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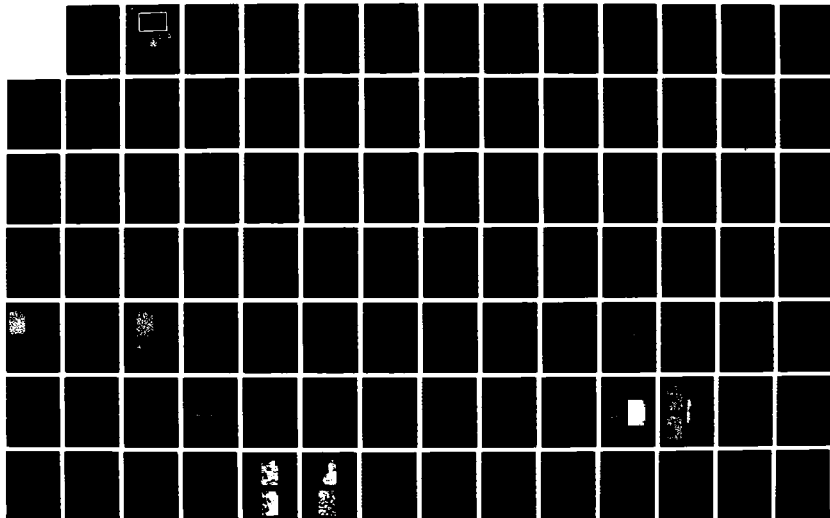
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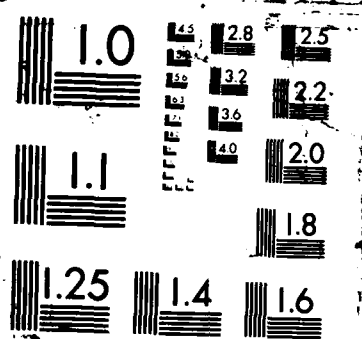
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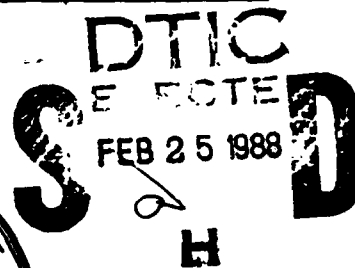
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INTERIM TECHNICAL REPORT

**SUBMITTED TO
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AIR FORCE OFFICE OF SCIENTIFIC RESEARCH
ELECTRONIC AND SOLID STATE SCIENCES
BOLLING AIR FORCE BASE
WASHINGTON, DC 20332**

**FOR THE PERIOD
DECEMBER 1, 1986 - NOVEMBER 30, 1987**

**FUNDAMENTAL STUDIES ON HIGH TEMPERATURE DEFORMATION
RECRYSTALLIZATION , AND GRAIN GROWTH
OF TWO - PHASE MATERIALS**

**BY
S.ANKEM, G.GREWAL, AND M.N.VIJAYSHANKAR**

AFOSR - 85 - 0367

JANUARY 1988

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REPORT DOCUMENTATION PAGE

1a. REPORT SECURITY CLASSIFICATION Unclassified		1b. RESTRICTIVE MARKINGS	
2a. SECURITY CLASSIFICATION AUTHORITY		3. DISTRIBUTION/AVAILABILITY OF REPORT Approved for public release, distribution unlimited	
2b. DECLASSIFICATION/DOWNGRADING SCHEDULE		Approved for public release,	
4. PERFORMING ORGANIZATION REPORT NUMBER(S)		5. MONITORING ORGANIZATION REPORT NUMBER(S) AFOSR-TR-88-0055	
6a. NAME OF PERFORMING ORGANIZATION University of Maryland	6b. OFFICE SYMBOL (if applicable)	7a. NAME OF MONITORING ORGANIZATION AFOSR/NE	
5c. ADDRESS (City, State, and ZIP Code) Dept. of Chemical & Nuclear Engineering University of Maryland College Park, Md. 20742-2111		7b. ADDRESS (City, State, and ZIP Code) Bldg. U10, Bolling Air Force Base Washington, D.C. 20332-6448	
8a. NAME OF FUNDING/SPONSORING ORGANIZATION AFOSR/NE	8b. OFFICE SYMBOL (if applicable)	9. PROCUREMENT INSTRUMENT IDENTIFICATION NUMBER AFOSR - 85 - 0367A	
8c. ADDRESS (City, State, and ZIP Code) Same 7b		10. SOURCE OF FUNDING NUMBERS PROGRAM ELEMENT NO. 61102F PROJECT NO. 2306 TASK NO. A1 WORK U ACCESSIC	
11. TITLE (Include Security Classification) Fundamental Studies on High Temperature Deformation, Recrystallization and Grain Growth of Two-Phase Materials.			
12. PERSONAL AUTHOR(S) S. Ankem, G. Grewal and M.N. Vijayshankar			
13a. TYPE OF REPORT Interim	13b. TIME COVERED FROM 12/01/86 TO 11/30/87	14. DATE OF REPORT (Year, Month, Day) February 4, 1988 JAN	15. PAGE COUNT 130
16. SUPPLEMENTARY NOTATION			

17. COSATI CODES			18. SUBJECT TERMS (Continue on reverse if necessary and identify by block number)
FIELD	GROUP	SUB-GROUP	

19. ABSTRACT (Continue on reverse if necessary and identify by block number)
During this period, work was concentrated on two areas: (1) Particle coarsening kinetics and (2) High temperature deformation studies. Isothermal particle growth studies were carried out on several Ti-Mn and Ti-V alloys at 973 K, 1023 K and 1073 K. It was found that, at any given temperature, the particle growth behavior could be expressed in terms of simple relationships between volume percent of phases and time. The growth of alpha and beta phases was attributed to a two way diffusion mechanism of the solute and solvent. Considerations of bulk diffusion alone did not explain the particle coarsening kinetics and hence the alternate mechanism of a mixed grain boundary plus bulk diffusion was proposed. Studies of the particle growth kinetics at different temperatures indicated that the temperature dependency of the particle sizes of the alpha and beta phase, at any given volume fraction, could be expressed in a simple Arrhenius type relationship. The studies seem to suggest that as the temperature of annealing is increased, there is a gradual change in mechanism from a predominantly grain boundary diffusion at 973 K to predominantly bulk diffusion at 1073 K.

20. DISTRIBUTION/AVAILABILITY OF ABSTRACT ☐ UNCLASSIFIED/UNLIMITED ☐ SAME AS RPT ☐ DTIC USERS		21. ABSTRACT SECURITY CLASSIFICATION U	
22a. NAME OF RESPONSIBLE INDIVIDUAL K. Shankar		22b. TELEPHONE (Include Area Code) (202) 761-4933	22c. OFFICE SYMBOL 87C

The high temperature deformation behavior at 973 K was found to be similar to that at 923 K for the six Ti-Mn alloys tested. Strain softening was observed in beta as well as in the alpha-beta alloys. The magnitude of flow stress drop (softening) increased with increasing beta volume percent as well as with an increase in the strain rate for a given beta volume percent. The high temperature (973 K and 923 K at a strain rate of 1.1×10^{-4} /Sec) flow stresses of Alloy 4, which has nearly equal volume percents of alpha and beta, were significantly lower than those expected by the law of mixture rule. This was attributed to a change in the deformation mechanism from slip in the alpha and beta single phase alloys to predominantly interface sliding in the alpha-beta alloys. This suggestion needs to be confirmed. It was further interesting to note that the flow stresses of Alloy 4 actually decreased with a decrease in temperature from 973 K to 923 K. This appears to be associated with a refinement in the size of the alpha and beta phases resulting in increased interface sliding activity. Experiments are under way to confirm this suggestion. The results to date indicate that the high temperature strength of alpha-beta alloys need not decrease with increase in temperature and depends on the volume percents and size of the constituent phases.

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1. INTRODUCTION

Two-phase materials are technologically important because optimum properties can be obtained by a proper combination of the two phases. Among these materials, two phase Titanium alloys are of particular interest for high temperature aerospace applications. However, there is a lack of understanding in the areas of high temperature deformation, recrystallization and grain-growth behavior of two-phase alloys in terms of the properties of the component phases. Such an understanding is essential to develop new titanium alloys with greater high temperature strength and stability for high temperature applications.

The lack of understanding is due to the complex deformation behavior of these two-phase materials. Whenever a material comprising two or more phases is subjected to stress, the component phases deform differently and this results in inhomogeneous strain and stress distributions. In addition, 'interaction stresses' develop as a result of interactions between the deforming phases. For these reasons, the deformation behavior of two-phase materials cannot be explained by the simple 'law of mixture rule'. This is due to the fact that the law of mixture rule always assumes either constant stress or constant strain and as mentioned before, this never happens in reality.

It is known that recrystallization is a nucleation and growth process and this process is promoted by increased cold work and high temperature. Therefore, the recrystallization process in single phase materials is somewhat straightforward. However, in two-phase alloys, where both the phases can deform, the

recrystallization behavior is not easy to understand. In these materials, the softer phase deforms more than the harder phase and therefore it can be argued that the recrystallization process is much faster in the softer phase than in the harder phase. However, the recrystallization process is also temperature dependent which implies that diffusion plays an important role. Therefore, if the diffusivity is higher in the harder phase, then it is difficult to predict which phase recrystallizes faster. This is particularly true in the case of some of the alpha-beta Titanium alloys where the beta phase is harder but its diffusivities are higher than that of the alpha phase. To further complicate this matter, the nature and magnitude of strain inhomogeneity depends also on the volume percent of the phases. To date, there is no model or mechanism which explains the recrystallization of two-ductile phase alloys in terms of the volume percent of the phases, strain inhomogeneities, and diffusivities of the component phases.

For many single phase materials, iso-thermal grain growth data can be represented by the empirical equation of the form

$$D = Kt^n$$

Where D is the mean grain diameter, t is the time, K is a constant of proportionality and n is the grain growth exponent. It is also known that impurity atoms in solid solution and impurities in the form of inclusions and second phase particles retard grain growth. Furthermore, in two-phase materials, size and volume fraction of the second phase is also known to effect the grain growth characteristics. However, there is no significant information as

to the empirical relationships which can predict grain growth behavior in two-phase materials. Recently, Anken and Margolin derived empirical relationships for grain growth of two-phase alloys in terms of the volume fractions of phases. Their derivations are based on the experimental results on alpha-beta titanium alloys. They found that under identical conditions, alpha-phase retards grain growth of the beta-phase much more effectively than vice-versa. Reasons were suggested for this behavior, but never proven. In addition, no atomistic or any other physical model was suggested for the grain growth of two-phase materials.

The aim of this program is to systematically study the high temperature deformation, recrystallization and grain growth behavior of two-phase materials. While this study will be focussed on two-phase materials where both the phases can deform, it is expected that the information and mechanisms to be developed can be applied to any two-phase material system. It is hoped that the outcome of this study will be useful in improving processing (forging, rolling, superplastic forming, etc.) methods, obtaining optimal microstructures for improved mechanical properties and increasing the high temperature capability of two-phase materials. Such a fundamental knowledge will be of great help in developing new titanium alloys for high temperature applications.

2. OBJECTIVES OF THE PROGRAM

1. Determine the high temperature (923 - 1023 K) deformation behavior of two-phase Titanium-Manganese and Titanium-Vanadium alloys. The factors to be varied include volume percent and nature of second phase (strength, diffusivity), temperature and strain rate.
2. Determine the high temperature deformation mechanisms in the two-phase titanium alloys: Slip, interface sliding, etc.
3. Study the recrystallization behavior of these two-phase titanium alloys by taking into account the inhomogeneous strain distributions and diffusivity differences between the component phases.
4. Study the particle coarsening and grain growth mechanisms and develop grain-growth relationships for two-phase alloys in terms of the volume fraction of phases, diffusivities, temperature, etc.
5. Theoretically Predict the stress-strain curves and stress and strain distributions in various two-phase materials.
6. Based on this information, propose models for high temperature deformation, recrystallization and grain growth of two-phase materials.

3. RESEARCH PROGRESS

In this report, research progress made during the second period of the program, December 1, 1986 - November 30, 1987, is presented. During this period work was concentrated on two areas

(A) Particle Coarsening Studies

(B) High Temperature Deformation Studies

The results of these studies are presented in separate sections and each section is entirely self contained in all respects.

(A) PARTICLE COARSENING STUDIES

(a) ISOTHERMAL PARTICLE GROWTH IN TWO PHASE TITANIUM ALLOYS

G. Grewal* and S. Ankem **

Isothermal particle growth studies were carried out on several Ti-Mn and Ti-V alloys consisting of varying amounts of α and β phases at 973⁰ K. It was found that the particle growth kinetics of the α and the β phases, in titanium alloys, could be represented by simple equations in terms of time and volume percents of the phases. The growth process is presented as a two-way diffusion process where the solvent and the solute atoms move in opposite directions resulting in conversion of α to β and β to α leading to the growth of the particles. The LSW theory, which considers only solute diffusion for growth of particles, is slightly modified to incorporate the diffusion of solute and solvent into the growth equation. The observation of the growth kinetics indicated that under identical conditions the growth of α particles in a β matrix was faster than the growth of β in an α matrix. Furthermore, for identical conditions, the growth kinetics of Ti-Mn are faster than those of the Ti-V alloys. While the faster kinetics are consistent with higher bulk interdiffusivities, the magnitude of the differences could not be explained on the bulk diffusivities alone. This was attributed to a mixed mode diffusion mechanism consisting of bulk diffusion as well as grain boundary diffusion controlling the growth process at 973⁰ K. Details of the investigation and various growth models of α - β Titanium alloys are presented.

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I. INTRODUCTION

For many single phase materials, isothermal grain growth data can be represented by equations of the form ---

$$D = K t^n \text{-----} (1)$$

where D is the mean grain diameter, t is the time, K is the constant of proportionality and n is the grain growth exponent. It is also known that impurity atoms in solid solution and impurities in the form of inclusions or second phase particles retard grain growth. Though it is known that the size and the volume fractions of the second phase affects the growth in two phase alloys (1 - 5), up until recently there was no significant empirical information available on the growth relationships in these alloys (6). Ankem and Margolin (6) derived empirical relationships for grain growth of the two - phase alloys in terms of the volume fractions of the phases. Their derivations, which are based on experimental work on two - phase α - β Titanium alloys, showed that under identical conditions, α retards grain growth of β phase much more effectively than vice - versa. Though reasons based on diffusivity considerations were given, they were not proven. Their study was limited to grain growth. They did not consider the particle coarsening kinetics. It is necessary to understand the particle coarsening kinetics to rationalize the grain growth kinetics since particles at the grain boundaries control the growth of grains.

To date, various two phase systems have been investigated in the literature to understand the particle coarsening kinetics. The two phase systems which have been theoretically investigated are the typical ostwaldian type of systems. A typical ostwaldian system is a two - phase system constituting a solute lean matrix dispersed with solute rich particles. A statistical dispersion of particles in a matrix is thermodynamically unstable due to the excess free energy associated with their interfacial surface area and thus the system tends to decrease this free energy by the process of particle coarsening. The solubility or the dissolution of these particles depends on their radii of curvature and is described by the Gibbs - Thompson relation (2).

$$C_r = C_{\infty} (1 + 2 \gamma V_m / R T r) \text{-----} (2)$$

where:

C_r = Concentration in matrix of solute near a particle of size r

C_{∞} = Equilibrium solute concentration in matrix.

γ = Surface free energy

V_m = Molar volume

R, T = Gas constant and temperature respectively

The fundamental theory of particle coarsening in a supersaturated solid solution decomposing by diffusion has been developed by Lifshitz and Slyosov (7) and independently

by Wagner (8), and is referred to as the LSW theory. The basic growth mechanism of the LSW theory is assumed to be a long range concentration gradient driven diffusion of solute through the matrix, between the second phase particles. The concentration gradients are created locally in the matrix near the particle matrix boundaries as a result of the dependence of the equilibrium solute concentration, $C(r)$ on the radius of the particles. The LSW theory was derived for the limiting case of the second phase tending to zero (7, 9) and is thus generally not valid for finite volume fractions of the second phase.

To bring real, finite second phase volume fraction systems into the framework of the LSW theory, modifications were made by Ardell (9). This modified LSW theory, known as the MLSW theory predicts that while the basic $t^{1/3}$ kinetics of the LSW theory are maintained, the coarsening rate increases with increasing volume fraction. This result is valid at even very small volume fraction of the second phase (9).

The Ardellian modification is based on the assumption that the diffusion of the solute to a growing particle will depend on a distance characteristic of the spatial distribution of particles in the matrix. In addition, the assumption of steady state diffusion under conditions of spherical symmetry are assumed (9). A somewhat different approach than that used by Ardell (9) was used by Davies, et al., (10) to incorporate the effect of finite volume fraction second phases into the framework of LSW theory. The rationale of the Davies (10) modification is the observation that the Ardellian approach predicts a strong volume fraction dependent growth kinetics which are generally not experimentally observed. The central idea of the Davies approach is that of "encounters" between growing particles. A growing particle encounters another growing particle, and a strong diffusion interaction results as a consequence of which the particles spontaneously coalesce to become a single particle. This theory, as modified by Davies, et al. is referred to as the LSEM theory (10). A more recent work by Brailsford, et al. (11) developed a theory of particle coarsening, which assumes that the growth rate of particles of one size class is a function of the entire size distribution and the existing concentration gradients around the particles. The growth rate predicted by this theory is more sensitive to volume fraction of the second phase than the theory proposed by Davies but is less sensitive to the theory proposed by Ardell.

Apart from the above theories, various models have been developed to explain the shape of the particle size distributions curves observed in real finite volume Ostwaldian type of systems. These models essentially use different geometrical approaches to obtain expressions for the radius of the influence sphere. The expression for the influence sphere is then substituted into the standard LSW theoretical formulation and the kinetic expression is evaluated by solving the resulting equation. An excellent paper which develops six such models is that of Tsumuraya, et al. (12). The particle coarsening theories reviewed above do not consider the diffusion of the Solvent. Furthermore, diffusion through the particles is assumed not to occur. In addition, the starting point of the LSW theory assumes a spherically symmetrical concentration gradient around a growing or shrinking particle and diffusion of solute occurs to the particle under steady state conditions (9).

In the two phase alloys, such as α - β titanium alloys, the geometry and distribution of the phases is complex and hence the conditions of spherical symmetrical symmetry cannot

be made. Furthermore, neither of the two phases can be considered as solute rich or solute lean and thus for the growth process to occur the diffusion of the solute as well the solvent have to be considered. In addition, as the volume percent of the 'second phase' becomes considerable, the growth process occurs by the diffusion of the solute and the solvent through both the phases. This is in marked contrast to particle coarsening where diffusion is not allowed to occur through the second phase dispersoids.

The aim of this study was to systematically study the effect of volume percents of phases, temperature and diffusivity of the alloying elements on the particle growth kinetics of $\alpha - \beta$ titanium alloys. It is hoped that such an understanding will be helpful for optimizing the microstructure of $\alpha - \beta$ titanium alloys through thermo-mechanical treatments and in developing new titanium alloys with increased microstructural stability for high temperature applications.

II. EXPERIMENTAL

A. MATERIALS

For this study six *Ti-Mn* and six *Ti-V* were used. The alloys were melted as 13.6 kg (30 - lb) ingots at RMI company. The aim alloy compositions are indicated in Figures 1 and 2 respectively. The actual chemistries of the ingots are given in Tables 1 and 2. Note that the actual chemical compositions of the ingots are close to the aim compositions in most of the cases. All of the Ingots were processed to 1.74 cm diameter by forging and rolling. The final step (rolling) was carried out in the $\alpha - \beta$ field at 973°K and a 60 % reduction in area was given for all the alloys. This work was found to be sufficient to recrystallize all of the alloys within 6 hours at 973°K.

B. HEAT TREATMENT

The heat treatments were carried for 25, 50, 100, 200 and 400 hours at 973°K, followed by water quenching. The treatments were carried out in vacuum encapsulated quartz tubes. The specimen were sealed, in batches of twelve, at a pressure better than or equal to 10^{-5} torr

C. OPTICAL METALLOGRAPHY

Standard polishing and etching techniques were used. The *Ti-Mn* alloys were etched with A- etch for 10 seconds, followed by etching with R- etch for about 10 seconds. This two - step etching was found to be beneficial for $\alpha - \beta$ microstructures. The compositions of A- etch and R- etch are given below:

A- etch: 25 ml HF 50%
 25 ml HNO₃ conc
 50 ml Glycerine

R- etch: 18.5 gm (17 ml) Benzalkonium
Chloride (50%)
35 ml Ethanol

The *Ti-V* alloys were etched with the following etchant for 6 to 10 seconds.

4 ml HNO_3 conc.
2 ml HF 50%
Bal. H_2O

The particle sizes and the volume fractions of the phases were measured by the linear intercept method (13).

III. RESULTS

The volume percents of alpha, for the $\alpha - \beta$ *Ti-Mn* alloy system, varied from 10% for alloy no.# 5 to 79% for alloy no.# 2. The volume percents of alpha for the $\alpha - \beta$ *Ti-V* alloy system varied from 7% for alloy no.# 11 to 70% for alloy no.# 7. It is pertinent to note that though the volume fractions of the alloys studied are different, the chemical compositions of the alpha and the beta phases is the same in all the alloys in each of the two systems studied, since all of them were heat treated at the same temperature, see Figs. 1 and 2. The chemical composition of the phases, in the two system studied is shown by the ends of the tie lines drawn for 973°K, in Figs. 1 and 2 respectively.

The heat treatments resulted in equiaxed $\alpha + \beta$ microstructures of *Ti-Mn* and *Ti-V* alloys. For example the microstructures, of the $\alpha - \beta$ *Ti-Mn* alloys, heat treated for 200 hours at 973°K are shown in Fig. 3. Fig. 4 (a) shows the time evolution of a *Ti-Mn* alloy. In Fig. 4(b) a comparison of particle sizes is made between a *Ti-Mn* and a *Ti-V* alloy of same volume fractions of the α and the β phases. As expected, the α and the β sizes increased as a function of time. The particle sizes of the α and β phases plotted as a function of time for alloys 2, 3, 4 and 5 of the *Ti-Mn* system and for alloys 7, 8, 9, 10 and 11 of the *Ti-V* system are shown in Figs 5 to 8

It was found that in the *Ti-Mn* system, particle growth of the α and β phases could be expressed by equations of the form

$$D_{\alpha}^{\alpha-\beta} = K_{\alpha}^{\alpha-\beta} t^n \text{-----} (3)$$

$$D_{\beta}^{\alpha-\beta} = K_{\beta}^{\alpha-\beta} t^n \text{-----} (4)$$

where $D_{\alpha}^{\alpha-\beta}$, $D_{\beta}^{\alpha-\beta}$ are particle sizes of the α and the β phases respectively in microns, t is the time in hours, and n is the growth exponent. $K_{\alpha}^{\alpha-\beta}$, $K_{\beta}^{\alpha-\beta}$ are growth parameters

for the α , and the β phases respectively.

The growth exponent n gave a constant value of approximately .28 for the Ti-Mn alloys studied. It was further more found that the growth parameters $K_{\alpha}^{\alpha-\beta}$, and $K_{\beta}^{\alpha-\beta}$ depended on the volume percents of the respective phases. It was found that the growth parameters could be represented by equations of the form.

$$K_{\beta}^{\alpha-\beta} = K_{\beta} / V_{\alpha}^{m_{\alpha}} \text{-----} (5)$$

$$K_{\alpha}^{\alpha-\beta} = K_{\alpha} / V_{\beta}^{m_{\beta}} \text{-----} (6)$$

where K_{β} , K_{α} are constants of the β and the α phases respectively. V_{α} , V_{β} are volume percents and m_{α} and m_{β} are known as the retardation exponents as per the terminology introduced by Ankem and Margolin (6). Plots of $K_{\alpha}^{\alpha-\beta}$ versus V_{β} and $K_{\beta}^{\alpha-\beta}$ versus V_{α} are shown in Fig. 9(a). The numerical form of the equations (5), and (6) obtained for the Ti-Mn system are:

$$K_{\beta}^{\alpha-\beta} = 105 / V_{\alpha}^{1.146} \text{-----} (7)$$

$$K_{\alpha}^{\alpha-\beta} = 23 / V_{\beta}^{.738} \text{-----} (8)$$

It was found that equations identical to (3), (4), (5) and (6) could be written for the Ti-V system, with only the constants having different values. The growth exponent n gave an approximately constant value of .24 for all the Ti-V alloys studied. Plots of $K_{\alpha}^{\alpha-\beta}$ versus V_{β} and $K_{\beta}^{\alpha-\beta}$ versus V_{α} are shown in Fig. 10(a). The numerical form of equations (5) and (6) for Ti-V system is given below

$$K_{\beta}^{\alpha-\beta} = 89 / V_{\alpha}^{1.120} \text{-----} (9)$$

$$K_{\alpha}^{\alpha-\beta} = 22 / V_{\beta}^{.757} \text{-----} (10)$$

IV. DISCUSSION

Comparison of equations (7) and (8), which correspond to the Ti-Mn system, show two differences. The first difference pertains to the numerators K_{β} and K_{α} respectively in

equations (7) and (8). The K_β value is larger than the K_α value by a factor of almost 4.5. The second difference is in the exponents m_α and m_β . We note that m_α is larger than the value of m_β . The isothermal growth parameters $K_\beta^{\alpha-\beta}$ and $K_\alpha^{\alpha-\beta}$, in equations (7) and (8) respectively, are scaled measures of the sizes of the β and the α phases respectively. For values of V_α greater than about 50%, the isothermal growth parameter $K_\beta^{\alpha-\beta}$, in equation (7), is a measure of the size of particulate β in an α matrix. Similarly for values of V_β greater than about 50% the isothermal growth parameter $K_\alpha^{\alpha-\beta}$ in equation (8) is a measure of the size of particulate α in a beta matrix.

Thus the values of $K_\beta^{\alpha-\beta}$, and $K_\alpha^{\alpha-\beta}$ in the right hand side of Fig. 9(a) are a measure of particulate β and α , respectively. The values of $K_\beta^{\alpha-\beta}$ and $K_\alpha^{\alpha-\beta}$ on the left hand side of Fig 9(a) correspond to the mean spacing of the matrix. The corresponding sizes of matrix phases and the particulate phase of each of the alloys studied in the *Ti-Mn* system have been indicated by dashed arrows in Fig. 9(b). From Fig. 9(a) or 9(b) one infers that for any given volume percent of the matrix phase the particle size of the α phase is larger than the particle size of the β phase. The ratio $K_\alpha^{\alpha-\beta}/K_\beta^{\alpha-\beta}$ has a maximum value of 1.43 for the limiting case of $V_\alpha = V_\beta = 100\%$. Furthermore, as the volume percent of the matrix phase is continuously reduced, the ratio $K_\alpha^{\alpha-\beta}/K_\beta^{\alpha-\beta}$ decreases and becomes unity at $V_\alpha = V_\beta = 42\%$.

Analysis of the *Ti-V* system shows a trend similar to the *Ti-Mn* system. Thus, an examination of equations (9) and (10) show that K_β is larger than K_α , and the value of M_α is larger than the value of M_β . The right half of Fig. 10(a), which represents the isothermal growth parameters of particulate β and α , shows that for any given volume percent of the matrix phase, the particle size of the α phase is larger than the particulate size of the β phase. The corresponding sizes of the matrix and the particulate phase for each of the *Ti-V* alloys studied are indicated by dashed lines in Fig. 10(b). The ratio $K_\alpha^{\alpha-\beta}/K_\beta^{\alpha-\beta}$ of the Isothermal growth parameters for particulate α to particulate β respectively, attains its maximum value of 1.31 for the limiting case of $V_\alpha = V_\beta = 100\%$. Furthermore, as the volume percents of the matrix phase is continuously reduced, the ratio $K_\alpha^{\alpha-\beta}/K_\beta^{\alpha-\beta}$ decreases and becomes unity at $V_\alpha = V_\beta = 47\%$.

A. RATIONALE FOR THE GROWTH PROCESS OCCURING IN TWO PHASE ALLOYS

A two phase alloy, with a fine dispersion of the two phases is thermodynamically unstable due to its large interfacial area. To reduce the excess free energy associated with the interphase interfaces, the two phase system tends toward a state where the surface area is minimized. The process occurs by an exchange of fluxes of the constituent components

between nearest neighbor interfaces. Thus the interfaces move by the process of conversion of one phase to the other. Thus in an α - β two phase system, regions of α phase convert to β and regions of β convert to α and the interfaces move. Since the volume fractions of the two constituent phases remains invariant with time, the rate of conversion of α to β and vice versa for the whole system is zero. The mutual exchange of fluxes between the α - β interfaces can only occur by the process of long range diffusion. Depending on whether the diffusion process predominantly occurs in the α , the β or in both the α and the β Phases, three distinct classes of the coarsening processes can be identified. These three classes, together with the characteristics of the type of alloys which belong to them are discussed in detail below. The *Ti-Mn* alloy system has been used as the model system for the sake of the exposition.

CLASS 1: DIFFUSION PROCESSES OCCURRING PREDOMINANTLY THROUGH THE β PHASE.

The diffusion processes, resulting in the conversion of phases at the interfaces, will predominate in the β phase when the total bulk volume fraction of the α phase is small. Thus the two phase alloy microstructures belonging to class 1 are those which contain small volume fractions of α distributed in β matrix.

Fig. 11(a) shows a small volume fraction of α particles distributed in a matrix of β . The α particles are mostly distributed at beta boundaries. To bring out the details of the atomistic process resulting as a consequence of flux exchange between interfaces, we consider a two particle system in Fig. 11(b).

From Figs. 1 and 2 we note from the tie lines at $973^{\circ}K$ in the two systems that the β phase is solute rich and the α phase is solute lean. If one assumes that the growth of particle *C* at the expense of particle *A* does not involve the creation of a vacancy concentration higher than the already present equilibrium vacancy concentration, then the growth process will involve the following steps.

- (1) Atoms of *Ti* in the shrinking α particle *A*, jump across the α - β boundary and diffuse to the β regions near the growing α particle *C*.
- (2) Atoms of *Mn* in the beta matrix near the growing particle *C* diffuse towards α particle *A* and jump across the α - β boundary of particle *A*.
- (3) Atoms of newly formed α near the interface of particle *C* jump across the α - β interface into particle *C*.

As a consequence of these series of events α particle *A* shrinks with a part of it transforming to β and the α particle *C* grows by a part of the β matrix adjacent to it transforming to α . The diffusion process thus involves the following fluxes—



The subscripts in (I) and (II) above indicate the source of flux, the arrow points towards the sink, and the superscript indicates the medium in which the diffusion occurs. The flow process has been shown schematically in Fig. 11(c). A schematic showing the transformed regions is shown in Fig. 17(d). Finally, Fig. 17(e) shows the net result of the diffusional exchange process.

Using simple conservation of mass and flux balancing, it can be shown that the $N_{Ti}^{\alpha \rightarrow \beta}$, the number of moles of atoms of *Ti* per unit volume moving out of the shrinking α particle *A* (where regions of α are transforming to β) is given by

$$N_{Ti}^{\alpha \rightarrow \beta} = \Omega (X_{Ti}^\alpha - X_{Ti}^\beta) \text{-----} (11)$$

where:

Ω : molar atomic density. It is assumed α and β have same number of atoms per unit volume.

X_{Ti}^α : mole fraction of *Ti* in α phase

X_{Ti}^β : mole fraction of *Ti* in β phase

obviously $N_{Ti}^{\beta \rightarrow \alpha}$ the number of moles atoms of *Ti* per unit volume moving into the growing particle *C* is also given by equation (11). It can be similarly shown, that $N_{Mn}^{\beta \rightarrow \alpha}$, the number *Mn* atoms per unit volume moving out from the β region near the growing α particle *C* is given by

$$N_{Mn}^{\beta \rightarrow \alpha} = \Omega (X_{Mn}^\beta - X_{Mn}^\alpha) \text{-----} (12)$$

where :

X_{Mn}^β : mole fraction of *Mn* in β phase

X_{Mn}^α : mole fraction of *Mn* in α phase

obviously $N_{Mn}^{\alpha \rightarrow \beta}$, the number of atoms of *Mn* per unit volume moving into the shrinking particle *A* is also given by equation (12). From (11) and (12) one can note that

$$N_{Mn}^{\beta \rightarrow \alpha} = N_{Ti}^{\alpha \rightarrow \beta} \text{-----} (13)$$

or in other words conversion of α to β occurs by the process of replacing a given number of Ti atoms by an equal number of Mn atoms. Similarly for the conversion of β to α , a given number of Mn atoms are replaced by an equal number of Ti atoms. It is thus apparent that the growth process occurring by mutual nearest neighbors interparticle flux exchange will preserve the volume fractions of the α and the β phases.

CLASS 2 : DIFFUSION PROCESSES OCCURING PREDOMINANTLY THROUGH THE α PHASE.

The diffusion processes, resulting in the conversion of phases at the interfaces, will predominate in the α phase when the bulk volume fraction of the β phase is small. Thus the two phase alloy microstructure belonging to class 2 are those which contain small volume percents of β distributed in a matrix of α .

Fig. 12(a) shows a small volume fraction of β particles distributed in a matrix of α . The β particles are mostly distributed at α boundaries. To bring out the details of the atomistic process resulting as a consequence of flux exchange between interfaces, we consider a two particle system in Fig. 12(b). In a manner similar to the previous case discussed (Class 1), the mutual growing and shrinking process will involve the following steps.

- (1) Mn atoms from the smaller β particle A jump across the $\alpha - \beta$ interface and diffuse to α matrix regions adjacent to the growing β particle C .
- (2) Ti from the α matrix regions near the growing β particle C diffuse to the shrinking particle A and jump across the $\alpha - \beta$ interface.
- (3) Atoms of newly formed β near the interface of particle C jump across the $\alpha - \beta$ interface into particle C .

As a consequence of this occurrence, the β particle A shrinks with a part of it transforming to α and the β particle C growing by a part of the α matrix adjacent to it transforming to β . The diffusion process thus involves the following fluxes.



The notation used above has the same meaning as in Class 1. It may be noted that the direction of fluxes of the two constituent species in class (II) is in a direction opposite to their respective flux directions in the class 1 case discussed before.

The flow process has been shown schematically in Fig. 12(c). A diagram showing the transformed regions is shown in Fig. 12(d). Finally, Fig. 12(e) shows the net result of the diffusional exchange process.

$N_{Ti}^{\alpha \rightarrow \beta}$, the number of atoms of Ti per unit volume and $N_{Mn}^{\beta \rightarrow \alpha}$, the number

of atoms of Mn per unit volume required for $\alpha \rightarrow \beta$ and $\beta \rightarrow \alpha$ conversions are given by equations (11) and (12) respectively .

CLASS 3: DIFFUSION PROCESSES OCCURING THROUGH THE α AS WELL AS THE β PHASES

The type of alloys in which the class (3) diffusion process will operate are those in which neither of two phases can be considered as small. Thus alloys with nearly equal volume percents of the α and the β phases fall into this category. In such an alloy, motion of the $\alpha - \beta$ interfaces will result by diffusion through α and β . Thus the atomistic processes described in Class (1) and Class (2) will occur simultaneously. A typical microstructure in which the Class (3) diffusion process occurs has been shown schematically in Fig. 13.

The particle growth mechanisms discussed above assume a mechanism of bulk diffusion for the transport of the component species between particles. An alternate mechanism of transport is grain boundary diffusion. Thus for Class 1 microstructures grain boundary diffusion will occur predominantly via the $\beta - \beta$ boundaries. In Class 2, the diffusive paths will be the $\alpha - \alpha$ boundaries, while in class 3 the predominant paths will be the $\alpha - \alpha$ and the $\beta - \beta$ boundaries. This alternate mechanism has been schematically shown in Fig. 14.

B. THE DRIVING FORCE FOR GROWTH

Consider a schematic of a typical two phase network in a two phase alloy, as shown in Fig. 15(a). The network comprises of $\alpha - \alpha$ grain boundaries, $\beta - \beta$ grain boundaries, and $\alpha - \beta$ interphase interfaces. It is worthwhile to note that, in general, the network will tend towards a state where the total area comprising of the $\alpha - \alpha$, $\beta - \beta$, and the $\alpha - \beta$ boundaries is minimized. But since the problem being addressed in this study is the time evolution of the mean particle size, we will be primarily concerned with the $\alpha - \beta$ interfaces.

From thermodynamic consideration it is known that only 3 boundaries can intersect at a point. Thus in Fig. 15(a) each node can have only three vertices. From Fig. 15(b) it can be seen that any terminating node, of any given interphase interface, is also a terminating node for a grain boundary (can be $\alpha - \alpha$ or $\beta - \beta$) and another $\alpha - \beta$ interphase interface. Fig. 15(c) is an exploded view of two nearest neighbor $\alpha - \beta$ interfaces AB, and ED. The interfaces exchange flux with each other through the Beta matrix. The new position of the interfaces after a time Δt is shown by dashed lines. It is assumed that the boundary moves parallel to itself and that the nodes of the interface track on adjacent grain boundaries as shown in Fig. 15(c). Making a first order approximation that the only changes occurring in the geometry of network is the displacement of interfaces, AB and CD, an expression for the driving force for flux exchange is presented below.

