# THERMO MECHANICAL TREATMENT OF TITANIUM-MOLYBDENUM ALLOYS

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*By*

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# CERTIFICATE

Certified that the dissertation entitled "THERMOMECHANICAL TREATMENT OF TITANIUM-MOLYBDENUM ALLOYS", which is being submitted by Shri Mahendra Kumar Mittal in partial fulfilment for the award of Degree of Master of Engineering in Metallurgical Engineering (Physical Metallurgy) of the University of Roorkee, Roorkee, is a record of his own work carried out by him under my supervision and guidance. The matter embodied in this dissertation has not been submitted for award of any degree or diploma.

This is further to certify that he has worked for a period of eight months from January 1972 to August 1972 for preparing this dissertation at this University.

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# ABSTRACT

In the present investigation thermomechanical treatment comprised of cold working of Ti-Mo alloys in the quenched condition from solution treating temperature followed by subsequent aging treatment at 400° C and 500°C respectively. Two alloys Ti-10% Mo and Ti-23% Mo are studied with the help of hardness measurements and optical microscopy to observe the mechanical properties as well as the structural changes during this treatment. It was observed that the peak hardness value of either alloy at 500° C aging temperature is higher than the corresponding value at 400° C. For Ti-23% Mo alloy it is found that there is no variation in the period of hardness peak on aging at either temperature. However, the peak hardness values of Ti-10% Mo alloy are found less than that of Ti-23% Mo alloy for either aging temperature. It is also observed that with increase in prior cold deformation, there is an increase in the hardness peak value upto 50 per cent deformation for Ti-10% Mo alloy but for Ti-23% Mo alloy the prior cold deformation upto 30 per cent is

effective to increase the peak hardness value. The basis of strengthening effect of thermal treatment is the appearance and growth of omega phase, stress induced martensite and the retained beta phase. During this study the presence of omega phase has not experimentally been observed.

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# CHAPTER 1

# INTRODUCTION

One of the principal needs leading to the development of new alloys is the desire to operate at higher temperatures than is possible with existing materials. This trend is illustrated by the substitution of titanium for aluminium alloys in the cooler, parts of gas turbine engines, of Zirconium for magnesium in nuclear reactors and, potentially, by the development of a new material to replace nickel-base alloys in the hotter parts of the turbine. New high temperature alloys are also needed to make possible many rocket and space developments.

The nature of the alloy development problem, and the choice of starting materials, obviously depends on the application concerned, but in each case it is a question of combining desired mechanical properties with some other characteristics. For titanium alloys the need has been to develop maximum strength in the temperature range 0-500°C while maintaining the density of the material at the lowest possible level. Success already achieved in this field has moved the critical temperature range to 500-600°C and has got subsequently added oxidation resistance to the specification which must be met by new titanium alloys.

For titanium extensive studies have shown that best creep strength at temperatures upto 500°C is given by alloys in which the major proportion of the alloying elements stabilize the low temperature hcp form of titanium, alpha. Unfortunately addition of alpha-stabilizing alloying elements results in embrittlement due to complex ordering reactions(1) at fairly low concentration levels, and solid solution strengthening is consequently limited. Strengthening due to alpha stabilizers must, therefore, be supplemented by the addition of beta stabilizing elements, which results in the formation of the bcc beta phase. This may or may not be stable and heat treatments based on betaa decomposition are the basis for several high-strength alloys in the current service temperature range. Nevertheless, the need to add beta as well as alpha stabilizers to achieve the degree of high temperature strength needed limits the service temperature level to which this type of alloy can be developed and it is becoming increasingly apparent that for service temperatures appreciably above these now in use titanium alloys based on the fully stabilized bcc beta phase will have to be considered.

The attempts in regard to the study of various compositions of Ti-Mo alloy have been done so far by many investigators but a little attention is paid on study of the effect of thermomechanical treatment. This treatment can be defined as the prior cold deformation on quenching the specimens after solution treatment which may provide the improved mechanical properties on subsequent tempering.

The attractive properties of Ti-alloys such as high strength to density ratio, excellent corrosion resistance at temperatures below 500°C in any media, relatively high melting point among the light metals, good fatigue strength and toughness make them a suitable material for many engineering applications such as in air craft industry, internal combustion engines etc.

The Ti-Mo as binary alloys have little commercial applications till now and mainly in aircraft industry as structural material with view of decreasing raw material cost. Ti-14Mo alloy$^{(2)}$ is supplied in solution treated condition as commercial alloy in the market with U.K. Specification VES(A1)507(domestic specification). ~~Although having a little interest in commercial applications so far~~ The Ti-Mo alloys have a great deal in expanding scope of research field being heat treatable alloys.

# CHAPTER 2

# LITERATURE REVIEW

At 882.5° C pure titanium metal undergoes an allotropic transformation from a hexagonal close-packed α-structure, stable at low temperature, to a body centred cubic β-form which persists upto the melting point, 1668°C(3). The occurrence of allotropic transformation in pure titanium controls the type of structure which can be produced by heat treatment of titanium-rich alloys and hence increases the range of their mechanical properties.

When the alloying elements dissolved in titanium have a great influence on its allotropic modifications, α and β then for a better understanding of thermal treatment of such alloys it is essential to have a sufficient knowledge of equilibrium diagram and phase transformation.

## 2.1 Constitution of Ti-Mo Alloys

The alloying nature of titanium is such that it forms extensive solid-solution alloys with all of the transition elements. Molybdenum, having bcc structure, forms a substitutional solid solution on alloying with titanium and stabilizes the high-temperature bcc form, β, of titanium. Thus, molybdenum may be called as beta stabilizer in alloying history of titanium.

### 2.1.1 Equilibrium Diagram of Ti-Mo System

The Ti-Mo system is characterized by complete miscibility between the bcc allotrope of titanium and molybdenum - hence it is termed a beta isomorphous systems as shown in Fig. 2.1 worked out by Hansen et al(4). The melting range is increased by molybdenum, the alpha-beta field is relatively broad, and the alpha solubility is limited, slightly less than 0.5 per cent. The temperature of transformation from the beta field to the alpha-plus-beta field, beta transus, decreases linearly with increasing molybdenum content.

The critical stabilizing concentration for molybdenum(56) is as 5.8 at pct (11.0 wt. pct) which is necessary to obtain a single phase structure of β-solid solution in a metastable state at room temperature.

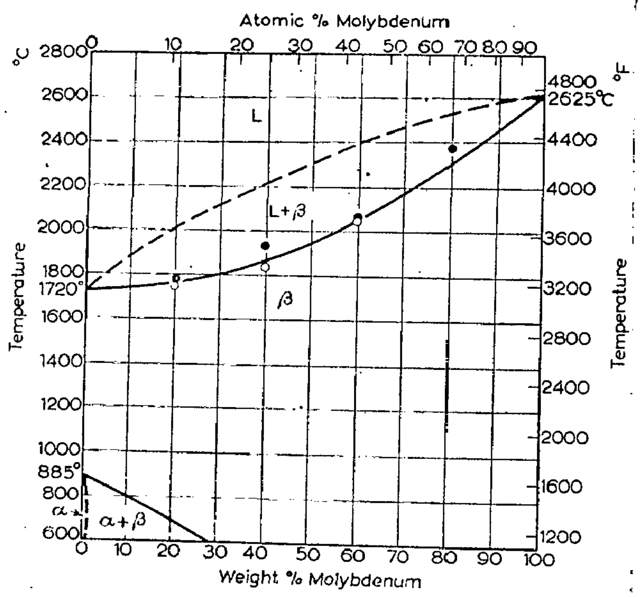


Fig. 2.1 Phase diagram of Ti-Mo System[4]

## 2.2 Some Aspects of Phase Transformation in Ti-Mo Alloys

The $\beta$ to $\alpha$ transformation in Ti-Mo alloys has been studied by Delazaro and Coworkers(7, 8). The time-temperature transformation characteristics have been studied by observations on the microstructures after quenching from the beta field to various temperatures and holding for various lengths of time. The curves of the start and finish of the $\beta$ to $\alpha$ reaction exhibit the characteristic 'C' shape, which is the net result of the low rate of nucleation at temperatures close to the equilibrium temperature and the increasing rate of nucleation at lower temperatures being offset by decreasing diffusion rates such that a maximum reaction rate occurs at an intermediate temperature. Fig. 22, adapted from Delazaro et al, shows the variation with molybdenum content of 'C' curves denoting the first appearance of alpha. As the molybdenum content increases, the nose of the 'C' start curve occurs at decreasing temperature. This effect would appear to result from the decreasing transus temperature. Also, there is a trend toward increasing time to start, which would be expected because of increased sluggishness of the transformation with higher beta alloy contents. The alpha phase nucleates predominantly at the beta grain boundaries, and only in later stages within the

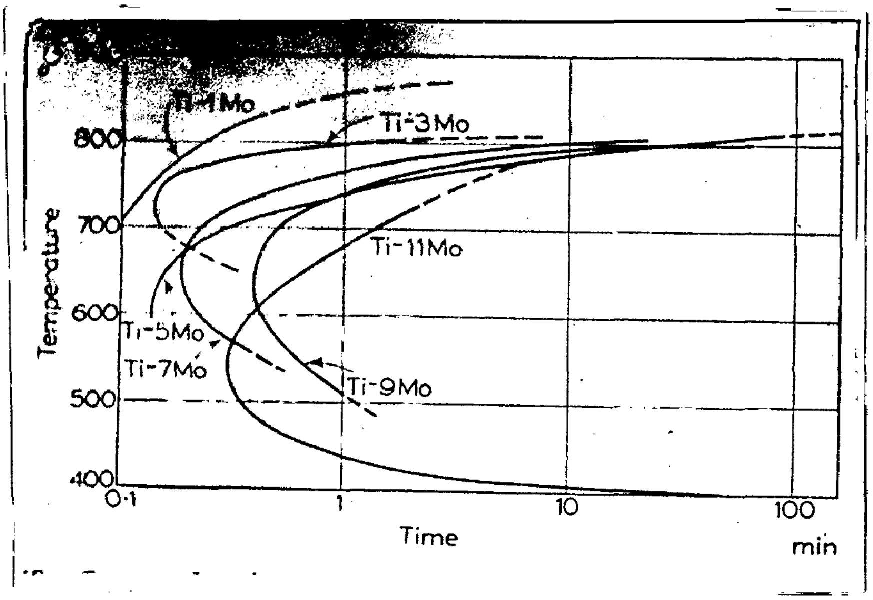


Fig. 2.2 TTT-Curves showing the effect of composition on the start of the beta- to alpha transformation in Ti-Mo alloys[7].

interior of the grains. As the temperature of transformation decreases, the alpha platelet size becomes progressively smaller and finer until, below the nose of the 'C' curve, the precipitated alpha is practically unresolvable. Delazaro and Rostoker(8) also studied that the rate of $\beta \longrightarrow \alpha$ reaction increases as the oxygen content increases.

In Ti-Mo alloys, the $\beta$ phase is retained at room temperature when quenched in water from the $\beta$-field. A number of phase transformations can occur in such metastable $\beta$-phase in favourable conditions as temperature, time and composition. These possible phase transformations are listed below :

(a) The $\omega$-phase can form on quenching(9)

(b) Marten site can be produced by subzero cooling or deformation(10)

(c) On aging at temperatures lower than nearly 550° C the $\omega$-phase is formed before the $\alpha$-phase(11)

(d) On aging at temperatures above 550°C. The $\alpha$-phase is formed(11).

(e) The martensite can be tempered. It has been reported that the $\alpha$-phase rather than the $\beta$-phase is precipitated during tempering(12).

## 2.3 Martensitic Transformation in Ti-Mo alloys

The structure of quenched titanium alloys may be that of unstable β-phase, sometimes called β', and/or the metastable forms of alpha designated as α' and α'' and are called the martensitic phases. The α' phase is supersaturated solid solution of α-titanium with a hexagonal crystal lattice. The presence of the α' phase considerably increases the hardness of the alloys. The α'' phase differs from the α' phase by the fact that it has an orthorhombic rather than a hexagonal lattice, as a result of the supersaturation of the α-solid solution with alloying elements. This phase is formed only in Ti-alloys containing those transition elements, the atomic radii of which are close to the atomic radii of titanium (V, Nb, Ta, W, Mo and Re). Alloys containing an α'' phase are considerably softer than those with α' phase(13).

### 2.3.1 Compositional Range for Martensitic Transformation

As reported by Jaffee(5) the alloy containing less than 7.0 wt. per cent molybdenum will consist the complete martensite phase on quenching from β-field Fig. (2) whereas the alloy containing molybdenum in the range of 7-12 wt. per cent may provide the martensitic structure along with the retained β-phase, β', but beyond this limit of molybdenum content, complete retained β-phase will be present.

### 2.3.2 Ms Temperature

The Ms temperature, the start of the martensite transformation from beta to alpha, decreases with increasing molybdenum content until at 11-12 per cent molybdenum the beta phase is completely retained after quenching(14). Ms temperatures obtained by Duwez(14) for a number of titanium alloys including Ti-Mo are shown in Fig. 2.3. The martensite arrest temperature was found to be relatively independent of cooling rates upto about 10,000°C/Sec. The Ms temperatures in Ti-Mo alloys are increased by oxygen, about 10°C for each 0.1 per cent of oxygen. For Ti-Mo alloys the Ms curves passes from 12 wt per cent Mo and Mf curve passes from 7 wt% Mo compositions.

### 2.3.3 OBientation Relationship during martensite transformation

Lui and Margolin(15) found two martensite habit planes in the Ti-Mo alloys after water-quenching, predominantly $\{334\}_\beta$ and to a lesser extent $\{344\}_\beta$. A transition occurring between 10.5 and 11.5% is found for both $\{384\}_\beta$ and $\{344\}_\beta$ martensites(16,17) the $\{344\}_\beta$ martensite found in the higher Mo alloys is usually formed by deformation of the quenched alloy. In a deformed

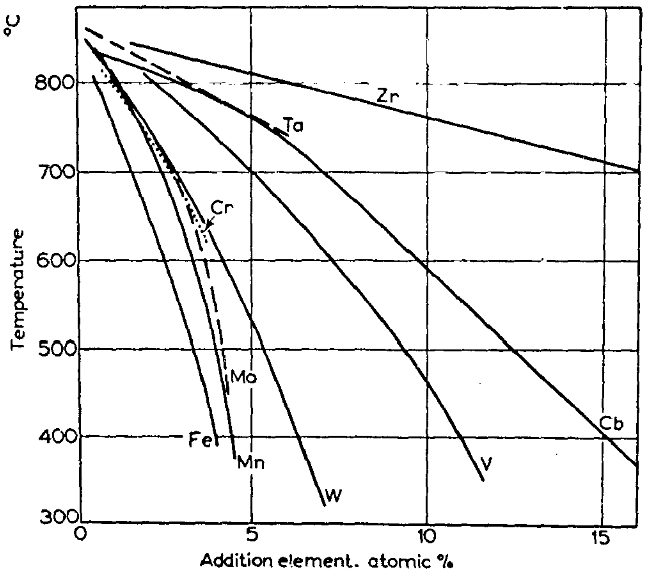


Fig. 2.3 $M_s$ curves for titanium binary alloys[14]

Ti-11.6% Mo alloy; it has been seen that the 'martensite' plate as transformation product consists of a complex array of twins(18). In lightly deformed specimens the habit plane is near $\{112\}_\beta$ and moves towards $\{111\}_\beta$ as the deformation increases. However, it is not proved possible, as yet, to account for the observed $\{344\}_\beta$ habit plane by a double or multiple twinning process. The twinned structure is produced by deformation and the observed magnitude of the shear is 0.3(16), thus it is difficult to account for this value unless the twinning shears take place in opposite directions. The Burgers Orientation relation for martensite in Ti-Mo and other titanium alloy martensites of the $\{334\}_\beta$ habit was suggested by Lui and Margolin(15),

$$(0001)_\alpha \parallel (110)_\beta$$

$$[1120]_\alpha \parallel \{111]_\beta$$

and was one of two orientations suggested(15) for martensite {344} formed from beta by cold work.

