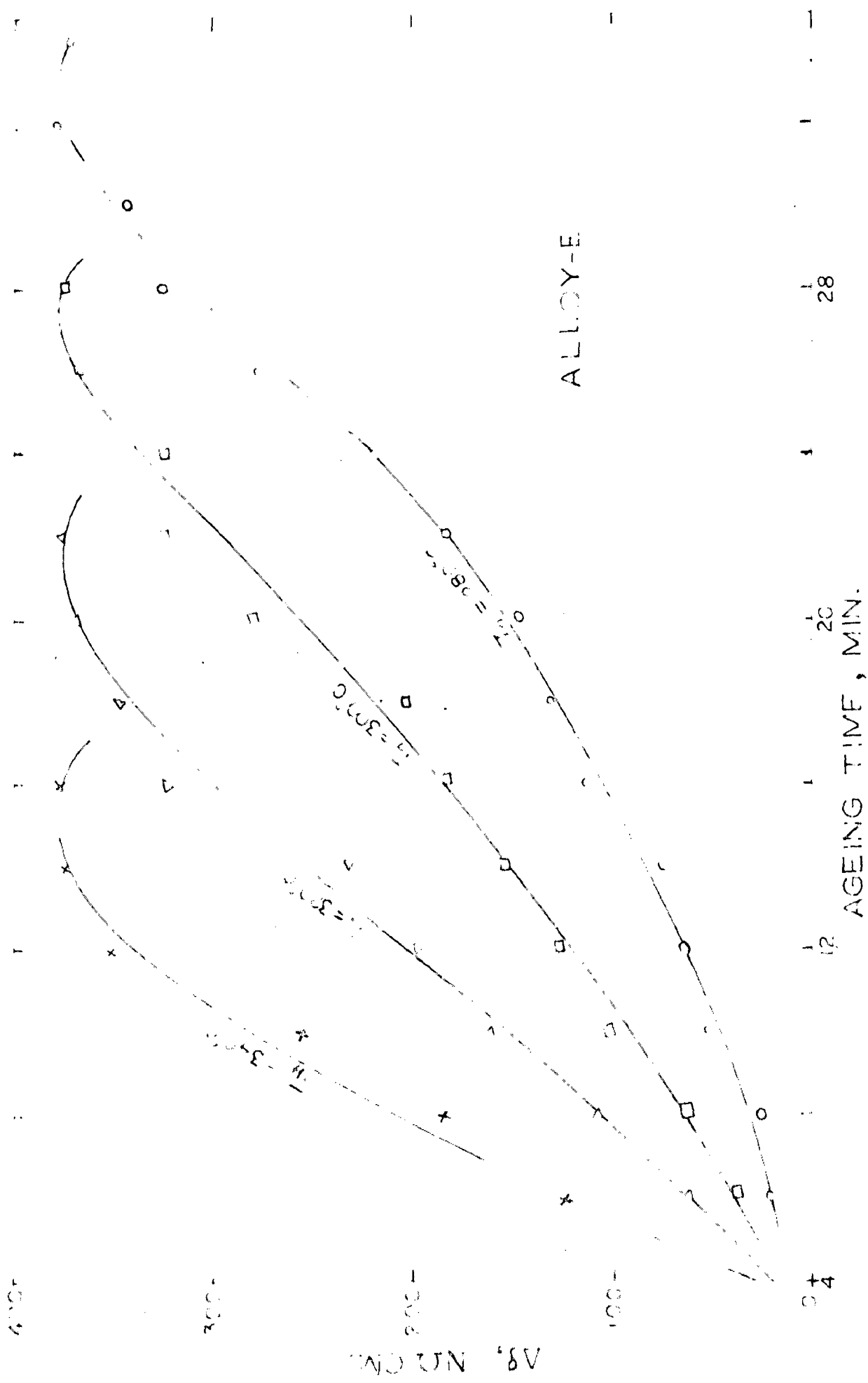
FIG. 3.1

FIG. 3.2

AGEING TIME, MIN

As-Q.N.CH.

100

200

300

400

0

100

200

300

400

4

6

8

10

12

14

TH = 420°C

TH = 400°C

TH = 380°C

TH = 360°C

As-Q.N.CH.

FIG. 3.3

FIG. 3.

ALLOY-E

AGEING TIME, MIN.

Ag, N/CM.

FIG. 3.5

FIG. 3.1

ALLOY-D

$T_H = 360°C$

$T_H = 380°C$

$T_H = 400°C$

$T_H = 420°C$

Δ3, NΩCM.

AGEING TIME, MIN.

FIG. 3.2

ALLOY-E

$T_H = 280°C$

$T_H = 300°C$

$T_H = 320°C$

$T_H = 340°C$

Δg, $\mu\Omega$ CM.

AGEING TIME, MIN.

FIG. 3.4

ALLOY-E

$T_H = 440°C$

$T_H = 460°C$

$T_H = 480°C$

$T_H = 500°C$

$\Delta \rho$, $\mu\Omega$.CM.

AGEING TIME, MIN.

FIG. 3.6

ALLOY - F

$T_H = 280°C$

$T_H = 300°C$

$T_H = 320°C$

$T_H = 340°C$

AGEING TIME, MIN.

ΔR, $\mu\Omega$ CM.

FIG. 3.7

FIG. 3.8

Alloy-F; ageing curves showing $\Delta\rho$ (nΩ.cm) versus ageing time (min) at ageing temperatures $Th = 300^\circ C$, $Th = 380^\circ C$, $Th = 400^\circ C$, and $Th = 420^\circ C$.

ALLOY-F

$\Delta \rho,\ \mu \Omega\ CM$

AGEING TIME, MIN.

$T_H = 440°C$
$T_H = 460°C$
$T_H = 480°C$
$T_H = 500°C$

FIG. 3.9

FIG. 3.10

FIG. 3.11

ALLOY - D1