In Fig. 15(c) let the Interface AB Convert dV_1 volume of α to β and let the interface ED convert dV_2 volume of β to α , by interacting with each other through the process of diffusional exchange of Ti and Mn through the β matrix. If this occurs in a time, dt , then invariance of

volume fractions demands

$$\begin{aligned} & dV_1 + dV_2 = 0 \\ \text{or } & (dV_1 + dV_2) / dt = 0 \\ \text{from above we obtain} \\ & dV_1 / dt = -dV_2 / dt \text{-----} (14) \end{aligned}$$

In two dimensions we replace volumes by areas and areas are replaced by linear dimensions. Thus equation (14) becomes

$$dA_1 / dt = -dA_2 / dt \text{-----} (15)$$

where A_1 and A_2 are the respective areas associated with interfaces AB and ED . If b_1 and b_2 are the instantaneous lengths of interfaces AB , and ED respectively, and v_1 and v_2 are the respective velocities of interfaces AB and ED , then from (15) we obtain

$$v_1 b_1 = -v_2 b_2 \text{-----} (16)$$

Now the condition of spontaneous growth is

$$\begin{aligned} & d(b_1 + b_2) / dt \leq 0 \\ \Rightarrow & db_1 / dt + db_2 / dt \leq 0 \end{aligned}$$

If l_1 and l_2 are the instantaneous distances of interfaces AB , and ED respectively from their initial position at $t = 0$ then the above equation becomes

$$\begin{aligned} & db_1 / dl_1 \cdot (dl_1 / dt) + db_2 / dl_2 \cdot (dl_2 / dt) \leq 0 \\ \Rightarrow & db_1 / dl_1 \cdot v_1 + db_2 / dl_2 \cdot v_2 \leq 0 \text{-----} (17) \end{aligned}$$

Plugging values of v_2 from equation (16) into equation (17) we obtain

$$\begin{aligned} & db_1 / dl_1 \cdot v_1 - v_1 \cdot (b_1 / b_2) \cdot db_2 / dl_2 \leq 0 \\ \Rightarrow & v_1 (b_2 \cdot db_1 / dl_1 - b_1 \cdot db_2 / dl_2) \leq 0 \end{aligned}$$

If v_1 defined positive in the above equation above, the condition of net reduction in surfaces

becomes

$$b_2 \cdot (db_1/dl_1) - b_1 \cdot (db_2/dl_2) \leq 0 \text{-----} (18)$$

Now, the net change of free energy per unit movement of interface, AB is the driving force, F , for growth, or

$$F = - dG/dl_1 = - dG/dt \cdot dt/dl_1 \text{-----} (19)$$

$$\Rightarrow F = - dG/dt \cdot dt/dl_1 = - \gamma \cdot (d(b_1 + b_2))/dt / (dl_1/dt)$$

$$= - \gamma \cdot v_1 \cdot (db_1/dl_1 - b_1/b_2 \cdot (db_2/dl_2)) / v_1$$

$$\Rightarrow F = - \gamma (db_1/dl_1 - b_1/b_2 \cdot (db_2/dl_2)) \text{-----} (20)$$

The above expression gives the driving force for the diffusional exchange between any two "nearest neighbor" interacting interfaces.

The driving force results in the flow of constituent species between the two interfaces which result in a net decrease in the surface area. For diffusion to occur through the bulk, the expected inter-particle concentration profiles for the growth of β particles in α matrix and for α particles in β matrix are shown in Figs. 16(a) and 16(b) respectively. The concentration gradients arise by a probable modulation of concentration of the respective components in the matrix about their mean value. A constant concentration gradient has been shown in Figs. 16(a), and 16(b), though in reality this is not going to be the case.

C. IDENTIFICATION OF THE RATE CONTROLLING STEP

There are two distinct dynamical processes involved in the particle growth of two phase titanium alloys. These are.

- (I) Migration of atoms across the "source" and "sink" interfaces.
- (II) Long range diffusion of solute and solvent atoms between interfaces.

The question that arises at this stage is as to which of the above two processes is the rate controlling step in the coarsening of particles in α - β alloys. We will consider the possibility of an interface control first.

(I) INTERFACE CONTROL

It is known that for certain type of coherent or semi-coherent interphase interfaces the boundary mobilities are extremely low. Since the velocity of an interphase is proportional to the interface mobility there is the possibility of a certain measure of interface control to exist (15). In the α - β Titanium alloys studied in the present work, the particle phase is located at either triple junctions or at the grain boundaries. For particles located at triple junctions it has been shown (17) that on the average only one interphase out of three is a coherent or a semi coherent boundary. For particles located at grain boundaries it is known (16) that generally either both the interfaces are incoherent or one interface is incoherent and the other is coherent. Thus, as a lower bound, 50% of interfaces will be incoherent interfaces. A typical α - β titanium microstructure with coherent and incoherent interfaces is shown in Fig. 17. Since on an average 50% of the interfaces are incoherent interfaces it is reasonable to assume that the kinetics of the coarsening process is not interface controlled. This is also confirmed by looking at the right side of Figs. 9(b) and 10(b). From these figures it can be seen that, for any given volume fraction of the matrix phase, the particle size of α in β matrix is larger than vice-versa. However if the kinetics were to be interface controlled, then the size of α particles would have been equal to the particle size of the β phase. Another pointer against the possibility of interface control existing comes from particle coarsening theories. These theories predict that if the slowest step in the coarsening process is the transfer of atoms across the interfaces (22) then the mean particle size evolves as the square root of time, $(\bar{D} \propto t^{1/2})$. Since the actual observed exponents are far less than .5, it is reasonable to conclude that the particle coarsening process is not interface controlled. However it is important to note that in the widmanstatten α and β structures there is the possibility that the process may be interface controlled, but these structures are not considered in this paper.

Having thus discounted the possibility of the kinetics being interface controlled, we move on to the process of long range diffusion.

(II) LONG RANGE DIFFUSION

Long range diffusion processes can occur by either of the following three (or in combinations) mechanisms.

- (a) dislocation (pipe) diffusion
- (b) bulk diffusion
- (c) grain boundary diffusion

(a) DISLOCATION (PIPE) DIFFUSION

It is known that the contribution of diffusion through dislocation becomes significant below $.5 T_m$ (16). The apparent diffusion coefficient D_{app} is approximately related to the diffusivities of lattice and pipe by the following expression —

$$D_{app} = D_l + g D_p$$

where D_l = Diffusivity of lattice

g = a geometrical factor (cross section area of pipe per unit area of matrix)

D_p = Diffusivity through dislocation "pipes". Because the activation energy for pipe diffusion is much smaller than bulk diffusion (16), at sufficiently low temperatures, the term $g D_p$ becomes so large that

$$D_{app} \approx g D_p \text{-----} (21)$$

in such a case the kinetics is said to be controlled by pipe diffusion. In the α - β Titanium alloys investigated in this study, it has been shown (17) that the β phase has a extremely low dislocation density in comparison to the α phase. Thus if the kinetics were purely controlled by Pipe diffusion then the sizes of β particles in α matrix would have been larger than the size of α particles in β matrix. But the results of Figs. 9(b) and 10(b) show an exactly reverse trend. The possibility of dislocation control existing is thus remote.

(b) BULK DIFFUSION

In a binary alloy where diffusion occurs by a substitutional mechanism, the conservation of lattice sites requires

$$J_A + J_B + J_V = 0 \text{-----} (22)$$

where J_A = flux of component A

J_B = flux of component B

J_V = flux of Vacancies

Then from (22), above, and Fick's first law we have

$$J_V = (D_A - D_B) \cdot \delta C_A / \delta X \text{-----} (23)$$

from equation (23) we get

$$\delta J_V / \delta X = \delta ((D_A - D_B) \cdot \delta C_A / \delta X) / \delta X = \delta C_V / \delta t$$

or in other words a flux gradient of a component will lead to the generation or consumption of vacancies. If an equilibrium concentration of vacancies is to be maintained, then a mechanism of vacancy consumption (or production) must operate in the alloy. Under such conditions the Fick's first law for diffusion in substitutional binary alloys can be written as (16)

$$J_A = \tilde{D} \delta N_A / \delta X \text{-----} (24)$$

and $J_B = \tilde{D} \delta N_B / \delta X$

where $\tilde{D}$ is interdiffusion coefficient, and is defined as

$$\tilde{D} = D_A X_B + D_B X_A \text{-----} (25)$$

where $D_A, D_B =$ are respective intrinsic diffusivities of components A and B

$X_A, X_B =$ are respective mole fractions of components A and B.

In the limiting cases of α particles distributed in a β matrix (class 1 microstructure) and β particles distributed in an α matrix (class 2 microstructures), the kinetics of the growth process controlled by bulk diffusion should be explainable in the framework of the LSW theory. Since the LSW theory primarily concerns itself with the diffusion of a single species (the solute), minor modifications have to be made to apply it to the present situation of a two way interdiffusion process. These modifications are considered below for class 1 and class 2 microstructures. Again the Ti-Mn system has been used for the discussion.

Consider a class 1 microstructure of α particles distributed in a matrix of β . The molar free energy plots of α , and β phases as a function of mole fractions of Ti, X_{Ti} is shown in Fig. 18(a). Invoking the thermodynamic conditions of equilibrium it can be shown that

$$X_{Ti}^{\beta^*} = X_{Ti}^{\beta} \cdot (\exp (2 \cdot \gamma^{\alpha-\beta} \cdot \Omega_{Ti}^{\alpha} / R \cdot T \cdot r)) \text{-----} (26)$$

where

$X_{Ti}^{\beta^*}$ = equilibrium concentration of Ti in matrix, near an α particle of radius r

X_{Ti}^{β} = equilibrium concentration of Ti in matrix in the absence of interfaces.

$\gamma^{\alpha-\beta}$ = α - β interface energy

Ω_{Ti}^{α} = partial molar volume of Ti in α phase

R, T = gas constant, and temperature respectively

r = radius of α particle

The concentration gradient at the particle matrix interface is given (following LSW theory) by

$$dC/dR (R=r) = C^I - C_r / r \text{-----} (27)$$

where C^I is the concentration at a very large distance from the particle - matrix interface and C_r is the concentration in matrix near a particle of radius r . Writing equation (27) in terms of

mole fractions instead of concentrations, we obtain

$$dC_{Ti}/dr = \Omega^\beta (X_{Ti}^{\beta'} - X_{Ti}^{\beta*})/r \text{ -----(28)}$$

where Ω^β is the number of atoms per unit volume in the β phase. The rate of growth of a particle of radius r can now be obtained as per LSW theory as

$$dr/dt = -(\tilde{D}^\beta \cdot U \cdot \Omega^\beta \cdot ((X_{Ti}^{\beta'} - X_{Ti}^\beta) + (X_{Ti}^\beta - X_{Ti}^{\beta*}))) / r \text{ -----(29)}$$

The terms in equation (29) are defined later. On plugging in the value of $X_{Ti}^\beta - X_{Ti}^{\beta*}$ from equation (26) into equation (29), we obtain

$$dr/dt = -(\tilde{D}^\beta \cdot U \cdot \Omega^\beta \cdot ((X_{Ti}^{\beta'} - X_{Ti}^\beta) - 2 \cdot X_{Ti}^\beta \cdot \gamma^{\alpha-\beta} \cdot \Omega_{Ti}^\alpha / R \cdot T \cdot r)) / r \text{ -----(30)}$$

The solution of equation (30), following the approach of LSW theory results in the kinetic expression

$$\bar{r}^3 - \bar{r}_0^3 = (6 \cdot \gamma^{\alpha-\beta} \cdot \Omega^\beta \cdot \Omega_{Ti}^\alpha \cdot U \cdot \tilde{D}^\beta \cdot X_{Ti}^\beta / R \cdot T) / t \text{ -----(31)}$$

where

$\bar{r}$ = mean particle radius of α at time, t

$\bar{r}_0$ = mean initial particle radius

Ω^β = no. of atoms per unit volume in β phase

Ω_{Ti}^α = partial molar volume of Ti in α phase.

$\tilde{D}^\beta$ = interdiffusion coefficient in β phase.

$1/U$ = no. of moles of Ti or Mn atoms per unit volume for $\alpha \rightarrow \beta$ and $\beta \rightarrow \alpha$ transformation.

X_{Ti}^β = equilibrium mole fraction of Ti in β matrix in the absence of interfaces.

t = time

The molar free energy plot of β particles in an α matrix (class 2 microstructures) as a function of mole fraction of Mn , X_{Mn} is shown in Fig. 18(b). In a manner equivalent to that for class 1 microstructures it can be shown that for class 2 microstructures the kinetic expression is

$$\bar{r}^3 - \bar{r}_0^3 = (6 \cdot \gamma^{\alpha-\beta} \cdot \Omega^\alpha \cdot \Omega_{Mn}^\beta \cdot U \cdot \tilde{D}^\alpha \cdot X_{Mn}^\alpha / R \cdot T) / t \text{ -----(32)}$$

where

Ω^α = no. of atoms per unit volume in α phase.

Ω_{Mn}^β = partial molar volume of Mn in β phase.

$\tilde{D}^\alpha$ = interdiffusion coefficient in α phase

X_{Mn}^α = equilibrium concentration of Mn in α matrix and other terms are, as already explained in equation (31). From equations (31) and (32), we expect for the $Ti - Mn$ system (for class 1 and class 2 microstructures).

$$(D_\beta^{\alpha-\beta} / D_\alpha^{\alpha-\beta})_{Ti-Mn}^3 = (\Omega^\alpha \cdot \Omega_{Mn}^\beta \cdot \tilde{D}^\alpha \cdot X_{Mn}^\alpha) / (\Omega^\beta \cdot \Omega_{Ti}^\alpha \cdot \tilde{D}^\beta \cdot X_{Ti}^\beta) \quad (33)$$

where $D_\beta^{\alpha-\beta}$, $D_\alpha^{\alpha-\beta}$ are particle sizes of the β and the α phases, respectively. $\tilde{D}^\alpha$, $\tilde{D}^\beta$ are the respective interdiffusion coefficients in the α and β phases. A similar expression should hold for the $Ti-V$ system. Thus we expect the following relationship to hold.

$$(D_\beta^{\alpha-\beta} / D_\alpha^{\alpha-\beta})_{Ti-V}^3 = (\Omega^\alpha \cdot \Omega_V^\beta \cdot \tilde{D}^\alpha \cdot X_V^\alpha) / (\Omega^\beta \cdot \Omega_{Ti}^\alpha \cdot \tilde{D}^\beta \cdot X_{Ti}^\beta) \quad (34)$$

Furthermore the growth of particulate α in $Ti-Mn$ and $Ti-V$ systems should be related by:

$$((D_\alpha^{\alpha-\beta})_{Ti-Mn} / (D_\alpha^{\alpha-\beta})_{Ti-V})^3 = K_{Ti-Mn} / K_{Ti-V} \quad (35)$$

where $K = \tilde{D}^\beta \cdot X_{Ti}^\beta \cdot \Omega^\beta \cdot \Omega_{Ti}^\alpha \cdot U$. The subscripts on the K 's in equation (35) are the alloy systems for which the expression for K is evaluated.

A similar expression as equation (35) should hold for the β phase in the $Ti-Mn$ and the $Ti-V$ systems. Numerical checks for equations (33), (34), and (35) will be made using tracer diffusivities. But first a rationale for using tracer diffusivities is being presented below--

The tracer diffusivity of Titanium D_{Ti}^β at 1223° K (18) is

$$D_{Ti}^\beta = 9.27 \cdot 10^{-10} \text{ cm}^2 / \text{sec}$$

The self diffusivity of Titanium in a 10 at % V titanium alloy, $D_{Ti}^{\beta-V}$ at 1223° K is reported by Murdock et.al (19) as

$$D_{Ti}^{\beta-V} = 5.41 \cdot 10^{-10} \text{ cm}^2 / \text{sec}$$

since the difference between D_{Ti}^{β} and $D_{Ti}^{\beta-V}$ is small interpolated values of the tracer self diffusivity of Titanium at 973° K has been used in the calculations for the *Ti-Mn* and the *Ti-V* systems.

The tracer diffusivity of V in β titanium, D_V^{β} at 1223° K is given by Askill (18) as

$$D_V^{\beta} = 5.97 * 10^{-10} \text{ cm}^2 / \text{sec}$$

The diffusivity of vanadium, at 1223° K, for a 10 at % V titanium alloy, $D_V^{\beta-V}$ is reported by Murdock et.al. (19) as:

$$D_V^{\beta-V} = 4.49 * 10^{-10} \text{ cm}^2 / \text{sec}$$

Again one finds that $D_V^{\beta} \approx D_V^{\beta-V}$. Thus interpolated values, at 973° K, of the tracer diffusivity of V has been used for calculations.

Similarly, the tracer diffusivity of Mn in titanium at 1171° K is reported by Askill (18) as

$$D_{Mn}^{\beta} = 3.24 * 10^{-9} \text{ cm}^2 / \text{sec}$$

A value of diffusivity of Mn in a 9.7 at % titanium alloy at 1171° K is given by Santos et.al (20) as

$$D_{Mn}^{\beta-Mn} = 1.86 * 10^{-9} \text{ cm}^2 / \text{sec}$$

Since $D_{Mn}^{\beta} \approx D_{Mn}^{\beta-Mn}$ it is reasonable to use the tracer diffusivities of Mn in a β titanium interpolated to 973° K, for all calculations.

The various tracer diffusivities interpolated to 973° K using data of Askill(18) are

$$D_{Mn}^{\beta} = 1.56 * 10^{-10} \text{ cm}^2 / \text{sec}$$

$$D_{Ti}^{\beta} = 3.31 * 10^{-11} \text{ cm}^2 / \text{sec}$$

$$D_{Ti}^{\alpha} = 6.944 * 10^{-14} \text{ cm}^2 / \text{sec}$$

$$D_V^{\beta} = 1.726 * 10^{-11} \text{ cm}^2 / \text{sec}$$

Fanny Dymant (21) states that the tracer diffusivity of Mn does not change significantly as one passes the $\beta \rightarrow \alpha$ transformation point. Thus one can safely assume the tracer diffusivity of Mn in α phase to be the tracer diffusivity of Mn in β phase. Thus

$$D_{Mn}^{\alpha} = D_{Mn}^{\beta} = 1.56 * 10^{-10} \text{ cm}^2 / \text{sec}$$

It can furthermore be assumed that the tracer diffusivity of titanium is the same in the α phases of the *Ti-Mn* and the *Ti-V* systems respectively (it being so for the β phase). Thus we can take

$$D_{Ti}^{\alpha-V} = D_{Ti}^{\alpha-Mn} = 6.944 * 10^{-14} \text{ cm}^2 / \text{sec}$$

using the mole fractions of the components from Figs. 1 and 2 we obtain the interdiffusion coefficients as-

for the *Ti-Mn* system, we have

$$\tilde{D}_{Mn}^{\beta} = 1.42 * 10^{-10} \text{ cm}^2 / \text{sec}$$

$$\tilde{D}_{Mn}^{\alpha} = 1.55 * 10^{-10} \text{ cm}^2 / \text{sec}$$

and for the *Ti-V* system we obtain

$$\tilde{D}_V^{\beta} = 1.94 * 10^{-11} \text{ cm}^2 / \text{sec}$$

$\tilde{D}_V^{\alpha}$ cannot be evaluated as the value of D_V^{α} is not available.

For a 75% matrix phase alloy, annealed for 200 hrs at 973° K, the left hand side of equation (33) takes on the value of .49 . The left hand side of equation (35) takes on the value of 2.75 Only equation (35) will be used for calculations , as all the quantities on the right hand side can be evaluated. This is not true of equation (33) or (34).

At 973° K, the right hand side of equation (35) gives a theoretical ratio of 9.67. Thus considerations based on bulk interdiffusion predicts a much larger difference in sizes of the α phases in the *Ti-Mn* and *Ti-V* systems than the ones actually observed. Thus the experimental results do not seem to support the hypothesis that the coarsening kinetics might be controlled by the interdiffusion coefficient. . The failure of experimental results to completely conform to any of the above mechanism thus seem to suggest that the volume diffusion is not the major controlling mechanism and that there are other mechanisms of long range diffusion (other than volume diffusion) which transport material between the interfaces. Since the particles are located at grain boundaries (triple points), an important mechanism to consider is grain boundary diffusion. This mechanism is considered in detail in the next section.

(c) GRAIN BOUNDARY DIFFUSION

Since the interphase interfaces are intersected by α - α and β - β grain boundaries it is pertinent to inquire about the possibility of such a mechanism to be the rate controlling mechanism in α - β two phase alloys. It is known that grain boundary diffusion becomes dominant below .75 - .8 *Tm* (16). The apparent diffusion coefficient, D_{app} is related to the grain boundary and bulk diffusivities by the following relation

$$D_{app} / D_l = 1 + D_b \cdot \delta / D_l \cdot d \text{-----} (36)$$

where

D_l = Diffusivity of lattice

D_b = Diffusivity of boundary

δ = Effective width of boundary

d = Effective mean grain size

Grain boundary control is said to exist if

$$D_b \cdot \delta \geq D_l \cdot d \text{-----} (37)$$

Though it is known that $D_b > D_l$ (16), the values of the grain boundary diffusivities for the alloys being studied are not available in literature. Hence equation (36) cannot be directly used to determine the possible existence of a grain boundary control of the growth kinetics. On the other hand in equation (37) a typical value of δ is .5 μm (16) and if one assumes a typical grain size of 1 μm , then one finds that the ratio D_b / D_l required to satisfy equation (37) is 2×10^3 . Thus if the grain boundary diffusivity is around 2000 times that of the bulk diffusivity, the kinetics will be controlled by grain boundary diffusion. Since a factor 2000 is a very reasonably possible number, the possibility of grain boundary control is likely. There is however, other indirect evidence which seem to suggest such a mechanism. The particle growth exponent for the Ti-V system at 973 K is .24. It has been shown (22) that under grain boundary controlled kinetics, the mean particle size evolves as

$$\bar{D}^4 - \bar{D}_0^4 = K \cdot t \text{-----} (38)$$

where

$\bar{D}$ = mean particle size at time , t

$\bar{D}_0$ = mean initial particle size

K = growth constant

t = time

The close correspondence of the growth exponent of the Ti-V system with the theoretical values (.25) seem to suggest that in the Ti-V system the growth kinetics may be grain boundary controlled. Thus the apparent diffusivity differences between the Ti-Mn and the Ti-V will be much smaller than the actual lattice diffusivity differences. This has been schematically shown in Fig. 19 (a). In plotting the schematic shown in Fig. 19 (a) the assumption has been made that the contribution of the grain boundary diffusivity term to the apparent diffusivity in equation (36) is the same for the Ti-Mn and the Ti-V systems. Though the grain boundary diffusivity is expected to be smaller in the Ti-V system than the Ti-Mn system , the grain size of the Ti-V system is also smaller than the Ti-Mn system and hence the assumption of same contributions of the grain boundary term in the two systems is reasonable. A typical grain boundary structure in a titanium alloy is shown in Fig. 19 (b).

V. CONCLUSIONS

1. It was found that the particle sizes of α and β phases in two phase Titanium alloys could be expressed in terms of simple relationships between Time , and Volume percents of the phases.

$$D_{\beta}^{\alpha-\beta} = (K_{\beta} / V_{\alpha}^{m_{\alpha}}) \cdot t^n$$

$$D_{\alpha}^{\alpha-\beta} = (K_{\alpha} / V_{\beta}^{m_{\beta}}) \cdot t^n$$

2. The growth of the α and β phases was attributed to a two way diffusion mechanism of the solute and the solvent.

3. Growth of α in β matrix is faster than vice versa. This was attributed to the larger grain boundary diffusivities in the β phase.

4. The particle growth kinetics cannot be explained on the basis of bulk interdiffusivity considerations alone. The growth rates predicted by considerations of bulk interdiffusion give particle size ratios between the Ti-Mn to Ti-V systems to be much larger than the ones actually observed. This was attributed to a mixed grain boundary plus bulk diffusion control in the Ti - Mn system and predominantly grain boundary diffusion control in the Ti - V system.

VI. ACKNOWLEDGEMENTS

Our thanks are due to Professors R. J. Arsenault , M. Wuttig and H. I. Aaranson for their encouragement and helpfull suggestions. The authors would like to express their appreciation to Messrs. D. J. McNeish, D. E. Thomas and S. R. Seagle and their associates of RMI company for making the alloys used in this investigation with great care at a nominal cost. The authors are indebted to Mr. I. L. Caplan of DTNSRDC and Dr. A. H. Rosenstein of AFOSR for their keen interest and constant encouragement throughout the course of this investigation. This work was supported by the Air Force office of Scientific Research on Grant AFOSR - 85 - 0367.

(b) TEMPERATURE EFFECTS

In continuation of the previous particle growth studies at 973^0 K , isothermal particle growth studies were carried out at 1023^0 K , and 1073^0 K . Furthermore, to confirm the validity of the experimental procedures additional particle growth studies were carried out at 998^0 K , and 1048^0 K . The results of the isothermal particle growth studies carried at 1023^0 K , and 1073^0 K are presented in detail below :---

ISOTHERMAL PARTICLE GROWTH STUDIES AT 1023^0 K

The volume percents of α , for the α - β Ti-Mn alloy system, varied from 18 % for alloy no. 4 to 74 % for alloy no. 2. The volume percents of α for the α - β Ti-V alloy system varied from 12 % for alloy no. 9 to 59 % for alloy no. 7. It is pertinent to note that though the volume fractions of the alloys studied are different, the chemical composition of the α and the β phases is the same in all alloys in each of the two systems studied. The chemical composition of the phases, in the two systems studied are shown by the ends of the tielines drawn for 1023^0 K , in figures (20) and (21), respectively. The heat treatments resulted in equiaxed α - β microstructures. Typical equiaxed α - β microstructures of Ti-Mn and the Ti-V alloys heat treated at 1023^0 K are shown in figures (22) and (23), respectively. As expected, the α and the β sizes increased as a function of time. The microstructural data for Ti-Mn and Ti-V systems is given in tables (3) to (5) and (6) to (8), respectively. The particle sizes of the α and the β phases plotted as a function of time for alloys 2, 3, and 4 of the Ti-Mn system and 7, 8, and 9 of the Ti-V system are shown in figures (24) to (27). It was found that in both the Ti-Mn and the Ti-V systems, the particle growth of the α and the β phases could be written in the form of equations (3) and (4). The particle growth exponents $K_{\alpha}^{\alpha-\beta}$, and $K_{\beta}^{\alpha-\beta}$ depended on the respective other phase volume percents. In both the Ti-Mn and the Ti-V systems, the particle growth exponents could be expressed by equations of the form (5) and (6). The numerical data for $K_{\alpha}^{\alpha-\beta}$ and $K_{\beta}^{\alpha-\beta}$ as a function of the respective other phase is shown in table (9) for the Ti-Mn system, while table (10) shows the corresponding data for the Ti-V system. The growth constants $K_{\alpha}^{\alpha-\beta}$ and

$K_{\beta}^{\alpha-\beta}$ were plotted as a function of the respective other phase. Such plots for the Ti-Mn system are shown in figure (28), while figure (29) shows the corresponding plots for the Ti-V system. The numerical forms of equations (3) to (6) for the Ti-Mn and the Ti-V systems are presented below---

(a) Ti-Mn SYSTEM

$$D_{\alpha}^{\alpha-\beta} = (117 / V_{\alpha}^{1.088}) \cdot t^{.31} \quad \gamma = .99$$

$$D_{\alpha}^{\alpha-\beta} = (48 / V_{\beta}^{.852}) \cdot t^{.31} \quad \gamma = .99 \text{---(39)}$$

(b) Ti- V SYSTEM

$$D_{\beta}^{\alpha-\beta} = (144 / V_{\alpha}^{1.175}) \cdot t^{.27} \quad \gamma = .99$$

$$D_{\alpha}^{\alpha-\beta} = (15 / V_{\beta}^{.592}) \cdot t^{.27} \quad \gamma = .99 \text{---(40)}$$

ISOTHERMAL PARTICLE GROWTH STUDIES AT 1073⁰ K

The volume percents of α , for the α - β Ti-Mn system, varied from 24 % for alloy no. 3 to 54 % for alloy no. 2 . The volume percents of α for the α - β Ti-V alloy system varied from 6 % for alloy no. 8 to 36 % for alloy no. 7 . The chemical composition of the phases , in the two systems , is shown by the ends of the tie lines drawn at 1073⁰ K in figures (20) and (21) respectively. Typical equiaxed α - β microstructures of Ti-Mn and Ti-V alloys heat treated at 1073⁰ K are shown in figures (30) and (31) respectively. The microstructural data for the Ti-Mn and the Ti-V systems is given in tables (11) to (14). The particle sizes of the α and the β phases plotted as a function of time for alloys 2 , and 3 of the Ti-Mn system and 7 , and 8 of the Ti-V system are shown in figures (32) to (35) .

It was found that in both the Ti-Mn and the Ti-V systems, the particle growth of the α and β phases could be expressed in the form of equations (3) and (4) . The exact numerical forms of these equations are given below for the two systems studied.

(a) Ti-Mn SYSTEM

(I) ALLOY NO. 2

$$D_{\alpha}^{\alpha-\beta} = 2.44 t^{.33} \quad \gamma = .99$$

$$D_{\beta}^{\alpha-\beta} = 2.08 t^{.33} \quad \gamma = .99 \text{---(41)}$$

(II) ALLOY NO. 3

$$D_{\alpha}^{\alpha-\beta} = 1.83 t^{.35} \quad \gamma = .99$$

$$D_{\beta}^{\alpha-\beta} = 5.85 t^{.33} \quad \gamma = .99 \text{---(42)}$$

(b) Ti- V SYSTEM

(I) ALLOY NO. 7

$$\begin{aligned} D_{\alpha}^{\alpha-\beta} &= 1.76 t^{.28} & \gamma &= .99 \\ D_{\beta}^{\alpha-\beta} &= 3.18 t^{.28} & \gamma &= .99 \text{-----} (43) \end{aligned}$$

(II) ALLOY NO. 8

$$\begin{aligned} D_{\alpha}^{\alpha-\beta} &= 1.63 t^{.29} & \gamma &= .99 \\ D_{\beta}^{\alpha-\beta} &= 22.7 t^{.29} & \gamma &= .99 \text{-----} (44) \end{aligned}$$

The studies carried out at 998⁰ K and 1048⁰ K were essentially used as controls to confirm the validity of the data at 973⁰ K, 1023⁰ K and at 1073⁰ K. The details of the particle growth studies at these temperatures are hence not reported here.

The parameters $\ln ((K_{\alpha}^{\alpha-\beta})^{1/n})$ and

$\ln ((K_{\beta}^{\alpha-\beta})^{1/n})$ for given volume percents of $\beta (V_{\beta})$, and $\alpha (V_{\alpha})$ respectively were plotted as a function of inverse temperature $(1/T)$. It was found that the normalized particle sizes $(D_{\alpha}^{\alpha-\beta})^{1/n} V_{\alpha}$ and $(D_{\beta}^{\alpha-\beta})^{1/n} V_{\beta}$ at given volume percents of $\beta (V_{\beta})$ and $\alpha (V_{\alpha})$, respectively could be written as

$$(D_{\alpha}^{\alpha-\beta})^{1/n} V_{\alpha} = K_{V_{\alpha}} \exp(-(Q/R.T)) . t \text{-----} (45)$$

$$(D_{\beta}^{\alpha-\beta})^{1/n} V_{\beta} = K_{V_{\beta}} \exp(-(Q/R.T)) . t \text{-----} (46)$$

The plots for the β and the α phases in a 24 % α - 76 % β and a 54 % α - 46 % β Ti-Mn alloys are shown in figures (36) and (37) respectively. A similar plot for a 36 % α - 64 % β Ti- V alloy is shown in figure (38). The activation energy , Q , determined from the above equations , for the alloys studied , varied from 32 Kcal /mole to 46 Kcal / mole . Though no clear trends could be discerned from the values of the activation energies obtained , it is pertinent to note that the activation energy values (32 Kcal/ mole - 46 Kcal/mole) obtained are close to the activation energies of solute and solvent lattice diffusivities in Titanium . Though the values of the activation energies obtained does not shed any light on the rate controlling mechanisms , the fact that the growth exponent , n , increases as the temperature of isothermal annealing is increased does seem to suggest that the growth kinetics in the Ti-Mn and the Ti-V tend toward " bulk control " as the temperature is increased. The growth exponent of .33 in the Ti-Mn system , for example, does indicate that at 1073⁰ K the long range diffusion occurs predominantly through the bulk. In addition the activation plots in Figs.

(36) and (38) indicate that there is a trend of increasing activation energy (slope of the plots) as the temperature increases . This is consistent with the fact that the diffusion mechanism may be shifting to bulk diffusion as the temperature is increased.

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TABLE 1: The Aim Compositions
and the Actual Ingot Chemistry
of Ti-Mn Alloys

Alloy No.	Compositions					
	AIM	ACTUAL COMPOSITIONS OF INGOTS				
	%Mn	Mn	Fe	O*	N	C
1	0.6	.4	0.02	0.071	0.01	0.02
2	2.7	3.0	0.01	0.086	0.01	0.02
3	4.9	4.3	0.01	0.082	0.006	0.01
4	7.1	6.0	0.01	0.083	0.008	0.01
5	9.3	9.4	0.02	0.100	0.10	0.02
6	11.5	13	0.01	0.116	0.012	0.02

* Oxygen content of final product.

TABLE 2: The Aim Composition
and the Actual Ingot Chemistry of Ti-V Alloys

Alloy No.	Compositions					
	AIM	ACTUAL COMPOSITIONS OF INGOTS				
	%V	V	Fe	O*	N	C
7	3.9	4.3	0.01	0.078	0.017	0.02
8	6.0	6.2	0.01	0.063	0.015	0.01
9	8.3	8.1	0.01	0.082	0.016	0.02
10	10.5	10.6	0.01	0.075	0.020	0.02
11	12.8	12.6	0.01	0.083	0.019	0.02
12	15.0	14.8	0.01	0.092	0.021	0.02

* Oxygen content of final product.

Table 3: Isothermal Growth Data for Alloy No.2 (74% α -26% β , Ti-Mn Alloy,
Annealed at 1023 K, WQ

Time t (hrs)	log t	D_{α} (μm)	log D_{α}	D_{β} (μm)	Log D_{β}
25.	1.3979	8.2390	.91587	2.8576	.45600
50.	1.6989	9.3975	.97301	3.2595	.51315
100.	2.	12.4879	1.0964	4.3314	.63662
200.	2.3010	15.4769	1.1896	5.3325	.72693
400.	2.6020	18.9278	1.27710	6.5651	.81724

$$D_{\alpha} \times 10^{-10} = 2.9276t^{.99} \quad = .99$$

$$D_{\beta} \times 10^{-10} = 1.01849t^{.99} \quad = .99$$

Table 4: Isothermal Growth Data for Alloy No.3 (49% α -51% β , Ti-Mn Alloy,
Annealed at 1023 K, WQ

Time t (hrs)	log t	D α (μ m)	log D α	D β (μ m)	Log D β
25.	1.3979	4.8527	.68597	4.9606	.6955
50.	1.6989	5.9712	.77606	6.10411	.7856
100.	2.	7.9472	.90021	8.1240	.9097
200.	2.3010	9.6489	.9844	9.8636	.9940
400.	2.6020	11.4124	1.0573	11.6663	1.06693

$$D_{\alpha}^{*10} = 1.7736t^{.99} = .99$$

$$D_{\beta}^{*10} = 1.81280t^{.99} = .99$$

Table 5: Isothermal Growth Data for Alloy No.4 (18% α -82% β , Ti-Mn Alloy.