### 2.3.4 Tempering (aging) Behaviour of Martensite

Domagala and Kostoker(19) studied the decomposition of martensite )presumably formed by strain induced transformation) in a Ti-13 Mo alloy and reported that equilibrium alpha was precipitated from the martensite plates. Tempering of martensite, $\alpha'$, involves diffusion of excess alloying elements, retained by quenching in $\alpha'$ in adjacent beta_phase. Domagala and Rostoker concluded that the predominant mechanism in tempering was precipitation of low-alloy-content alpha particles from martensite until it became sufficiently reverted to enriched beta by an undetermined process.

The tempering behaviour of orthorhombic martensite was studied by Williams and Hickman(20). They found that the tempering behaviour of this type of martensite is strikingly different from that reported for hexagonal martensite. The tempering study was done on Ti-4 at % Mo and given an X-ray diffraction pattern characteristic of the orthorhombic structure as,

$$a = 3.001\ A^{\circ} \qquad b = 4.998\ A^{\circ} \quad \text{and} \quad C = 4.657\ A^{\circ}$$

In Ti-Mo alloys aged at the higher temperatures, the $\beta$-phase reflections increase in intensity. Examination of this alloy, using optical and thin foil electron microscopy, showed the initiation of a cellular reaction at prior $\beta$-phase grain boundaries. The product of the cellular reaction is a lamellar $\alpha + \beta$ structure which consumes the $\alpha$ + orthorhombic martensite mixture . With further aging, cellular reaction went to completion. The reaction in the Ti-4 at % Mo alloy commenced after 15 minutes at 550°C and was $\geqslant$ 95 per cent complete after 16 hours at 550°C(20).

## 2.4 Omega Phase Precipitation

On quenching the Ti-Mo alloys with the slower rate as compared with the rate in case of martensite formation, from the solution treating temperature, or, on aging the quenched Ti-Mo alloys in lower temperature range than 550°C may result a marked increase in hardness. This embrittlement behaviour of alloys is due to the formation of submicroscopic complex hcp transition phase, Omega, along with retained $\beta$-phase where retained $\beta$-phase is enriched by the alloying elements compared with the initial $\beta$-phase of elevatdd temperature. This $\omega$-phase is of electron compound type and may appear in all the systems of titanium with a transition elements.

### 2.4.1 Detection Techniques for Omega Phase

Because the Omega phase is of submicroscopic size, it is not possible to detect it simply with the help of optical microscope. Subsequent studies using X-ray diffraction and, more recently, transmission electron microscopy and quantitative X-ray diffraction techniques have successfully defined most of the features of $\beta \rightarrow \beta' + \omega$ tfansformations(21 - 25). T. Yukawa et al(26) used the dilatometric technique to study the Omega phase precipitation in Ti-15 Mo and Ti-12 Mo alloys while the contraction occurs during the omega phase precipitation and expansion during alpha phase precipitation.

### 2.4.2 Effect of Composition on Omega Formation

According to Hickman(22) $\omega$-phase formation occurred upto higher temperatures in the 6 at % Mo alloy than in the 10 at % Mo alloy (500° compared with 450° C) and was stable for longer periods in the lower molybdenum alloy.

### 2.4.3 Structure and Morphology

The structure of the $\omega$-phase has been confirmed as hexagonal. Evidence for the hexagonality of $\omega$-phase is given by parameter measurements(27), as, if the c/a ratio is not the ideal value $\sqrt{3}/(2\sqrt{2})$ derived from a perfect cube, there is positive proof that  is hexagonal with axial ratio 0.612-0.613 space group $P\bar{3}\,m1$ and atom positions 000, $\pm$ 1/3  2/3  Z

where $Z = 0.480 - 0.495$

Evidence in favour of this identification is also given by a dark field micrograph taken using a $(10\bar{1}1)_\omega$ reflection from a quenched specimen and the corresponding diffraction pattern are taken by Silcock$^{(27)}$. The $\omega$-reflections are streaked and the two directions of streaking are evidence for the hexagonal structure of the phase. Further, the streaking indicates that the $\omega$-phase is not spherical but forms as ellipsoids, as observed by Blackburn and William$^{(23)}$ in the later stages of aging, or as plates on $\{111\}_\beta$. On aging, the $\omega$-phase grows in the Ti-Mo alloys and forms as ellipsoids. Trace analysis has shown that the major axis of the ellipsoids lie in $[1000]_\omega$ or $\langle 111\rangle_\beta$ and the minor axis in $(0001)_\omega$. The orientation relationship between $\omega$ and $\beta$ phases being $[0001]_\omega \parallel [111]_\beta$
$(11\bar{2}0)_\omega \parallel (110)_\beta$

## 2.4.4 Coherency of Omega Phase

Blackburn and Williams$^{(23)}$ suggested that the morphology of $\omega$-phase particles can be related to misfit stain. The linear lattice misfit, L, is given for the various alloys: Ti-10 at % Mo - The initial lattice misfit + 0.42% and final + 0.88%$^{(22)}$. These values were computed from the relationship,

$$L = \frac{1}{3} \frac{V_\omega - V_\beta}{V_\beta}$$

where, $V_\omega$ and $V_\beta$ are the unit cell volumes divided by the number of atoms per unit cell for the $\omega$-and $\beta$-phase respectively. Where the misfit is low (less than about 0.5%) the particles may assume with ellipsoidal morphology because the elastic strain energy is low and hence minimization of surface energy determines the particle shape, the precipitates being ellipsoidal presumbbly because of the anisotropic structure of the $\omega$-phase.

Blackburn and Williams(23) considered that the $\omega$-phase is coherent with the matrix to its maximum particle sizes. Blackburn(18) observed the maximum particle size, 1500 A°, in Ti-11.6% Mo alloy on aging at 400°C for more than 3000 hrs. Due to the high volume fraction, no strain contrast effects associated with coherent particles were observed in the $\beta/\omega$ interfaces under any contrast conditions.

One additional observation(23) relevant to the $\omega$ phase occurring only as a coherent precipitate is its behaviour during spontaneous transformation of thin foils. At small particle sizes, e.g., in quenched Ti-11.6 Mo alloy, it appears to be unchanged by this transformation (although it is destroyed during the strain induced martensitic transformation). It is, however, destroyed at large sizes when a region is spontaneously transformed, and in this case it appears to revert to the β-phase. Thus it is possible that passage of the interface which bounds the transformed regions renders the ω-phase incoherent and thus unstable. The ellipsoidal shaped particles in Ti-Mo, probably, arise from anisotropy of surface or strain energy in the precipitate rather than the matrix.

### 2.4.5 Composition of Omega Phase

Only one experimental observation can be related to the composition of the phase. Indirect evidence for the ω-phase composition in Ti-Mo alloys has been obtained using the data of Holden et al(28) who showed that the aging response at 400°C of a Ti-20 Mo alloy was marked whereas that of a Ti-24.5 Mo was very slight. Thus assuming the limit of ω-phase formation is 25 wt. per cent and knowing the volume fraction of ω-phase in Ti-11.6 Mo, the composition of omega phase was computed as $Ti_{97}Mo_3$ using the lever rule. Luke et al[29] suggested that the ω-phase was an electron compound.

Blackburn and Williams(23) have reported that in regions, which got spontaneously transformed during thin foil preparation, the ω-phase, is destroyed. It has been seen(23) that the structure is retransformed to the β phase, and that the phase maintains its composition difference.

### 2.4.6 Growth and Volume Fraction

The ω-phase particles grow on aging but it is virtually impossible to establish a growth law from thin foil observations. This is due to the very large volume fraction of ω, which leads to particle overlap, and to the large range of particle sizes. These factors also hinder the measurement of volume fraction of the phase.

As in many precipitation processes the initial growth rate of the particles is high and then falls as aging is continued. The maximum particle size that has been observed is nearly 3000 A° in Ti8Mo alloy which was measured(23) at the initiation of the β +ω→β +ω+ α stage of the transformation. Short aging treatments increase the quantity and size of ω particles which makes the alloy brittle.

It is postulated(23) that the limiting size of the precipitates is controlled by the mismatch between the ω-phase and β-phase. During the formation of the ω-phase the solute concentration of the matrix increases. Molybdenum decreases the lattice parameter of β-Ti and causes the contraction during omega phase formation.

~~Ea~~ Earlier works$^{(21-25)}$ have shown that the volume fraction of ω-phase can be controlled by :

(i) varying the aging temperature

(ii) varying the solute content of the β-phase either by changing the initial alloy content or by solution heating in the (α + β) phase field to give an enriched β-phase.

(iii) by ternary additions of such elements as Al, Sn and O.

## 2.4.7 Mechanism of Omega Formation

The mechanism of ω-phase formation during quenching has been the subject of much speculation. As discussed by Hatt and Roberts$^{(30)}$, it is possible to form the ω-structure from the β-structure by alternate shears of 0, g, $\bar{g}$ on the {112} planes in the $\langle 111\rangle_\beta$ direction where $g = \frac{1}{6}\sqrt{\frac{3}{2}}\,a_\beta$. They suggested that the faulting observed in the ω-structure is compatible with such a mechanism.

The fact, as shown in work of Hickman$^{(27)}$ that during aging the ω-phase can exist over a range of composition, means that during quenching of alloys within this composition range transformation to ω-phase can occur without a composition change.

The ω-phase particles can then nucleate and grow during quenching because no significant solute diffusion need occur. It is, therefore, not necessary to involve a shear mechanism, but on the basis of the available evidence it is not possible clearly to decide if the transformation to the very small particles ( < 50A°) of ω-phase observed in quenched materials occurs by a co-operative shear or by random atom suffles. The $\beta \rightarrow \omega$ transformation can be effected by displacements, magnitude 1/12 a [111] , of neighbouring rows of atoms in positive direction.

As the ω-phase is formed the β-phase becomes enriched in solute element which changes its lattice spacing, there is evidence that the ω-phase formed in the various systems exhibit similar type of lattice spacing variation and thus the mismatch is controlled to a first approximation by the beta phase(18).

Hickman(21) suggested that segregation of the β-phase into solute rich and solute poor regions occurs during quenching i.e. a spinoidal decomposition, and subsequently the solute poor regions transform into the ω-phase. Blackburn and Williams(23) suggests that this transformation could be induced by hydrostatic stress due to differences in thermal expansion between the solute rich and solute poor regions. An alternative explanation is that quenching can not suppress the short range diffusion required for the formation of the $\omega$-phase.

### 2.4.8 Decomposition of Omega Phase

Frost(31) suggested that in general during aging after quenching the omega phase breaks down forming alpha and enriched beta. In system with ellipsoidal $\omega$-phase, the $\alpha$-phase forms by a separate nucleation and growth process which may take the form of a cellular transformation or the growth of isolated plates probably from dislocations present in the original $\beta$-phase(32).

Explanation for the $\alpha$-phase formation from ($\beta$ + $\omega$) structure, and not directly from the $\beta$-phase given by Blackburn and Williams(23) was that in Ti-Mo alloy the alpha phase is formed either by a cellular reaction or by hetrogeneous nucleation of a number of $\alpha$-plates. In case of the first type behaviour a cell containing alternate lamellae of ($\beta$ + $\alpha$) grows from a grain boundary. The orientation relationship between the $\alpha$ and $\beta$ phase was found to have the expected Burgers and habit plane was to be near $\{110\}_\beta$ but not exact. In second mode of alpha formation it was considered that such alpha plates, i.e. isolated needles are nucleated at dislocations, however, such an association is difficult to prove conclusively, but it has been shown that such heterogeneous nucleation is in fact observed in deformed and aged specimens.

TTT curve for a Ti-13 Mo alloy by spachner and Rostoker(11), in which the transformation to omega was followed is shown in fig. 2.4. Three type of reactions may take place on aging in Ti-13 Mo alloy

(i) $\beta \rightarrow \beta + \alpha$ Over 575°C

(ii) $\beta \rightarrow \beta + \omega \rightarrow \beta + \omega + \alpha \rightarrow \beta + \alpha$ from 500 to 575° C

(iii) $\beta \rightarrow \beta + \omega$ below 500°C

As there is no simple relation lattice relation between the $\alpha$ phase and the $\omega$-phase (whereas there is between $\omega$ and $\beta$), the transformation can not be looked upon monotropic. Furthermore, the composition of the $\alpha$-phase appears to be different from that of the $\omega$-phase.

Resistometeric measurements were used primarily to determine the times for initiation and completion of transformation by Spochner and Rostoker(11) in which the current potential method of Graft et al(33) was used in the determination of the resistivity of heat treated specimens. The identity of time-temperature dependent phase fields between initiation and completion of transformation was accomplished largely by X-ray work.

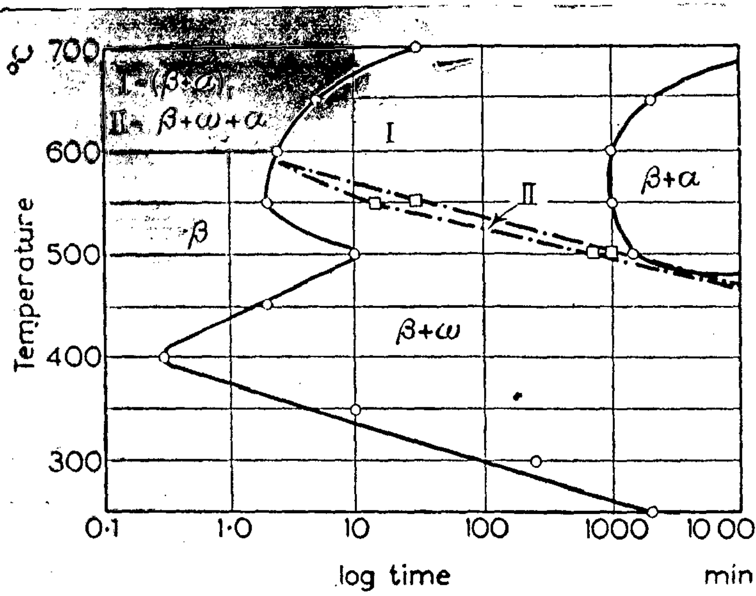


Fig. 2.4 TTT curve for a Ti-13% Mo alloy[23]

## 2.5 Thermomechanical Treatment of Ti-Mo Alloys

The thermomechanical treatment is usually defined as the use of deformation prior to or during an allotropic change so as to obtain an improvement in mechanical properties.

Although the work on phase transformation under cooling and heating has been done widely and satisfactorily but the thermomechanical treatment effects have not been yet studied extensively and conflicting evidences have been published. Presently, it has been discussed that the role of plastic deformation is very important one during structural changes on thermal treating such as aging.