Annealed at 1023 K, WQ

Time t (hrs)	log t	D α (μ m)	log D α	D β (μ m)	Log D β
25.	1.3979	3.1330	.49597	14.0909	1.14893
50.	1.6989	4.1762	.62078	18.7826	1.27375
100.	2.	4.5037	.65356	20.2558	1.30654
200.	2.3010	6.4057	.80656	28.81011	1.45954
400.	2.6020	7.9596	.900890	35.7987	1.55386

$$D_{\alpha} = 1.081t^{.987}$$

$$D_{\beta} = 4.864t^{.987}$$

Table 6: Isothermal Growth Data for Alloy No.7 (59% α -41% β , Ti-V Alloy,
Annealed at 1023° K, WQ

Time t (hrs)	log t	D α (μ m)	log D α	D β (μ m)	Log D β
25.	1.3979	3.91545	.5927	2.7277	.43579
50.	1.6989	4.72128	.674059	3.2890	.51708
100.	2.	5.69296	.75533	3.9659	.59834
200.	2.3010	6.8646	.83661	4.7821	.67961
400.	2.6020	8.2774	.91789	5.7663	.76089

$$D_{\alpha}^{*10} = 1.641t^{.327} = .99$$

$$D_{\alpha}^{*10} = 1.1438t^{.327} = .99$$

Table 7: Isothermal Growth Data for Alloy No.8 (39% α -61% β , Ti-V Alloy,
Annealed at 1023 K, WQ

Time t (hrs)	log t	D α (μ m)	log D α	D β (μ m)	Log D β
25.	1.3979	3.2439	.51106	4.9852	.69768
50.	1.6989	3.9387	.59535	6.0529	.78196
100.	2.	4.78246	.67965	7.3496	.86625
200.	2.3010	5.8068	.76393	8.9238	.95055
400.	2.6020	7.05063	.848220	10.8354	1.03484

$$D_{\alpha-\beta} = 1.3172t^{.99} \quad = .99$$

$$D_{\alpha-\beta} = 2.024t^{.99} \quad = .99$$

Table 8: Isothermal Growth Data for Alloy No.9 (12% α -88% β , Ti-V Alloy),
Annealed at 1023 K, WQ

Time t (hrs)	log t	D α (μ m)	log D α	D β (μ m)	Log D β
25.	1.3979	2.4820	.39480	17.779	1.2499
50.	1.6989	2.9928	.4760	21.4388	1.3312
100.	2.	3.6088	.55736	25.8546	1.4125
200.	2.3010	4.3515	.63863	31.1714	1.4937
400.	2.6020	5.2471	.719910	37.5861	1.5750

$$D_{\alpha} = 1.0408t^{.39} = .39$$

$$D_{\beta} = 7.455t^{.39} = .39$$

Table 9: Isothermal Growth Parameters K_{eff} and K_{eff} for the various Ti - Mn Alloys, at 1023°K

Alloy No.	%	log V_{eff}	K_{eff}	log K_{eff}	V_{eff}	log V_{eff}	K_{eff}	log K_{eff}
2	74.24	1.8706	1.018	.0079568	25.76	1.4109	2.92	.46538
3	49.45	1.6941	1.8128	.25834	50.55	1.7037	1.7736	.2488
4	18.19	1.2598	4.864	.6869	81.81	1.9128	1.081	.03382

$$K_{\text{eff}} = \frac{117}{V_{\text{eff}}^{1.00004}} = .99$$

$$K_{\text{eff}} = \frac{48}{V_{\text{eff}}^{.99210}} = .99$$

Table 10: Isothermal Growth Parameters K_{eff} and $K_{\text{eff}}^{\text{eff}}$ for the various Ti - V Alloys, at 1023°K

Alloy No.	V_L	$\log V_L$	K_{eff}	$\log K_{\text{eff}}$	V_H	$\log V_H$	$K_{\text{eff}}^{\text{eff}}$	$\log K_{\text{eff}}^{\text{eff}}$
7	58.94	1.7704	1.14385	.05836	41.06	1.6134	1.641	.2151
8	39.42	1.5957	2.024	.30621	60.58	1.7823	1.3172	.1196
9	12.25	1.0881	7.4559	.8725	87.75	1.9432	1.0408	.01736

$$K_{\text{eff}}^{\text{eff}} = \frac{144}{V_L^{1.1725}} = .99$$

$$K_{\text{eff}}^{\text{eff}} = \frac{15}{V_H^{0.99122}} = .99$$

Table 11: Isothermal Growth Data for Alloy No.2 (54% α -46% β , Ti-Mn Alloy,
Annealed at 1073 K. WQ

Time t (hrs)	$\log t$	$D\alpha$ (μm)	$\log D\alpha$	$D\beta$ (μm)	$\log D\beta$
12½	1.0969	5.8538	.76743	4.90887	.69098
25.	1.3979	6.9776	.843706	6.0030	.77836
50.	1.6989	9.1400	.960946	7.66456	.88448
100.	2.	10.4287	1.01823	8.74524	.941771
200.	2.3010	13.9757	1.14537	11.7196	1.06891
400.	2.6020	18.77964	1.2736	15.7462	1.19783

$$D_{\alpha} \times 10^6 = 3.4455 \times 10^{-11} \quad = .39$$

$$D_{\beta} \times 10^6 = 2.93719 \times 10^{-11} \quad = .39$$

Table 12: Isothermal Growth Data for Alloy No.3 (24% o-7s%₂, Ti-In 4.1%)
 Annealed at 1073 K, WQ

Time t (hrs)	log t	D _α (μm)	log D _α	D _β (μm)	Log D _β
12½	1.0969	4.0766	.610298	12.9874	1.1135
25.	1.3979	5.7854	.762333	18.4315	1.2655
50.	1.6989	6.74659	.829084	21.4936	1.3323
100.	2.	8.39247	.92388	26.73621	1.42709
200.	2.3010	11.2323	1.05046	35.78465	1.5536
400.	2.6020	13.30895	1.12414	42.47373	1.62736

$$D_{\alpha} \cdot t = 1.83653t^{1/2}$$

$$= .39$$

$$D_{\beta} \cdot t = 5.85049t^{1/2}$$

$$= .39$$

Table 13: Isothermal Growth Data for Alloy No.7 (36% α -54% β , Ti-7 Al-2.5 V)
Annealed at 1073 K. WQ

Time t (hrs)	log t	D α (μ m)	log D α	D β (μ m)	Log D β
12.	1.0969	3.6352	.5605	6.57751	.81805
25.	1.3979	4.4338	.64677	8.02250	.904309
50.	1.6989	5.49977	.740346	9.9533	.99795
100.	2.	6.5958	.8192	11.9343	1.07672
200.	2.3010	7.65418	.88389	13.85238	1.14152
400.	2.6020	10.1479	1.00637	15.3555	1.28401

$$D_{\alpha} = 1.76250t^{.39} \text{ } \mu\text{m}$$

$$D_{\beta} = 3.18875t^{.39} \text{ } \mu\text{m}$$

Table 14: Isothermal Growth Data for Alloy No.8 (6% α -94% β , Ti-V Alloy,
Annealed at 1073 K, WD

Time t (hrs)	log t	D α (μ m)	log D α	D β (μ m)	log D β
12½	1.0969	3.39113	.530366	47.3371	1.67520
25.	1.3979	4.1461	.617639	57.857	1.76235
50.	1.6989	5.0692	.70493	70.7381	1.8495
100.	2.	6.1979	.79224	86.46804	1.935756
200.	2.3010	7.5779	.87954	105.7451	2.02426
400.	2.6020	9.2648	.96683	129.2971	2.11158

$$D_{\alpha} = 1.63019t^{.499}$$

$$= .499$$

$$D_{\beta} = 22.744t^{.499}$$

$$= .499$$

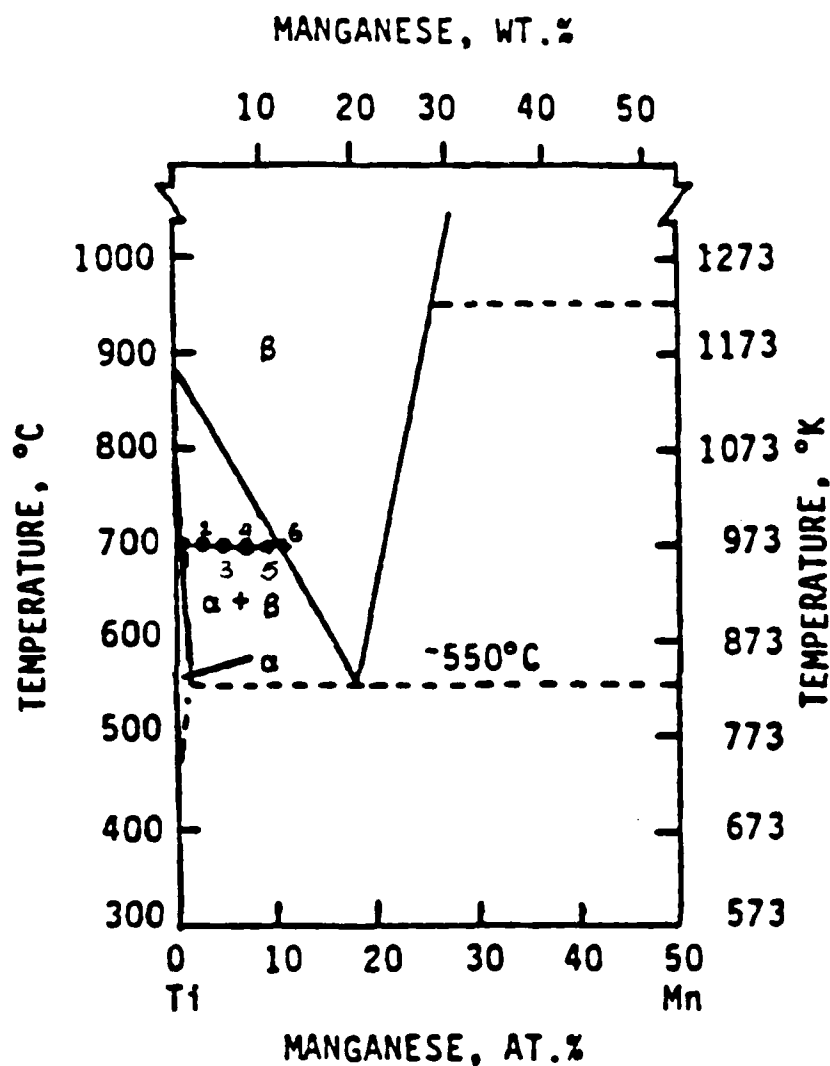


Fig. 1 : PARTIAL TI END OF THE Ti - Mn PHASE DIAGRAM. THE AIM COMPOSITIONS OF ALLOYS 1 - 6 ARE INDICATED WITH SOLID CIRCLES. [14]

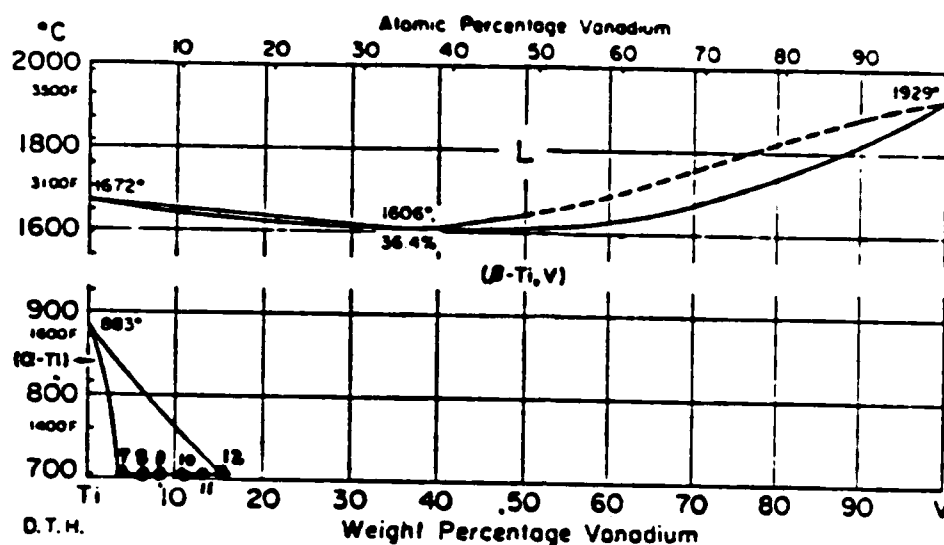
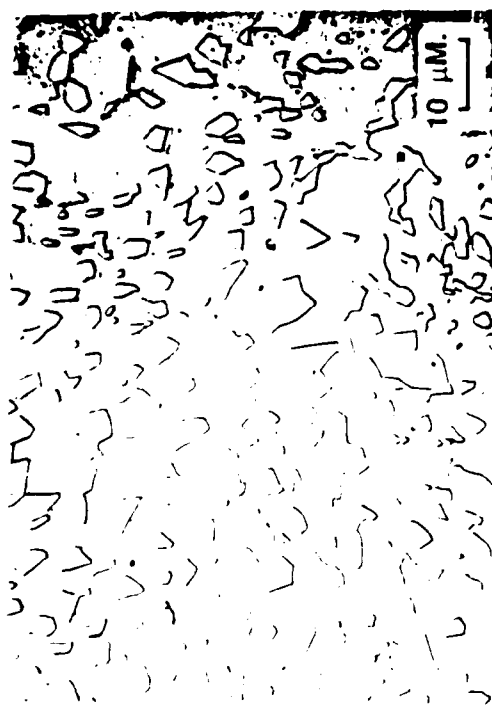


Fig. 2 : THE Ti - V PHASE DIAGRAM. THE AIM COMPOSITIONS OF ALLOYS 7 - 12 ARE INDICATED WITH SOLID CIRCLES.[15]



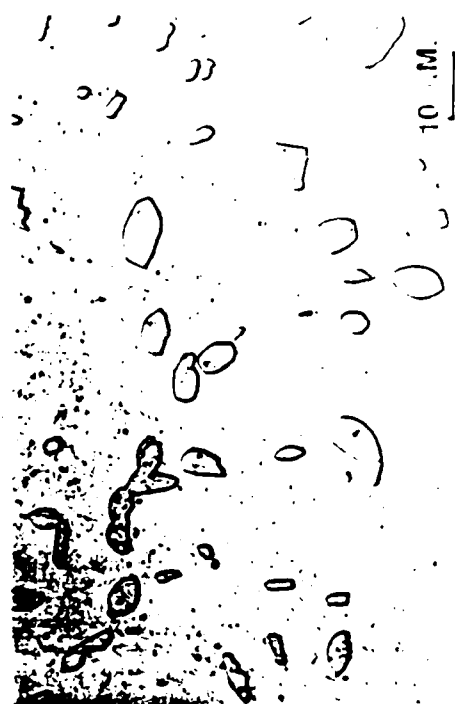
79% α - 21% β



63% α - 37% β



45% α - 55% β



10% α - 90% β

Fig. 3 : MICROSTRUCTURES OF VARIOUS α - β Ti - Mn
ALLOYS ANNEALED FOR 200 HRS. AT 973 K , W. Q.



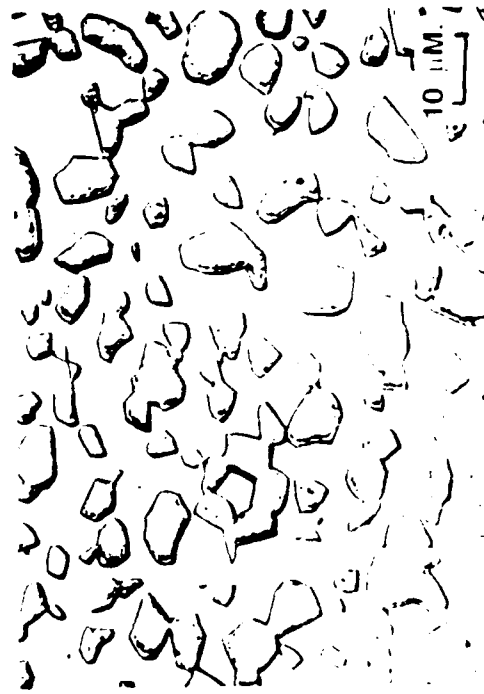
25 HRS.



50 HRS.



100 HRS.



400 HRS.

Fig. 4(a): EFFECT OF TIME ON PARTICLE SIZES OF A 45 % U
- 55 % B ALLOY , ANNEALED AT 973 K FOR VARI-
OUS TIMES.



45 % α - 55 % β Ti - Mn ALLOY



44 % α - 56 % β Ti - V ALLOY

Fig. 4(b): COMPARISON OF THE PARTICLE SIZES OF α β
 Ti - Mn AND Ti - V ALLOYS ANNEALED FOR 400
 HRS. AT 973 K , W. Q.

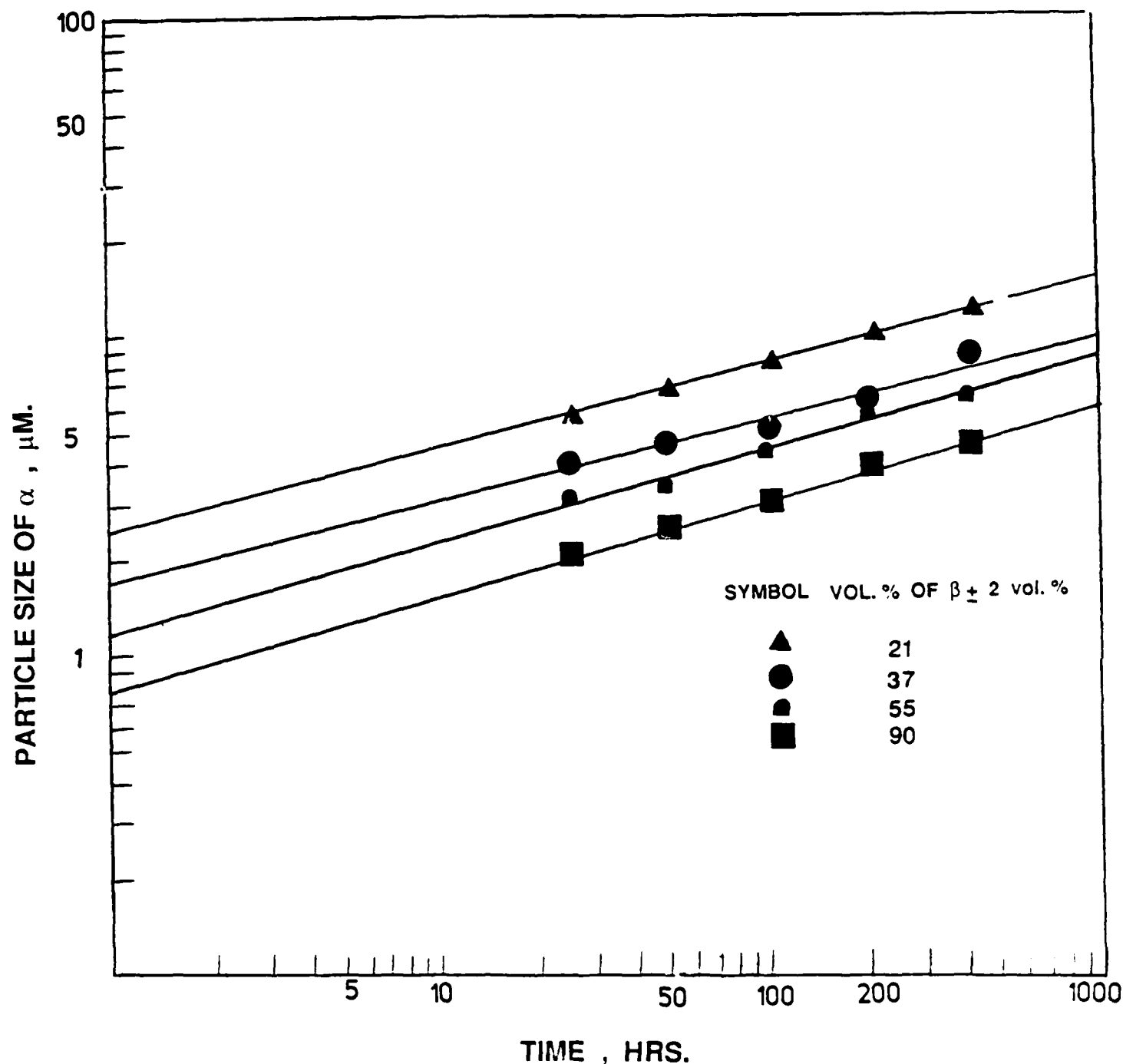


Fig. 5 : PARTICLE SIZES OF α - PHASE IN α - β Ti - Mn ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 973 K, W. Q.

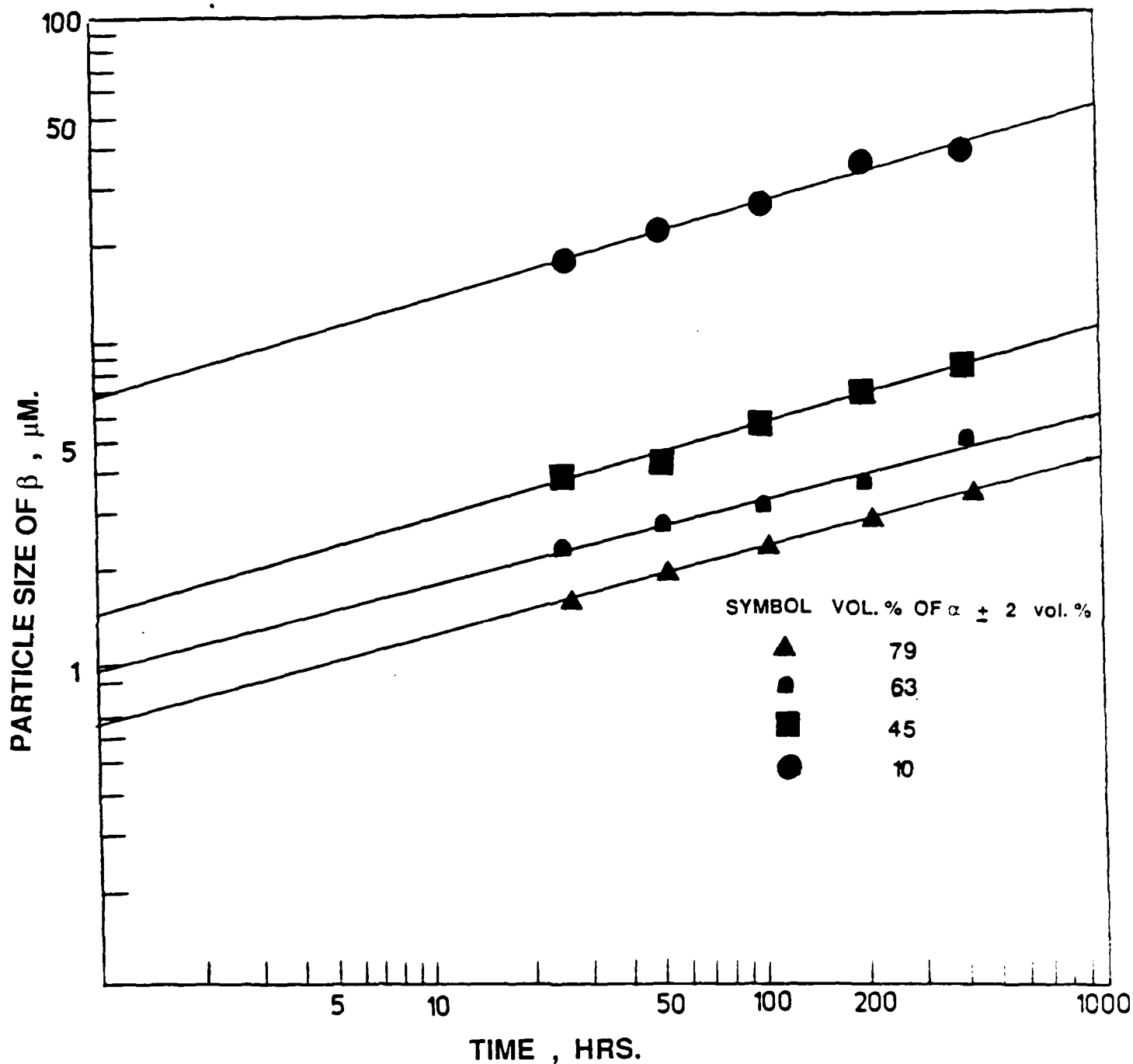


Fig. 6: PARTICLE SIZES OF β - PHASE IN α - β Ti - Mn ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 973 K, W. Q.

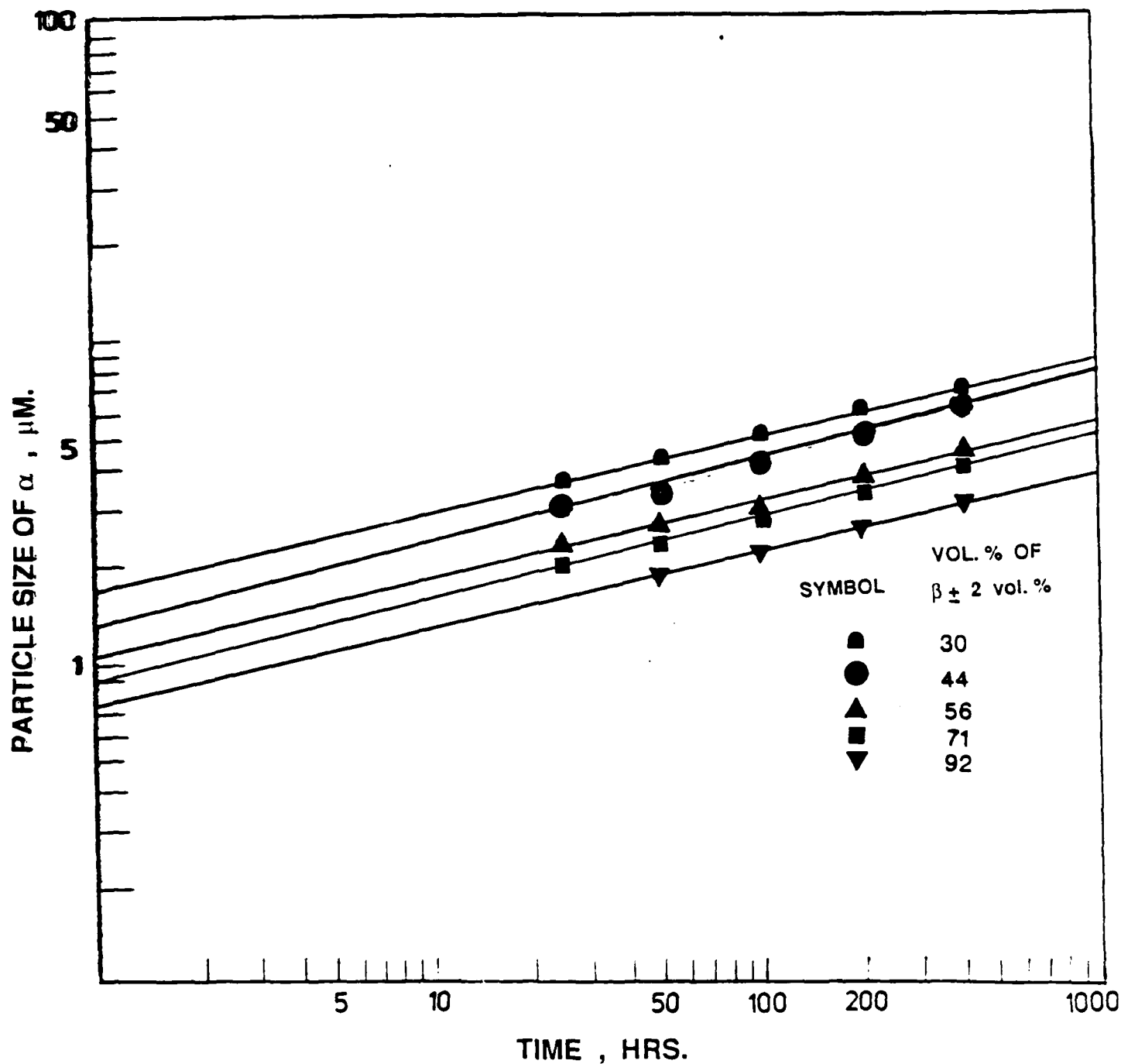


Fig. 7: PARTICLE SIZES OF α - PHASE IN α - β Ti - V ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 973 K , W. Q.

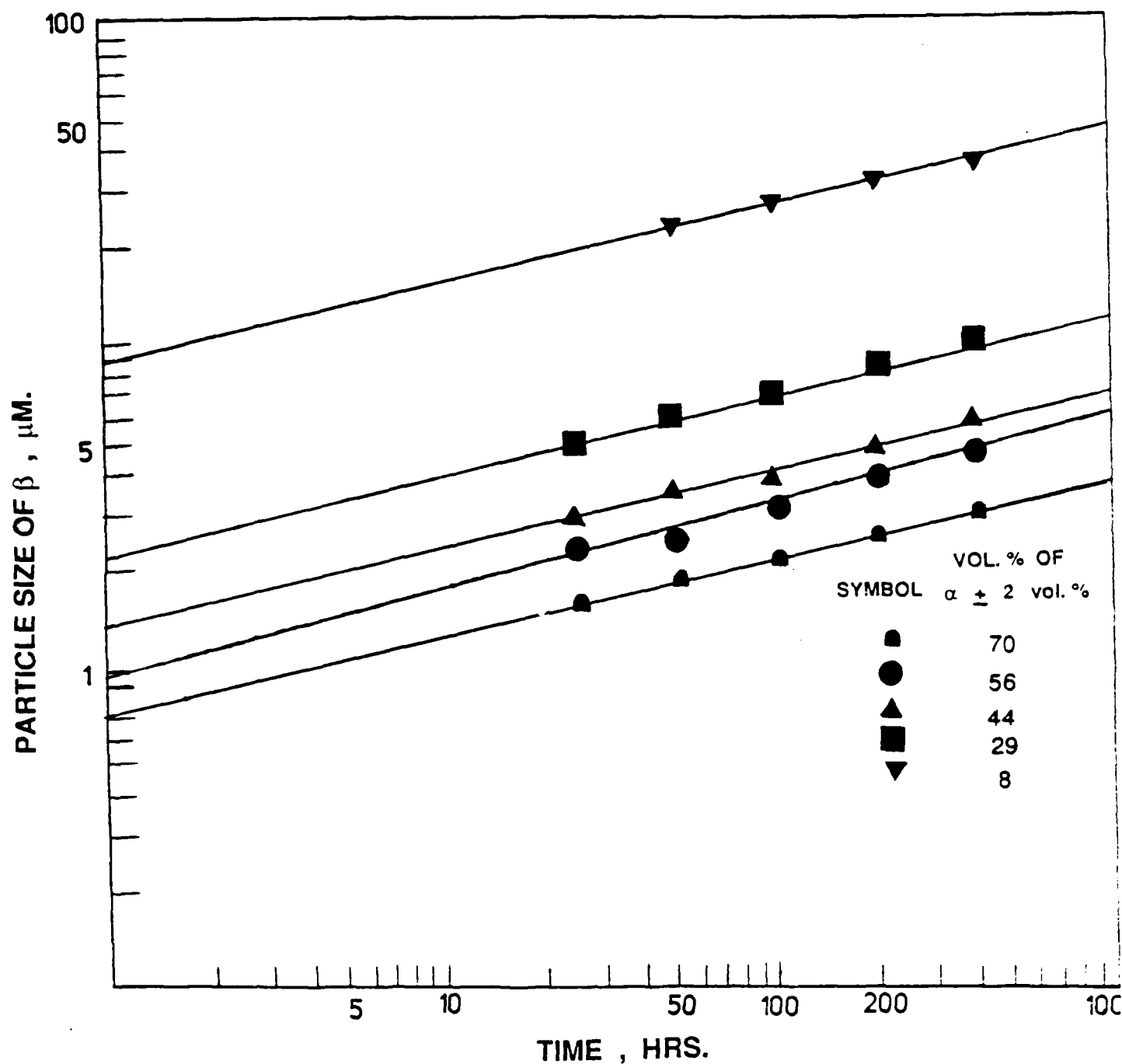


Fig. 8 : PARTICLE SIZES OF β - PHASE IN α - β Ti - V ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 973 K , W. Q.

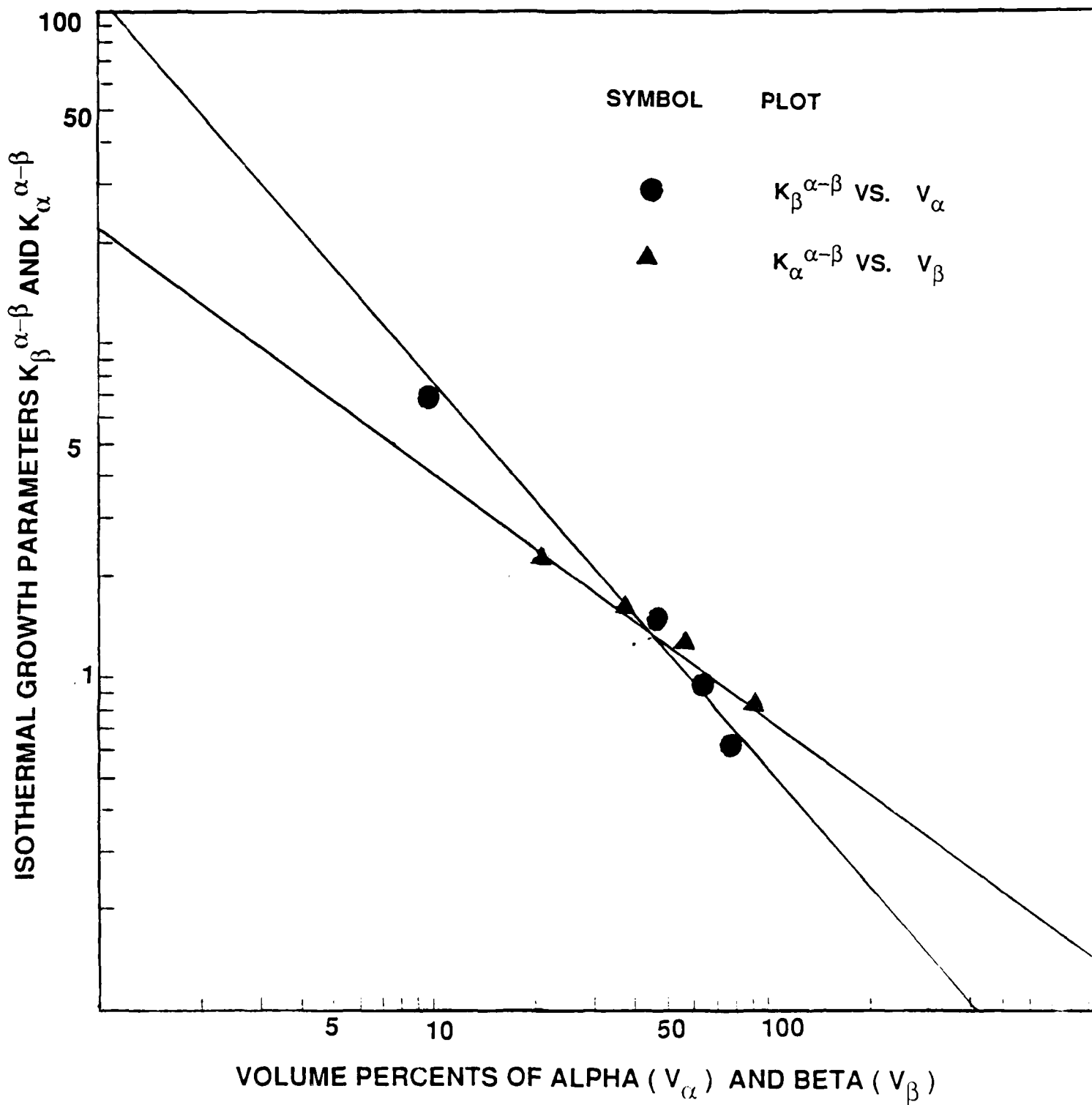


Fig. 9(a) : DEPENDENCY OF THE GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON VOLUME PERCENTS OF β (V_{β}) AND α (V_{α}), RESPECTIVELY, FOR THE Ti - Mn SYSTEM AT 973 K

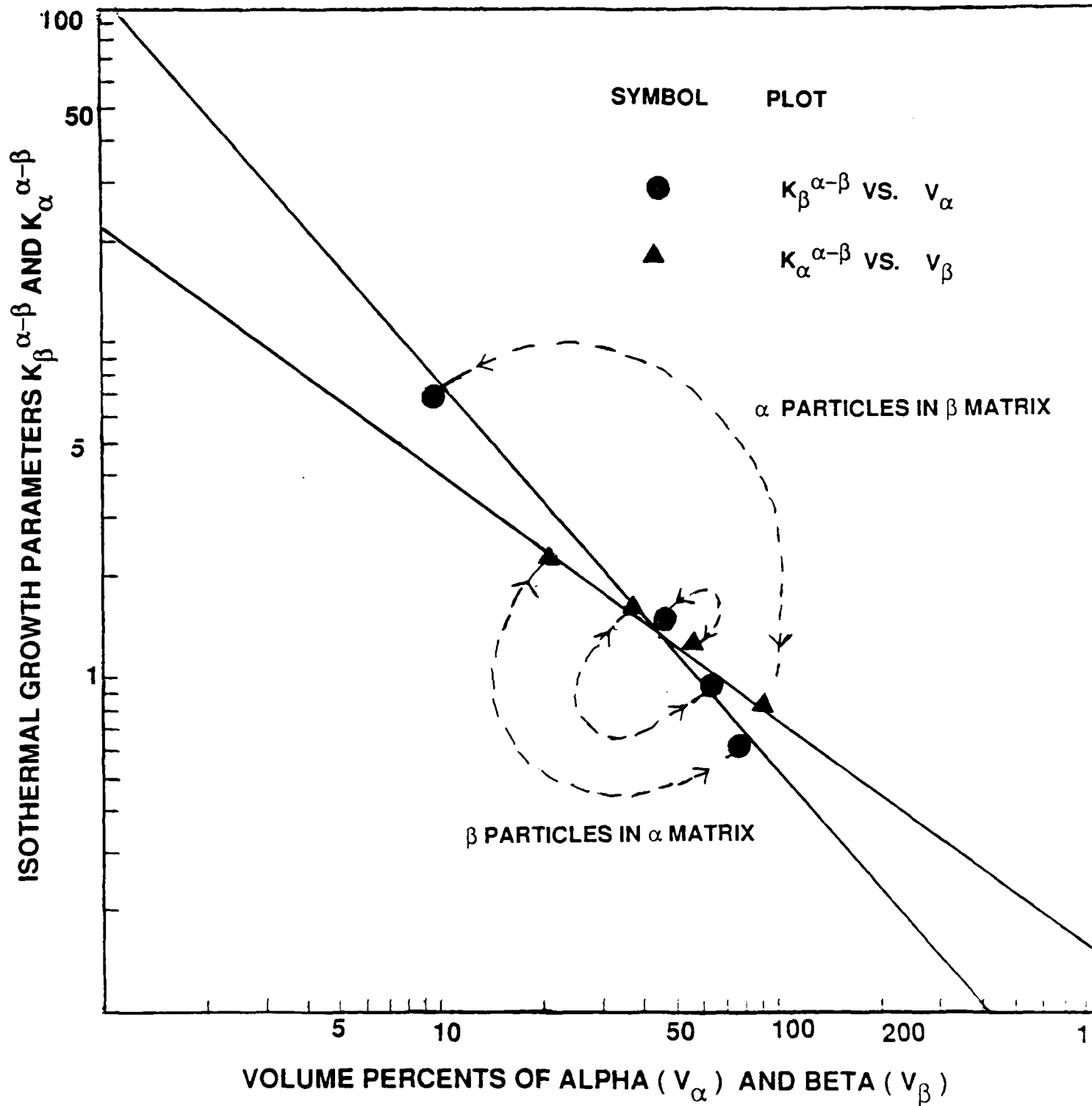


Fig. 9(b) : DEPENDENCY OF THE GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON VOLUME PERCENTS OF β (V_{β}) AND α (V_{α}), RESPECTIVELY, FOR THE Ti - Mn SYSTEM AT 973 K

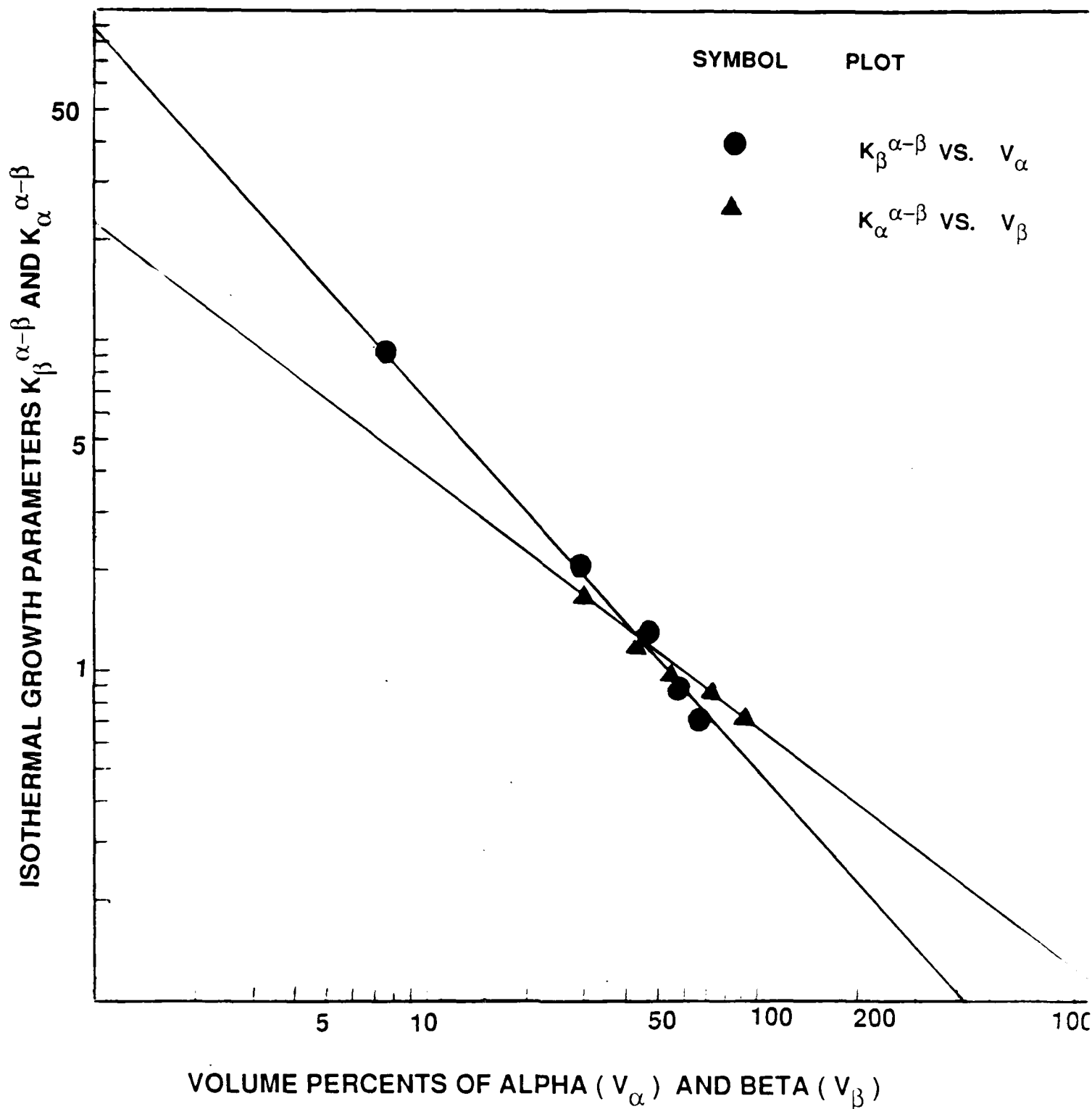


Fig.10(a) : DEPENDENCY OF THE GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON VOLUME PERCENTS OF β (V_{β}) AND α (V_{α}), RESPECTIVELY, FOR THE Ti - V SYSTEM AT 973 K

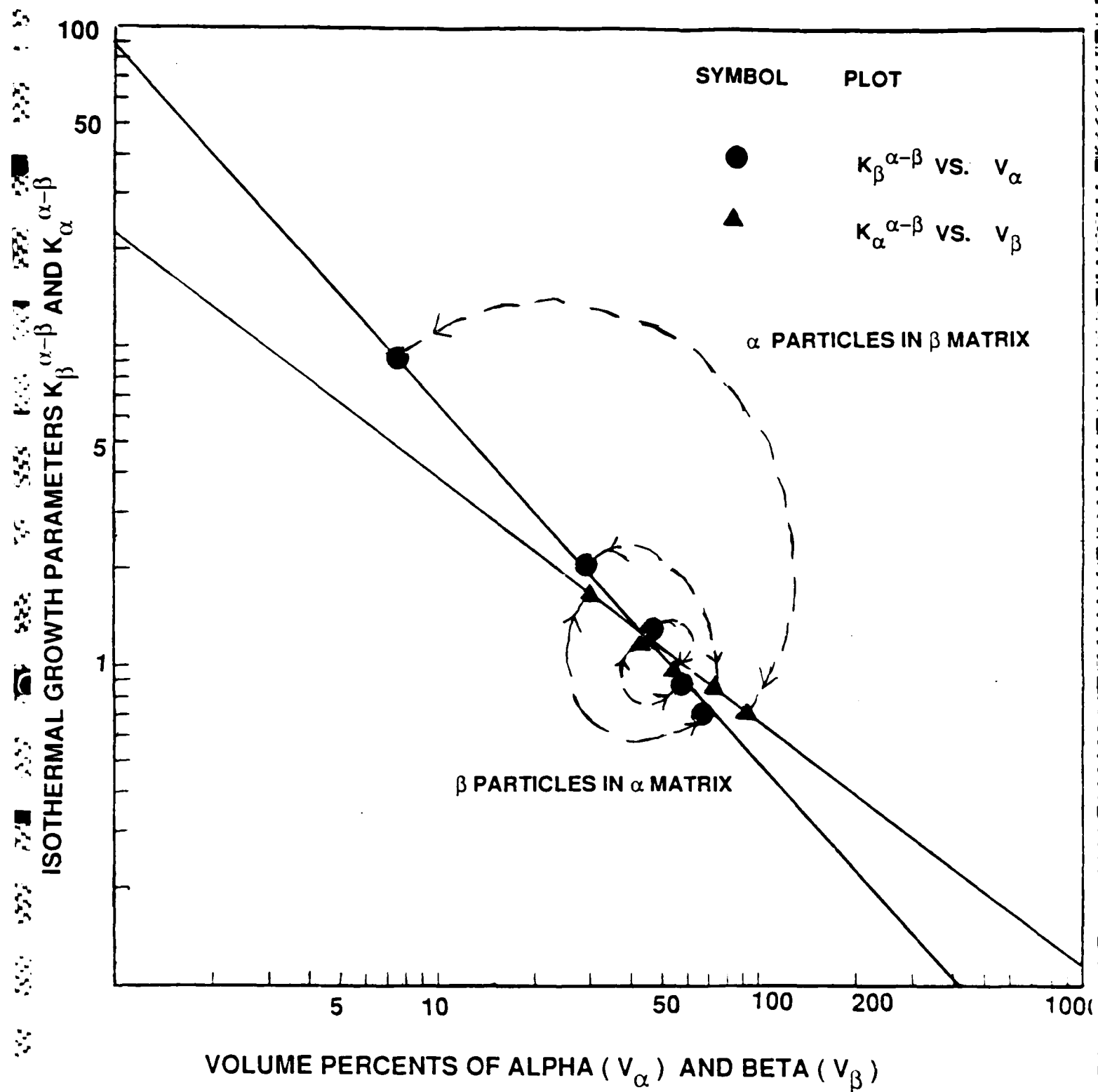


Fig.10(b): DEPENDENCY OF THE GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON VOLUME PERCENTS OF β (V_{β}) AND α (V_{α}), RESPECTIVELY, FOR THE Ti - V SYSTEM AT 973 K

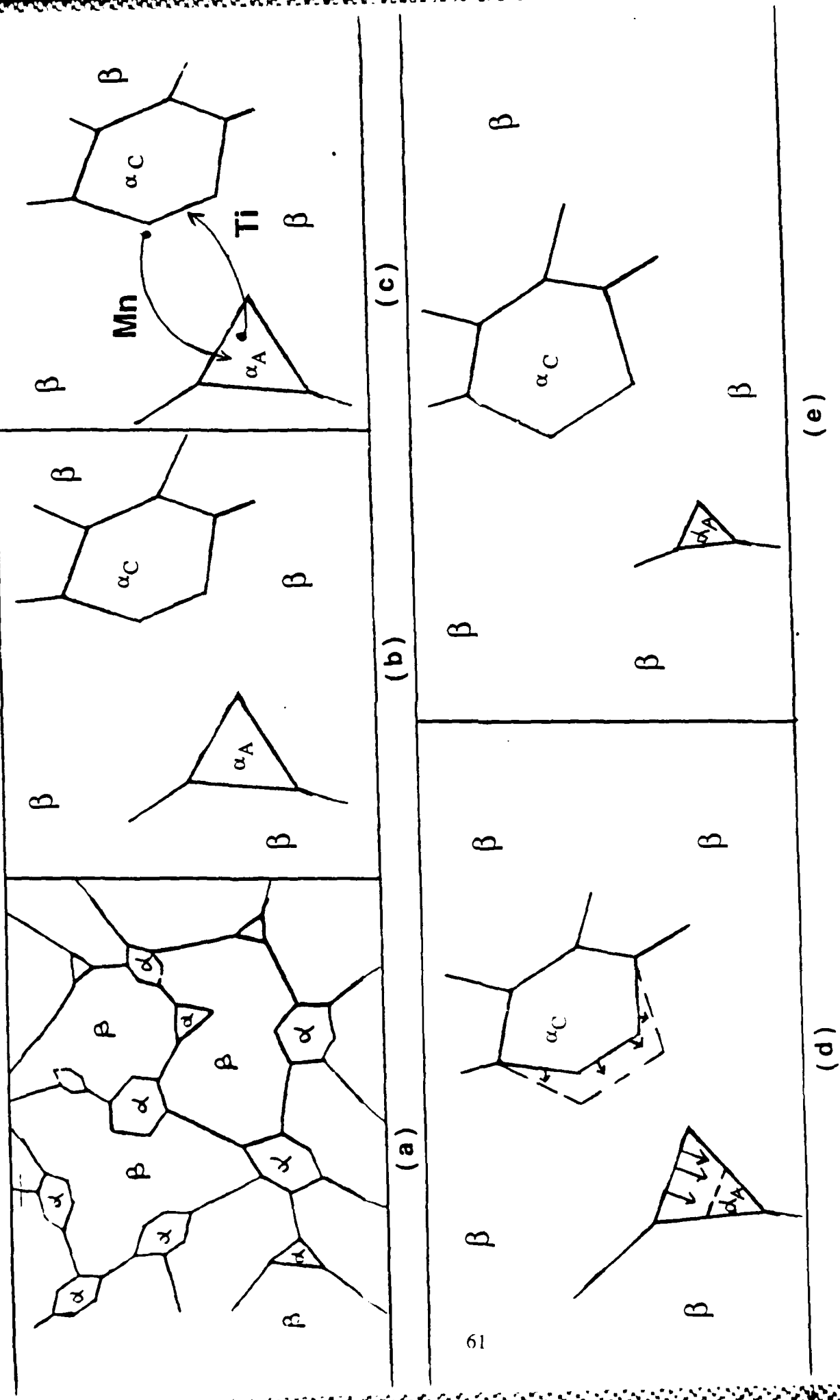


Fig. 11: A SCHEMATIC OF PARTICLE GROWTH IN TYPE 1
(ALPHA IN BETA MATRIX) MICROSTRUCTURES.

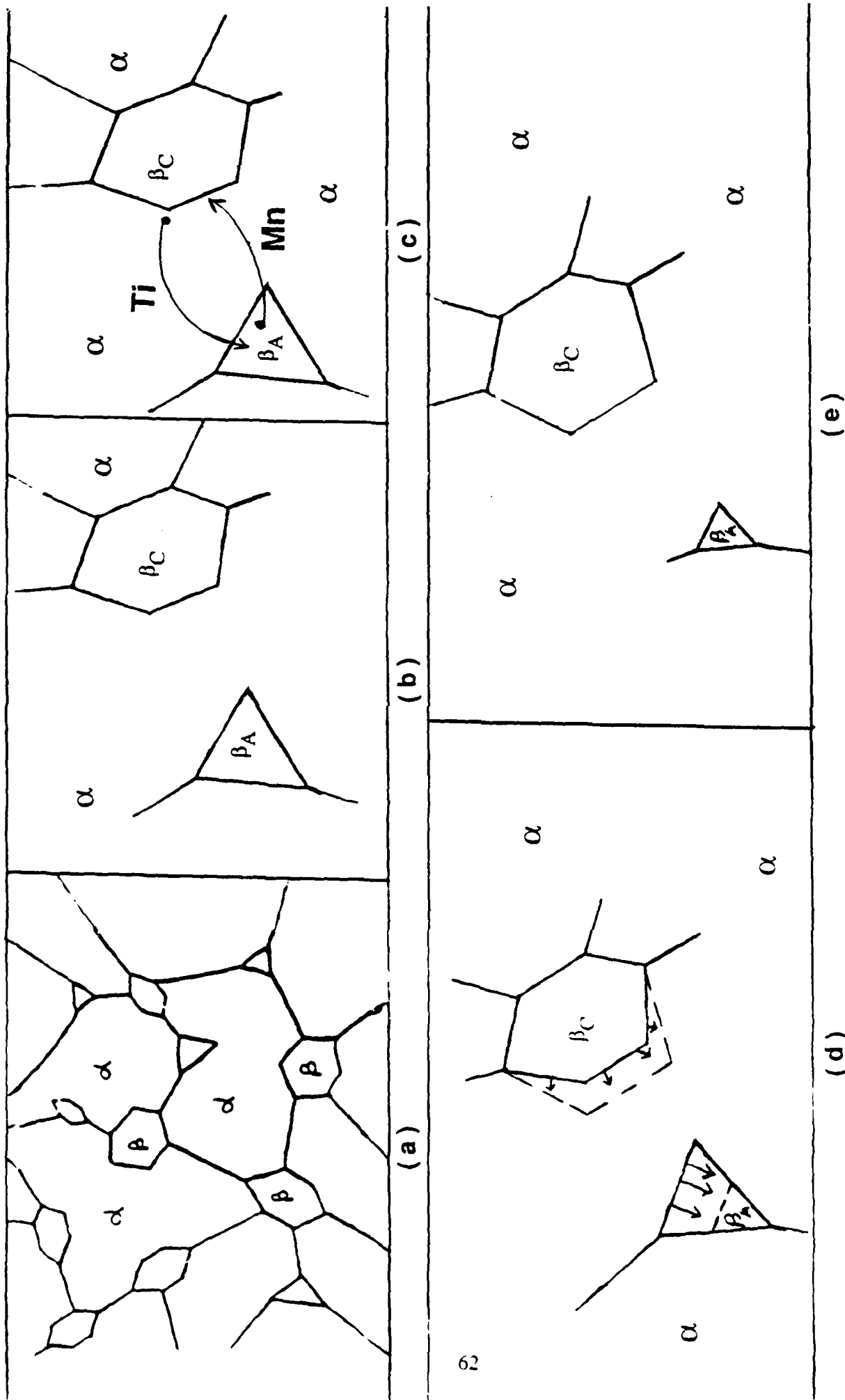
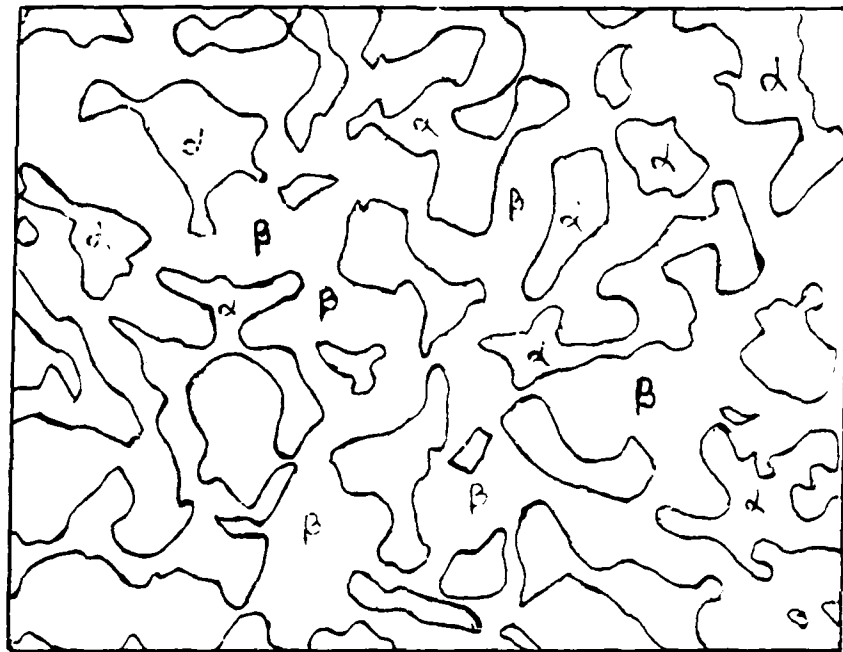
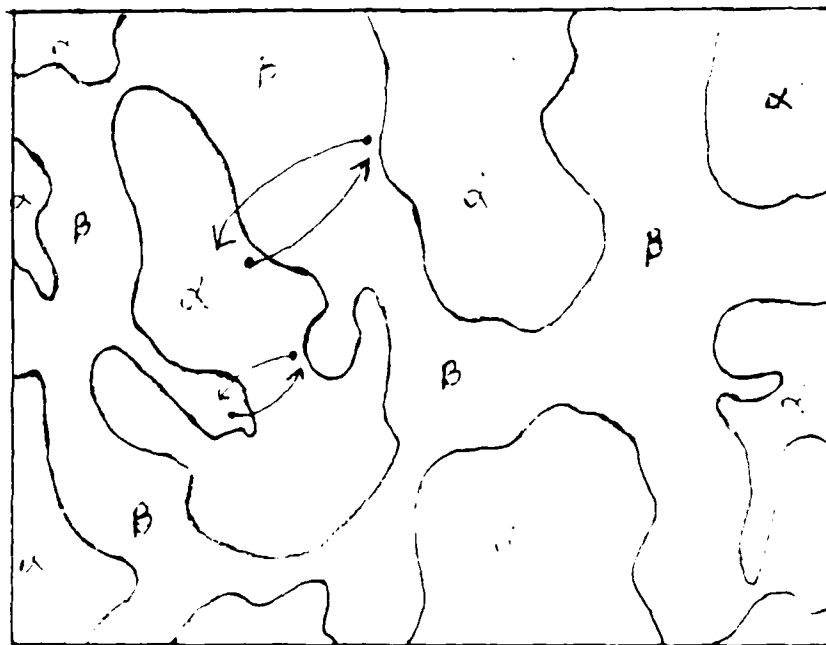


Fig. 12 : A SCHEMATIC OF PARTICLE GROWTH IN TYPE 2
(BETA IN ALPHA MATRIX) MICROSTRUCTURES.



(a)



(b)

Fig. 13 : A SCHEMATIC OF DIFFUSION PATHS IN TYPE 3
(NEARLY EQUIVALENT VOLUME PERCENTS OF α
AND β) MICROSTRUCTURES.

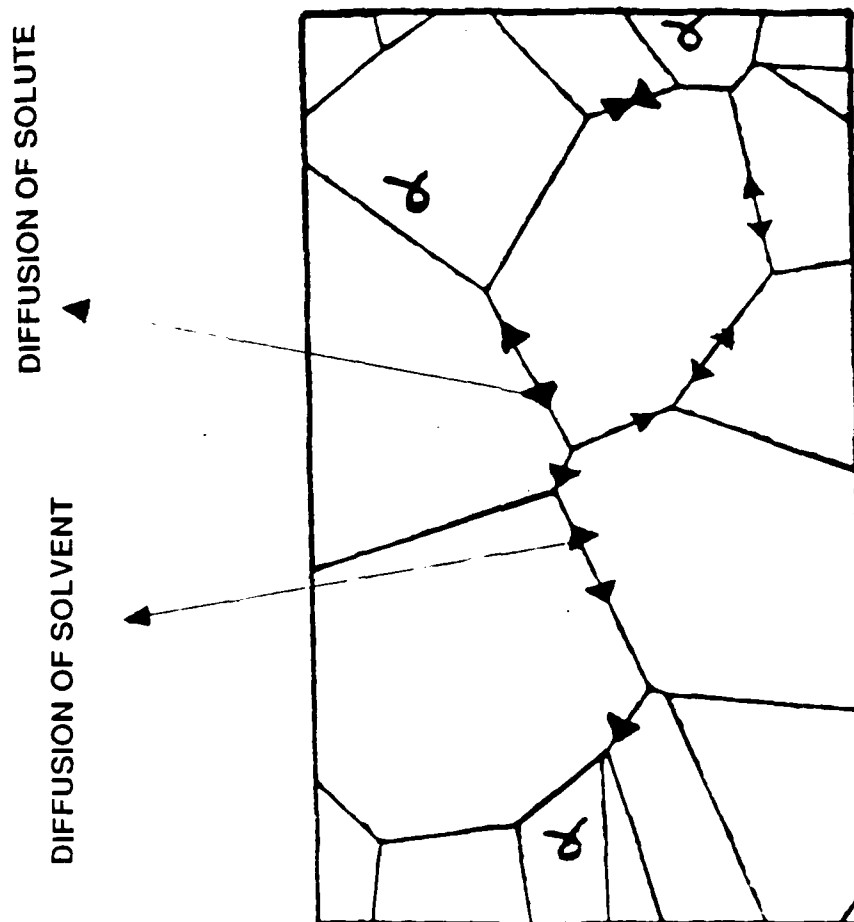
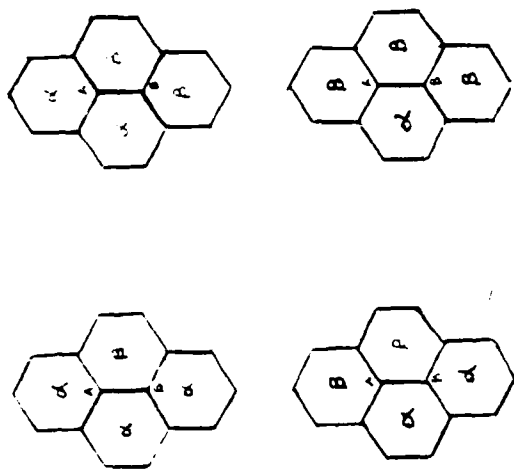


Fig. 14 : AN ALTERNATE MECHANISM OF BULK DIFFUSION VIA
 β GRAIN BOUNDARIES IN CLASS 1 MICROSTRUCTURES.



(a)



(b)

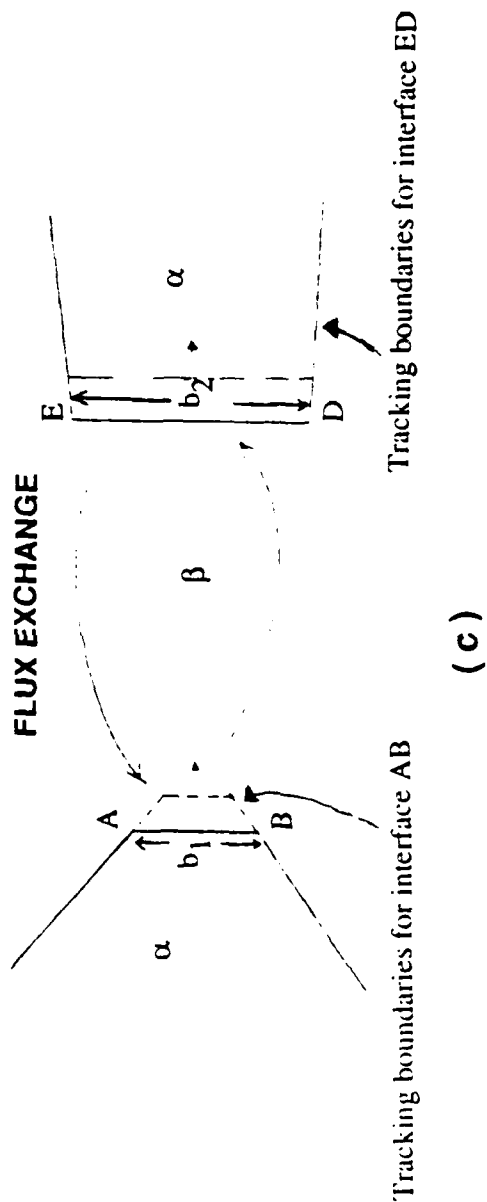


Fig. 15 : (a) A NETWORK OF α AND β GRAINS IN A $\alpha - \beta$ TWO PHASE ALLOY (b) TYPICAL NEAREST NEIGHBOURS IN A $\alpha - \beta$ TWO PHASE ALLOY. (c) AN EXPLODED VIEW OF TWO INTERFACES.

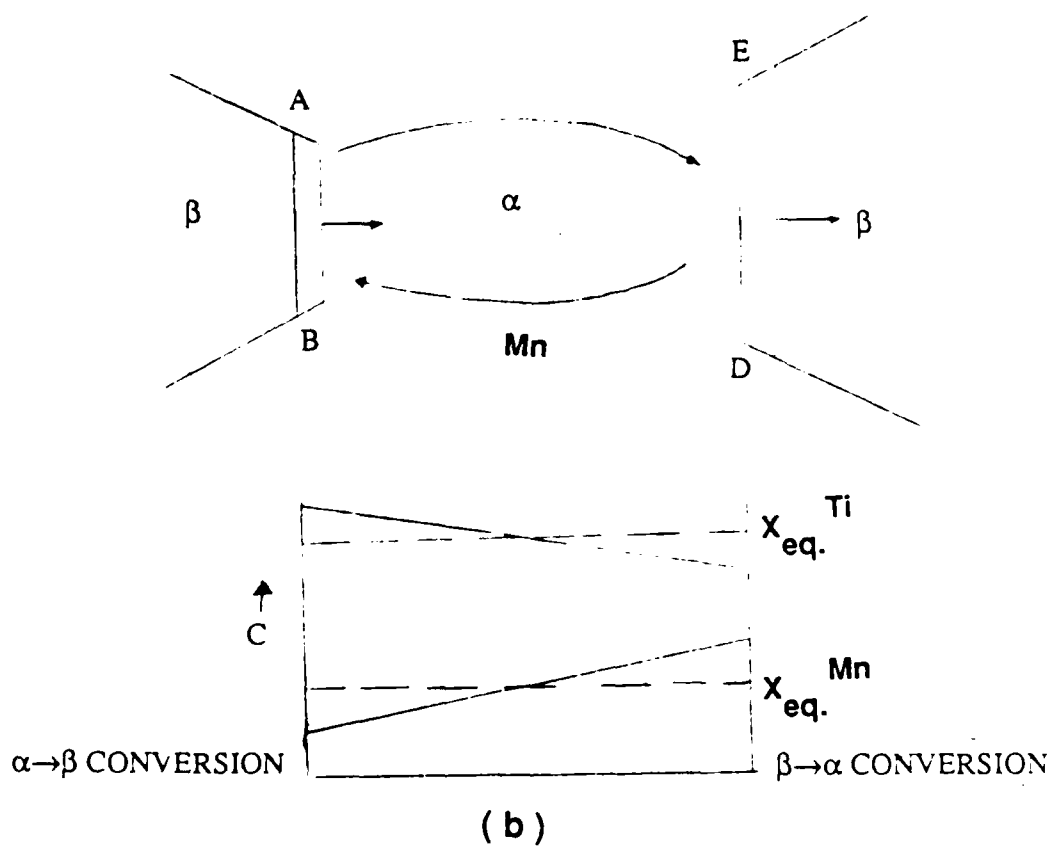
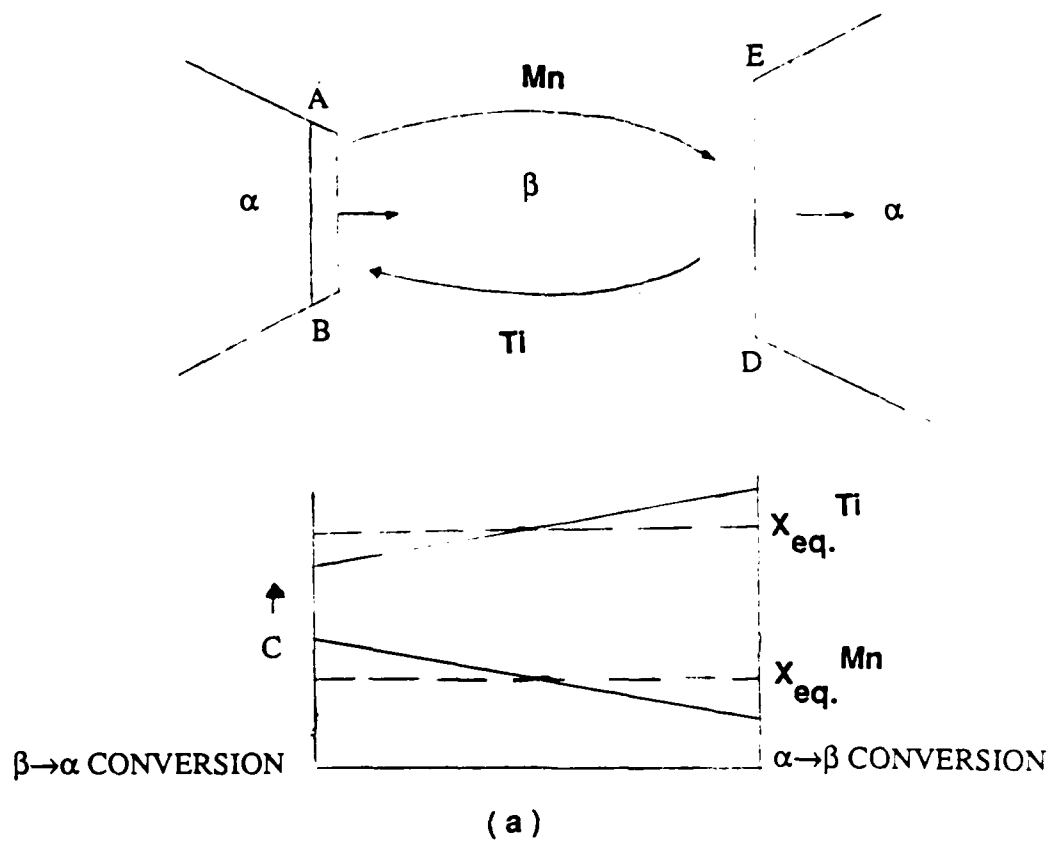


Fig. 16 : (a) TYPICAL CONCENTRATION GRADIENTS OF Ti AND Mn IN MATRIX IN CLASS 1 MICROSTRUCTURES. (b) TYPICAL CONCENTRATION GRADIENTS IN MATRIX IN CLASS 2 MICROSTRUCTURES.



(a)

(b)

Fig. 17 : TYPICAL PLANAR (a) AND CURVED (b) INTERFACES
IN α - β TITANIUM ALLOYS.

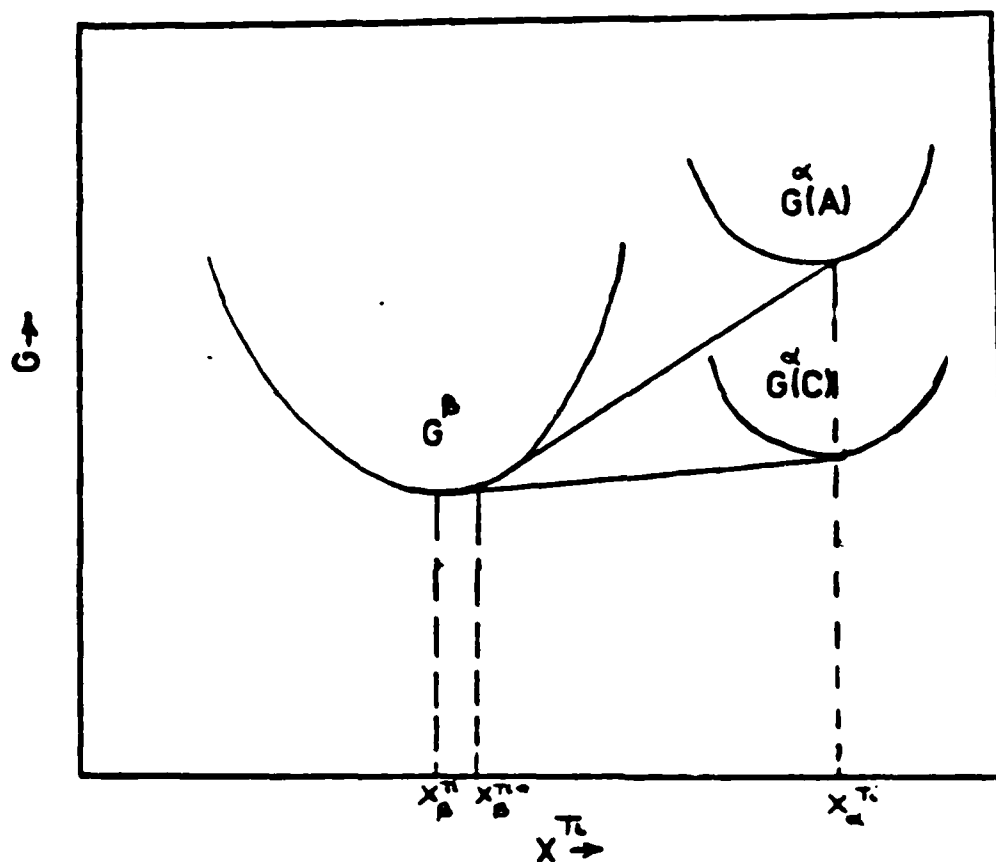


Fig. 18 (a): MOLAR FREE ENERGY PLOTS OF PARTICLE AND MATRIX PHASES IN CLASS 1 MICROSTRUCTURES.

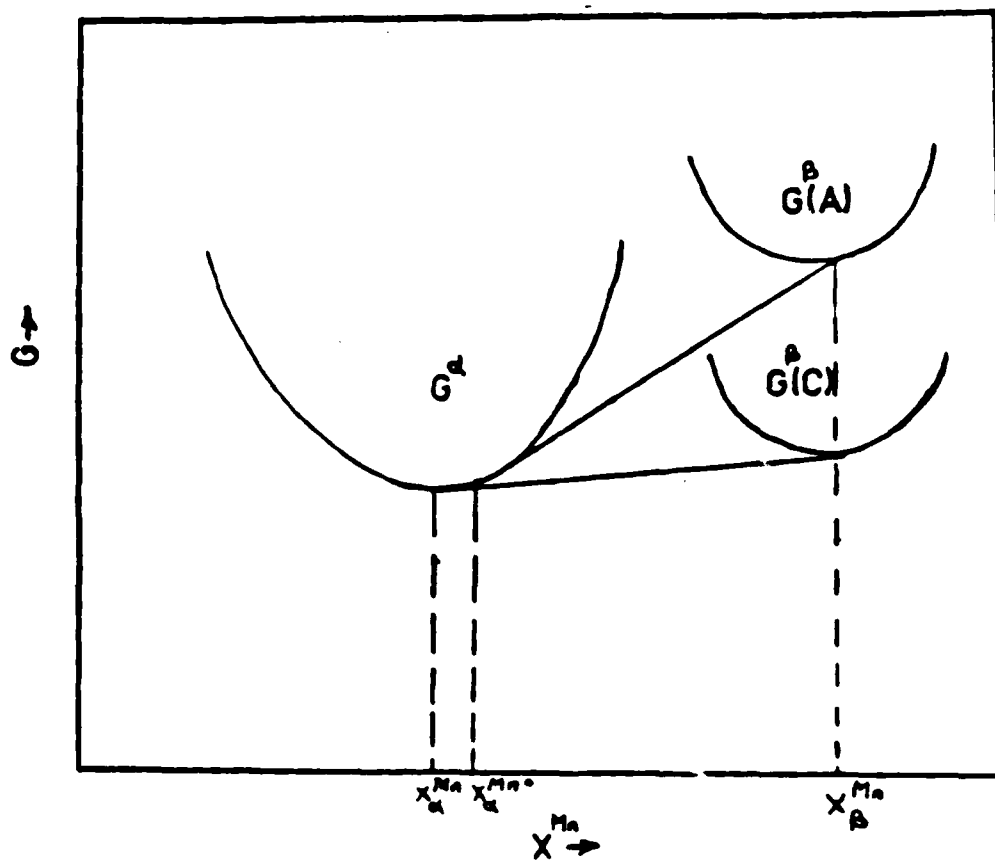


Fig. 18 (b): MOLAR FREE ENERGY PLOTS OF PARTICLE AND MATRIX PHASES IN CLASS 2 MICROSTRUCTURES.

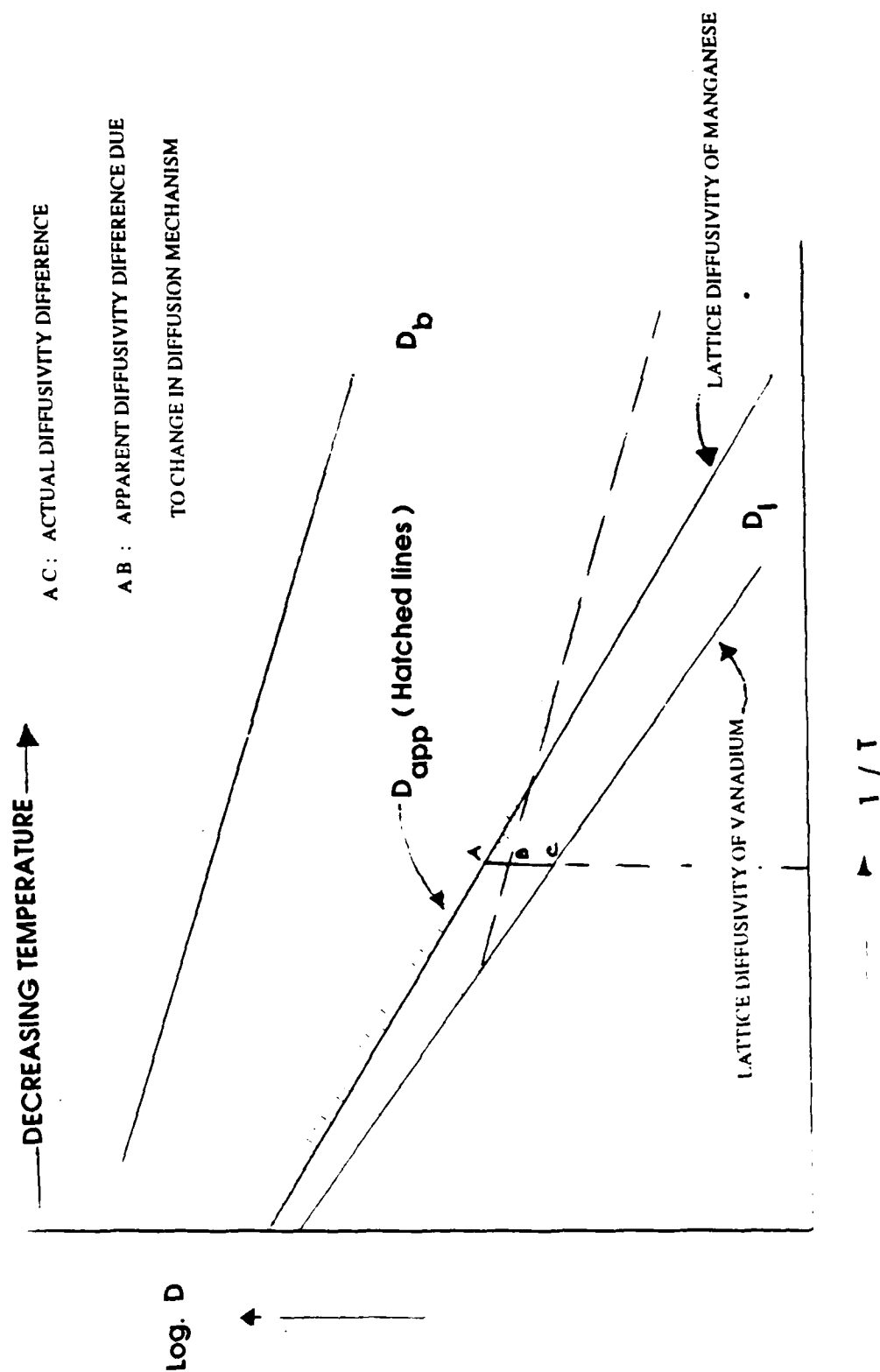


Fig. 19 (a) : ACTUAL AND APPARENT DIFFUSIVITY DIFFERENCES IN
Ti - Mn AND Ti - V ALLOY SYSTEMS.