### 2.5.1 Effect of Deformation

As the martensite can be produced by subzero cooling or deformation(10), deformation of the quenched Ti-Mo alloys occur by formation of stress induced 'Martensite' (or twins) and slip. This deformation induced martensite has been found to have a bcc or bct structure, rather than the previously reported hexagonal structure. The dislocation arrangements in the matrix after 10 per cent tensile deformation were found to be quite random in nature. This deformation weakens the intensity of phase reflections or causes them to completely disappear(23). Aging of quenched and deformed specimens at temperatures $< 550°$ C results in precipitation of the alpha phase within the martensite, precipitation of the $\alpha$-phase on dislocations and growth or nucleation and growth of the $\omega$-phase in the matrix.

The martensite formed during plastic deformation has been qualitatively examined(23). Martensite needles were readily detectable in polished and deformed samples by optical microscopy. However, this foil examination has shown that these apparent needles actually consist of stacks of thin twins and that the structure of these regions is bcc within the accuracy of selected area diffraction. The internal twinning system has been identified to occur on $\{112\}$. Trace analysis has shown that in lightly deformed specimens some transformed regions have a habit plane near $\{121\}_\beta$ and such plates exhibit little or no secondry twinning. In more heavily deformed specimens the structure of the transformed regions becomes more complex, two or more twinning systems being found within a plate. Trace analysis from these structures shows such a large scatter that they can not be conclusively identified as having habit planes near $\{344\}_\beta$ although most results lie near the $\{111\}$ pole. It has been shown that the spontaneous transformation of martensite is distinct from the deformation induced transformation of martensite and that no direct relationship exists between them. The structure of stress induced martensite as bcc or bct explains the extra reflections on bcc layer lines observed by Gaunt and Christian(17). Blackburn and Williams(23) propose two explanations, viz. a cooperative and rather complex twinning or a bcc to bct transformation in which the c/a ratio of the bct phase is nearly 1, further work is needed to test these suggestions. Some alpha phase plates have been observed to restrict their own growth by generating slip across the growth direction.[34] Analysis of slip line separation magnitudes in one of these instances indicated that a maximum stress, of the order of $10^{10}$ dyn/cm$^2$, is present at a plate tip and that stress decreases linearly with distance from point in the direction of the propagation of plate.

From the work of Holden et al(28) on Ti-Mo alloys it is believed that martensitic alpha formation is the first consequence of an applied stress when the molybdenum content is in the approximate range 7-16 per cent. Certainly this was observed for a Ti-12 Mo alloy(17).

## 2.5.2 Effect of Deformation on Omega phase

The beta-to-omega transformation can be effected by displacements, magnitude 1/12 $a\{111\}$, of neighbouring rows of atoms in positive directions. An ability to form independently in four crystallographic directions is implicit in the orientation relationship, $\{0001\}_\omega \parallel \{111\}_\beta$

$(11\bar{2}0) \parallel (110)$

This possibility was examined by Silcock(27) with a negative result and it was concluded that deformation does not provide nuclei for subsequent omega formation as might be expected if it occasioned the correct atom movements for production of embryonic omega phase. Her work, by inference from results obtained for Ti-13Mo, contradicts the findings of Brotzen et.al(35). It appears possible that the failure of Silcock to observe Omega formation in deformed specimens was a result of preferentially created martensitic alpha(28, 23).

Wood(34) drew the following conclusions as the result of cold deformation to the Ti-15 Mo alloy as in β-solution treated condition.

(i) When subjected to compressive stress, coarse and fine grained specimens of a Ti-15% Mo alloy in the beta solution treated condition deform initially by formation of martensitic alpha as plates with a $\{334\}_\beta$ habit.

(ii) In favourably oriented grains of coarse grained material, growing plates have an interfacial velocity of such a magnitude that the stress field associated with a plate tip can not be accommodated elastically by the surrounding metal and local deformation, manifest as a slip line array, takes place.

(iii) Stress in the surrounding metal is a maximum at a plate tip of the order of $10^{10}$ dyn/cm² near the point and decreases linearly therefrom with increasing distance in the direction of tip propagation.

(iv) Deformation of fine grained specimens in excess of about 5 per cent occurs by slip and is accompanied by formation of omega phase.

(vi) Deformation induced omega phase occurs in a plate-like form on $(111)_\beta$ planes.

X-ray diffraction results(34) have established that omega phase in this alloy forms in greater quantities as cold deformation proceeds, at least until about 40 per cent strain has been effected. The phase originates on octahedral planes in the lattice of the β-matrix and appears to grow on these planes.

## 2.6 Effect of Structural Changes on various Properties

The strength of interatomic bonds, the changes taking place in the elementary crystal unit and the changes in the forces of interatomic reaction occurred by alloying, heat treatment, mechanical or thermomechanical treatment influence many physicomechanical properties of alloys. A consideration of these relationships must form the basis for the development of new alloys with improved properties.

### 2.6.1 Mechanical Properties

In any investigation of the reactions of titanium with other alloying elements it is important to examine those characteristics which most closely reflect the relationship between the interatomic bonds and the chemical composition of alloy in the equilibrium and metastable conditions during both heating and cooling. Among these characteristics the most important are the elastic constants. The importance of interatomic binding forces in the evaluation of the elastic constants was indicated by the work of Kurnokov and Lazarev who expressed that the indentation hardness and modulus of elasticity are closely interconnected. In the work reported by Fedtov(13), it has been likewise found that the elastic properties of titanium alloys, both the Young modulus and the shear modulus, can be almost doubled (by alloying) or decreased (by hardening) as compared with those of pure metal. In the equilibrium condition both the elastic modulus and the shear modulus continuously decrease with the increase in the content of molybdenum in titanium alloy. This is also indicated(13) that the β-solid solution having a bcc structure has a lower elastic modulus than the α-solid solution with its h.c.p lattice. The relationship between elastic properties of hardened, i.e., quenched alloys, and their composition was quite different from that of alloys in the equilibrium condition.

In the hardened condition of Ti-Mo alloy at first the elastic constants sharply decrease (almost half) then with the same suddenness these constants are restored to the initial magnitude or even higher. There follows a further decrease and finally there is a continuous increase of elasticity in the β-field, apparently upto the level of the elastic constants of the pure metal i.e. Mo. The rather unusual nature of the variations in the elastic constants of hardened Ti-alloys is a consequence of their structure. The presence of the α' phase, hexagonal martensite, considerably increases the hardness of the alloys. The α" phase, orthorhombic martensite, is considerably softer than those with α' phase.

The Mechanical properties found by Holden et al(28) for Ti-Mo alloys quenched from the beta field are shown in fig. 2. Although beta phase is completelyretained at 11-12 per cent molybdenum, the minimum in yield strength occurs at 8 per cent molybdenum which has a structure of beta plus beta as quenched. The low yield strengths found in quenched alloys with 8-16 per cent molybdenum result from the low shear stresses required to initiate the beta-to-martensite transformation in this composition range. As the molybdenum content of the β-phase increases, the stress required to initiate the martensitic transformation increases until it finally becomes greater than the critical stress for slip.

The effect of martensite transformation on the mechanical properties of Ti-8% Mo has been investigated$^{(36)}$. At quenching temperatures close to β/(α+β) phase boundary, an adrupt transition in mechanical properties was found. As the quenching temperature is decreased through a range of 10°C, the elastic modulus, hardness, and tensile strength increase markedly and the internal friction decreases by a factor upto 100. The preceding data have been correlated with observations derived from ordinary microscopy, electron microscopy and X-ray diffraction work$^{(36)}$. Above the transition the microstructure consists of a mixture of β and martensite. As the quenching temperature is lowered into the transition region, equilibrium alpha begins to form thereby increasing the concentration of β-stabilizing elements in the β-phase. As a result, only a small amount of β decomposes martensitically. Below the transition temperature ω-forms in the stabilised β -phase. The transition temperature at which hardness increases adruptly on lowering the quenching temperature for Ti-8% Mo alloy is 815°C. A model was proposed by James and Moon(36) which explains thd transition in mechanical properties in terms of the stress, induced martensite transformation and also gives a possible explanation of the low yield strength of quenched Ti-Mo alloy it can be deduced that an increase in modulus requirds that a large fraction of the matrix must be converted to ω-phase during quenching as the quench temperature is lowered through the transition region. If an alloy is quenched from β-field the β-phase is partially transformed to martensite. The ω-phase is not formed in the untransformed beta because β-phase is not sufficiently rich in the β-stabilising elements . The brittle ω-phase formation results in a rapid increase in the yield strength and modulus and a decrease in the internal friction(36).

The Ti-10 % Mo alloy quenched from above the β-transus and aged exhibits measurable ductility. No ω-phase is formed in this alloy at aging temperatures higher than 500°C, since in this temperature range the equilibrium alpha phase forms directly. When the alpha phase is formed, the yield stress decreases in comparison to the samples which contain $\omega$-phase with attendant increase in elongation. In all heat treatment conditions the work hardening rate of Ti-Mo alloy was low. Thus the difference between ultimate tensile stress and yield stress was small(37).

Maximum strain induced transformation occurs in the Ti-11.7 Mo alloy. Since the martensite shear stress is not expected to increase much with transformation strain, it is estimated from the shape of the flow curve(28) that strain induced transformation takes place upto about 0.1 strain, and is followed predominently by slip upto about 0.3 strain. Less strain induced transformation occurs in the Ti-7.9 Mo alloy in which some thermal martensite is present.

It is apparent from fig. 2.5 that the combination of strength and ductility of the beta quenched Ti-Mo alloys is most favourable at the 12 per cent molybdenum composition, where maximum strain transformation occurs. It was also found that the notch toughness is a maximum at the same alloy content(28)

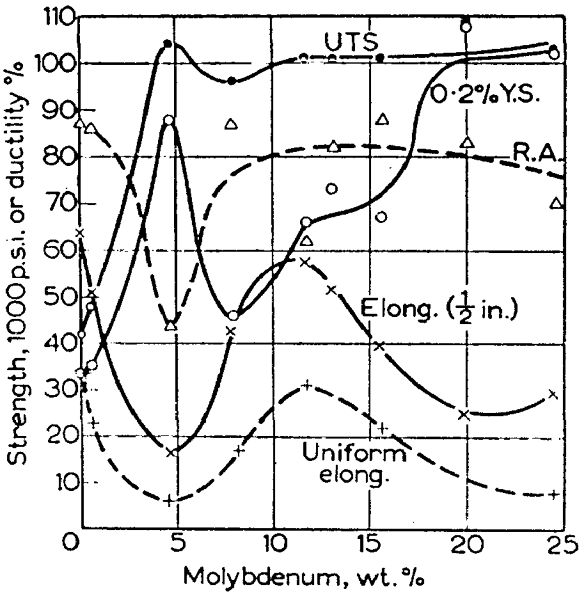


Fig. 2.5 Mechanical Properties of beta-quenched Ti-Mo alloys[28]

### 2.6.2 Physical Properties

Physical properties and structural phase stability was discussed by collings and Ho(38) for Ti-Mo system. With the Ti-10 at % Mo alloy some ω-phase has begun to form, and the lower concentration alloys consist of (β + ω). Both low temperature electronic specific heat coefficient, $\gamma$, and Debye characteristic temperature, $\theta_D$, showed broad turning points near 15 at % Mo and it is interesting to speculate on whether the relatively rapid changes in these quantities are induced by the presence of the ω-precipitate. It has been suggested(38) that in single-phase β material both low temperature electronic specific heat, $\gamma$, and magnetic susceptibility, $\chi_g$ should continue to increase as the molybdenum concentration decrease, until the transformation to $\alpha'$ occurs and that the rapid decreases encountered in $\chi$ and $\gamma$ are in fact due to the formation of the ω-phase.

Spachner and Rostokov(11) have used the physical properties measurement technique namely resistometric to determine the time for initiation and completion of transformation in Ti-13% Mo alloy. The measurement of such other physical properties may also be used to follow the rates at which transformation occur.

## 2.7 Scope of the Present Work

The work relating to the thermomechanical treatment of Ti-Mo alloys has been carried out very little, although it has a wide scope for new investigations and developments.

The primary purpose of the present work is to observe the effects of thermomechanical treatment on titanium molybdenum alloys. Two alloys Ti-10% Mo and Ti-23% Mo are selected for study since both these alloys have distinct transformation characteristics on quenching from solution treating temperature. The effect of mechanical working on these quenched alloys have a wide scope of study. It is a well known feature that the thermomechanical treatment contributes in enhanced strengthening.

The study of aging treatment will be definitely interesting since the appearance and proportion of different metastable phases, i.e. martensitic alpha, omega and retained beta under different conditions will show varying responses during such a treatment.

# CHAPTER 3

## EXPERIMENTAL PROCEDURE

### 3.1 Materials Under Study

The present investigation was carried out on the Ti-Mo alloys of compositions Ti-10 wt pct Mo and Ti-23 wt pct Mo with exact analytical compositions 9.43 wt pct Mo and 23 wt pct Mo respectively.

The alloys were supplied by the Metallurgical Division of Bhabha Atomic Research Centre, Trombay (India) in the form of buttons. These alloys were prepared by argon arc melting of sponge titanium and high purity molybdenum metals. The melting procedure was repeated many times over the hearth in order to ensure an uniform composition.

### 3.2 Specimens Preparation

Disc shaped specimens of 4 to 6 mm thick were prepared from the supplied buttons by hack saw cutting. Microscopic examination and hardness measurement of the as received specimens from the arc melted buttons were done.

## 3.3 Thermal and Mechanical Treatment

Titanium alloys are very active chemically at elevated temperatures and react readily with oxygen, nitrogen, and hydrogen in a furnace atmosphere. Therefore, it is important to take special precautions during thermal treatment.

### 3.3.1 Solution Treatment

The specimens, eight in numbers, of each alloy were sealed in two separate evacuated silica capsules. These two capsules containing eight specimens of each Ti-10 pct Mo and Ti-23 pct Mo alloys were placed in muffle furnaces with a temperature control of ± 5°C at 1000°C and 830°C respectively holding for 12 hours. After this solution treatment the capsules were quenched in water. One specimen of each capsule was polished, etched and examined under optical microscope. The hardness values of each specimen were also taken.

### 3.3.2 Prior Cold Deformation

The specimens of these two alloys after water quenching were compressed on 200 ton compression machine, with varying percentages of cold deformation as zero, 20, 30 and 50 per cent. Each percentage of deformation was given to two specimens of each alloy and finally eight specimens of each alloy were grouped into two . Each containing four specimens of zero, 20, 30 and 50 pct deformation respectively. Percentage cold deformation was calculated as :

$$\text{Percentage cold deformation} = \frac{\text{Change in thickness of specimen}}{\text{Original thickness of specimen}} \times 100$$

Hardness measurements after the cold deformation of the Ti-Mo alloy specimens were carried out and some of them were polished, etched and studied under the optical microscope.

### 3.3.3 Aging

Each grouped specimens, in four numbers, were sealed again in evacuated pyrex capsules. The capsules corresponding to either Ti-10% Mo or Ti-23% Mo alloy were placed in muffle furnace for aging at 400°C and 500°C respectively for 1 hr. and then quenched in water. Hardness measurements after quenching were carried out and some of them were polished, etched and studied under the optical microscope. After that again the same specimens were aged similarly for 2 hr, 3 hr, and 4 hr respectively. Each step was followed by the hardness measurements and microscopic study of same specimens.

## 3.4 Hardness Measurements

The hardness tests were performed in the Vicker's hardness Tester using a load of 10 kgs. An average of four to five readings was taken as the hardness value of the specimens.

## 3.5 Metallographic Studies

Mostly the structures were examined in the magnification range of 100-200 X. The etchant used for Ti-10 Mo alloy was 60 c.c. of glycerol, 20 c.c. of HF and 20 c.c. of $HNO_3$ solution, whereas for the Ti-23% Mo alloy it was 20 c.c. H.F., 20 c.c. $HNO_3$ and 40 c.c. glycerol solution (Vilella's etchant).