Fig. 19 (b) : A TYPICAL GRAIN BOUNDARY STRUCTURE IN AN $\alpha-\beta$ ALLOY.

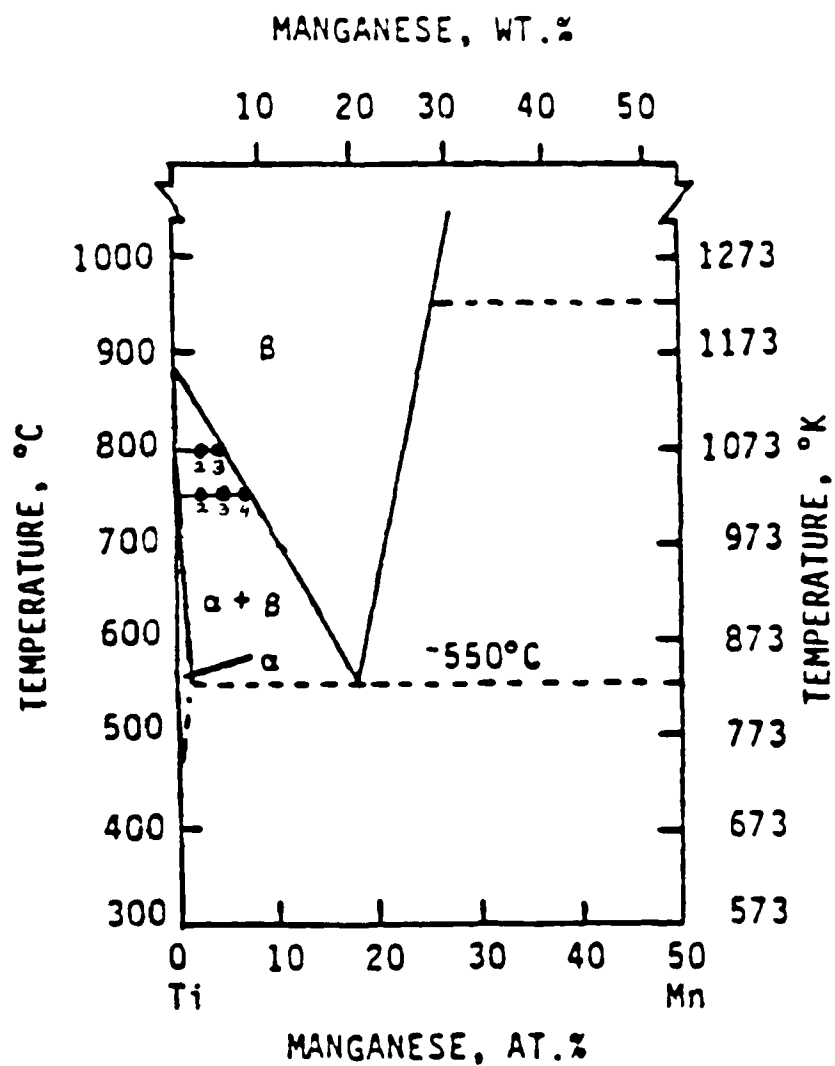


Fig. 20 : PARTIAL Ti END OF THE Ti - Mn PHASE DIAGRAM. THE AIM COMPOSITIONS OF ALLOYS ARE INDICATED WITH SOLID CIRCLES.

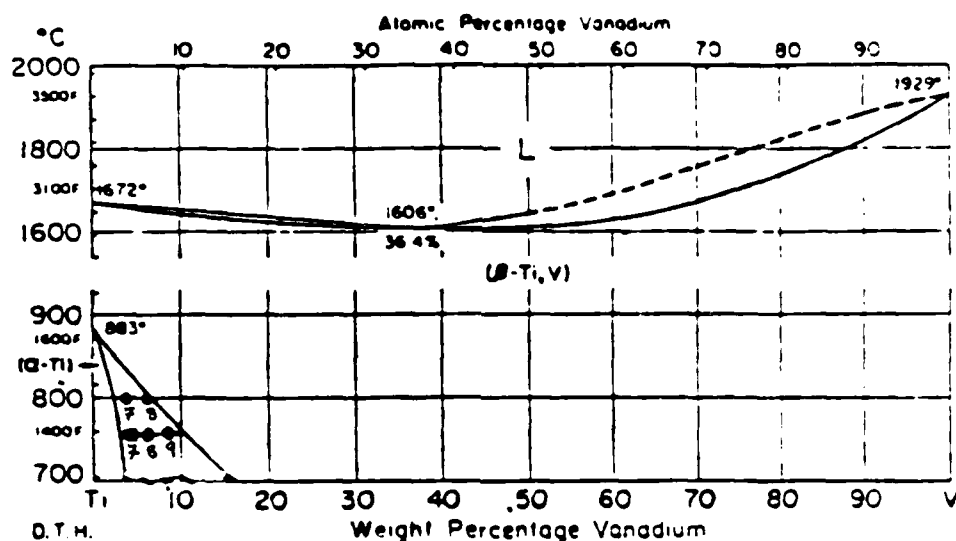
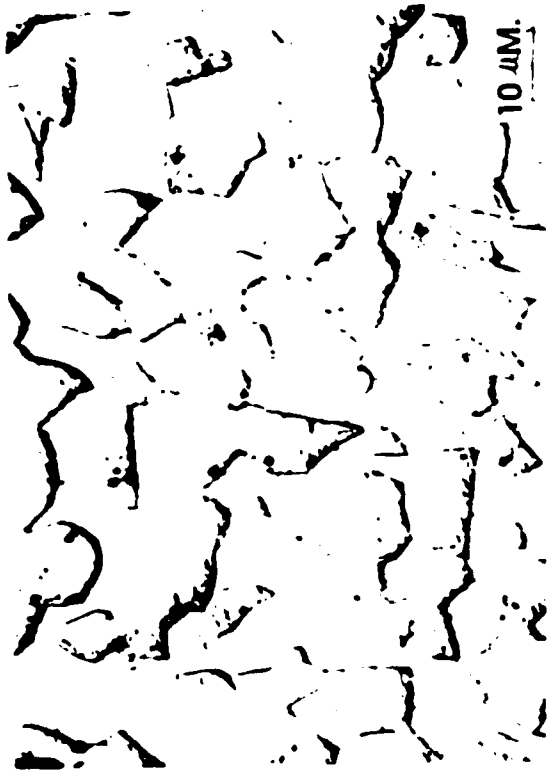


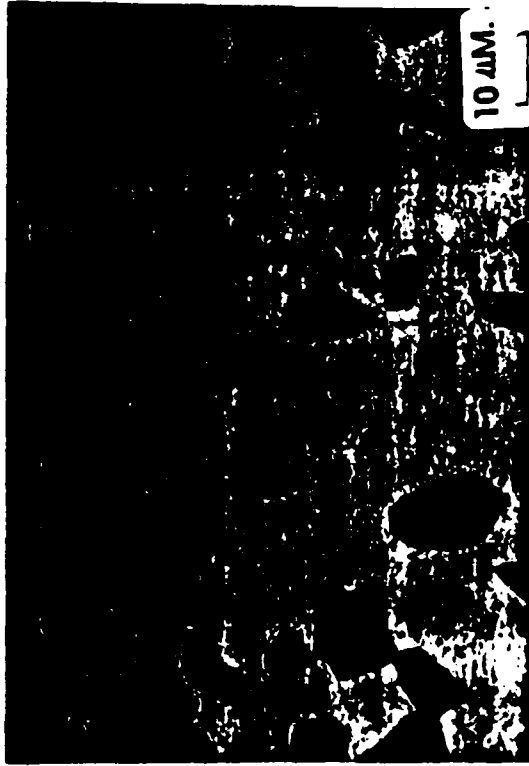
Fig. 21 : THE Ti - V PHASE DIAGRAM. THE AIM COMPOSITIONS OF ALLOYS ARE INDICATED WITH SOLID CIRCLES.



74 % α - 26 % β



49 % α - 51 % β



18 % α - 82 % β

Fig. 22 : MICROSTRUCTURES OF VARIOUS α - β Ti-Mn ALLOYS
ANNEALED FOR 200 HRS AT 1023° K



59 % α - 41 % β



39 % α - 61 % β



12 % α - 88 % β

Fig. 23 : MICROSTRUCTURES OF VARIOUS α β Ti-V ALLOYS
ANNEALED FOR 200 HRS. AT 1023° K

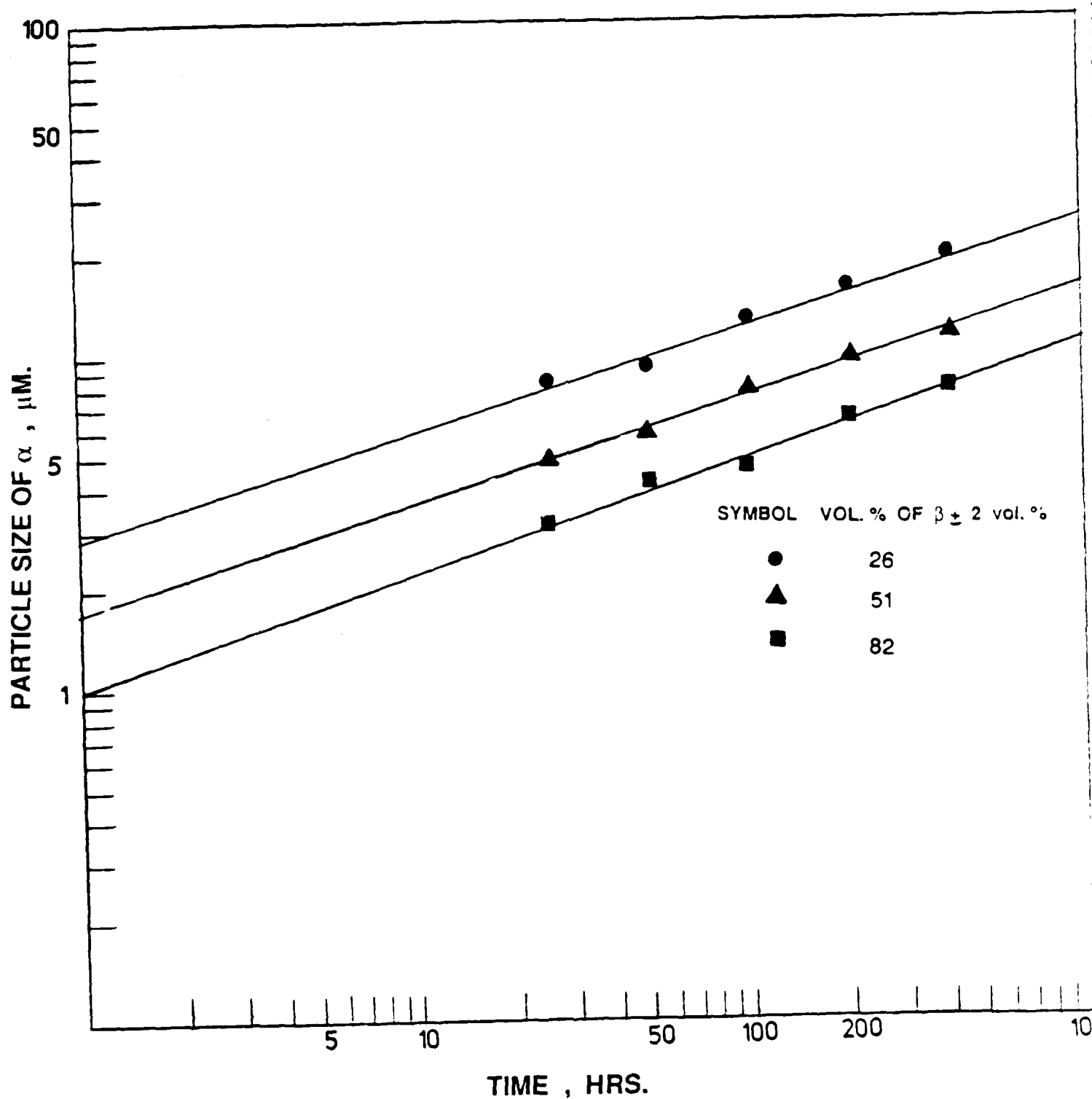


Fig. 24 : PARTICLE SIZES OF α - PHASE IN α - β Ti - Mn ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1023 K, W. Q.

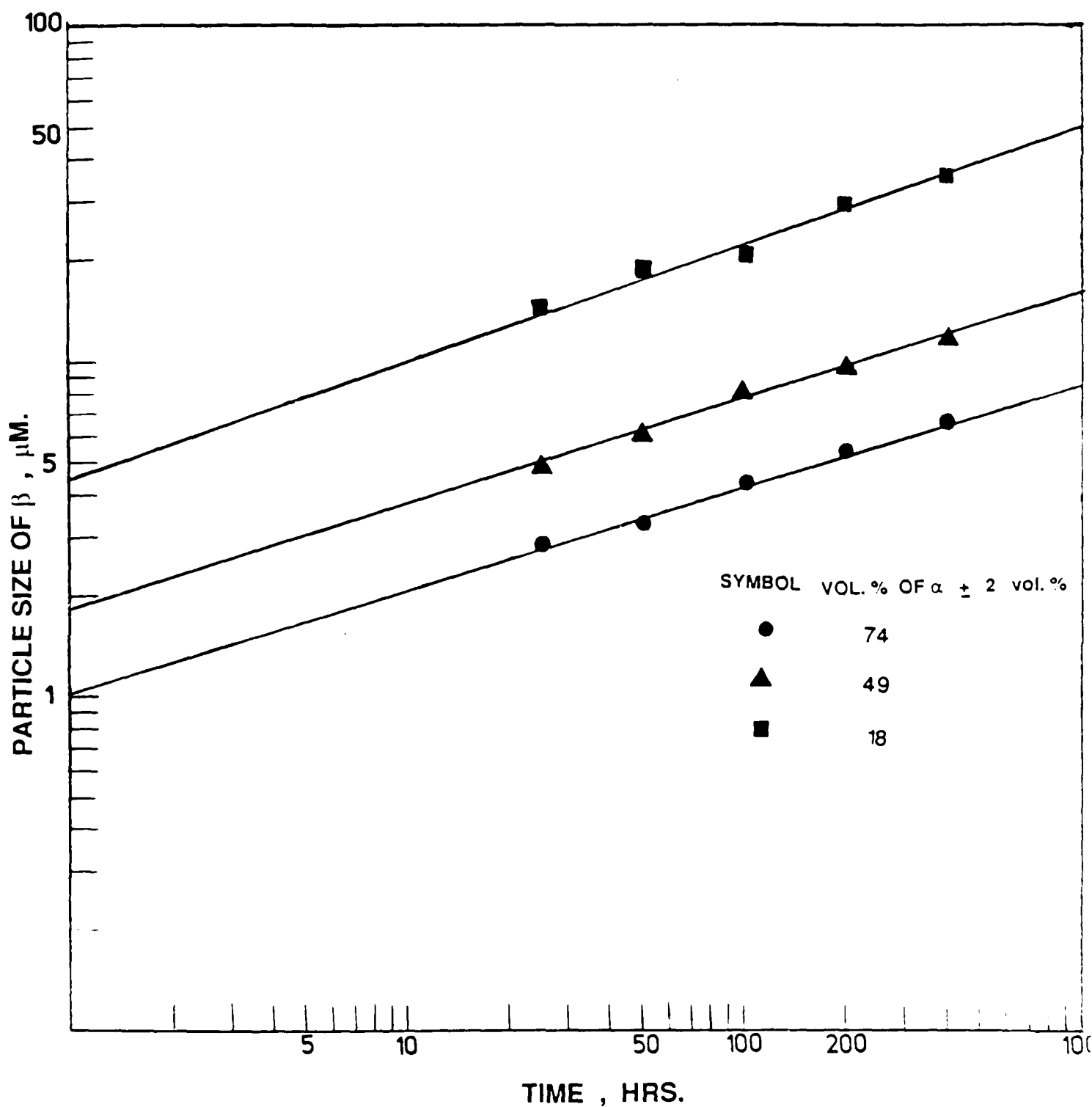


Fig. 25 : PARTICLE SIZES OF β - PHASE IN α - β Ti - Mn ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1023 K, W. Q.

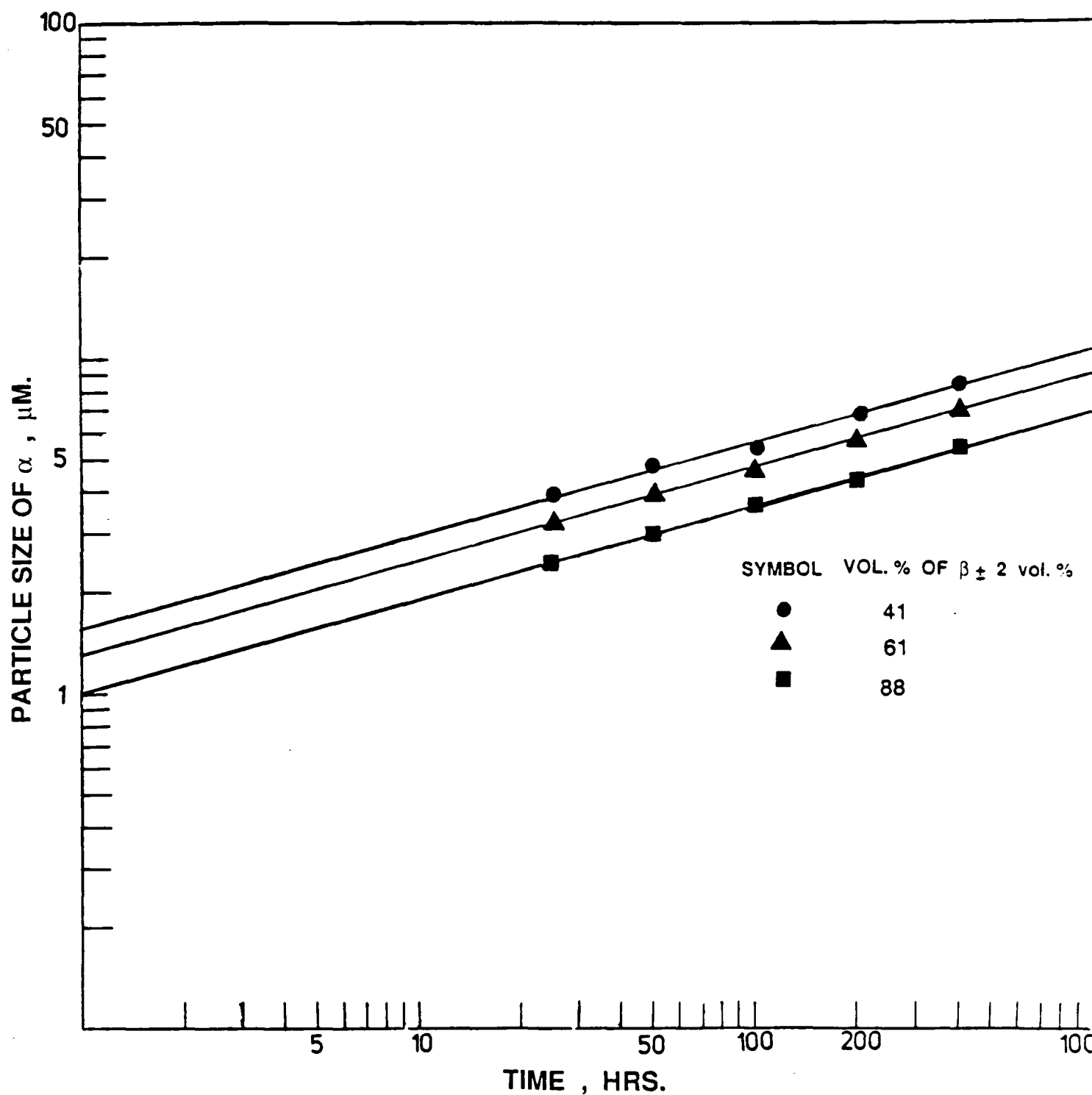


Fig. 26 : PARTICLE SIZES OF α - PHASE IN α - β Ti - V ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1023 K , W. Q.

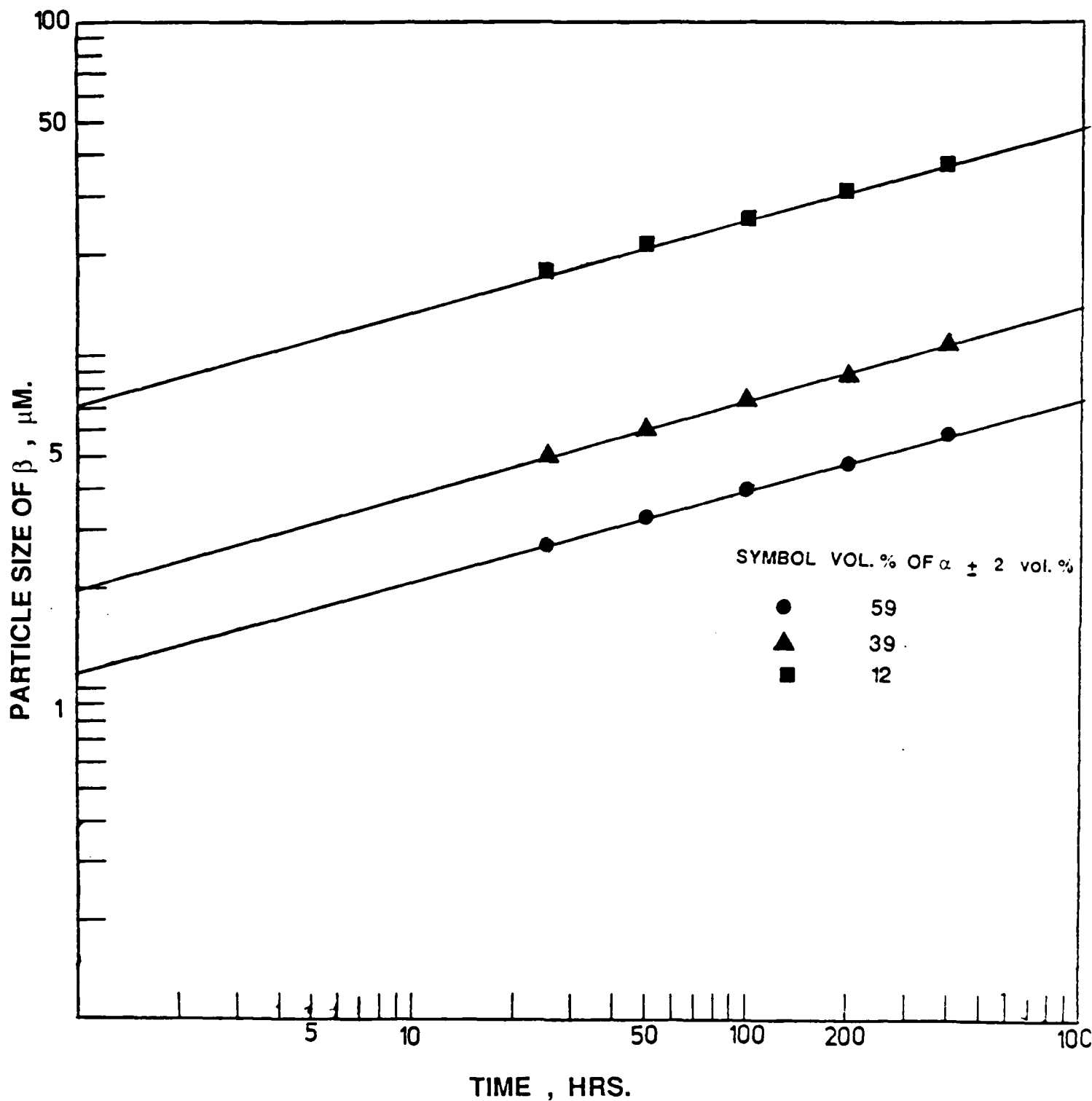


Fig. 27: PARTICLE SIZES OF β - PHASE IN α - β Ti - V ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1023 K , W. Q.

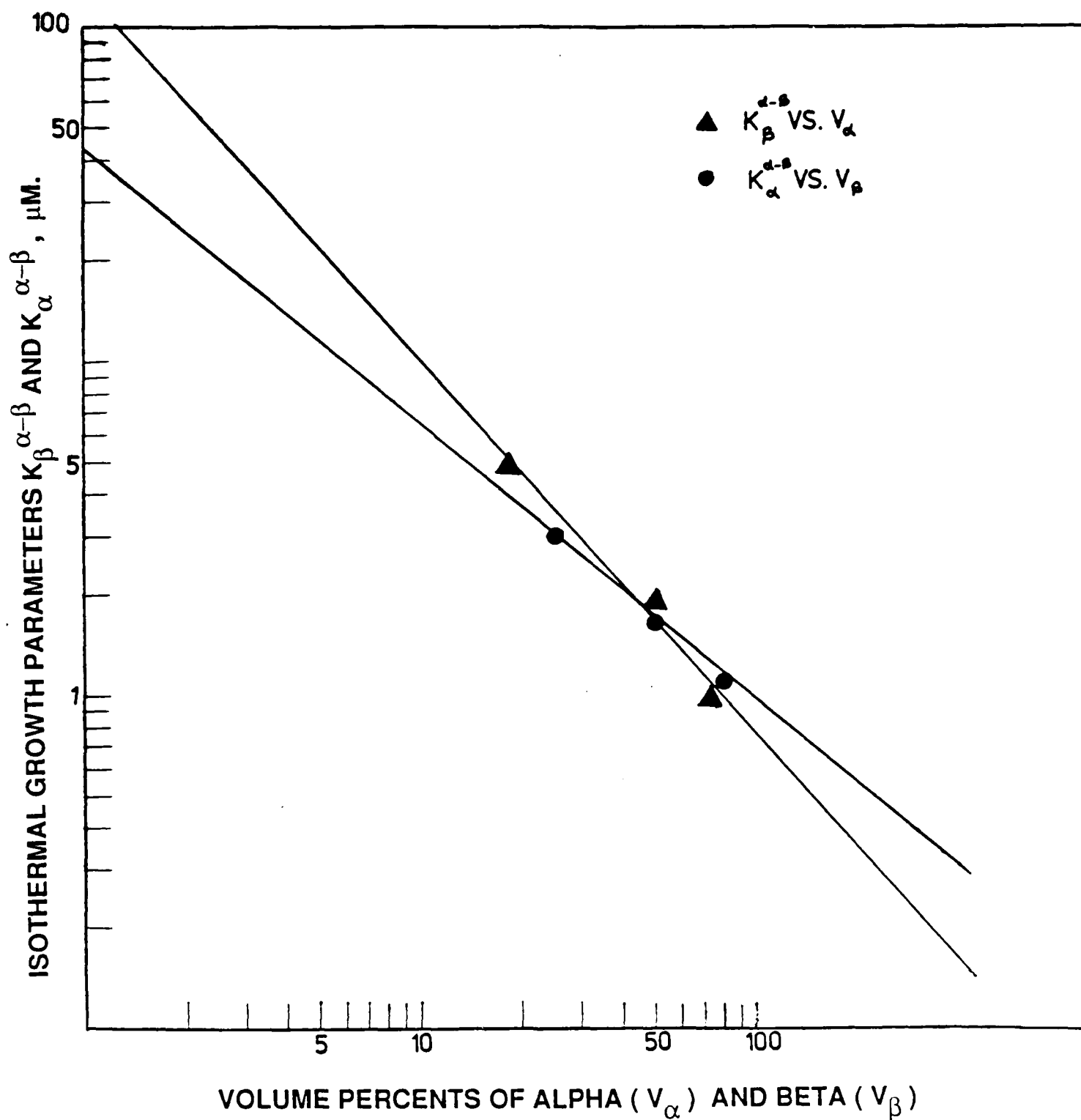


Fig. 28: DEPENDENCY OF THE GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON VOLUME PERCENTS OF β (V_{β}) AND α (V_{α}), RESPECTIVELY, FOR THE Ti - Mn SYSTEM AT 1023 K

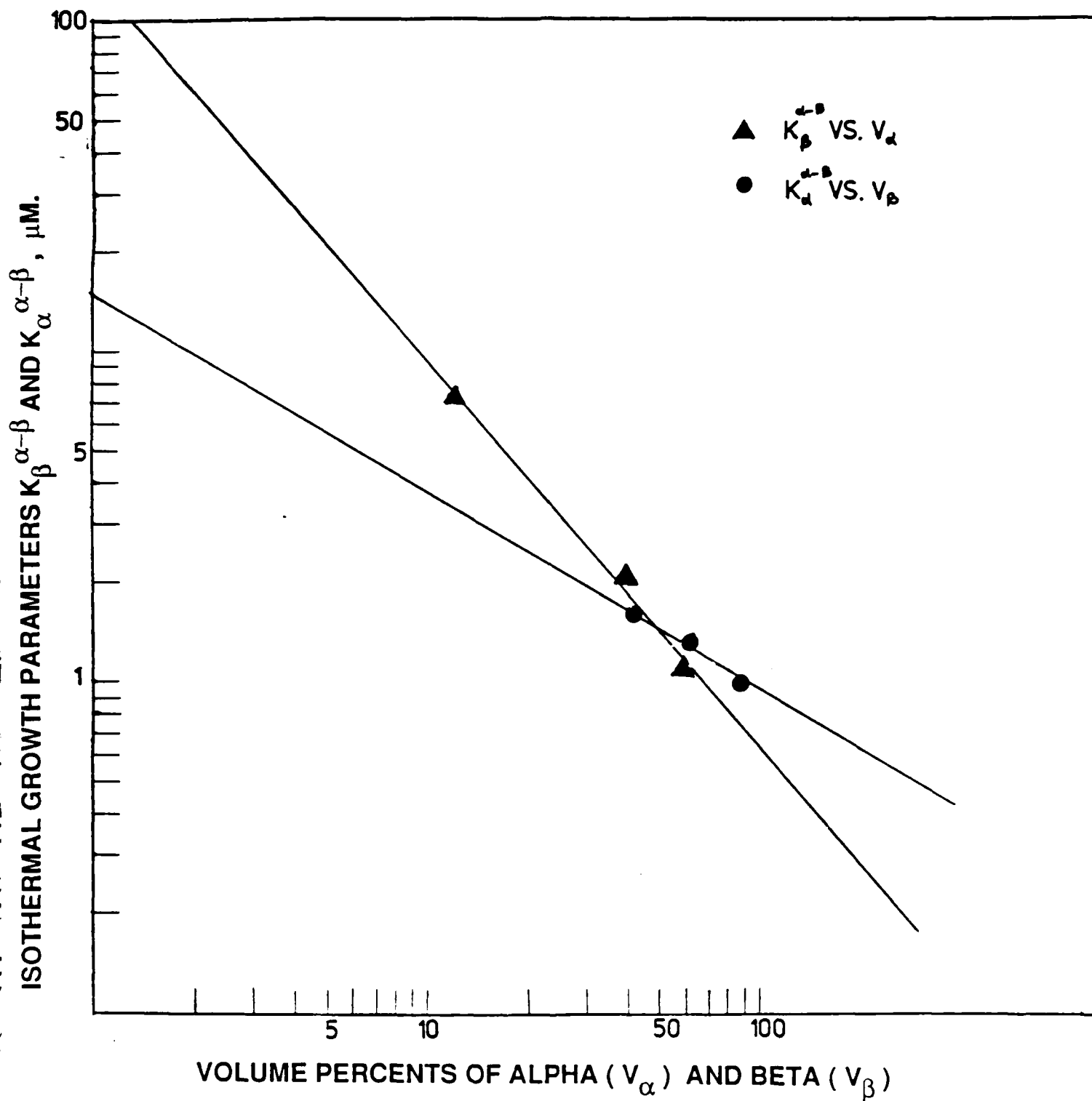
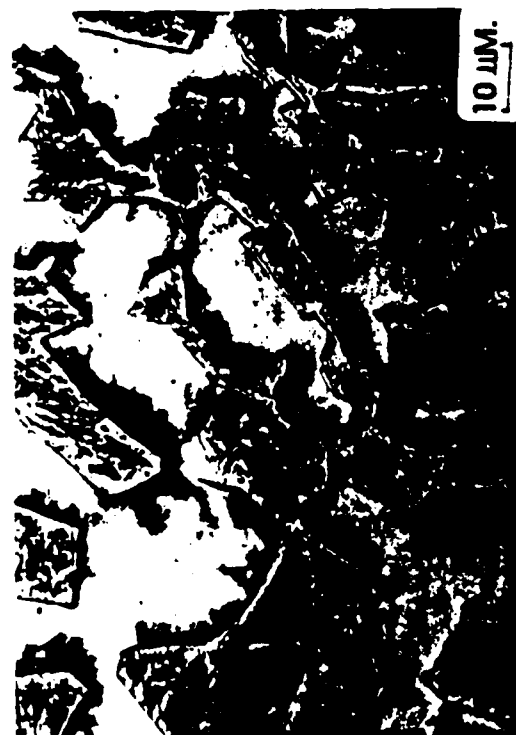


Fig. 29: DEPENDENCY OF THE GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON VOLUME PERCENTS OF β (V_{β}) AND α (V_{α}), RESPECTIVELY, FOR THE Ti - V SYSTEM AT 1023 K



54 % α - 46 % β



24 % α - 76 % β

Fig. 30: MICROSTRUCTURES OF VARIOUS α - β Ti-Mn ALLOYS
ANNEALED FOR 200 HRS. AT 1073° K



36 % α - 64 % β



6 % α - 94 % β

Fig. 31 : MICROSTRUCTURES OF VARIOUS α - β Ti-V ALLOYS
ANNEALED FOR 200 HRS. AT 1073° K

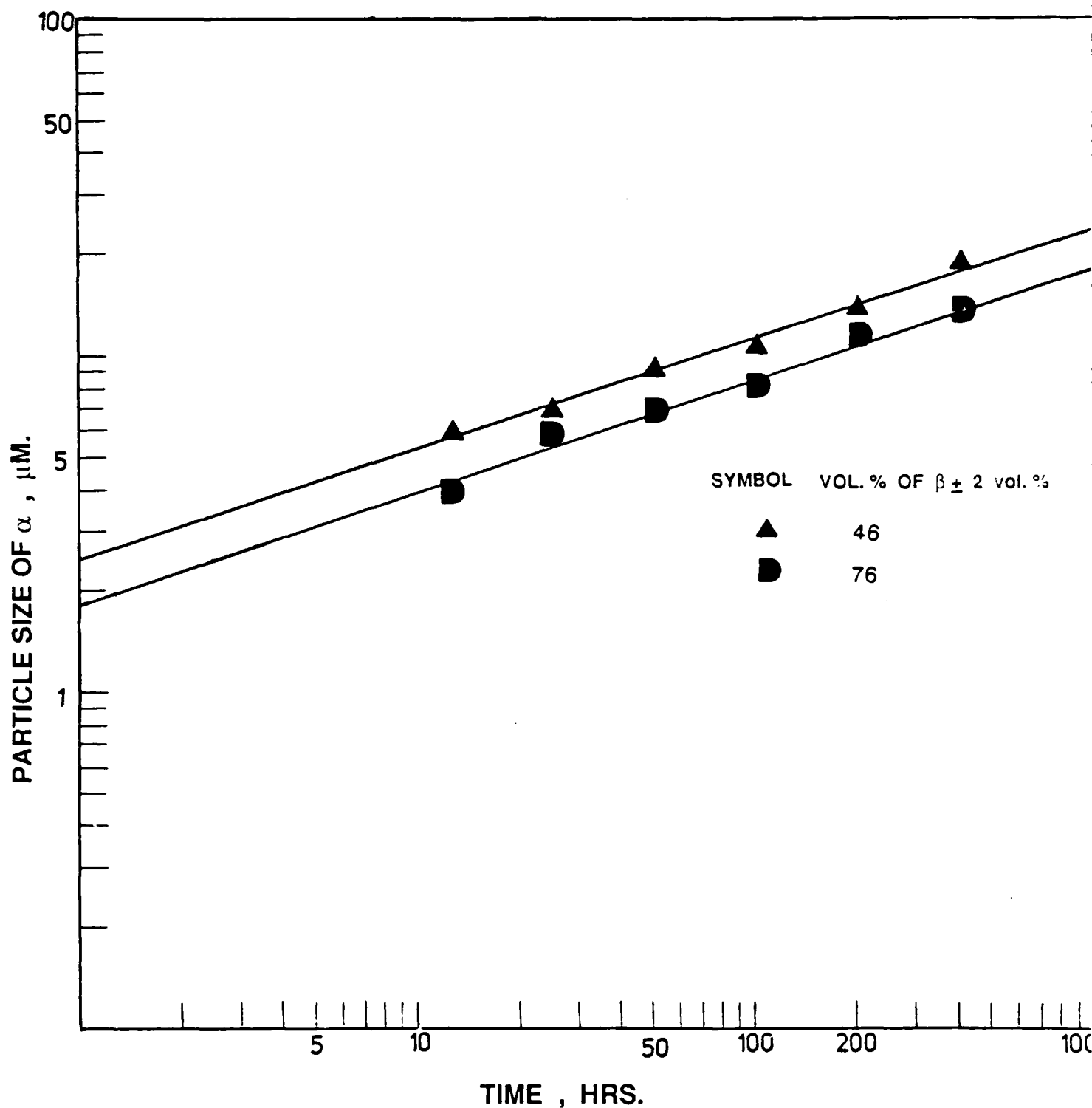


Fig. 32: PARTICLE SIZES OF α - PHASE IN α - β Ti - Mn ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1073 K , W. Q.

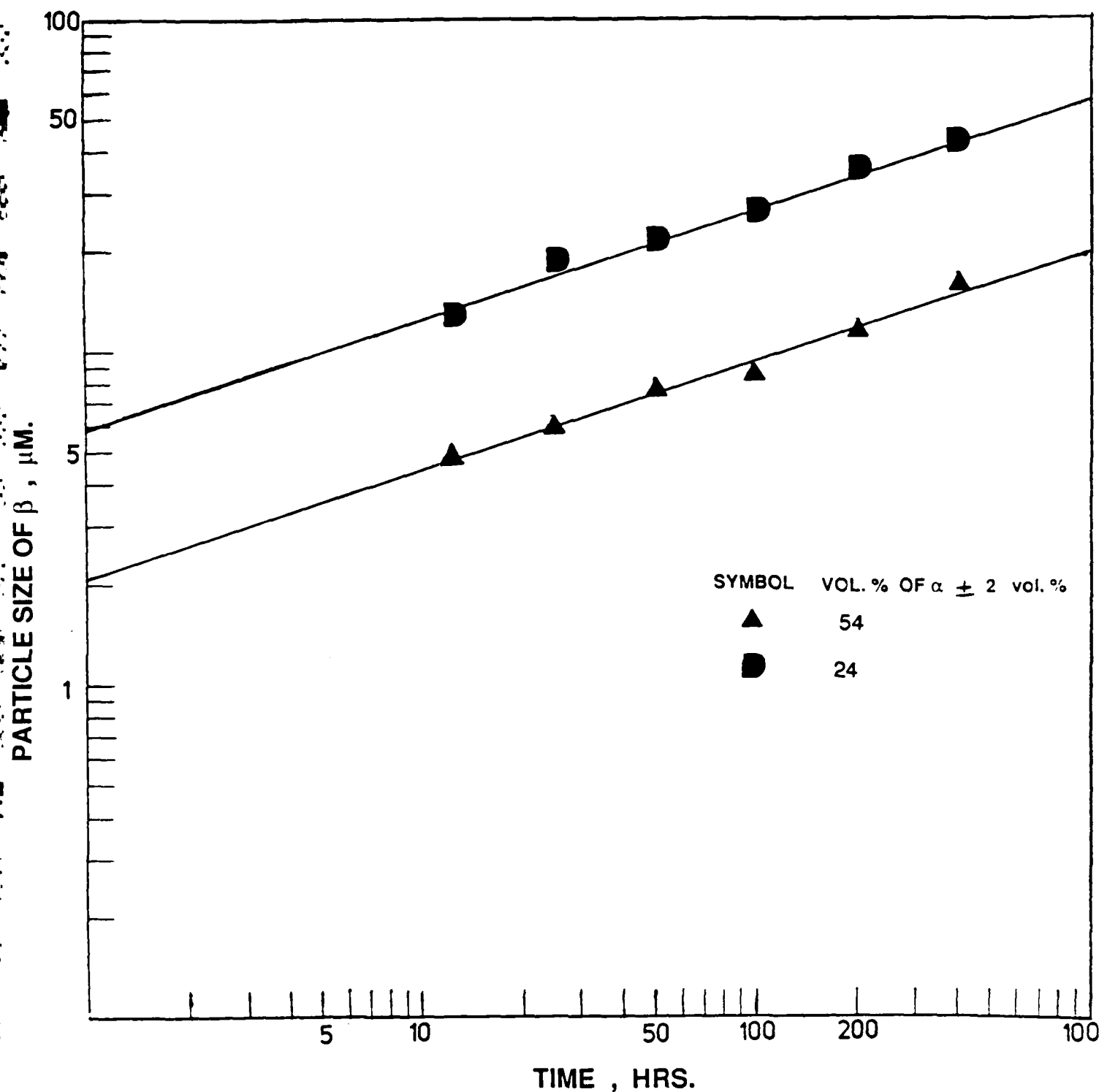


Fig. 33: PARTICLE SIZES OF β - PHASE IN α - β Ti - Mn ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1073 K , W. Q.

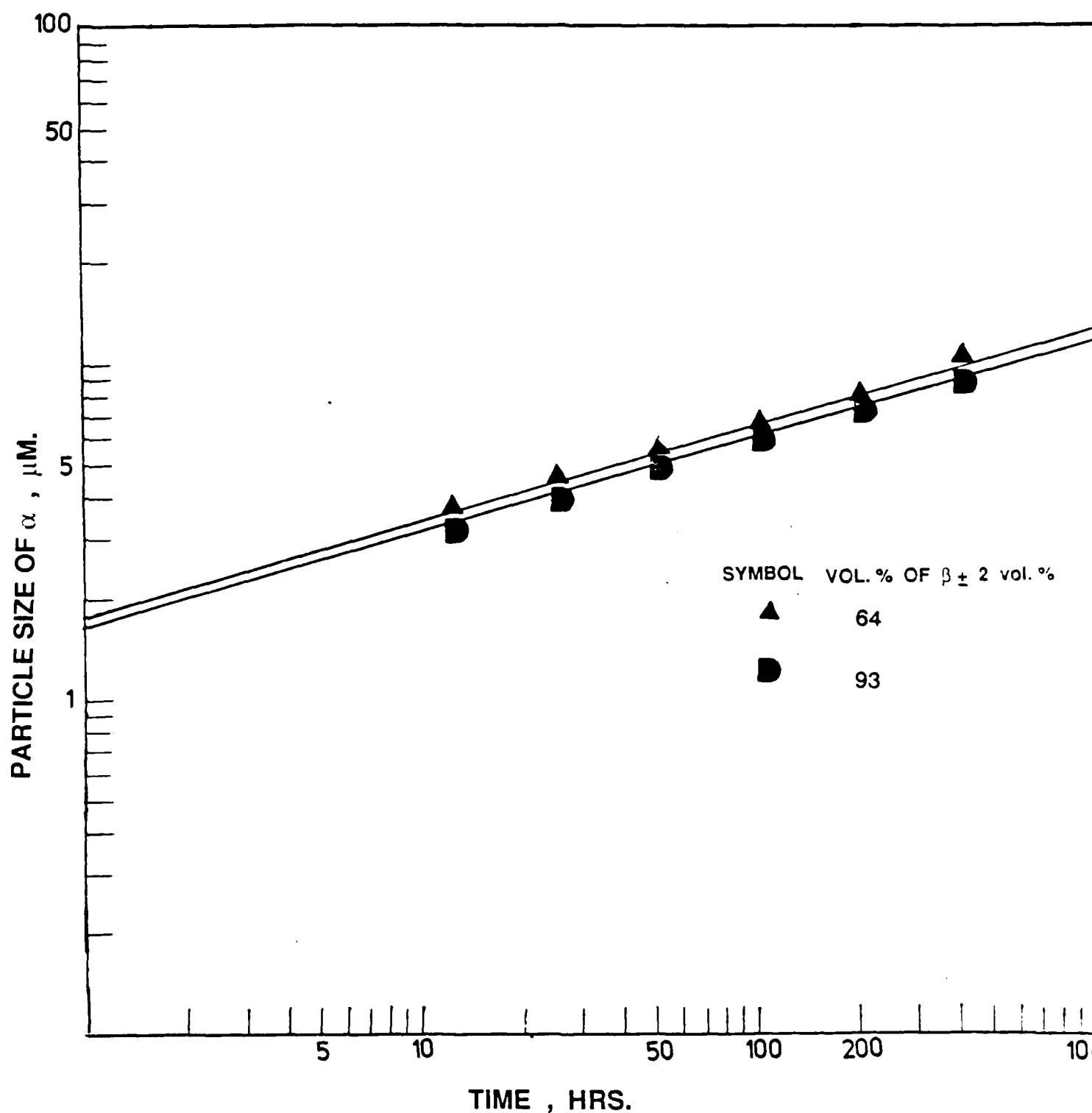


Fig. 34: PARTICLE SIZES OF α - PHASE IN α - β Ti - V ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1073 K , W. Q.

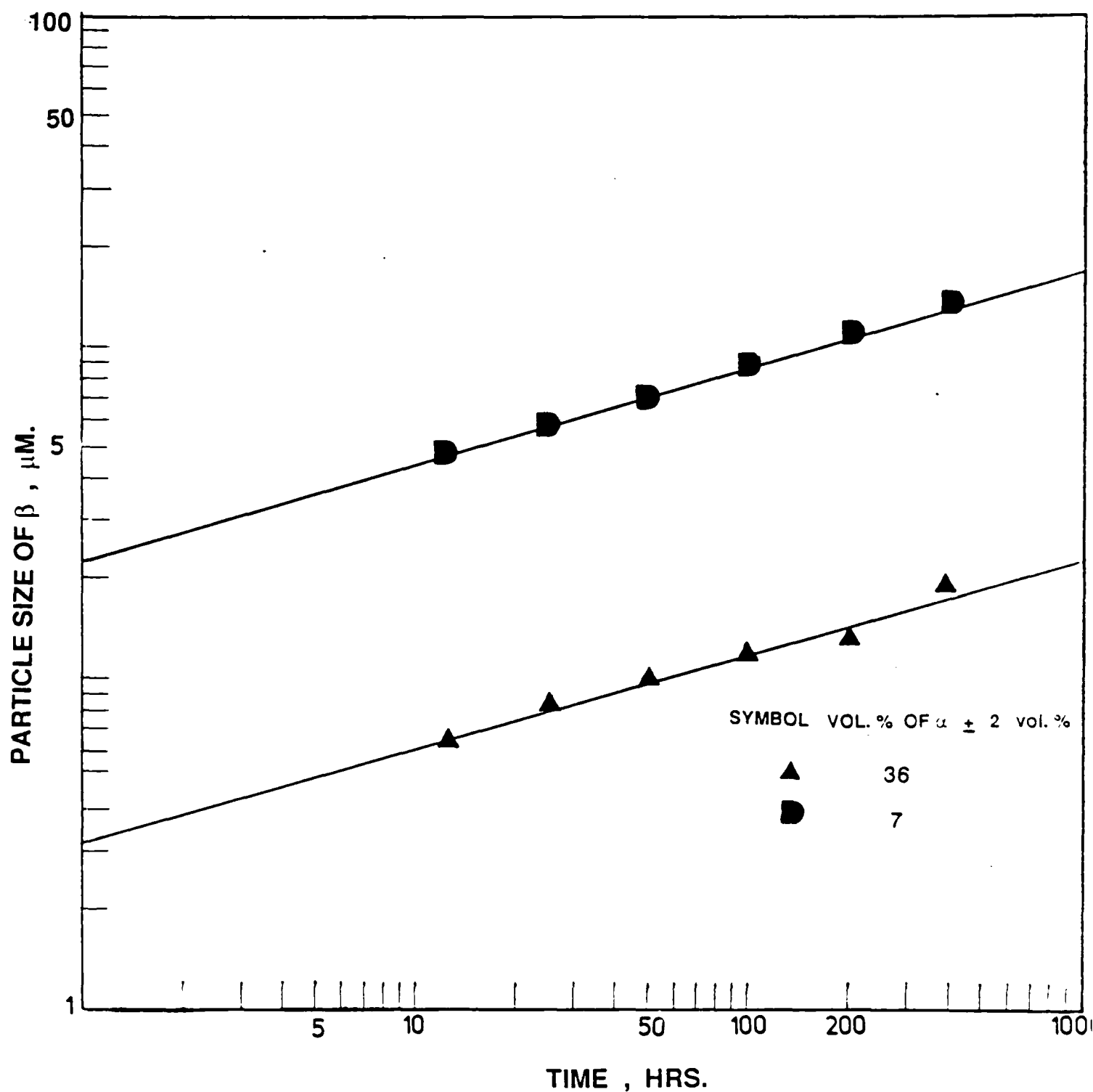


Fig. 35: PARTICLE SIZES OF β - PHASE IN α - β Ti - V ALLOYS. THE ALLOYS WERE ANNEALED FOR VARIOUS TIMES AT 1073 K , W. Q.

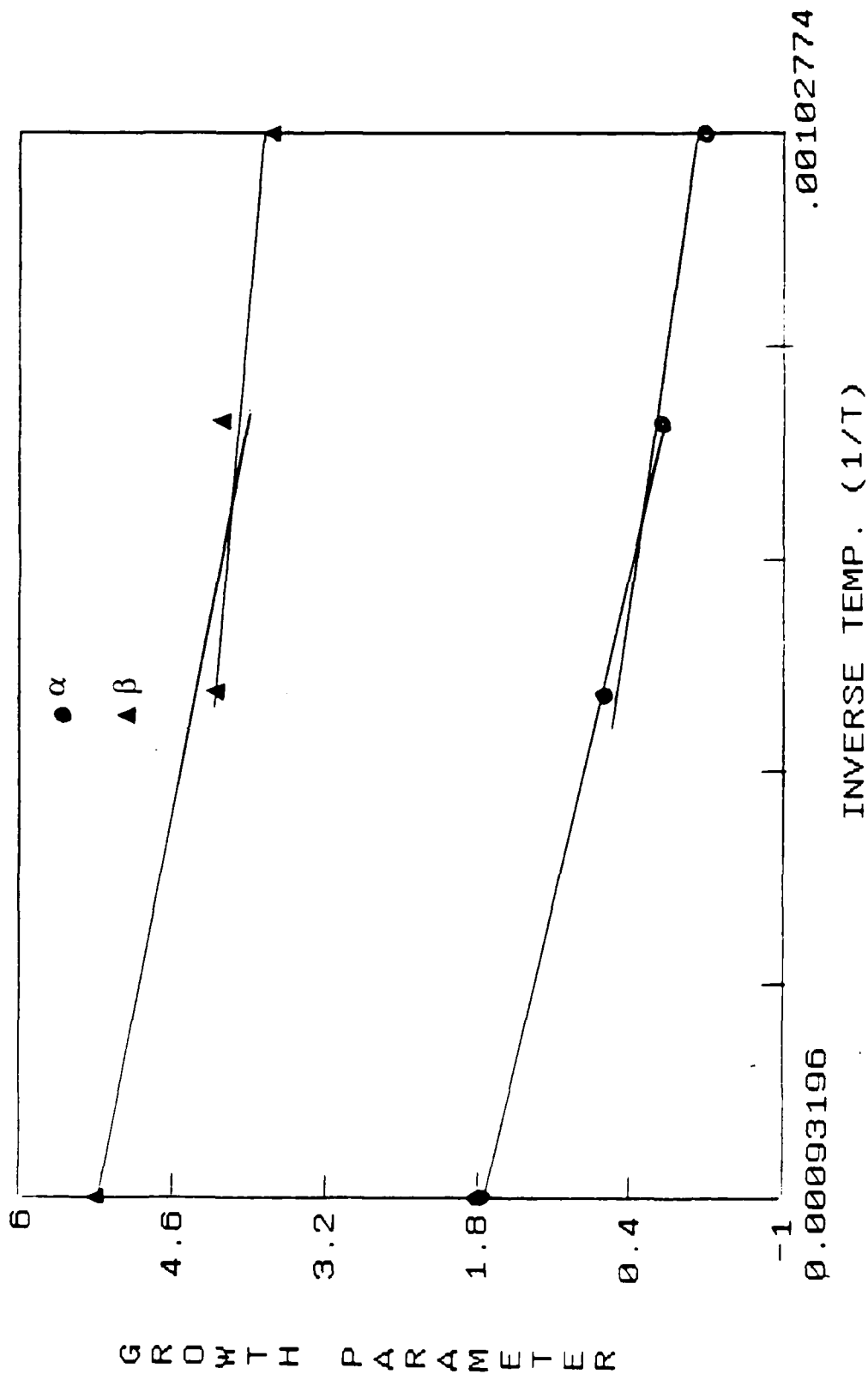


Fig. 36: DEPENDENCY OF GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON TEMPERATURE FOR A 24 % α - 76 % β Ti-Mn ALLOY.

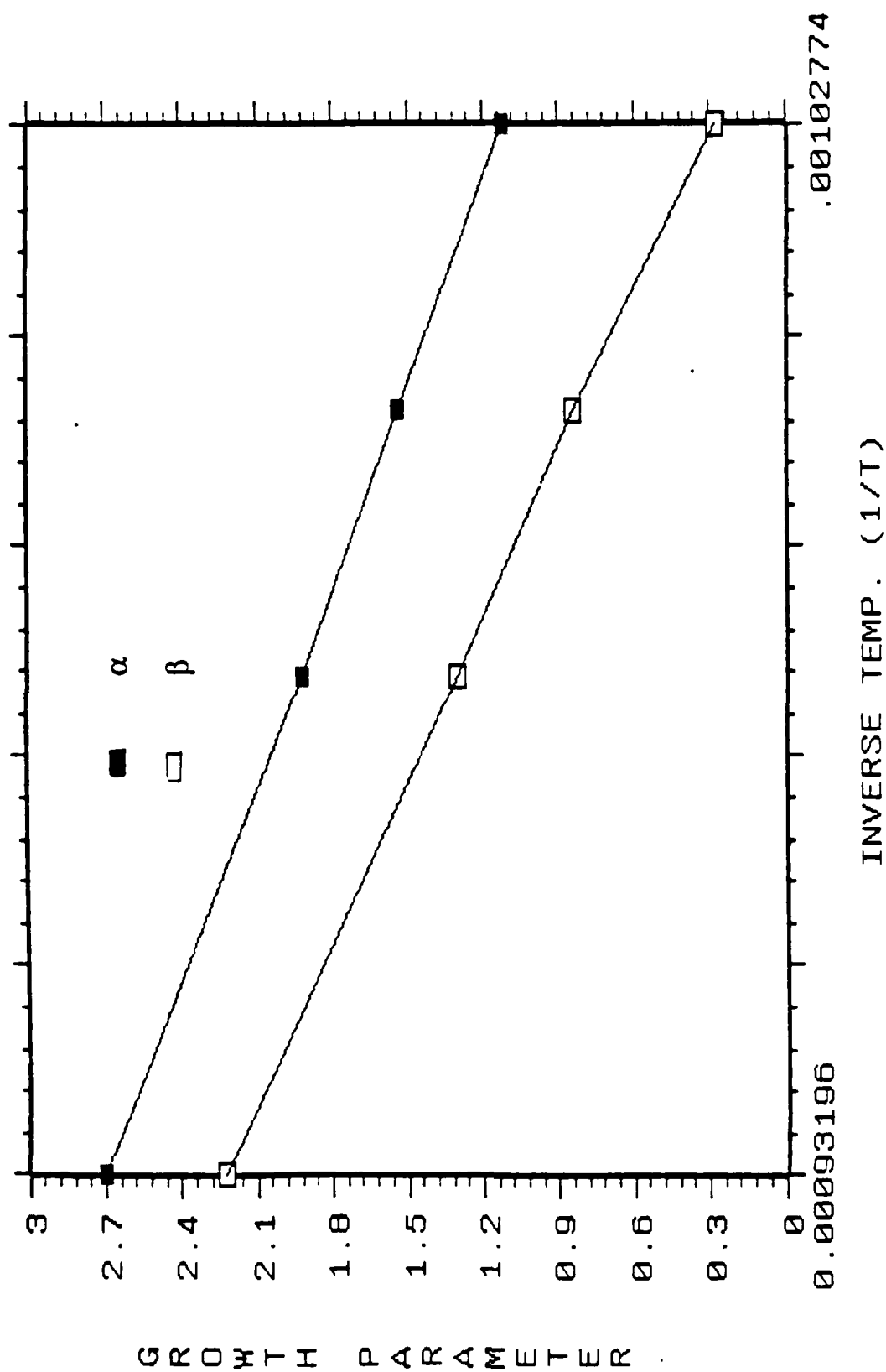


Fig. 37: DEPENDENCY OF GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON TEMPERATURE FOR A 54 % α - 46 % β Ti-Mn ALLOY.

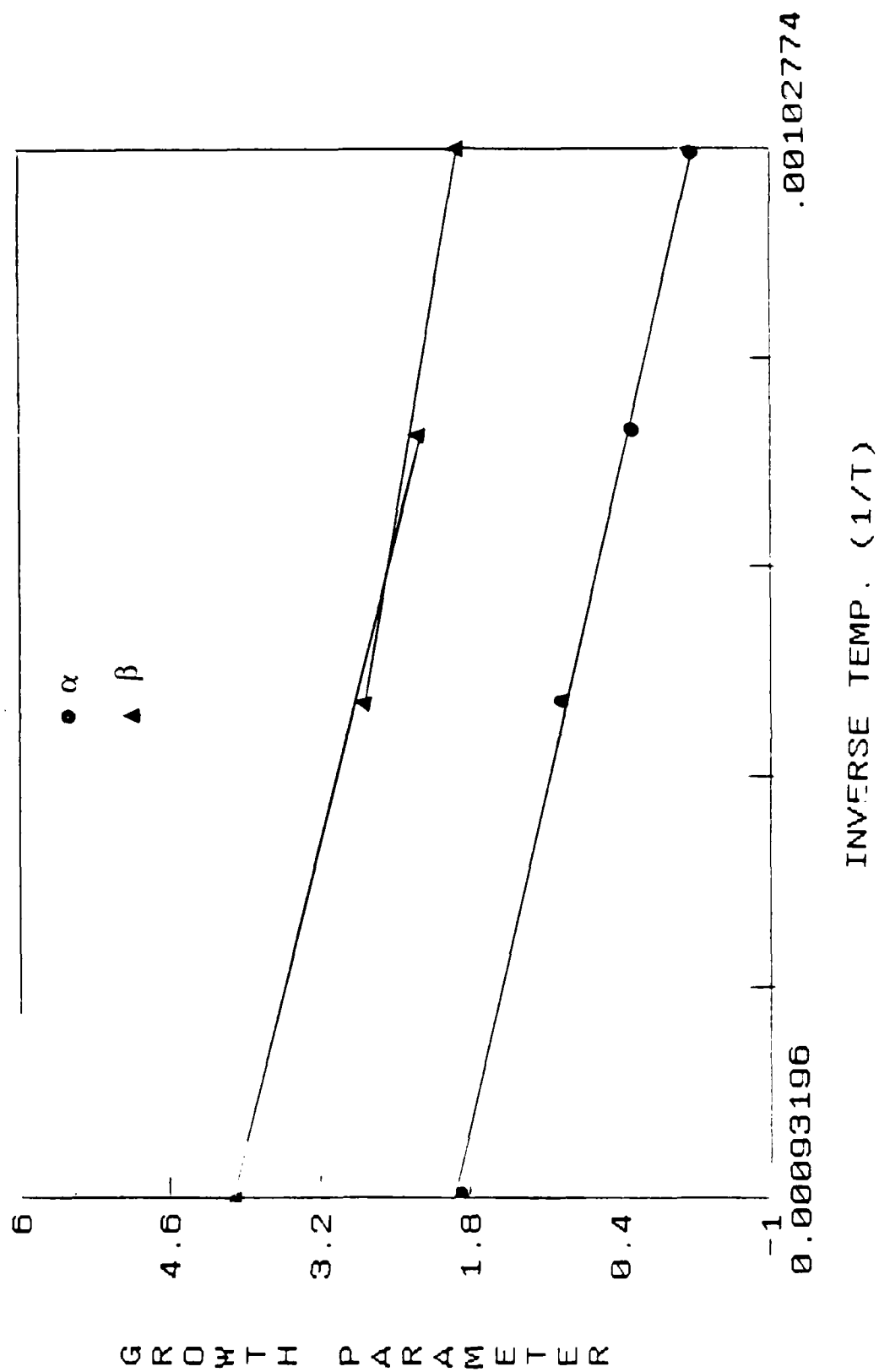


Fig. 38 : DEPENDENCY OF GROWTH PARAMETERS $K_{\alpha}^{\alpha-\beta}$ AND $K_{\beta}^{\alpha-\beta}$ ON TEMPERATURE FOR A 36 % α - 64 % β Ti-V ALLOY.

B. HIGH TEMPERATURE DEFORMATION STUDIES

(a). EXPERIMENTAL

Six Ti-Mn alloys were selected for the high temperature deformation study. The intended and actual compositions of the six alloys are listed in Table 1 of the previous section on Particle Coarsening Studies.

Heat Treatment and Microstructure

Specimens were cut from the as-received bars into sections about 3.5 inch (8.89 cm) in length. These were vacuum annealed in the alpha+beta field at 973 K and at 923 K for 200 hours followed by water quenching. This resulted in an equiaxed microstructure. Typical microstructures were presented in the previous section dealing with Particle Coarsening Studies. Vacuum was better than 10 millitorr during the annealing process. Upon completion of heat treatment helium gas was let into the chamber to produce a slightly positive pressure before being opened to air. The heat treatments were carried out so as to completely recrystallize the alloys, and to obtain equiaxed microstructures.

Specimen Preparation

The annealed bars were machined with braced carbide tools with a semi-cylindrical tip of 3/16 inch radius. Special grips for specimens complying with ASTM standards were prepared. Grips were fabricated using a high temperature material, 304H, which has a .2% YS about 8 times higher than the Ti-Mn alloy.

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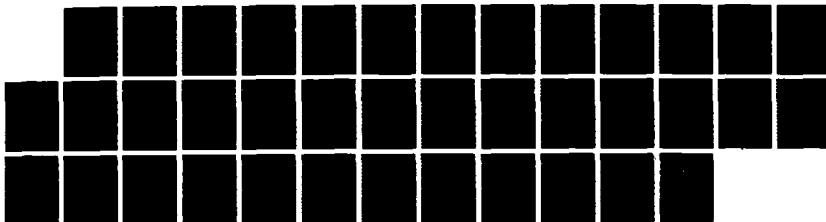
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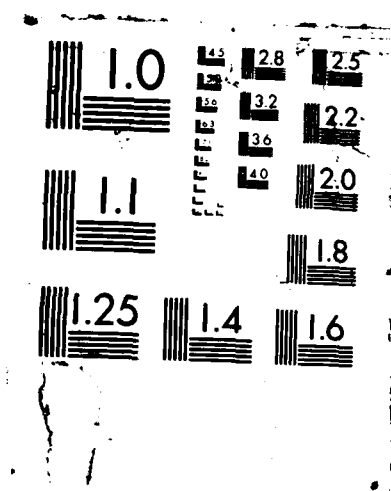
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High Temperature Tensile Test Apparatus

Tensile tests were carried out in an Instron testing machine capable of various constant cross-head velocities between 5 cm/min and 0.005 cm/min. Semi-cylindrical resistance heating elements with a constant temperature zone approximately 2 inch long were used. The furnace was surrounded by a ceramic fiber insulation to improve the heating efficiency and ensure temperature stability. Tensile tests were conducted at two different temperatures, 973 K and 923 K, in a vacuum chamber equipped with a set of mechanical and turbo pumps. Vacuum was better than 10 millitor. The temperature of the specimen was controlled by a feed back temperature controller. All the Ti-Mn alloys were tested at four different engineering strain rates, $1.1 \times 10^{-4}/\text{Sec}$, $2.6 \times 10^{-4}/\text{Sec}$, $2.6 \times 10^{-3}/\text{Sec}$, $2.6 \times 10^{-2}/\text{Sec}$. Each test was repeated under identical conditions to check for reproducibility. The variation was found to be in the range of + 5 percent. A schematic of the vacuum chamber used for the high temperature tensile test is shown in fig.1.

Computation of Experimental Data

The load-elongation plots obtained from the Instron machine was digitised on a Hicomscan, Hitachi digitiser. A computer program was developed for converting the load elongation data into true stress-true strain data. The program also calculated the offset flow stresses and the strain rate sensitivity. The alpha and beta volume fractions were determined by the linear intercept method (1) according to the procedure described in the previous section dealing with Particle Coarsening Studies.

(b). RESULTS AND DISCUSSION

As indicated in Tables 1 and 2, the beta volume percent increases as the Manganese content increases. Alpha phase is the matrix phase in alloy 2 and beta is the matrix in alloy 5. In the case of alloys 3 and 4 none of the two phases can be regarded as completely continuous.

Stress-Strain behavior in alloys 1 and 2

Alpha phase has a hcp structure with c/a ratio about 2% less than the ideal value of 1.633 (2). the operative slip systems in alpha Titanium alloys are $(1\bar{1}00)$ $[11\bar{2}0]$ (3, 4), (0001) $[11\bar{2}0]$ (5) and $(10\bar{1}1)$ $[11\bar{2}0]$ (6). In addition to the "grown-in" dislocation sources, there are some dislocation generators coming into play because of the interaction between the dislocations of different slip systems, i.e. multiple slip. Dislocations are generated in an accelerating manner with increasing strain which results in the strain hardening behavior of alloy 1. The extent of work hardening at elevated temperatures is not so evident as that of the alloys tested at room temperature (7). Longo (8) explained the temperature dependence of the work hardening in terms of a thermal softening superimposed on an athermal hardening process. The effect of the strain rate would vary the time available for the action of the recovery mechanism. This theory is substantiated from our experimental data given in fig. 2 (973 K) and fig. 14 (923 K) where the strain hardening is more evident at the higher strain rates.

The offset flow stress versus beta volume percent, figs. 8 to 12 (973 K) and figs. 20 to 24 (923 K) indicate that the offset

flow stress increases from alloy 1 to alloy 2 at all the strain rates for both 973 K and 923 K. This is as one would expect since alloy 2 contains the hard beta phase whereas alloy 1 is nearly 100 % alpha. In addition, the grain size of alpha decreases with the addition of beta and this also contributes to the higher flow stress of alloy 2.

Interface Sliding in Alpha-Beta Alloys

It has been found that interface sliding in alpha-beta titanium alloys may be an important model of deformation even at room temperature (9,10). Ankem and Margolin studied the deformation of alpha-beta titanium alloy specimens with aluminum grid lines on polished surface and found that the plastic straining of the alpha grains in the vicinity of the sliding is more evident than strain in the beta grains. They (9) also pointed out that an alpha-beta interface need not be a flat plane for sliding to take place. In alloy 3 and alloy 4, both at 973 K and 923 K, the alpha-beta interfaces per unit volume are much higher than that in alloys 1, 2, 5 and 6. Therefore, there is a greater chance for interface sliding to take place in alloys 3 and 4. Between alloys 3 and 4 the particles of alloy 4 are more equiaxed which is ideal for interface sliding. Also, alloy 4 has lower flow stresses at 923 K than at 973 K, compare fig. 5 with fig. 17 for a strain rate of 1.1×10^{-4} /Sec. This can be explained by the interface sliding mechanism. Optical microscopic observations indicated that the microstructure is finer at 923 K than it is at 973 K. Hence this enhances the interface sliding

activity. Since interface sliding was observed at room temperature, it is reasonable to assume that interface sliding will be more pronounced at elevated temperatures because thermally induced vibrations of the crystal atoms accelerate the diffusional process (11,12).

Ashby and Verrall (12) suggested a threshold stress which controls sliding behavior of materials at high temperatures. At low strain rates, there is sufficient time for diffusion to occur therefore, less external stress is required to initiate sliding. Higher stress must be applied before sliding comes into play at high strain rates. Our experimental data shown in fig.5 (973 K) and fig 17. (923 K) substantiates this theory. Also, the offset flow stress versus beta volume percent plots, figs.8 to 12 (973 K) and figs.20 to 24 (923 K) show a drop at alloy 4 at the lower strain rates. Moreover, as mentioned above, the offset flow stresses at 973 K are higher than that at 923 K for alloy 4 at lower strain rates. This strongly suggests that interface sliding must be the dominating mechanism in this alloy. Experiments are underway to check if indeed interface sliding occurs in this alloy at these temperatures and strain rates.

Sharp Flow Stress Drops in Near Beta Alloys

Beta-phase (bcc) is the dominant phase in alloy 5 and alloy 6 both at 973 K and 923 K. This implies that beta phase is responsible for the unusual behavior for the abrupt flow stress drops at small strains and the flow softening behavior. Flow softening has been observed in Ti-C (13), Zr-Nb (14) and Zr-Mo (15) systems. Jonas, Heritier and Luton (14) have shown that for

Zr-Nb (15% Nb) the flow softening is due to the declustering phenomenon. Paton and Hamilton (16) however attribute the flow softening and sharp drop in flow stress to the formation of subgrains due to the occurrence of dynamic recovery. They have studied the post deformation microstructure, both by optical and transmission electron microscopy, and have shown the existence of subgrains in the microstructure. Optical micrographs of deformed specimens does tend to indicate the existence of subgrains. However, transmission electron microscopy will have to be used to confirm that dynamic recovery is the operating mechanism in the case of near beta alloys. Flow stress of alloy 6 at both 973 K and 923 K, fig. 7 (973 K) and fig. 19 (923 K), is higher than that for alloy 5, fig. 6 (973 K) and fig. 18 (923 K). This is due to the higher beta content of alloy 6. The offset flow stress versus beta volume percent also indicate similar trends for the same reasons. Experiments are underway to determine the mechanism of flow softening in alloy 6.

Comparison of Calculated and Experimental Data

Theoretical calculations were made to predict the stress-strain behavior of alpha-beta alloys based on the iso-strain model for the four different strain rates, not shown here. All the calculated flow stresses of alloy 4 were consistently higher than the experimental stresses. This can be interpreted as being due to a negative contribution by the interaction term (17) due to significant amount of interface sliding.

The calculated stress-strain curves were sometimes lower and sometimes higher than the experimental data suggesting that the assumption of constant strain is not valid and interactions are

expected as suggested by Ankem and Margolin (17).

Strain Rate Sensitivity Determination

Strain rate sensitivity can be determined by various methods as outlined by Hart (18), Backofen, Turner and Avery (19), Hedworth and Stowell (20), and by Gibbs (21). However, the method adopted for determining strain rate sensitivity (m) in our case was similar to the one formulated by Hedworth and Stowell (20). The difference between their (20) method and ours is that they (20) determined the strain rate sensitivity by a velocity change experiment on a single specimen whereas in our case separate specimens were used for each data point.

Flow stress for a given total plastic strain, 0.001, was determined for each alloy for four different strain rates for both the test temperatures i.e. 973 K and 923 K. The slope of the plot of $\log$ (flow stress) versus $\log$ (strain rate) was the strain rate sensitivity, m . The correlation co-efficient obtained (> 0.97) indicate the validity of this method of determining strain rate sensitivity.

The strain rate sensitivity versus beta volume percent plots for both 973 K and 923 K, fig. 13 (973 K) and fig. 25 (923 K) indicates similar trends. The strain rate sensitivity shows a drop at about 20 volume percent beta and a peak at around 50 volume percent beta at 973 K as well as at 923 K. Moreover, the strain rate sensitivity at about 50 volume percent beta (alloy 4) at 923 K is higher than at 973 K.

(c). SUMMARY

The flow stress behavior of the Ti-Mn alloys investigated

indicated similar trends at both the test temperatures i.e. 973 K and 923 K. This implies that the operating mechanisms may be similar. At Manganese contents greater than 3 wt. % (alloys 3 to 6) the Ti-Mn alloys exhibited strain softening and flow stress drops at both test temperatures. The magnitude of the flow stress drop increased with increasing strain rate for a given beta volume percent. Also, the flow stress drop increased with an increase in the beta volume percent.

The fact that the offset flow stresses do not vary linearly with beta volume percent at the lower two strain rates indicates that a change in mechanism occurs with a change in beta volume percent. Smaller flow stresses in alloys with nearly equal volume percents of the alpha and beta phases can be utilized by the Titanium industry with great advantages for fabrication of these alloys.

Further work will be conducted to check if indeed interface sliding occurs in the alpha-beta alloys at low strain rates. The Ti-Mn alloys will be tested at one more temperature, 1023 K, at the four strain rates. Similar series of tests will be conducted on Ti-V alloys. The outcome of these investigations should enable us in understanding how the various factors namely strain rate, temperature and the volume percent of the second phase effect the high temperature deformation behavior of two phase alloys. Theoretical modelling will also done by the FEM method to predict the deformation behavior of various two phase materials.

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TABLE 1. VOLUME PERCENTS OF ALPHA AND BETA PHASES. H.T. 973 K,
200 HRS. W.Q.

=====

ALLOY NO.	ALPHA VOLUME PERCENT	BETA VOLUME PERCENT
1	100	0
2	79	21
3	63	37
4	45	55
5	11	89
6	0	100

=====

TABLE 2. VOLUME PERCENTS OF ALPHA AND BETA PHASES. H.T. 923 K,
200 HRS. W.Q.