# CHAPTER 4

# RESULTS

## 4.1 EFFECT OF AGING TREATMENT ON HARDNESS

### (A) Aging Period

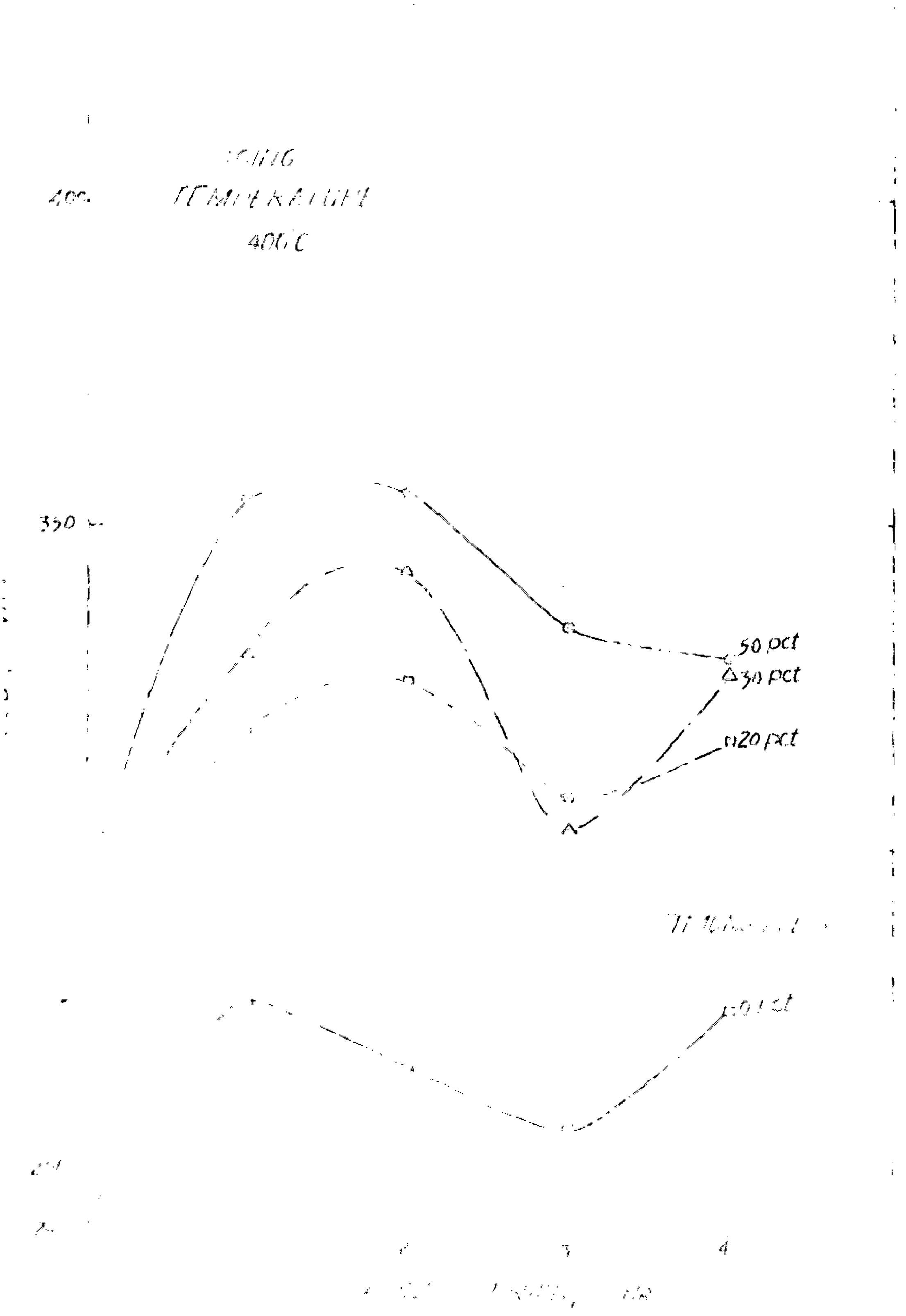
The plots of aging period versus hardness with varying cold deformation for Ti-10 % Mo and Ti-23% Mo alloys at 400°C and 500°C aging temperatures are shown in figures, 4.1 - 4.4. The aging curves are similar to those of typical age-hardening type ~~of-a~~ of alloys. For both aging temperatures the hardness value initially increases with increasing aging period, reaches a peak value and then drops continuously except only for Ti-23% Mo alloy where on aging at 400°C, the hardness value initially decreases and after that increases to the peak value. The maximum hardness value (425 VHN) among both the thermomechanically treated alloys is achieved for Ti-23% Mo alloy aged at 500°C for 3 hours when 30 per cent cold deformation is imparted. The Ti-10% Mo alloy shows the maximum hardness values between 375 VHN to 391 VHN, when aged at 500°C for 3 to 4 hours aging periods with 50 per cent cold deformation. In case of Ti-10% Mo alloy aged at 400°C, the peak hardness value increases with increase in degree of prior cold deformation. However, in case of aging at 500°C temperature, there is a departure from this sequence, since specimens with no deformation show higher hardness values.

### (B) Aging Temperature

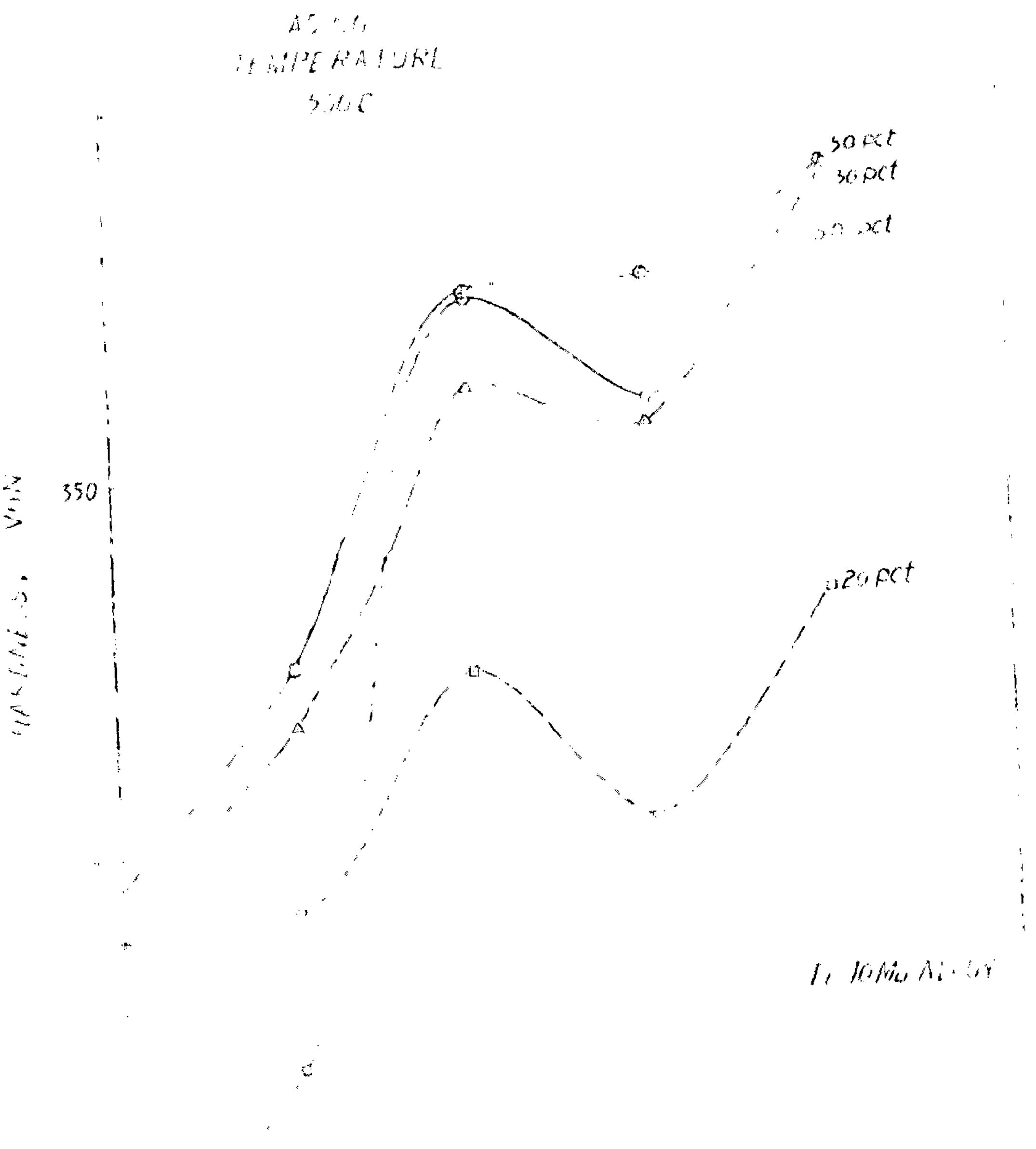
The behaviour of hardness peaks at different aging temperature for Ti-10% Mo and Ti-23%Mo alloys at various cold deformations is shown in figures (5.1 - 5.4) which are obtained after replotting the figures 4.1 - 4.4 It is evident that for either aging temperature peak hardness value of Ti-10 % Mo alloy is less th than that of Ti-23% Mo alloy. Aging at either temperature in case of Ti-23% Mo alloy does not vary the period of hardness peaks. The results shows that in case of Ti-10% Mo alloy, generally, there is a shift in hardness peak towards a higher period at 500°C as compared to that of 400°C. It is also observed that the peak hardness value of either alloy at 500°C aging temperature is higher than the corresponding value for 400°C.

## 4.2 EFFECT OF PRIOR COLD DEFORMATION ON HARDNESS

Figures 45-48 show the effect of prior cold deformation on hardness of quenched specimens of Ti-10% Mo and Ti-23% Mo

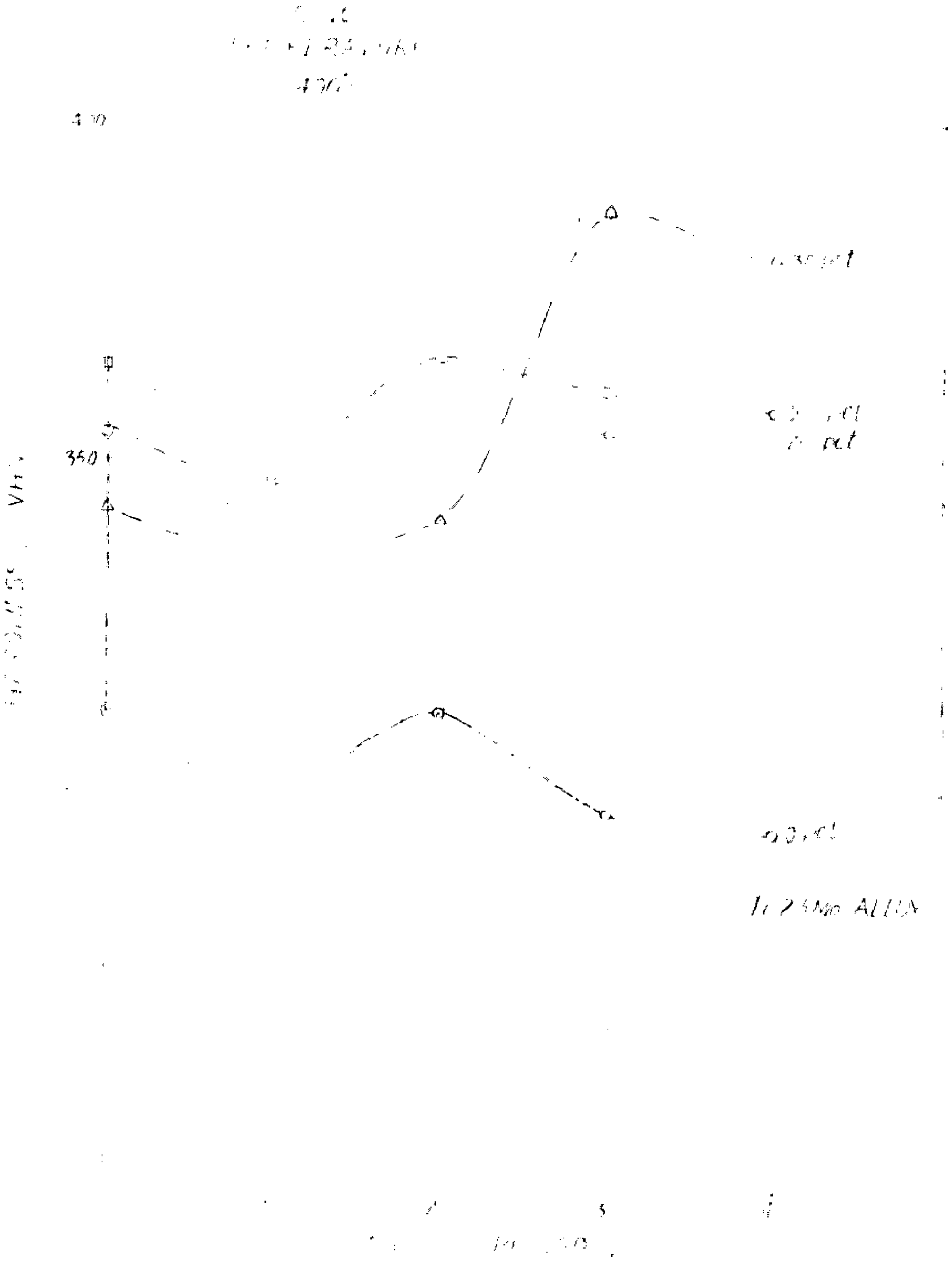

400
TEMPERATURE
400 C
350
50 pct
30 pct
20 pct
0 pct
2
3
4

[illegible]