=====

ALLOY NO.	ALPHA VOLUME PERCENT	BETA VOLUME PERCENT
1	100	0
2	95.6	6.4
3	68.4	31.6
4	56.7	43.3
5	24.4	75.6
6	12.1	87.9

VACUUM SYSTEM

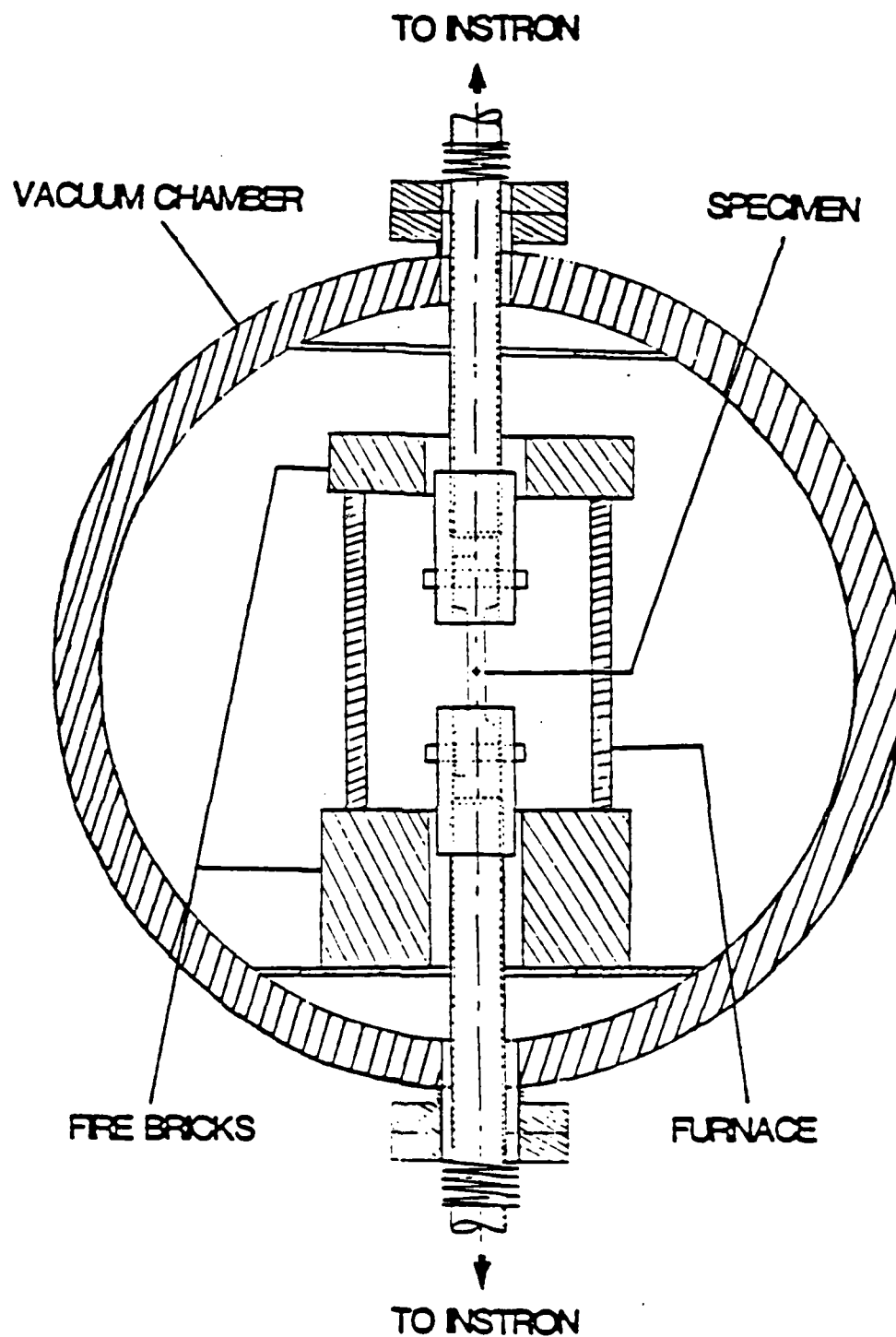


FIG 1: A SCHEMATIC OF THE HIGH TEMPERATURE VACUUM CHAMBER USED FOR THE TENSILE TEST.

FIG 2: TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 1 (100 % α , Ti-Mn ALLOY)
TESTED IN VACUUM AT 973 $^{\circ}$ K AT DIFFERENT STRAIN RATES.

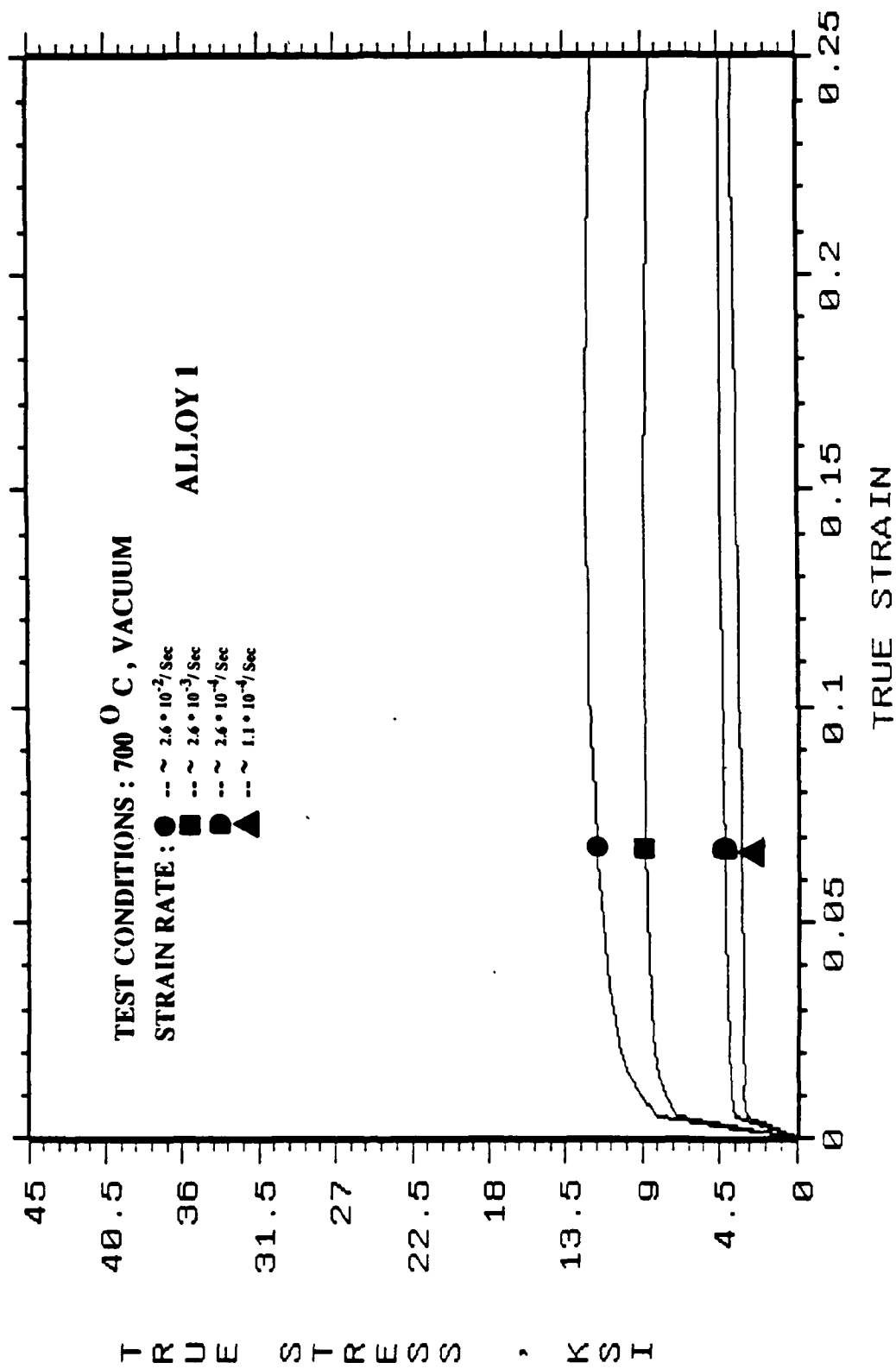


FIG 3: TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 2 (79 % α - 21 % β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 973 $^{\circ}$ K AT DIFFERENT STRAIN RATES.

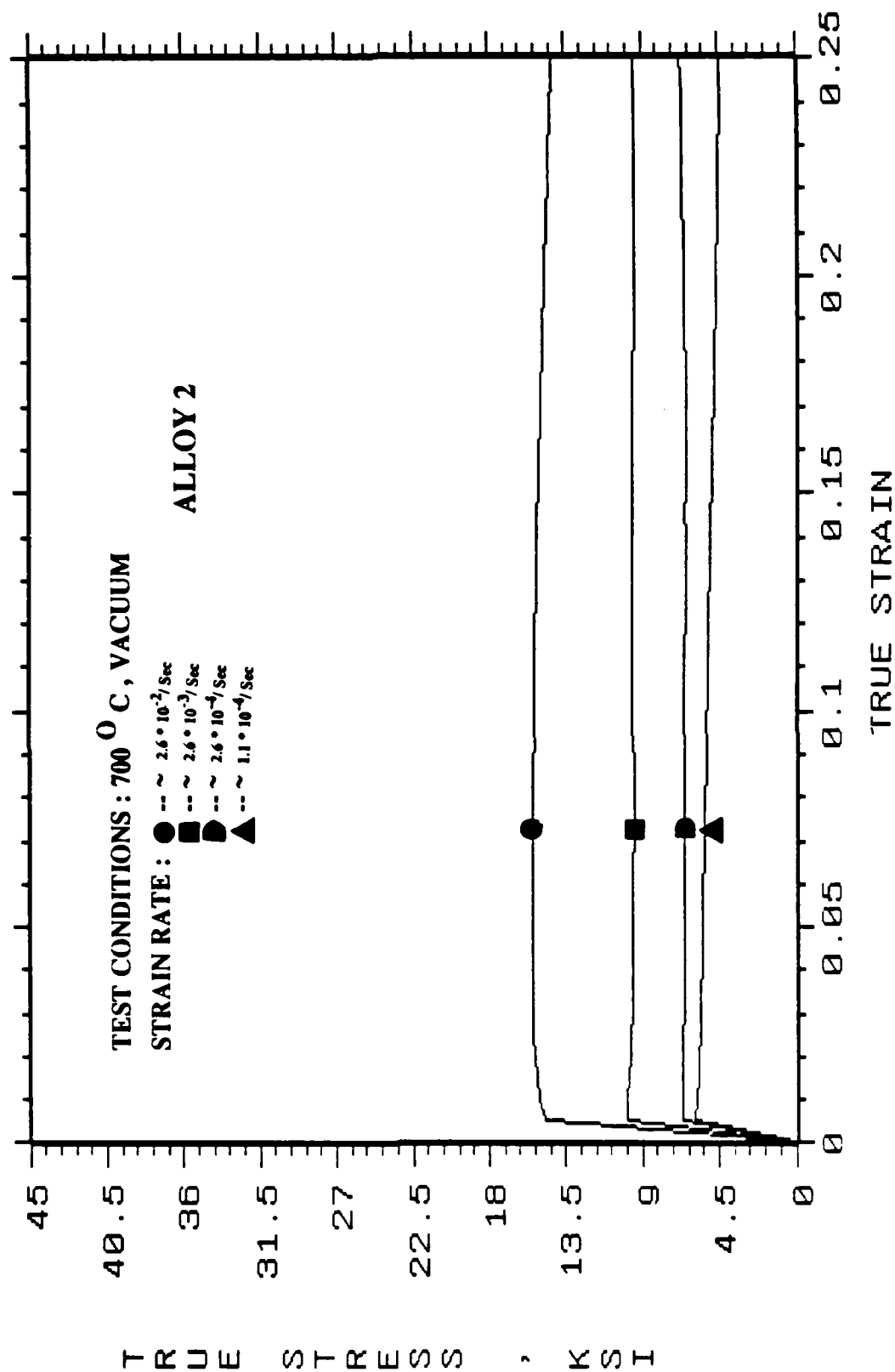


FIG 4: TRUE STRESS-TRUE STRAIN CURVES OF ALLOY 3 (63 % α - 37 % β , Ti - Mn ALLOY)
TESTED IN VACUUM AT 973°K AT DIFFERENT STRAIN RATES.

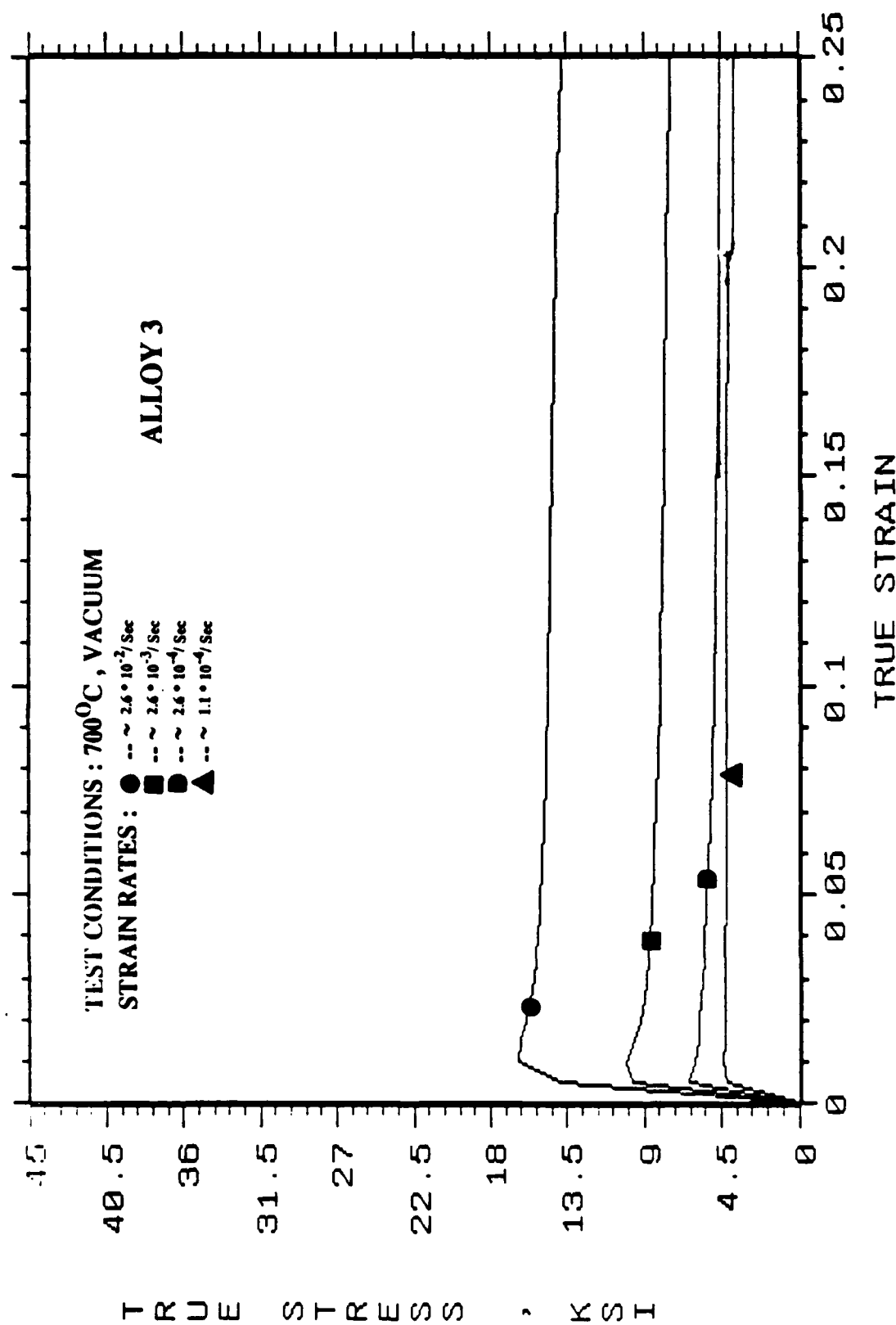


FIG 5: TRUE STRESS-TRUE STRAIN CURVES OF ALLOY 4 (45 % α -55 % β , Ti - Mn ALLOY)
TESTED IN VACUUM AT 973°K AT DIFFERENT STRAIN RATES.

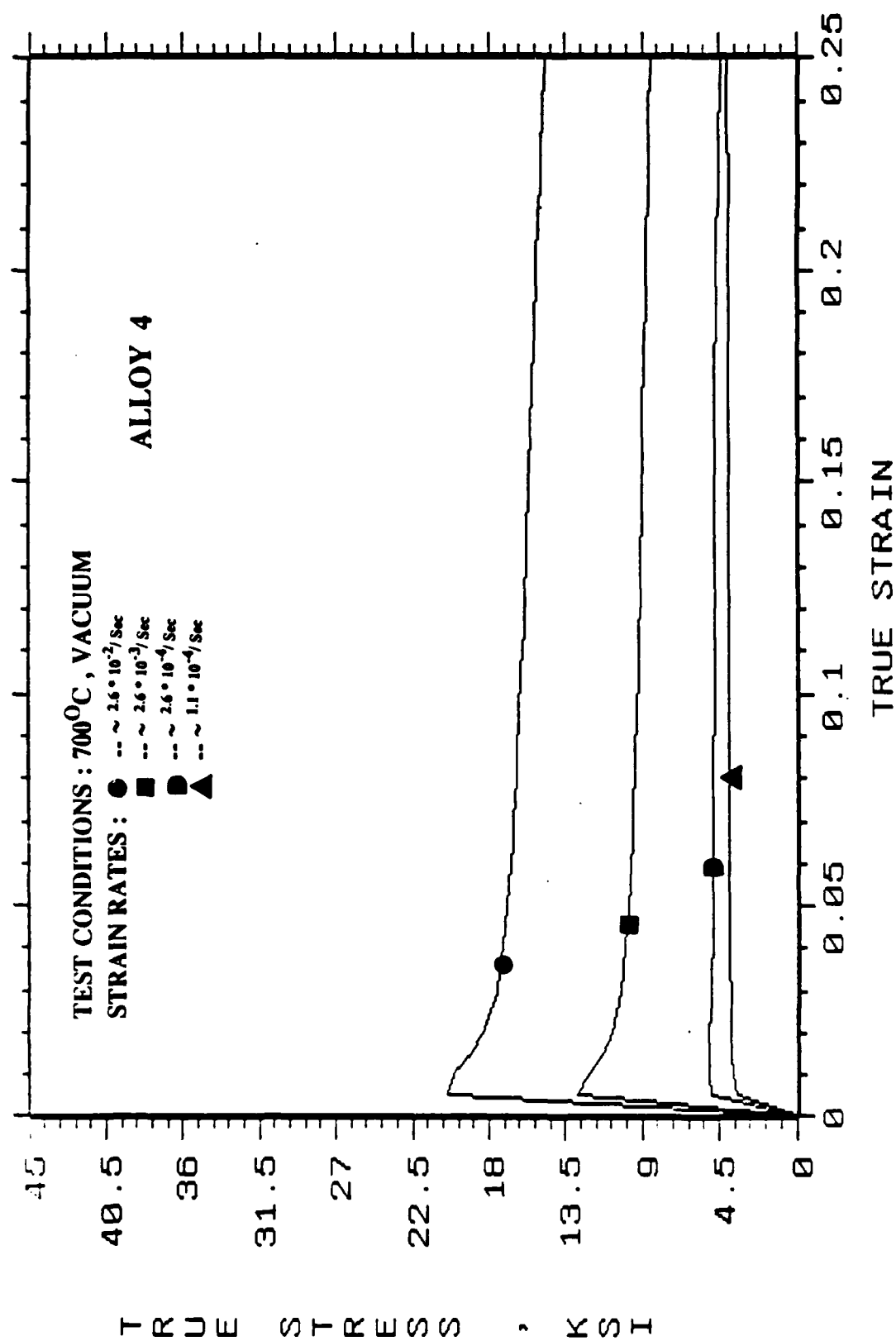


FIG 6 : TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 5 (10 % α - 90 % β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 973 °K AT DIFFERENT STRAIN RATES.

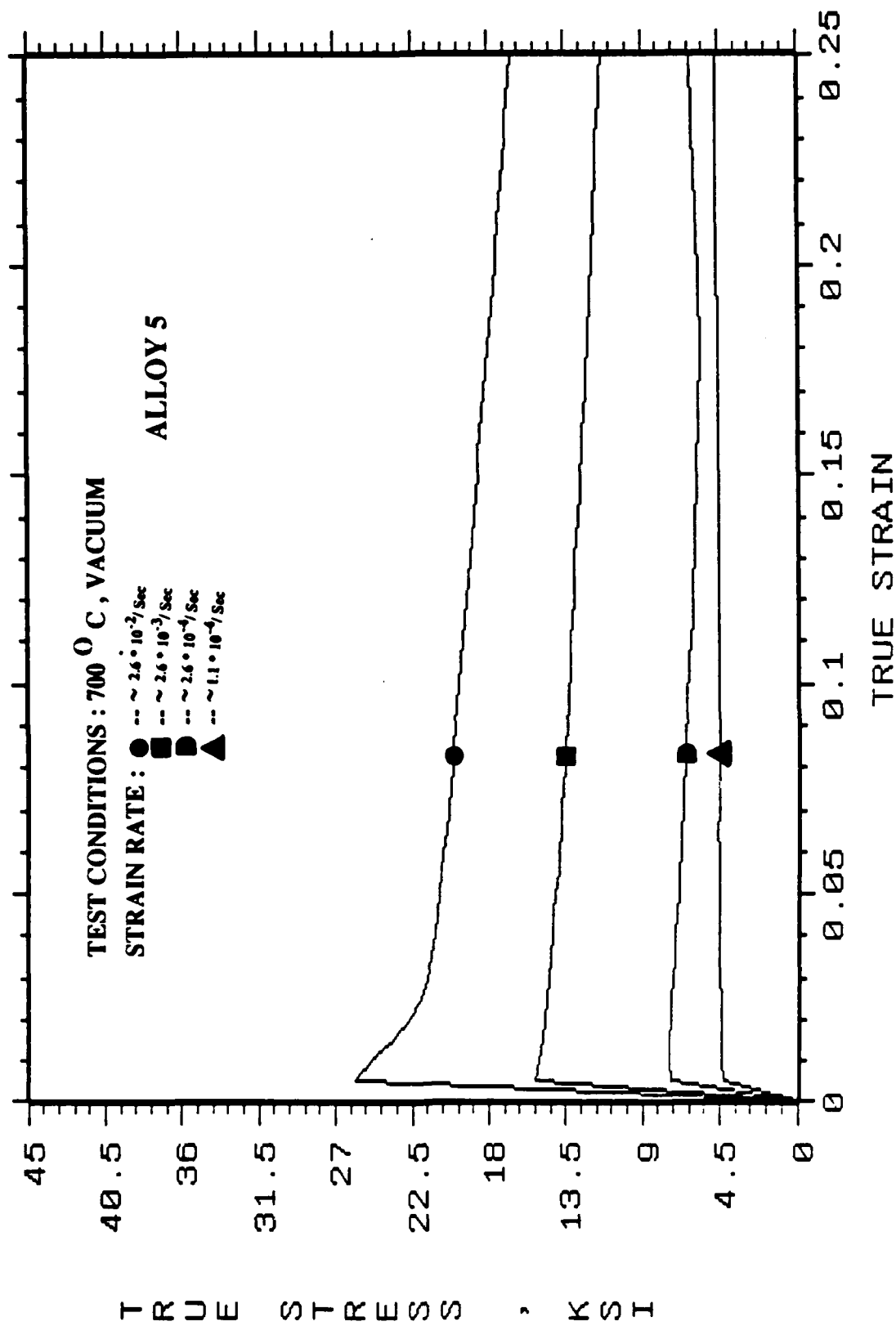


FIG 7: TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 6 (100% β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 973 °K AT DIFFERENT STRAIN RATES.

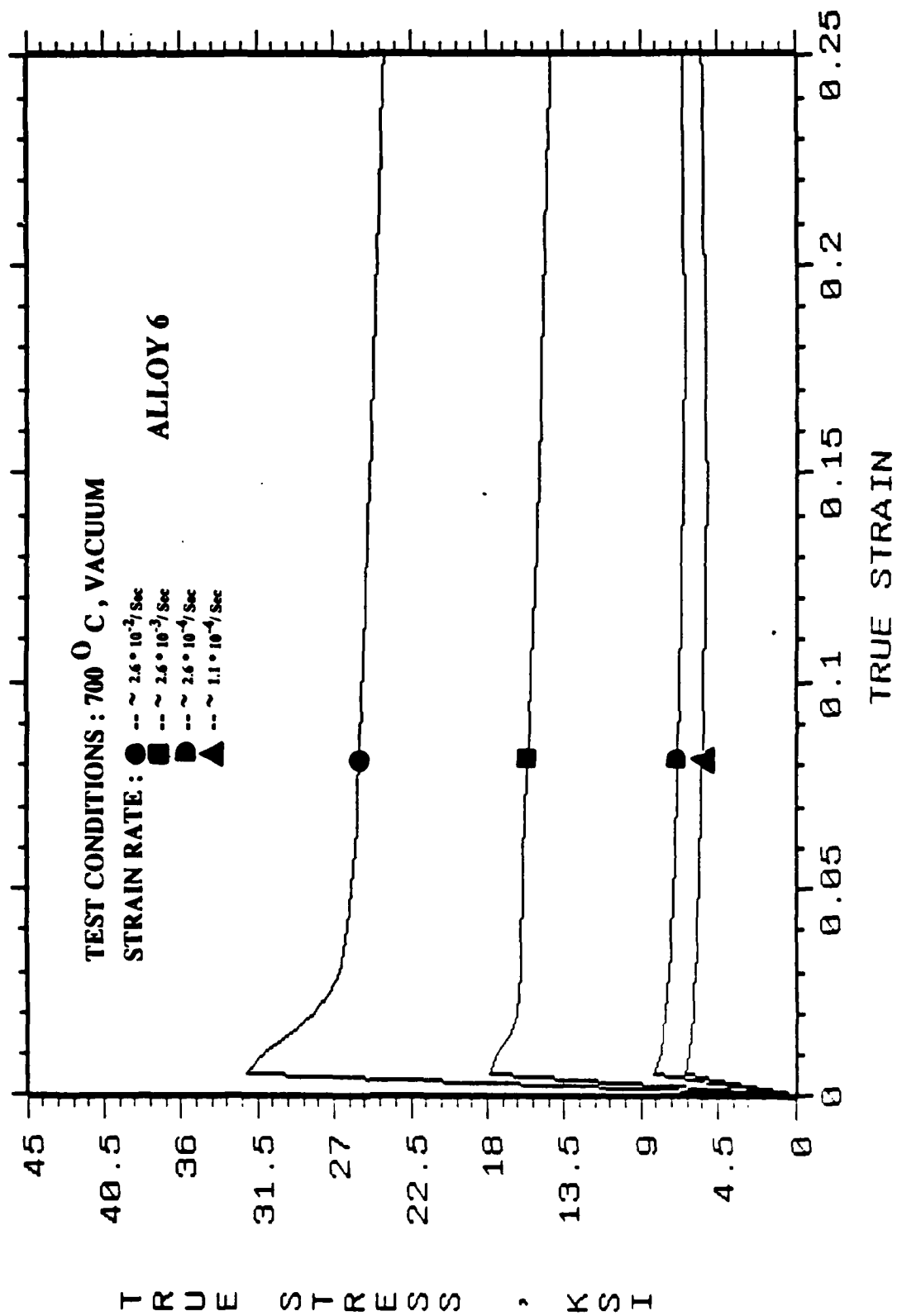


FIG 8 : FLOW STRESS V_s BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 0.2 % PLASTIC STRAIN

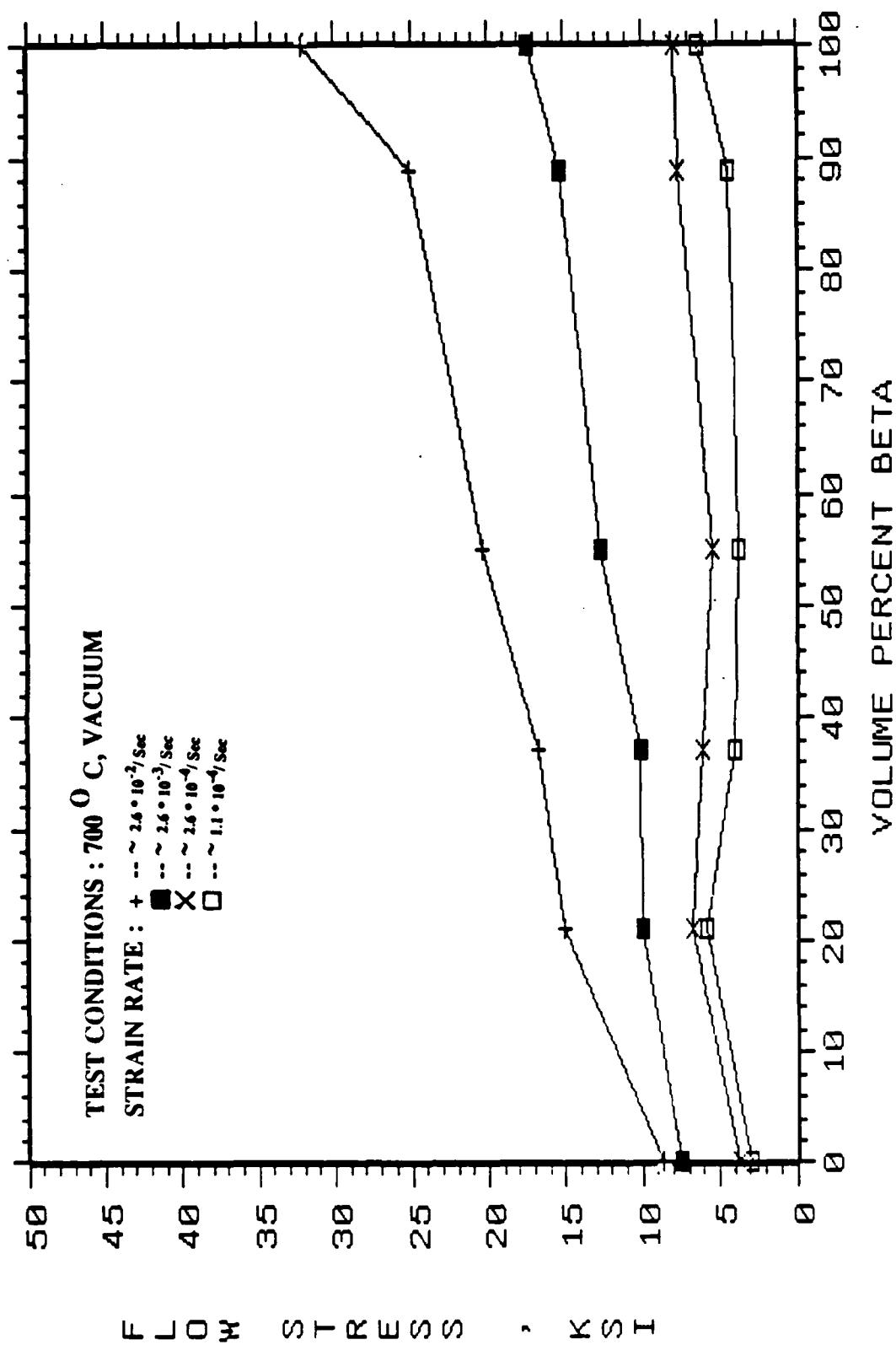


FIG 9: FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 0.5 % PLASTIC STRAIN

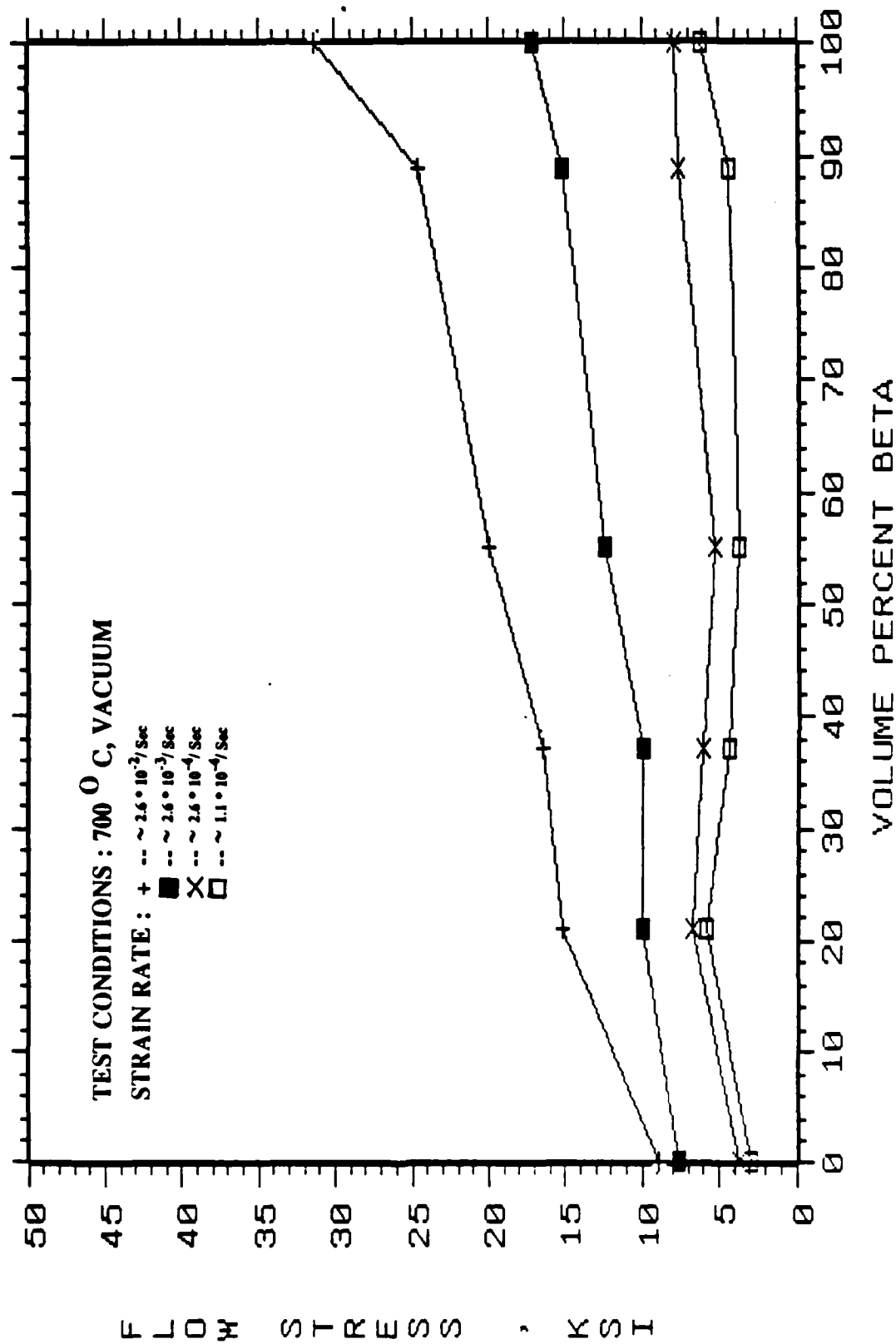


FIG 10 : FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 5 % PLASTIC STRAIN

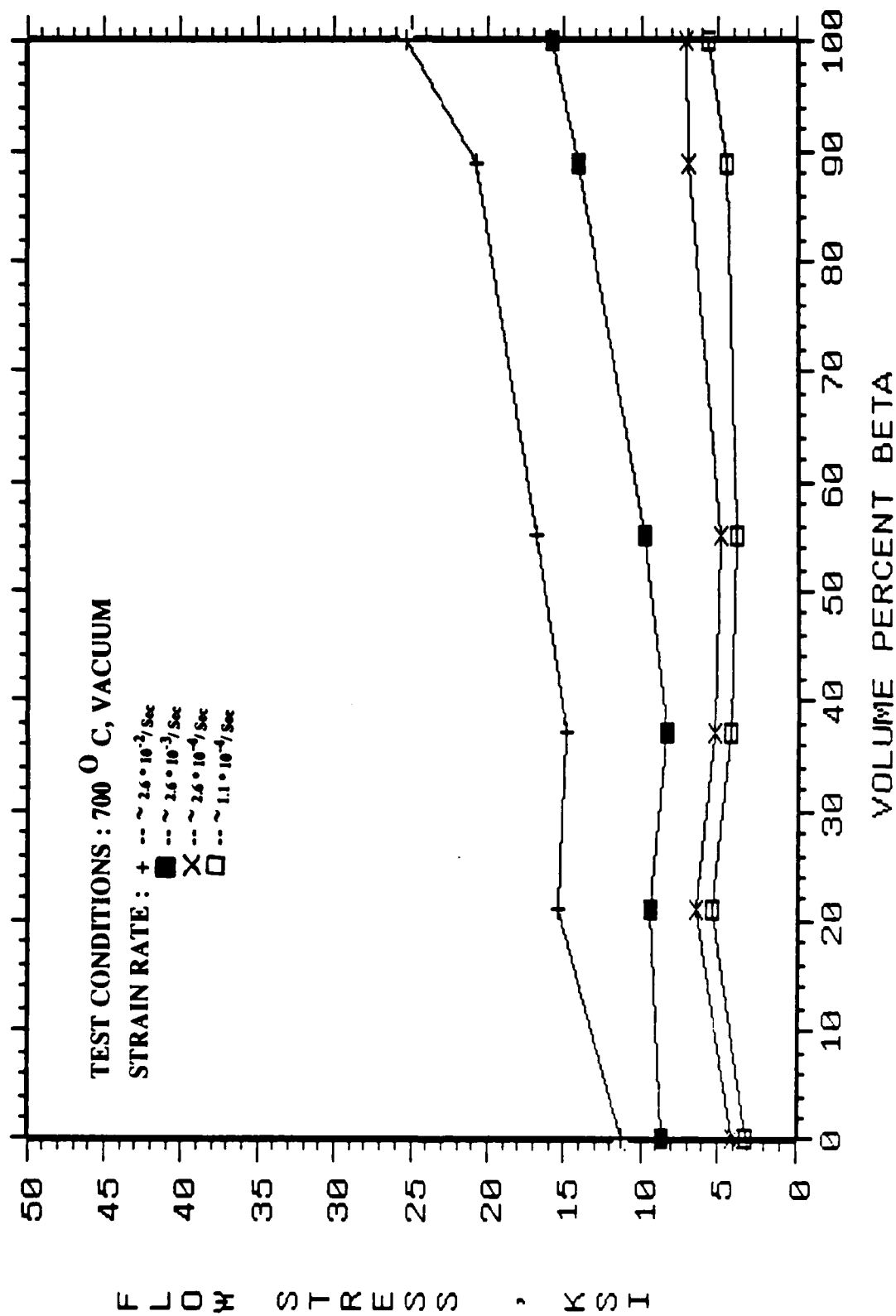


FIG 11 : FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
FLOW STRESS CORRESPONDS TO 10 % PLASTIC STRAIN

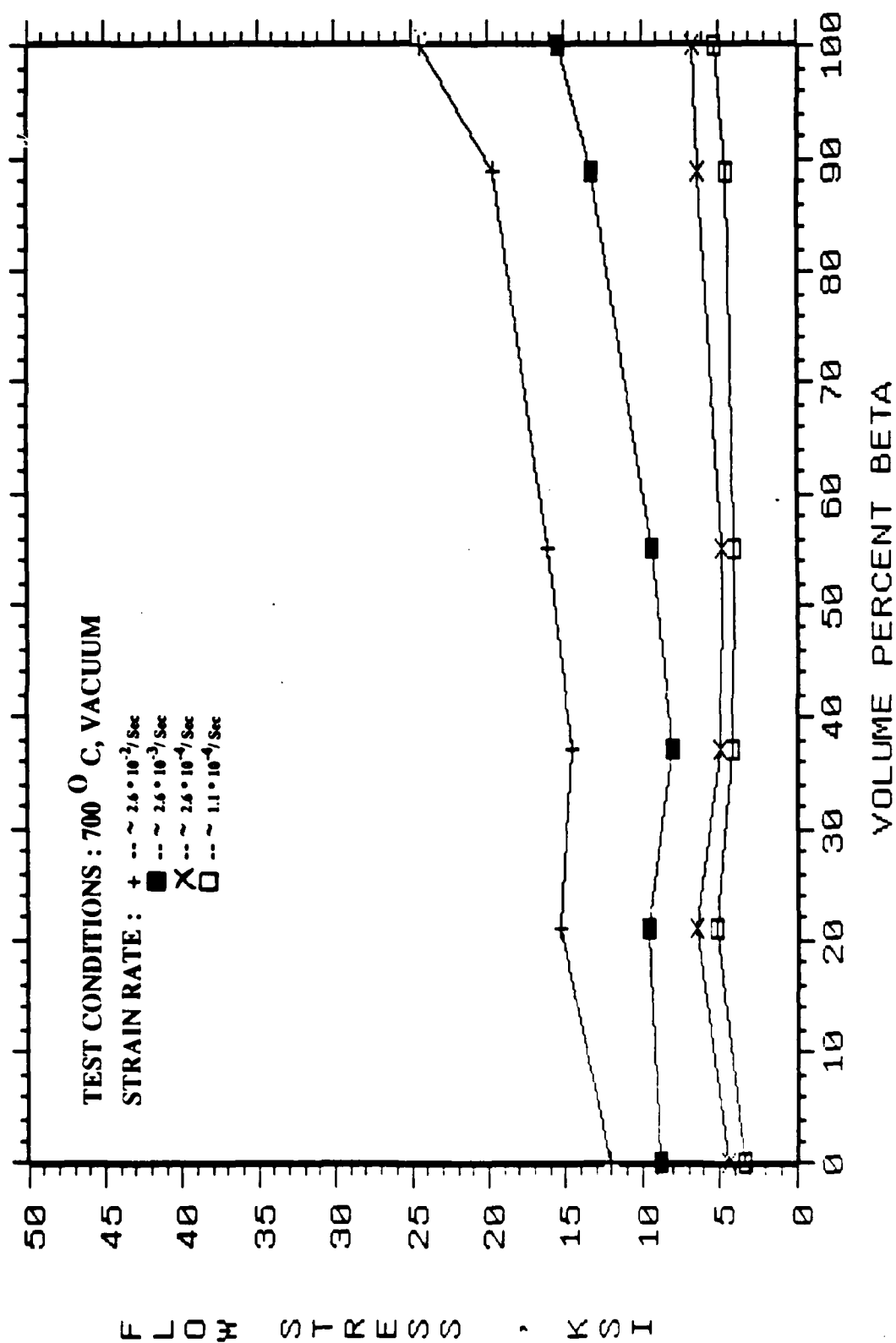


FIG 12: FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 20 % PLASTIC STRAIN