AGING
TEMPERATURE
500 C
50 pct
30 pct
pct
20 pct
350
HARDNESS, VHN
Ti-10Mo-Al-Cr

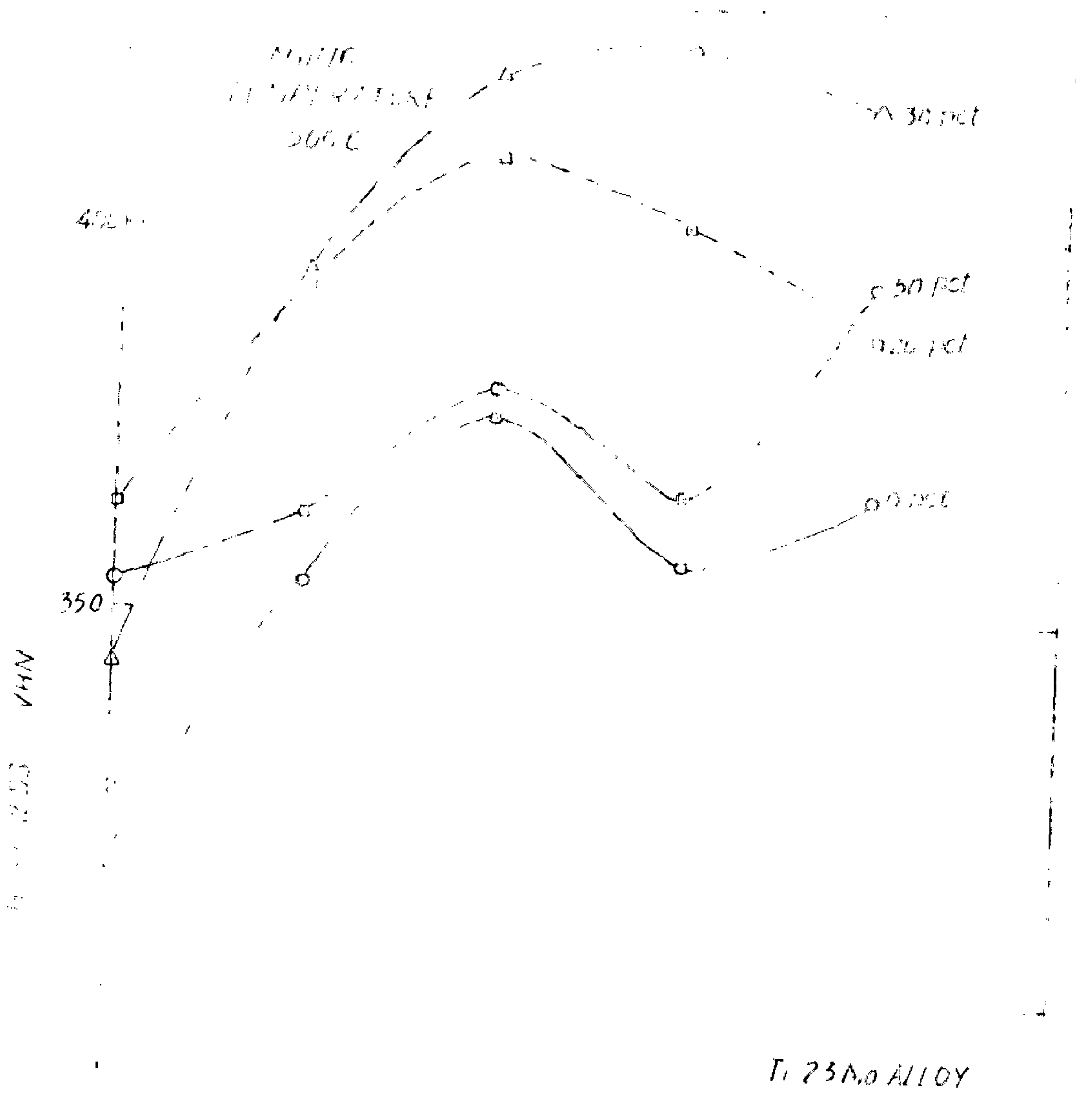
[illegible] N [illegible] F [illegible] IN

[illegible]



[illegible] EFFECT ON HARDNESS OF [illegible] IN

[illegible]


AGING
TEMPERATURE
500 C
400
350
VHN
30 pct
50 pct
0.00 pct
0 pct
Ti 23Mo ALLOY

[illegible]

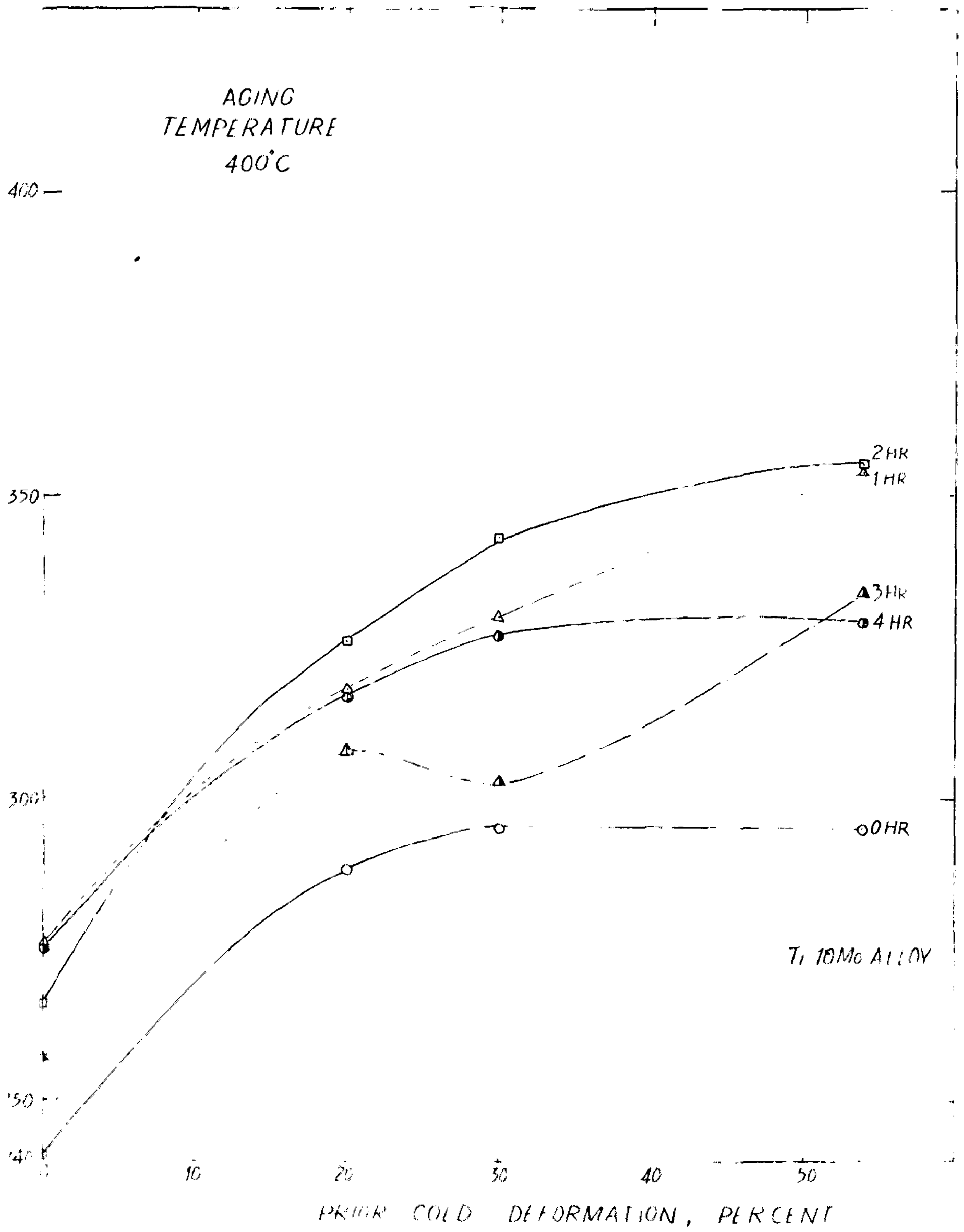
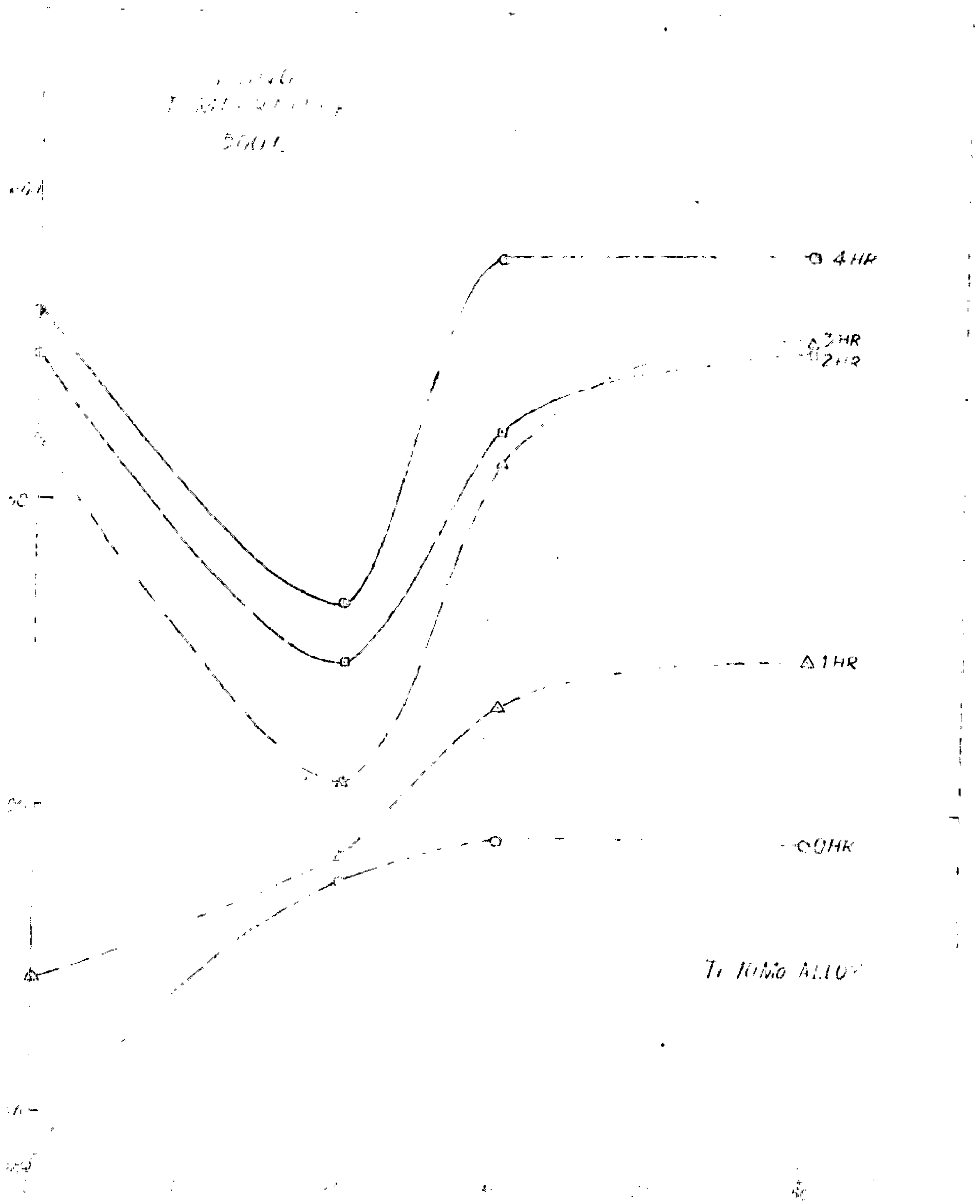
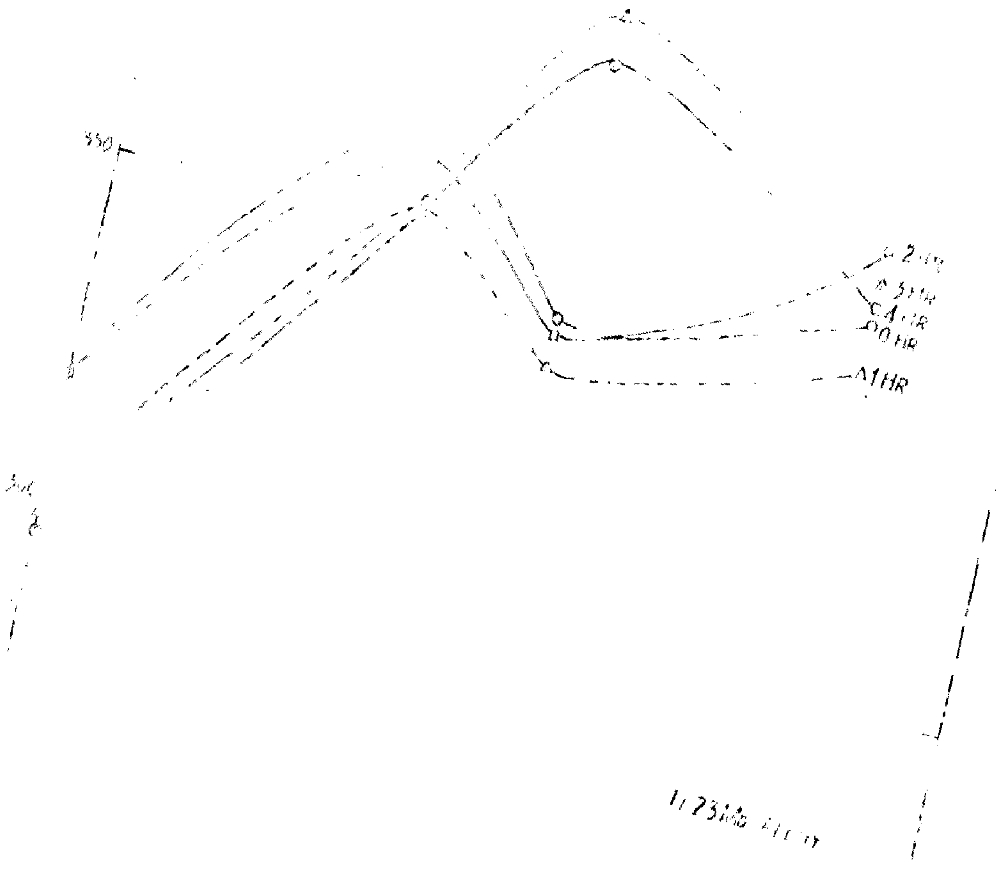


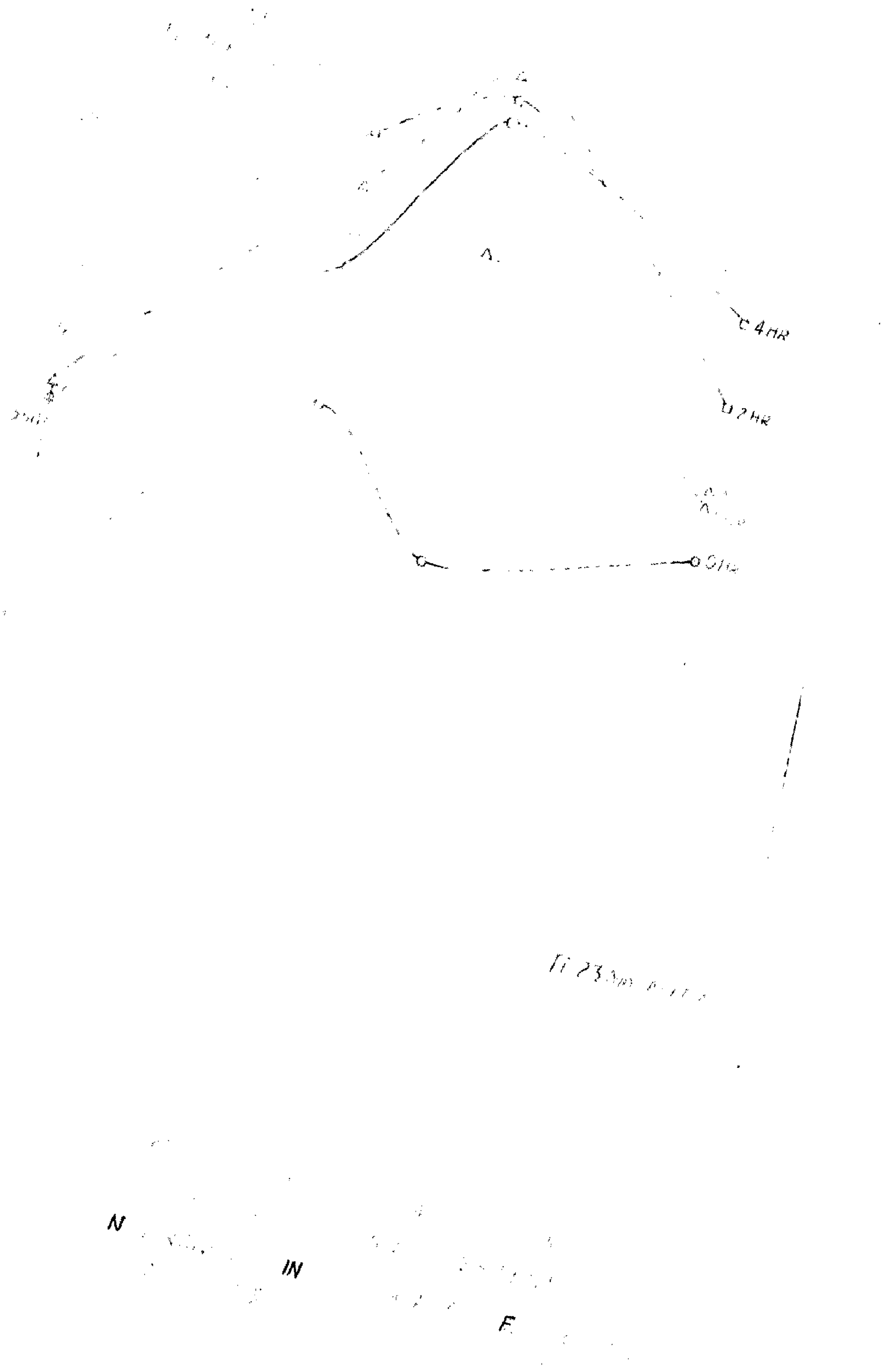
FIG 45 EFFECT ON HARDNESS IN Ti 10% Mo ALLOY OF PRIOR COLD DEFORMATION AGED AT 400°C FOR 0 4 HR DURATION


500°C
4 HR
3 HR
2 HR
1 HR
0 HR
Ti 10Mo ALLOY

[illegible]


450
2 HR
3 HR
4 HR
0 HR
1 HR

[illegible]


C 4HR
D 2HR
N
IN
F

alloys which were solution treated in single β-phase region at 1000°C and 890°C respectively. It is evident from these graphs that the observations of hardness are related with the degree of prior cold deformation, subsequent aging periods and aging temperatures.

In Ti-10% Mo alloys aged at 400°C hardness increases rapidly with increase in prior cold deformation upto 30 per cent and subsequantly increases slowly reaching practically a constant value (fig. 4.5). At 500°C aging temperature, the behaviour of hardness increment corresponding to 0 hr and 1 hr periods, was found similar to that observed for 400°C aging temperature.

In Ti-23% Mo alloy shifting of hardness peak values are found from 20 per cent to 30 per cent deformation further after 2 hr aging period at 400°C aging temperature (fig. 4.7). But in case of 500°C aging temperature the hardness peak occurs at 30 per cent prior cold deformation for practically all aging periods. In addition, the higher aging temperature e.g. 500°C promotes the values of hardness peaks as compared with lesser aging temperature (400°C) for any aging period. The maximum peak hardness value for Ti-23% Mo alloy occurs at 30 per cent prior deformation aged at 400°C of value 390VHN while at 500° C that occurs at 30 per cent deformation (425 VHN) at 3 hr aging period.

## 4.3 MICROSTRUCTURES

Optical photomicrographs of as-received specimens of Ti-10% Mo and Ti-23% Mo alloys are shown in fig. 4.9. It was found that Ti-10% Mo alloy contains two phases alpha and beta whereas the Ti-23% Mo alloy also has ~~only one single phase beta~~. ~~An~~ alpha and beta phases but the amount of beta being far greater than the previous alloy. n hetrogenety was also found under microscope which might have been due to entrapped gas.

The martensitic needles along with some retained beta phase are observed in case of Ti-10% Mo alloy when it was quenched in water from solution treating temperature. These martensitic needles are fine in shape having directionally orientations as shown in fig. 4.10(A). Moreover, it is found in case of plastically deformed quenched specimens that martensitic needles are much wider with 20 per cent prior cold deformation (Fig. 4.10.B) and these needles are going to be finer one with increase in degree of prior cold deformation (Fig. 4.10.C, D). These martensitic needles in all plastically deformed specimens are randomly distributed.

On aging the as-quenched specimens of Ti-10% Mo alloy it is found that as the aging proceeds the coarsening of

(A) Ti-10% Mo Alloy

(B) Ti-23% Mo Alloy

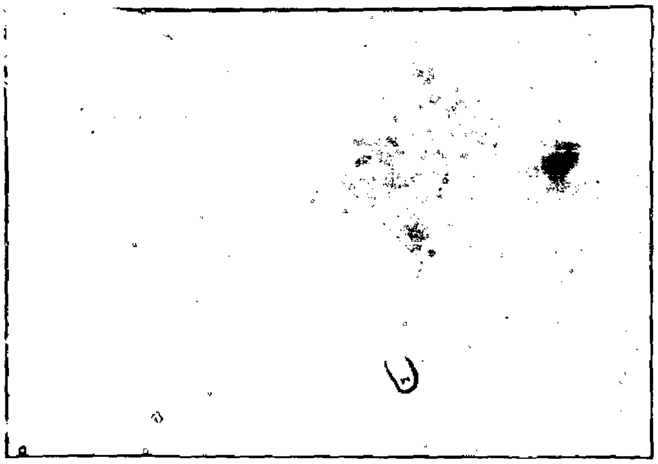
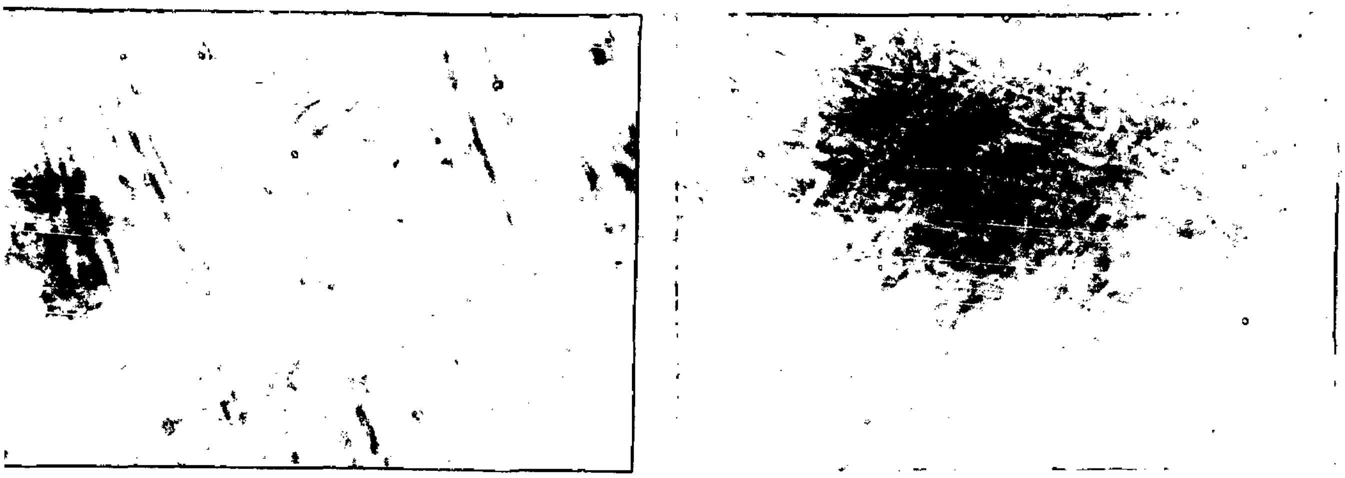


Fig. 4.9 Microstructures of Ti-10% Mo and Ti-23% Mo Alloys in the as-received condition (200 X)

(A) 0 pct. deformation (B) 20 pet deformation



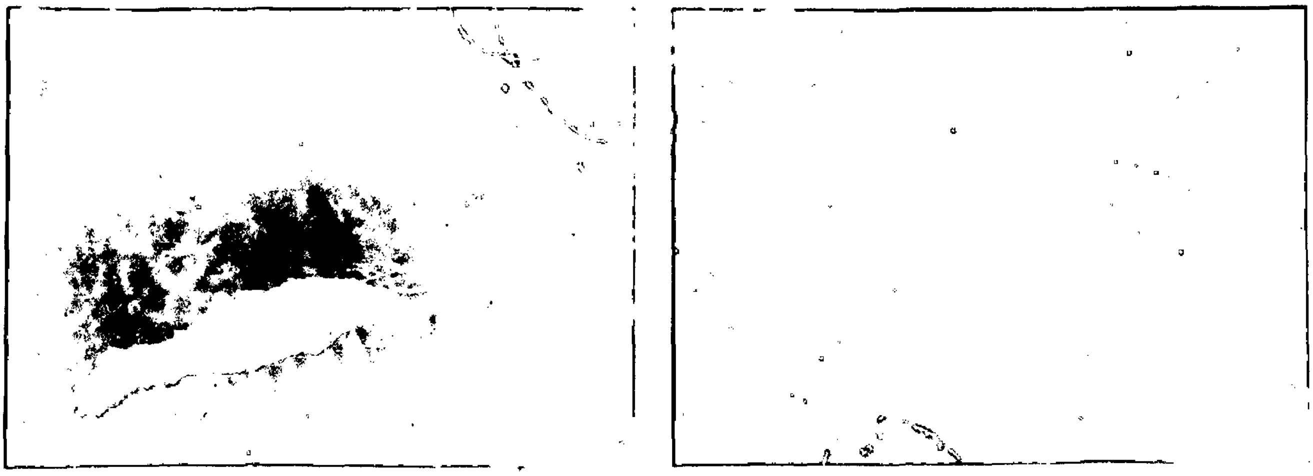
(c) 30 pct deformation (D) 50 pct. deformation

Fig. 4.10 Microstructures of water quenched Ti-10% Mo Alloy specimens with various prior cold deformations without aging (200X)

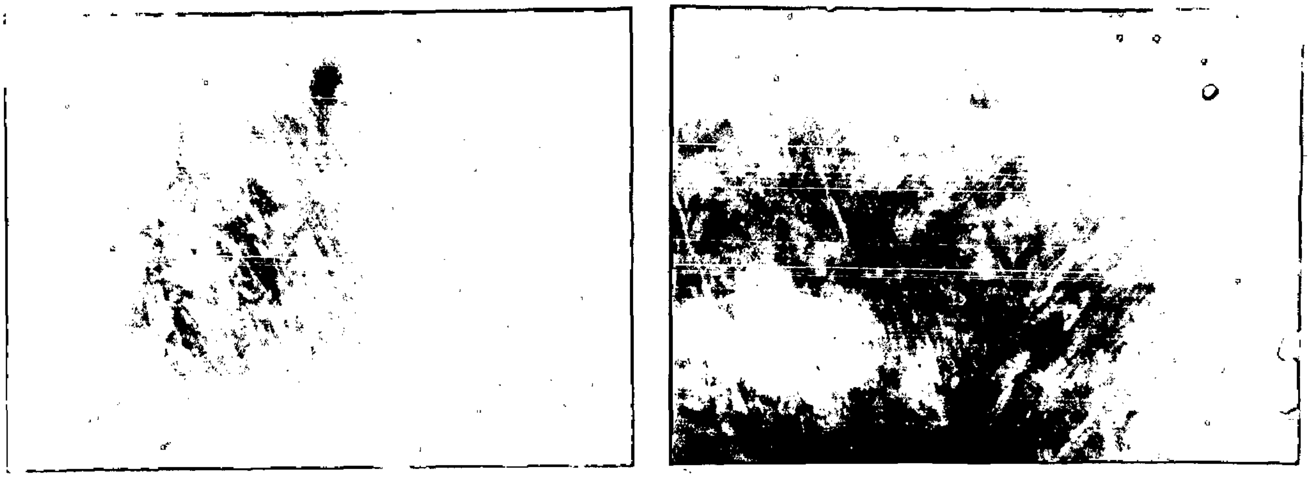
(A) 20 pct deformation

(B) 50 pct deformation

Fig. 4.11 Microstructures of Ti-10% Mo alloy specimens with various prior cold deformations after one hour aging at 400°C. (200X)



(A) 0 pct deformation (B) 20 pct deformation



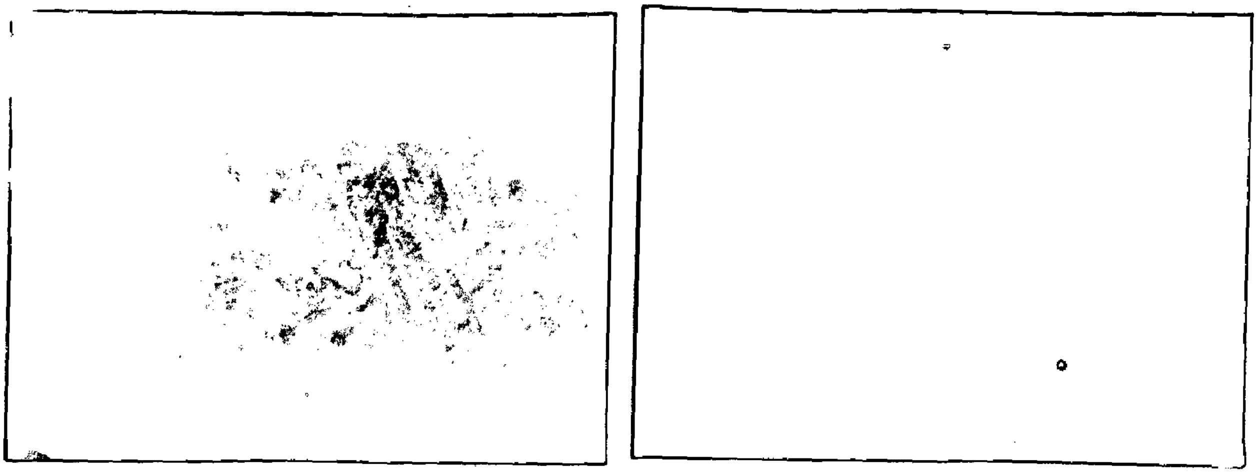
(C) 30 pct deformation (D) 50 pct deformation

Fig. 4.12 Microstructures of Ti-10% Mo alloy specimens with various prior cold deformations after one hour aging at 500°C (200X)

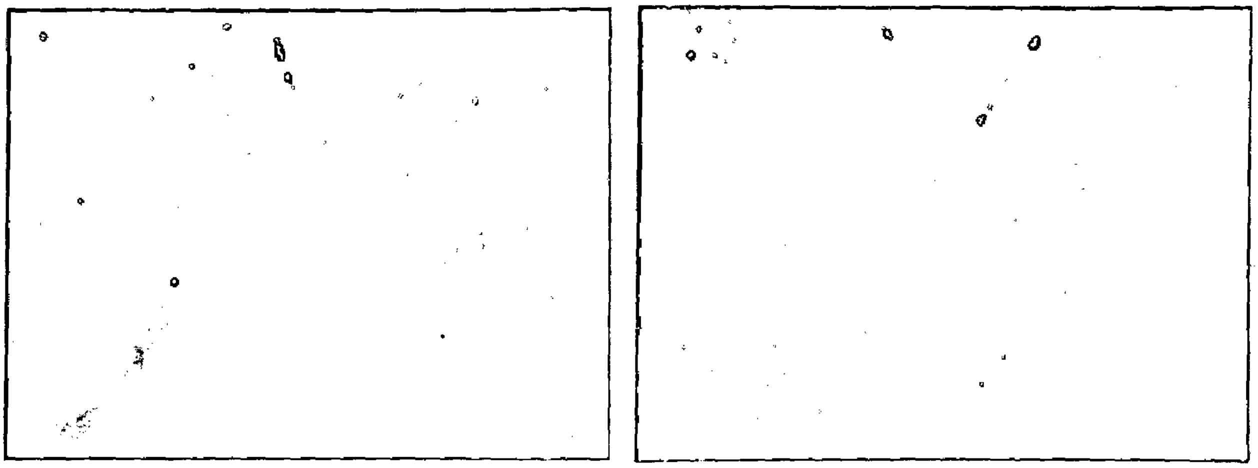
(A) 20 pct deformation

(B) 50 pct deformation

Fig. 4.13 Microstructures of Ti-10% Mo alloy specimens with various prior cold deformations after three hours aging at 400°C (200 X)

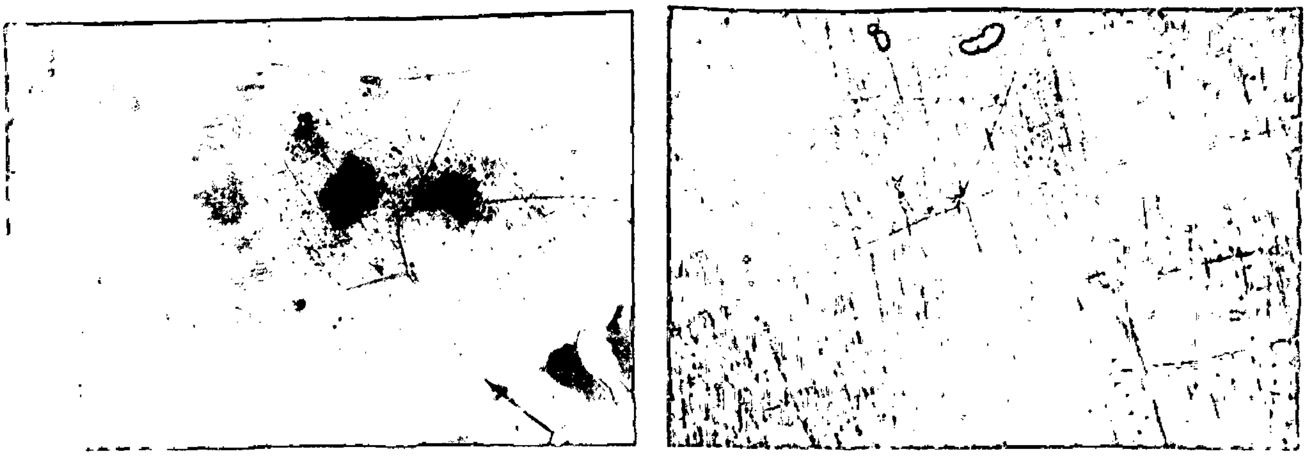


(A) 0 pct deformation (B) 20 pct deformation

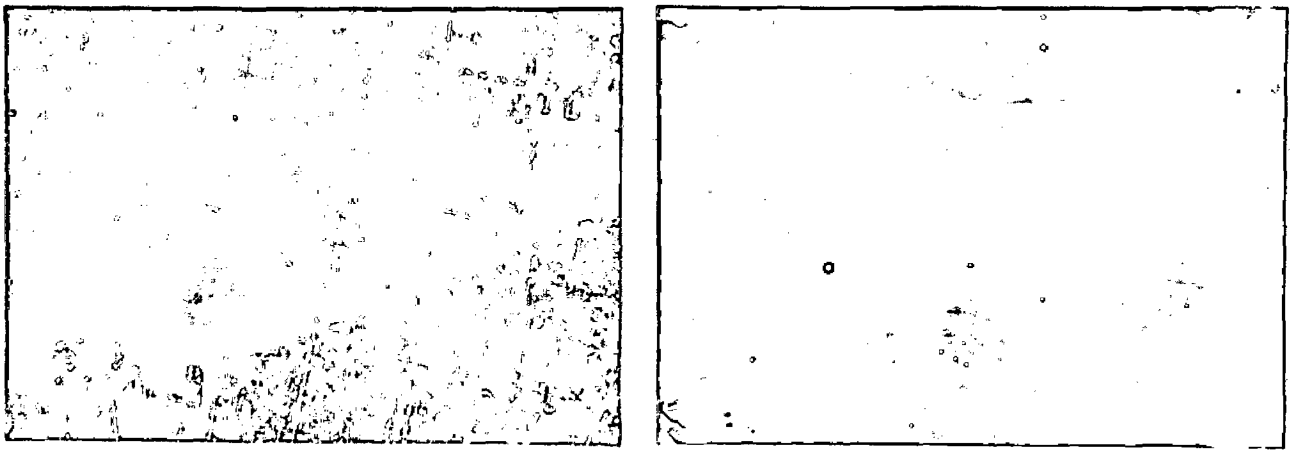


(C) 30 pct deformation (D) 50 pct deformation

Fig. 4.14 Microstructures of Ti-10% Mo alloy specimens with various prior cold deformations after three hour aging at 500°C (200X)

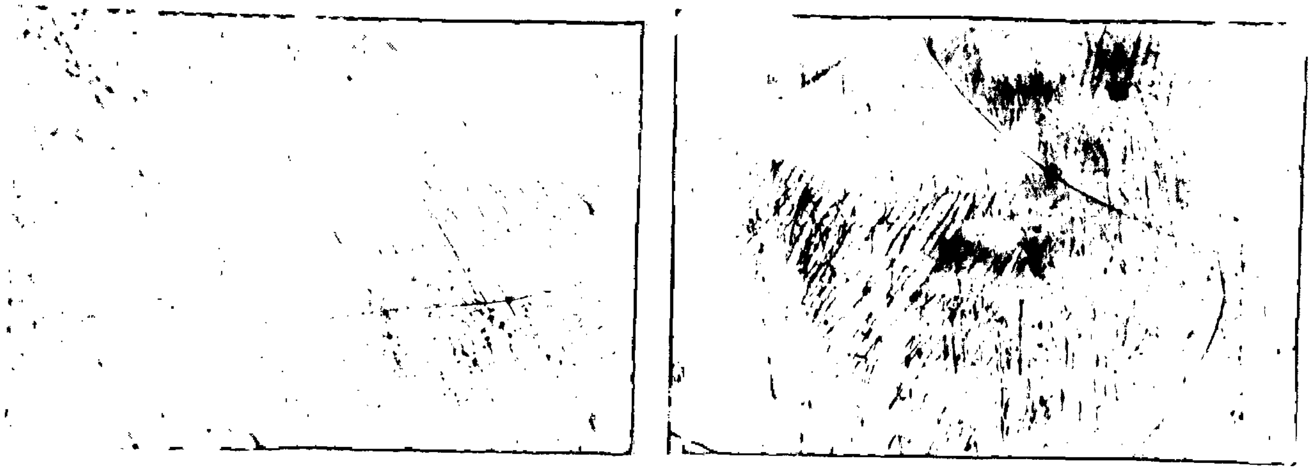


(A) 0 pct deformation (B) 20 pct deformation

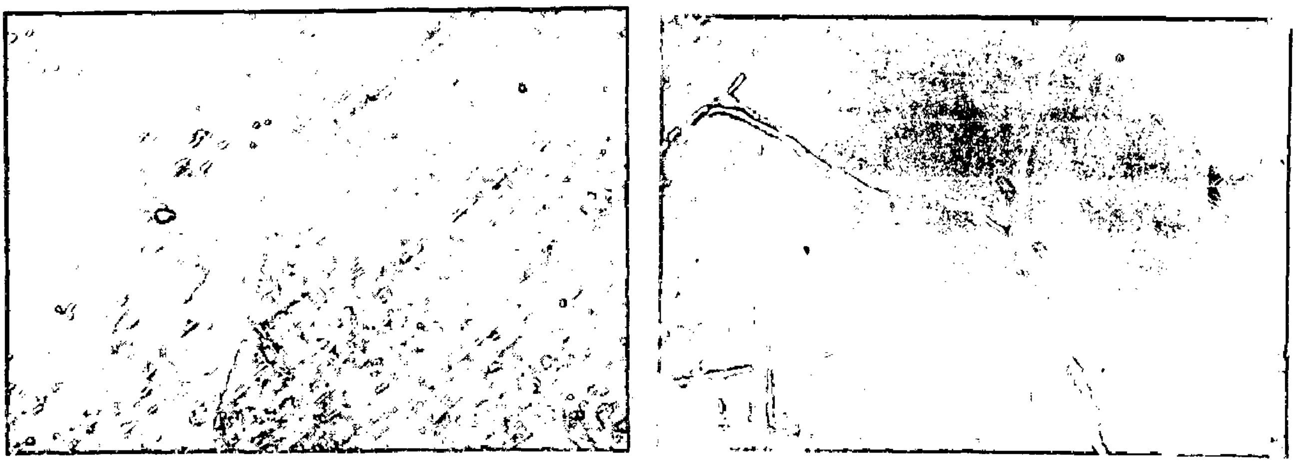


(C) 30 pct deformation (D) 50 pct deformation

Fig. 4.15 Microstructures of Ti-23% Mo alloy specimens with various prior cold deformations after four hours aging at 400°C (100 X)



(A) 0 pct deformation (B) 20 pct deformation



(C) 30 pct deformation (D) 50 pct deformation

Fig. 4.16 Microstructures of Ti-23% Mo alloy specimens with various prior cold deformations after four hours aging at 500°C (100 X) martensitic needles takes place at 500°C (fig. 4.12 A and 4.14 A).

In case of 20 per cent and30 per cent priorly cold deformed specimens it is found that with proceeding aging treatment martensitic needles dissolve with a faster rate at higher temperature (500°C) than at the lower temperature (400°C) (Fig. 4.11 - 4.14). However, the case with 50 per cent plastically deformed specimens is somewhat different. In this case as the aging proceeds the coarsening of martensitic needles takes place similarly as the case with the aging of as-quenched specimen of Ti-10% Mo alloy.

In case of Ti-23% Mo alloy, a single phase, retained $\beta$, is observed (fig. 4.15 - 4.16) which is common feature for all the specimens thermomechanically treated. Aging has not given any positive change in the microstructure.

# CHAPTER 5

## DISCUSSION

The essential feature of the quenched Ti-Mo alloys from solution treating temperature is the complete retained beta phase with higher concentration of molybdenum(6) ($>$11-12%) whereas below this concentration a martensitic transformed phase is also obtained and the sluggishness of transformation increases with increase in concentration of solute atoms(8). Plastic deformation may also lead to the formation of martensitic phase known as stress-induced martensite(10). The formation of omega phase(9) in quenched Ti-Mo alloys is of much importance where it increases the hardness of alloy markedly. The evidence of this omega phase formation may be felt in hardness vs aging period curves of our present investigation.

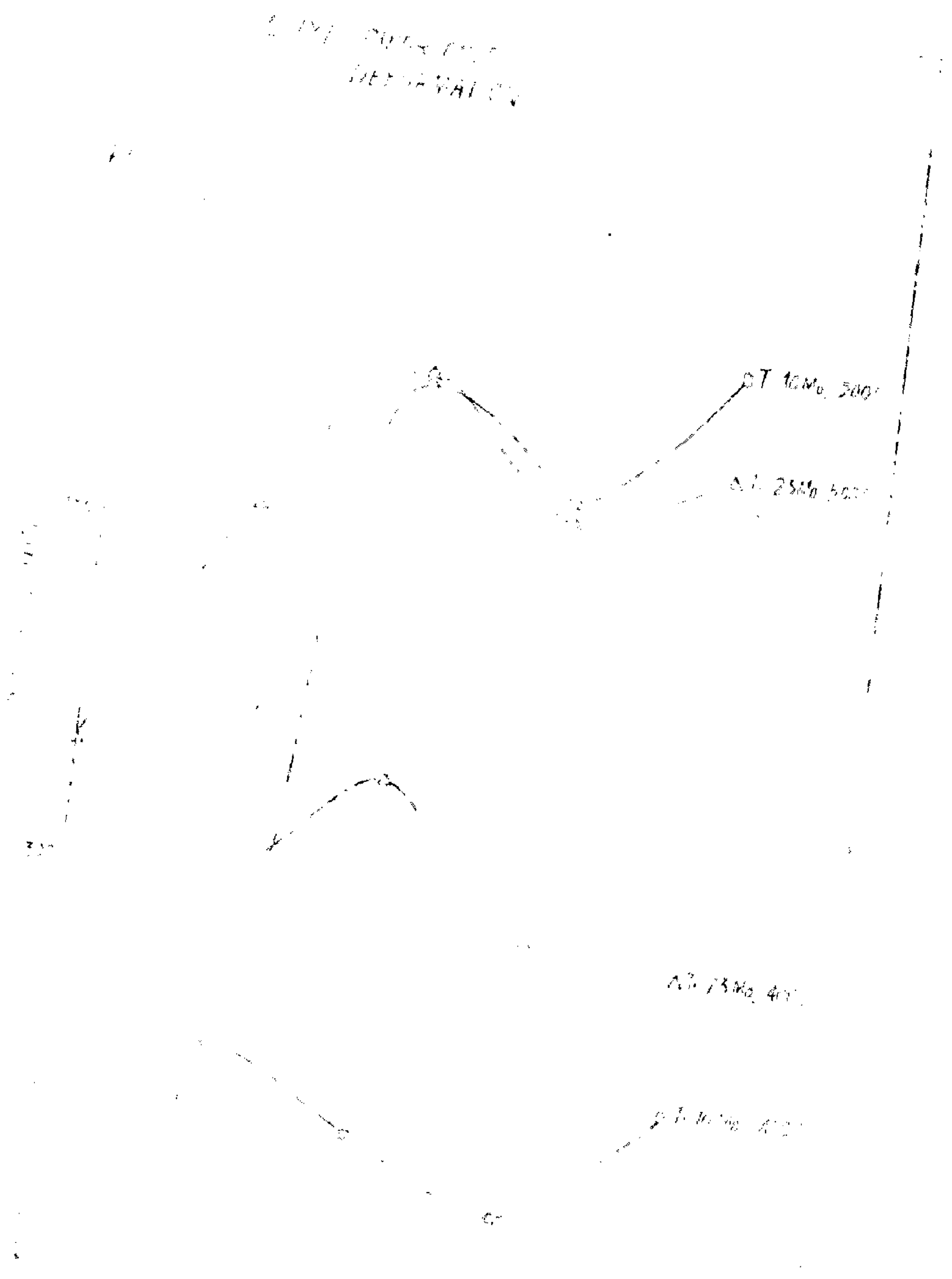
According to Fedetov(13) the hardness of martensitic phase is in between the hardness values of retained beta phase and omega phase and thus the rapid increment in hardness value is expected due to the presence of omega phase. According fig. 4.1 - 4.4, it is expected that initially the amount of omega phase is insignificant or ever absent just after quenching but during subsequent aging this phase increases in volume fraction(23). The evidence is an increase in hardness with increase in aging period. After a certain period it is seen that the hardness decreases. This according to Spachner et al(11) may be possible either due to the decomposition of omega phase into alpha and beta or due to the change in degree of coherency as there is no simple lattice relation between the alpha phase and the omega phase.

Wood(34) has established that $\omega$-phase forms in Ti-15% Mo alloy in greater quantities as cold deformation proceeds, at least until about 40 per cent strain has been effected when the quenched structure was $\beta$-phase. A similar behaviour has been noticed by us in case of Ti-23% Mo alloy, where hardness value increases as deformation increases in the range 0-30% deformation but it decreases at 50 per cent deformation. However, this case was not observed in Ti-10% Mo alloy, where the hardness values were continuously in increasing order with increase in deformation. This is expected to be due to the formation of martensite in large amount owing to the applied stress. In more heavily deformed specimens the structure of the transformed regions become more complex, two or more twining systems being found within a plate. The evidence is the micrographs of Ti-10% Mo alloy specimens deformed after quenching, fig. (4.10 - 4.14). These micrographs show the broad and twinned needles of martensite. Along martensite $\omega$-phase is also expected to form within the retained beta region in case of Ti-10% Mo alloy. The combining effect of decomposition of martensite and omega phase during aging is manifested in an increase in hardness with deformation.

In Ti-10% Mo alloy where martensitic phase is also present, there is expected the presence of omega phase in lesser amount than that in Ti-23% Mo alloy. Therefore, the hardness values of Ti-10% Mo alloy are lesser than the hardness values of Ti-23% Mo at either aging temperature (fig. 5.1. - 5.4).

The hardness peak periods of Ti-23% Mo alloy are observed after the corresponding peaks in Ti-10% Mo alloy. This delay in hardness peak periods may be expected only due to the sluggishness of transformation during aging with higher concentration of molybdenum.

The values of hardness in case of aging at 500° C are higher than the hardness values observed during aging at 400°C



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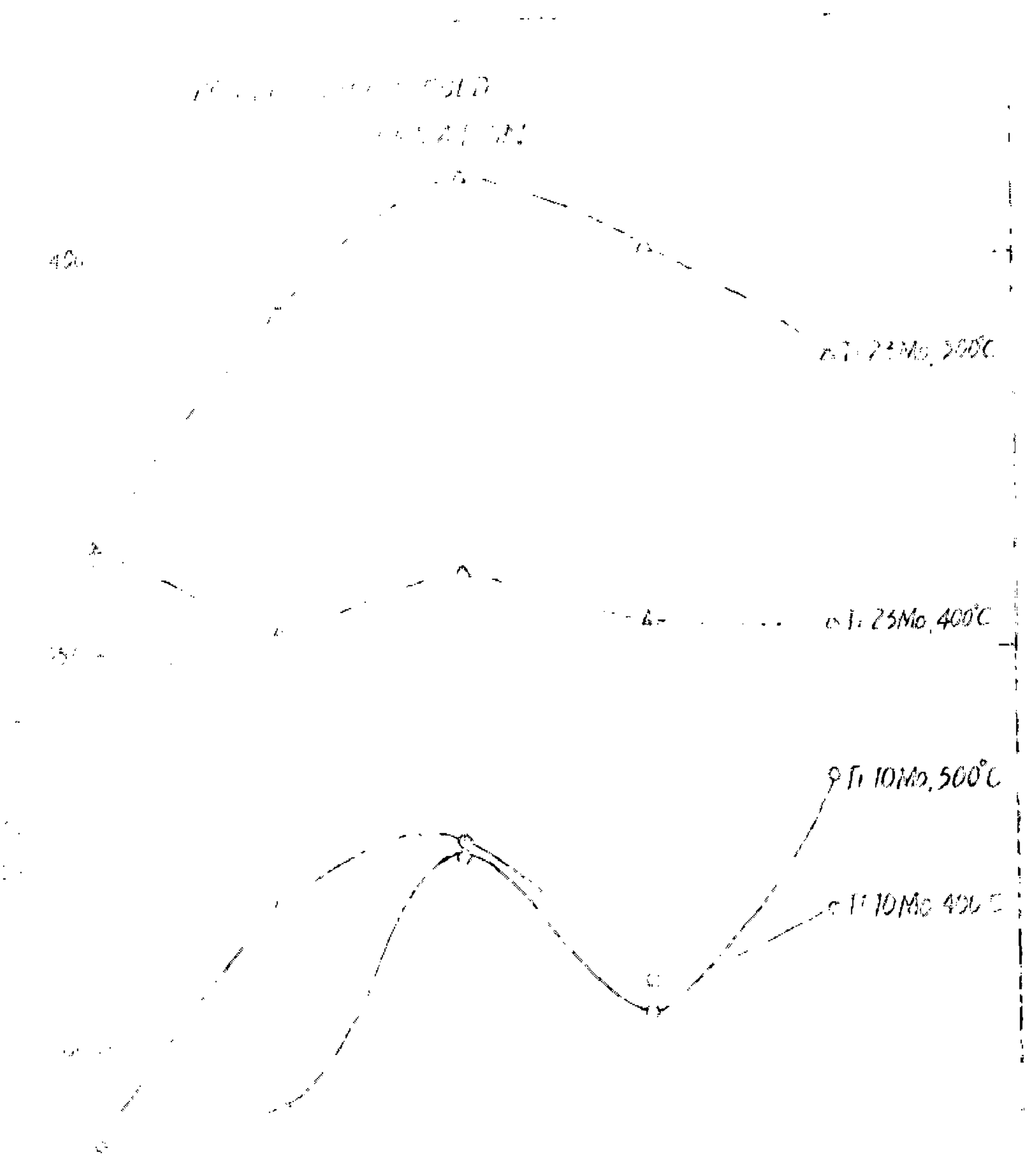
400
Ti 25Mo, 500°C
Ti 25Mo, 400°C
Ti 10Mo, 500°C
Ti 10Mo 400°C

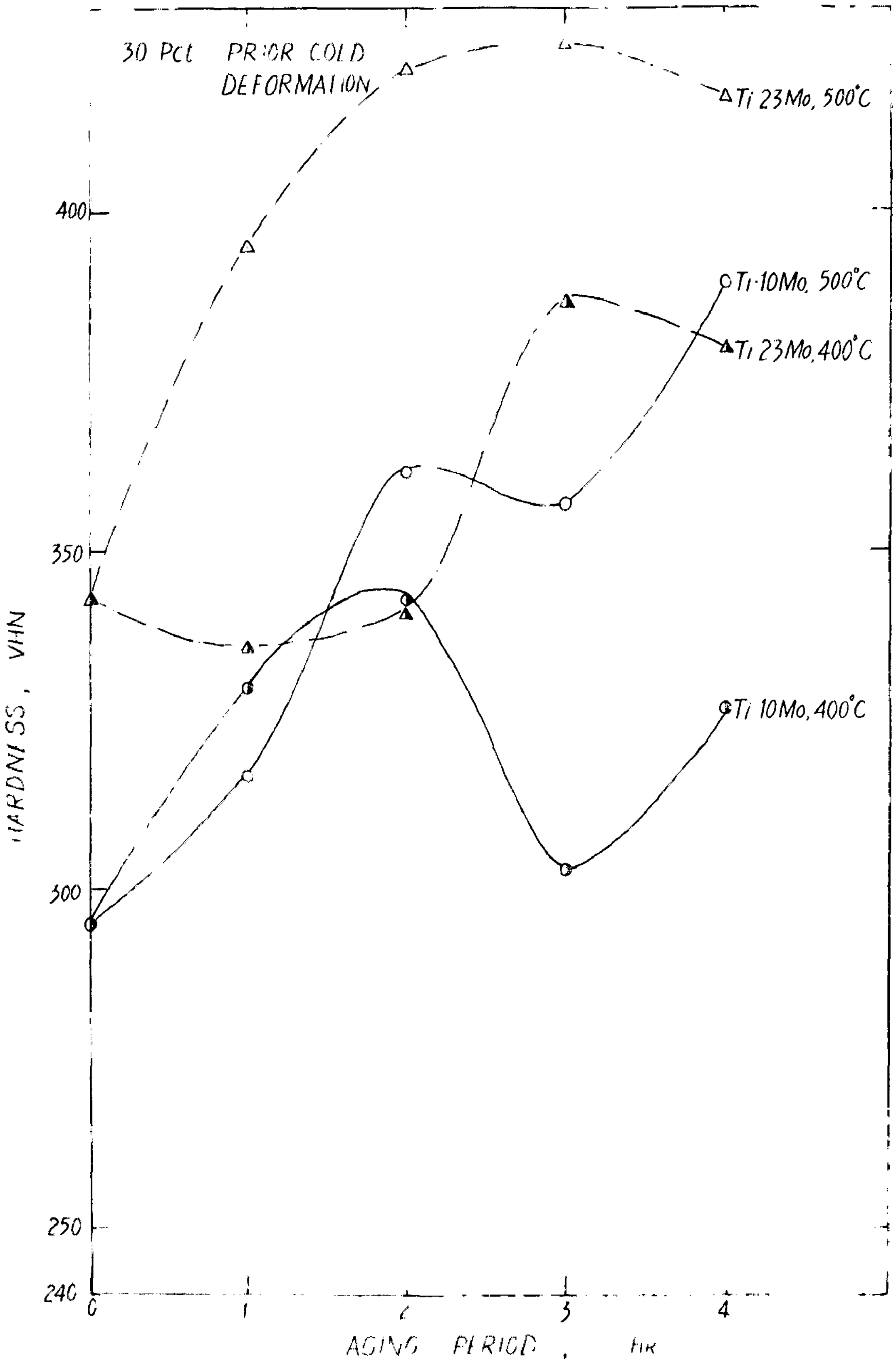

FIG. 5.5 EFFECT ON HARDNESS OF Ti 10% Mo AND Ti 23% Mo ALLOYS WITH AGING PERIOD AT 400°C AND 500°C RESPECTIVELY, 30 PER CENT DEFORMATION

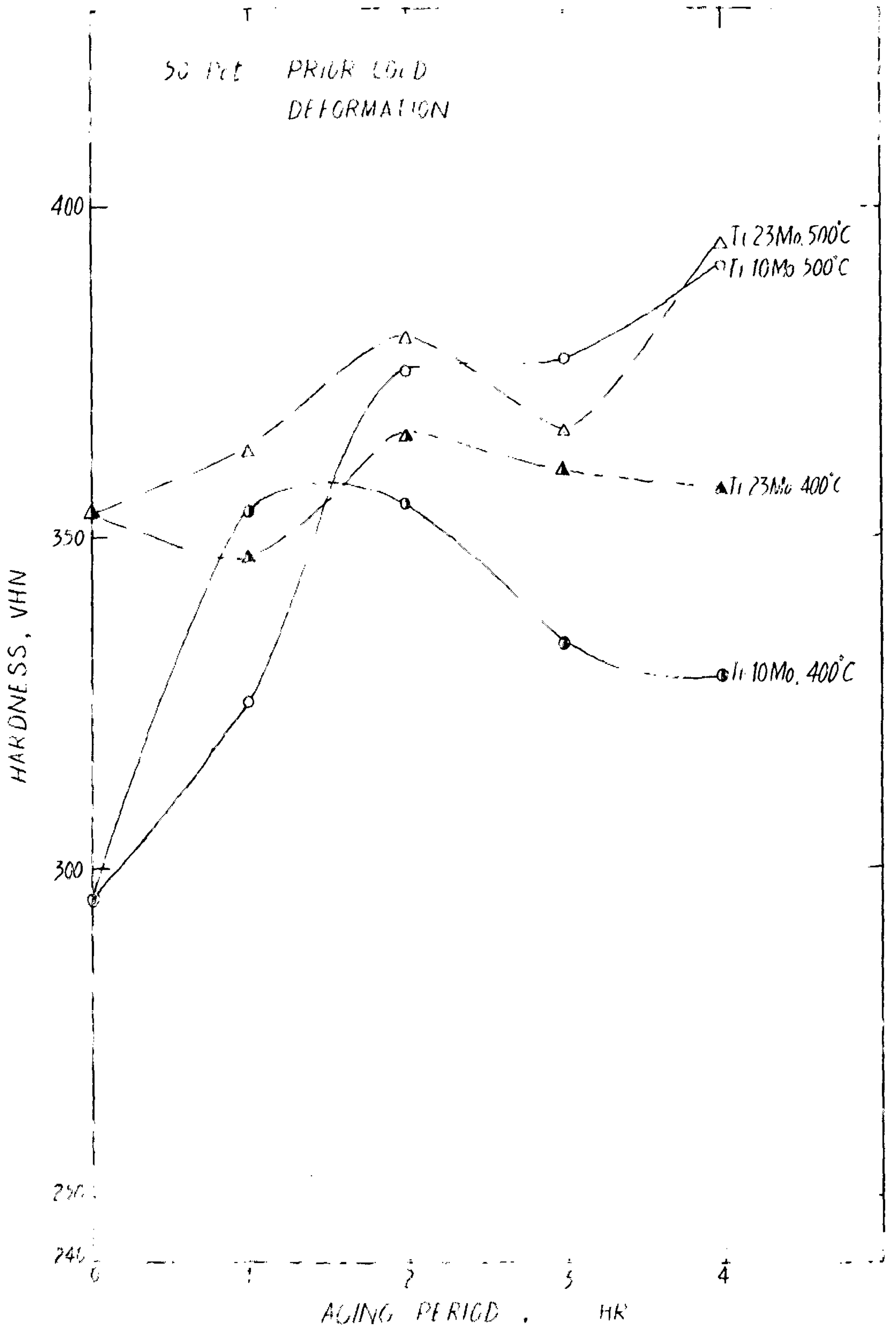

FIG. 5.4 EFFECT ON HARDNESS OF Ti 10%Mo AND Ti 23%Mo ALLOYS WITH AGING PERIOD AT 400°C AND 500°C RESPECTIVELY. 50 PER CENT DEFORMATION

for either alloys, which indicates that the transformation during aging is thermally activated one. According to Blackburn and Williams$^{(23)}$ on aging at quite higher temperature e.g. more than 500°C in case of Ti-4 at%Mo alloy (Ti-7.7 wt % Mo alloy) the omega phase may be absent. It appears from the general trend of the curves of our investigations that for shorter aging period the omega phase is in higher proportion at 500° C as-compared to 400°C.

The initial decrement in hardness values of the order of 10VHN for 1 hr aging period in case of Ti-23% Mo alloy (fig. 4.3) may be due to the more sluggishness of transformation for $\omega$-phase at a relatively lower temperature i.e. 400° C. The hardness decreases initially due to relief of thermal stresses produced during quenching, which presently appears to be predominant than the effect of omega phase formation.

# CHAPTER 6

# CONCLUSIONS AND SUGGESTIONS

## 6.1 CONCLUSIONS

1. For either aging temperature (400°C, 500°C) peak hardness value of Ti-10% Mo alloy is less than that of Ti-23% Mo alloy.

2. Aging at either temperature (400°C, 500°C) in case of Ti-23% Mo alloy does not vary the period ~~of~~ required to reach the hardness peak.

3. In case of Ti-10% Mo alloy, generally, there is a shift in hardness peak towards a higher period at 500°C as compared to that of 400°C aging temperature.

4. The peak hardness values of either alloys at 500°C aging temperature are higher than the corresponding values for 400°C.

5. The results established suggest the presence of maximum hardness for Ti-23% Mo alloy which was prior cold worked at the level of 30% and aged at 500°C for 3 hr.

## 6.2 SUGGESTIONS FOR FURTHER WORK

On the basis of the experimental results already obtained it is evident that scope of thermomechanical treatment of Ti-Mo alloys is very extensive. However, a number of parameters are still necessary to understand the picture completely and give a clear cut base. In this regard following types of work may be further extended.

1. A systematic study of phase identification by electron microscopy technique is required which may be very well extended to quantitatively estimate the volume fraction and size of omega phase at initial as well as later stages of aging.

2. X-ray study is also desirable to study the orientation relationship and crystal structure of stress induced martensite.

3. Resistivity measurements could be used to study the kinetics of phase transformation.

4. A detailed microhardness study is also desirable in order to evaluate the possible presence of different microconstituents.

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