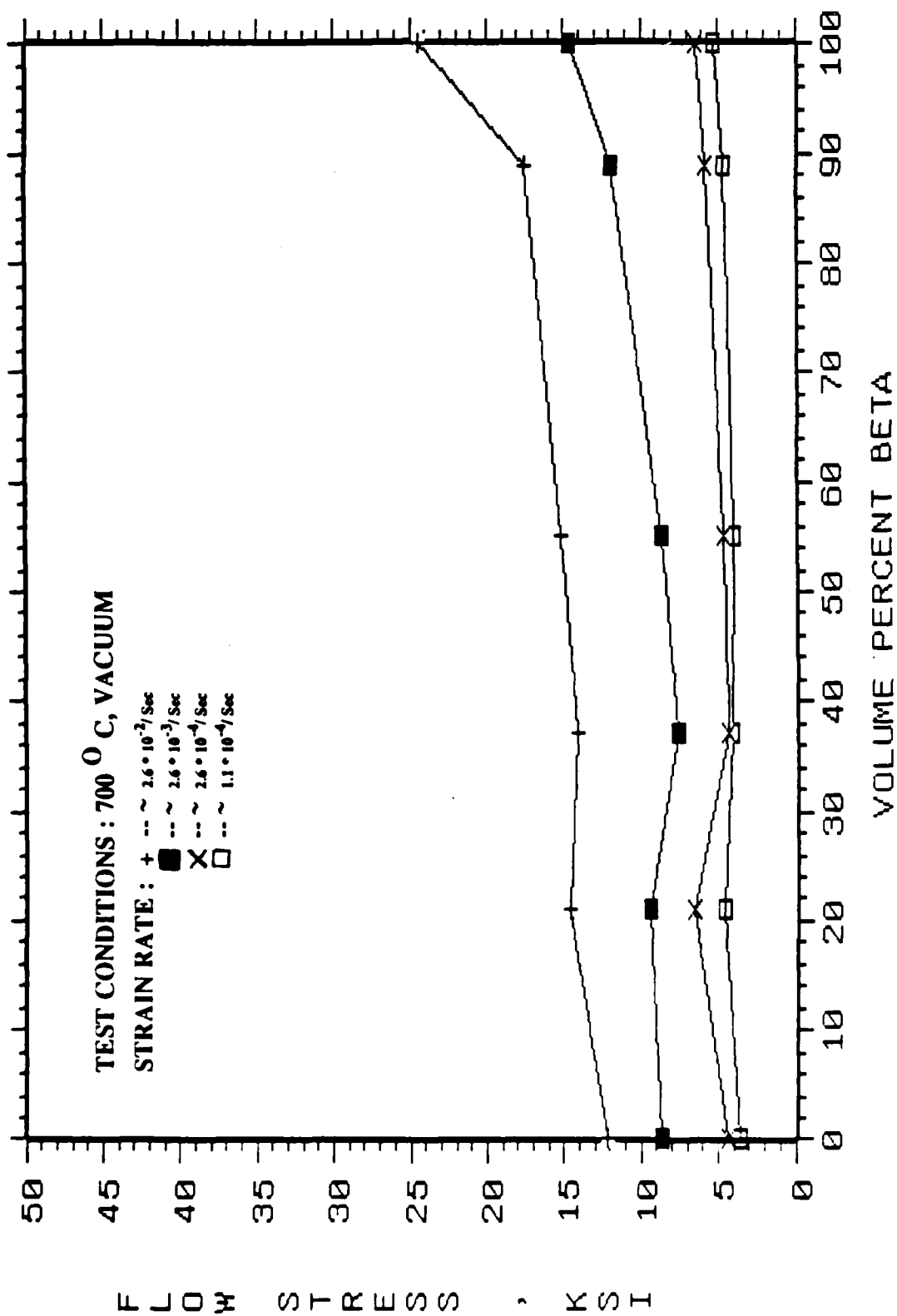


FIG 13: STRAIN RATE SENSITIVITY (m) Vs BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS DETERMINED FROM $\log \sigma$ Vs $\log \epsilon$ PLOT. THE SLOPE OF THE PLOT IS THE STRAIN RATE SENSITIVITY (m).

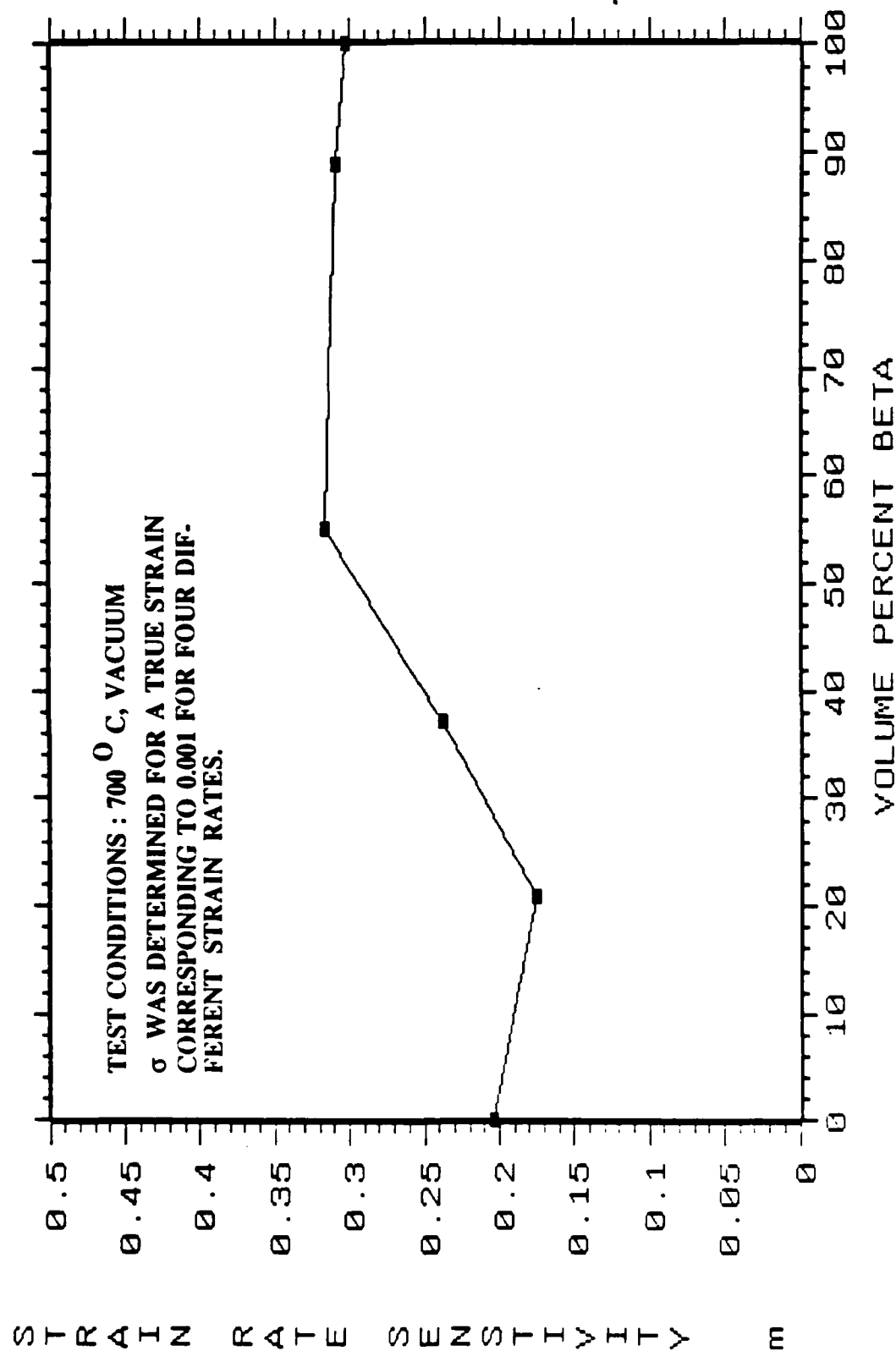


FIG 14: TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 1 (100 % α , Ti-Mn ALLOY) TESTED IN VACUUM AT 923 °C AT DIFFERENT STRAIN RATES.

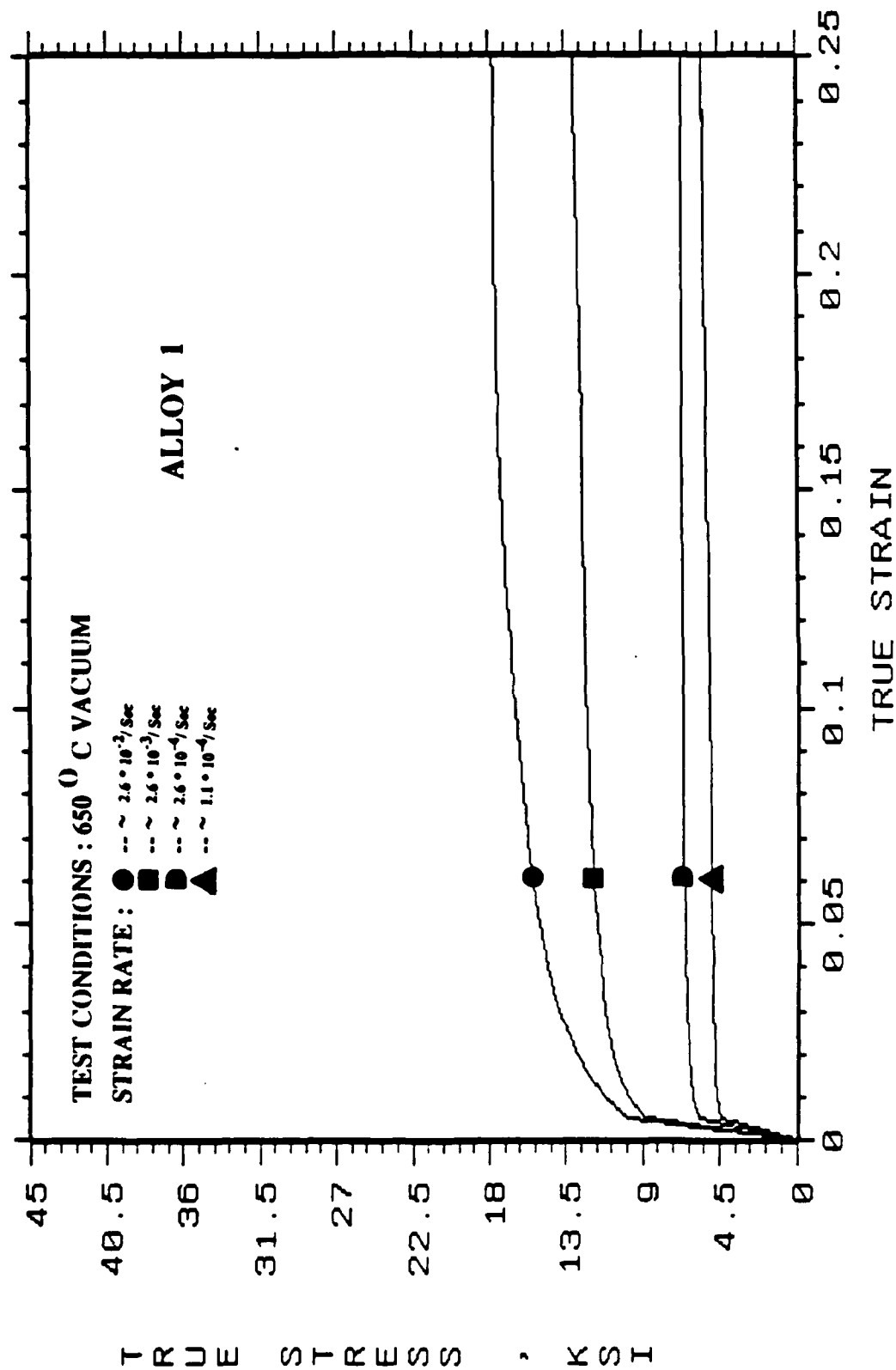


FIG 15: TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 2 (95.6 % α - 6.4 % β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 923 °K AT DIFFERENT STRAIN RATES.

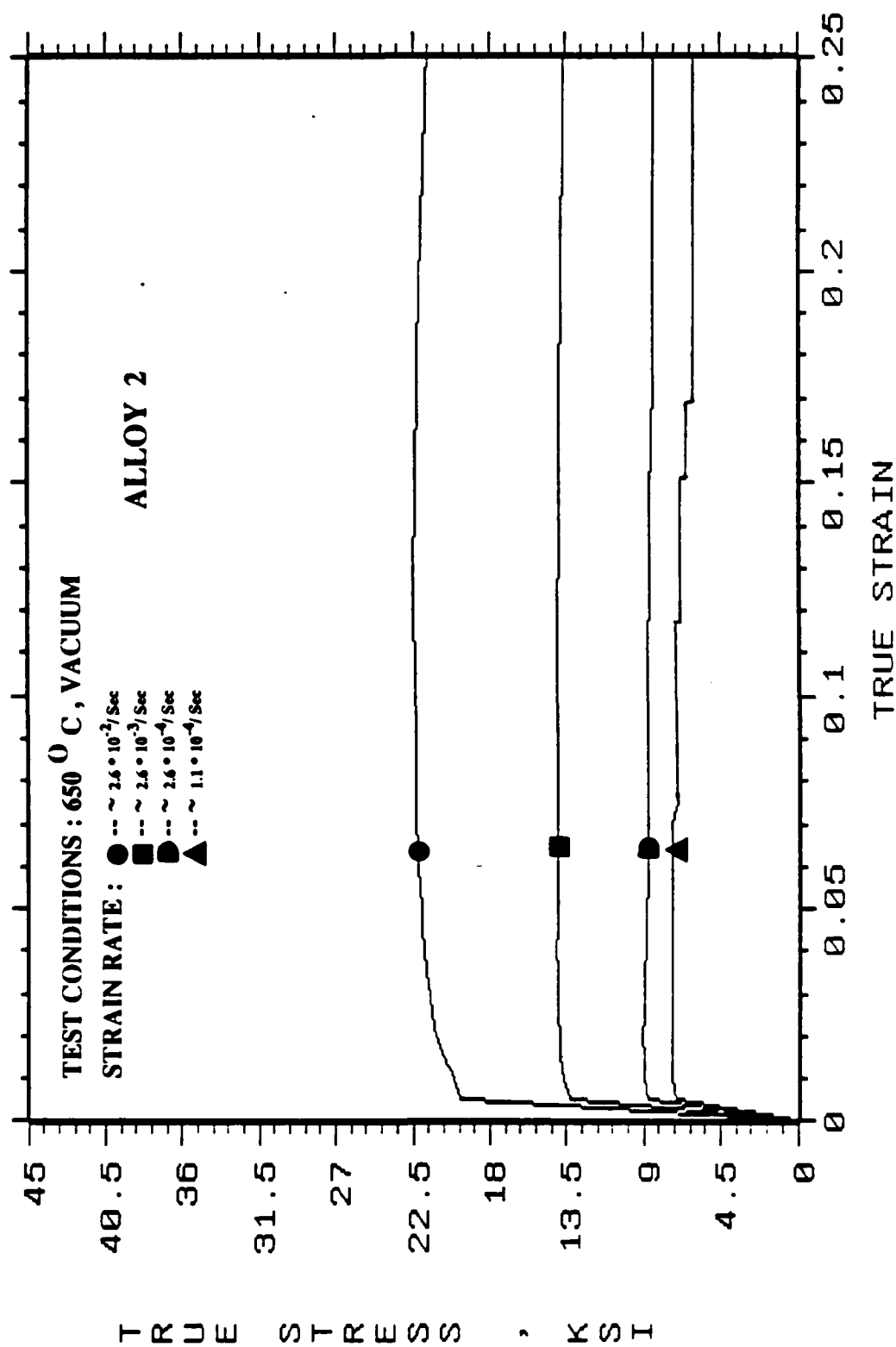


FIG 16: TRUE STRESS - TRUE STRAIN CURVES OF ALLOY3(68.4 % α - 21.6 % β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 923 $^{\circ}$ K AT DIFFERENT STRAIN RATES.

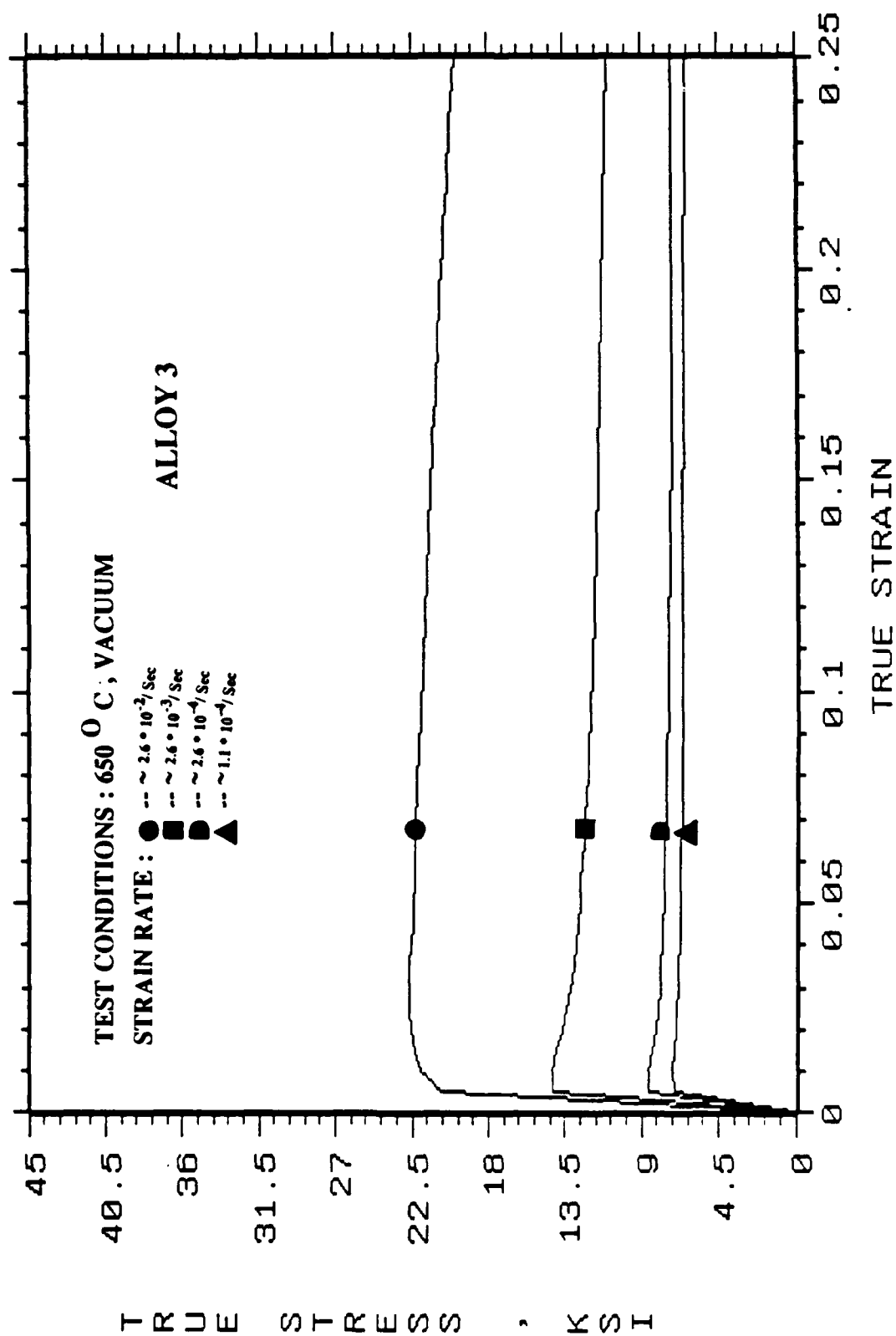


FIG 17 : TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 4 (56.7 % α - 43.3 % β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 923 $^{\circ}$ K AT DIFFERENT STRAIN RATES.

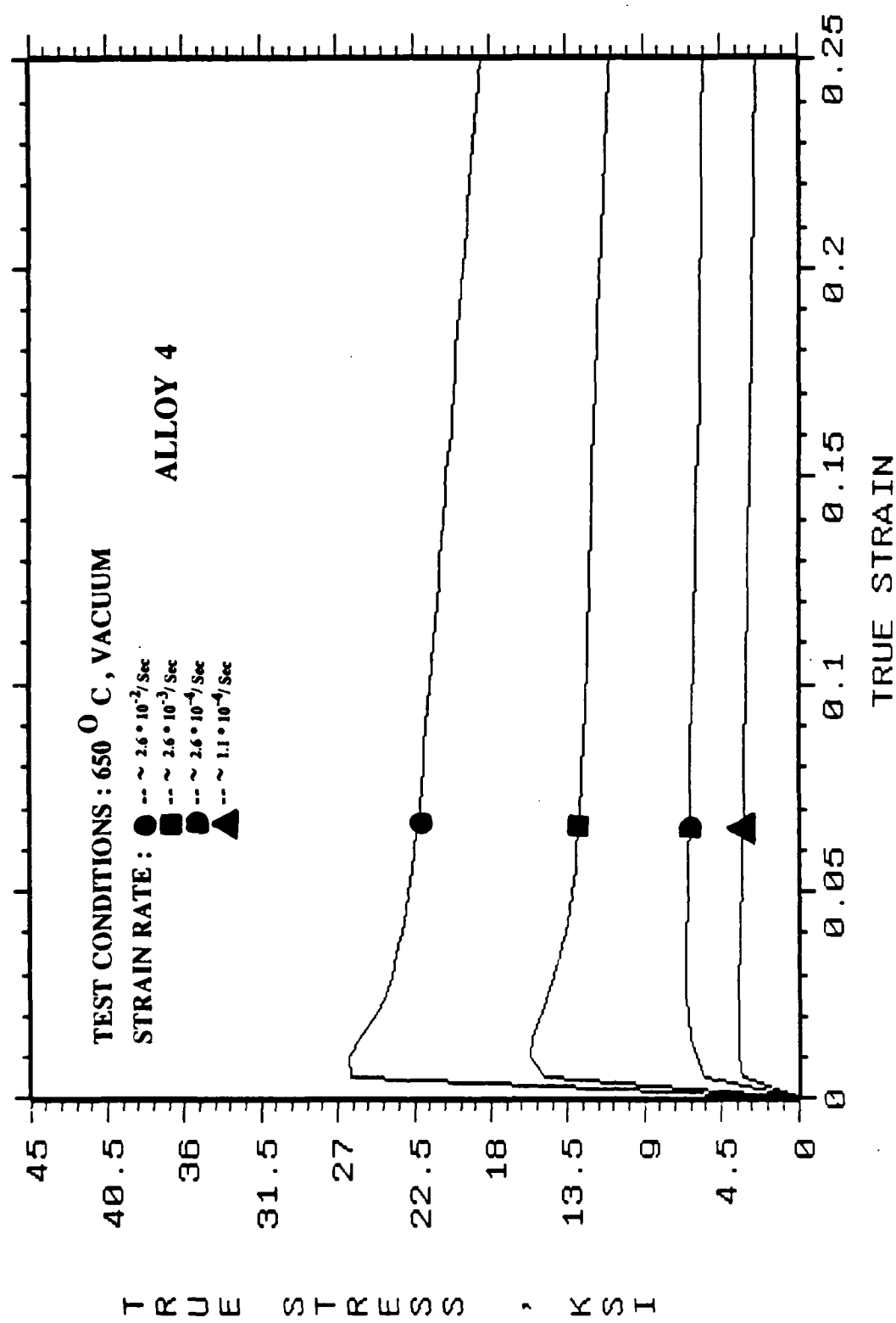


FIG 18 : TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 5 (24.4 % α - 75.6 % β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 923 °K AT DIFFERENT STRAIN RATES.

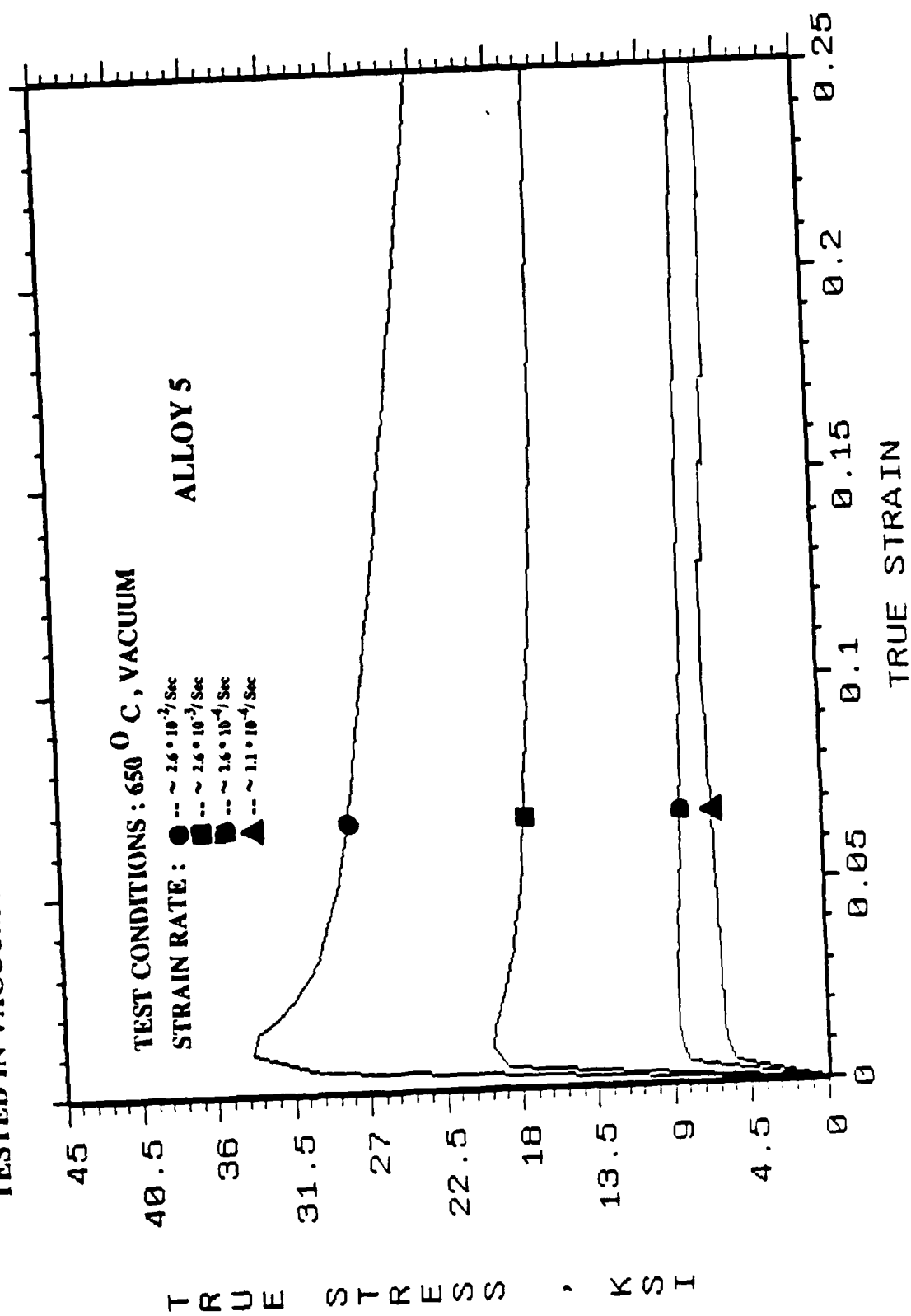


FIG 19 : TRUE STRESS - TRUE STRAIN CURVES OF ALLOY 6(12.1 % α - 87.9 % β , Ti-Mn ALLOY)
TESTED IN VACUUM AT 923 $^{\circ}$ K AT DIFFERENT STRAIN RATES.

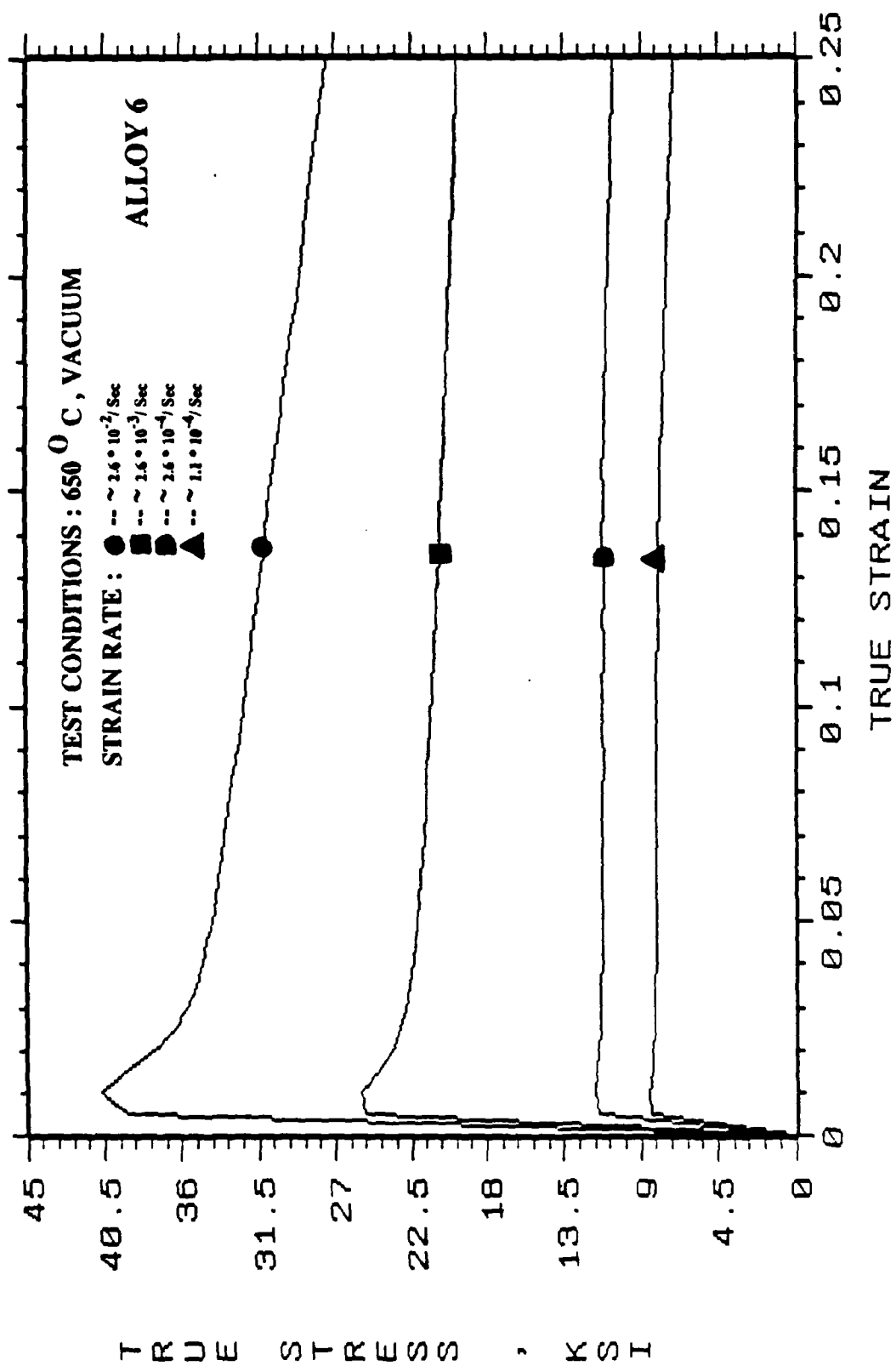


FIG 20 : FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 0.2 % PLASTIC OFFSET

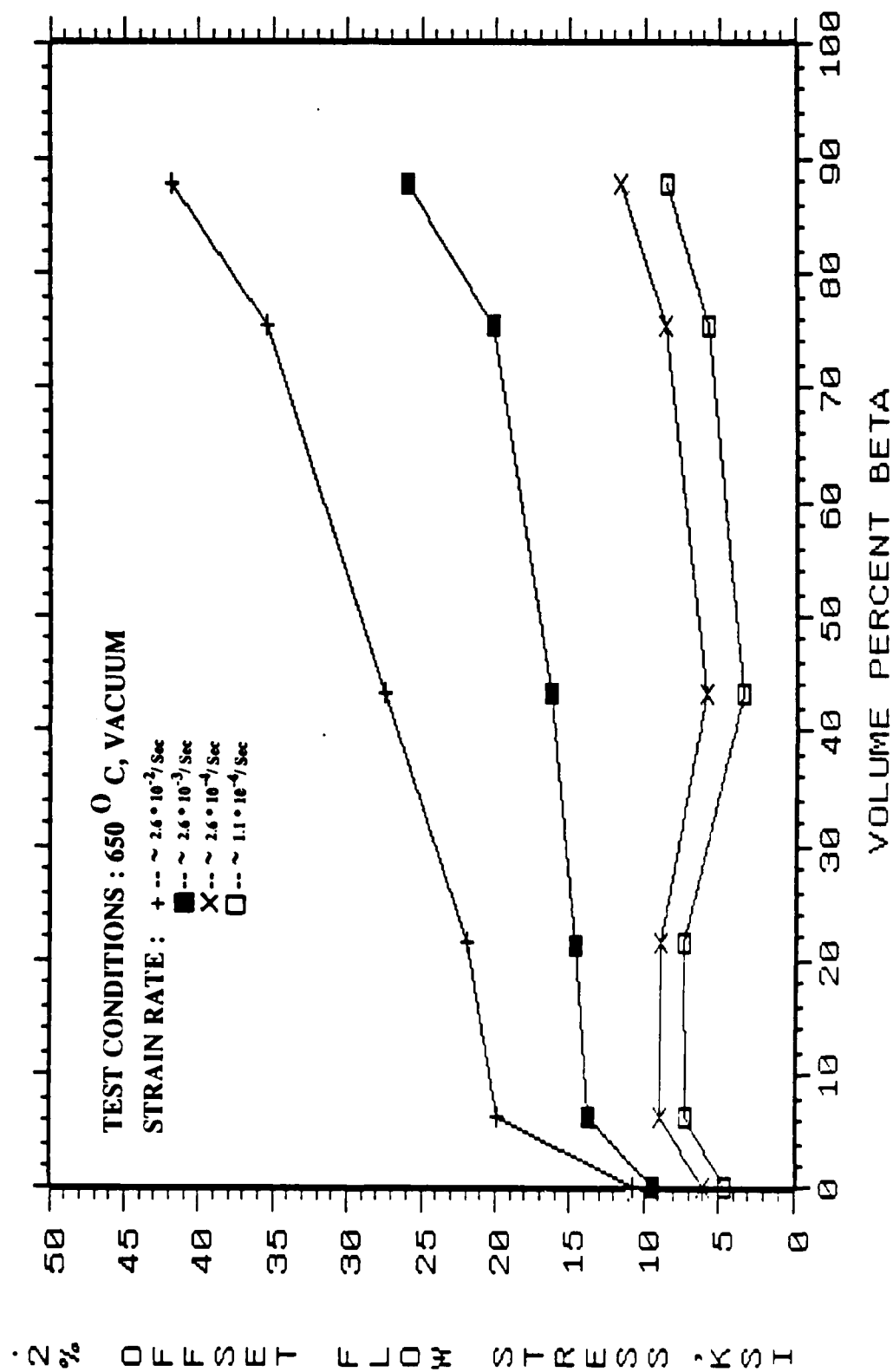


FIG 21: FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 5 % PLASTIC STRAIN

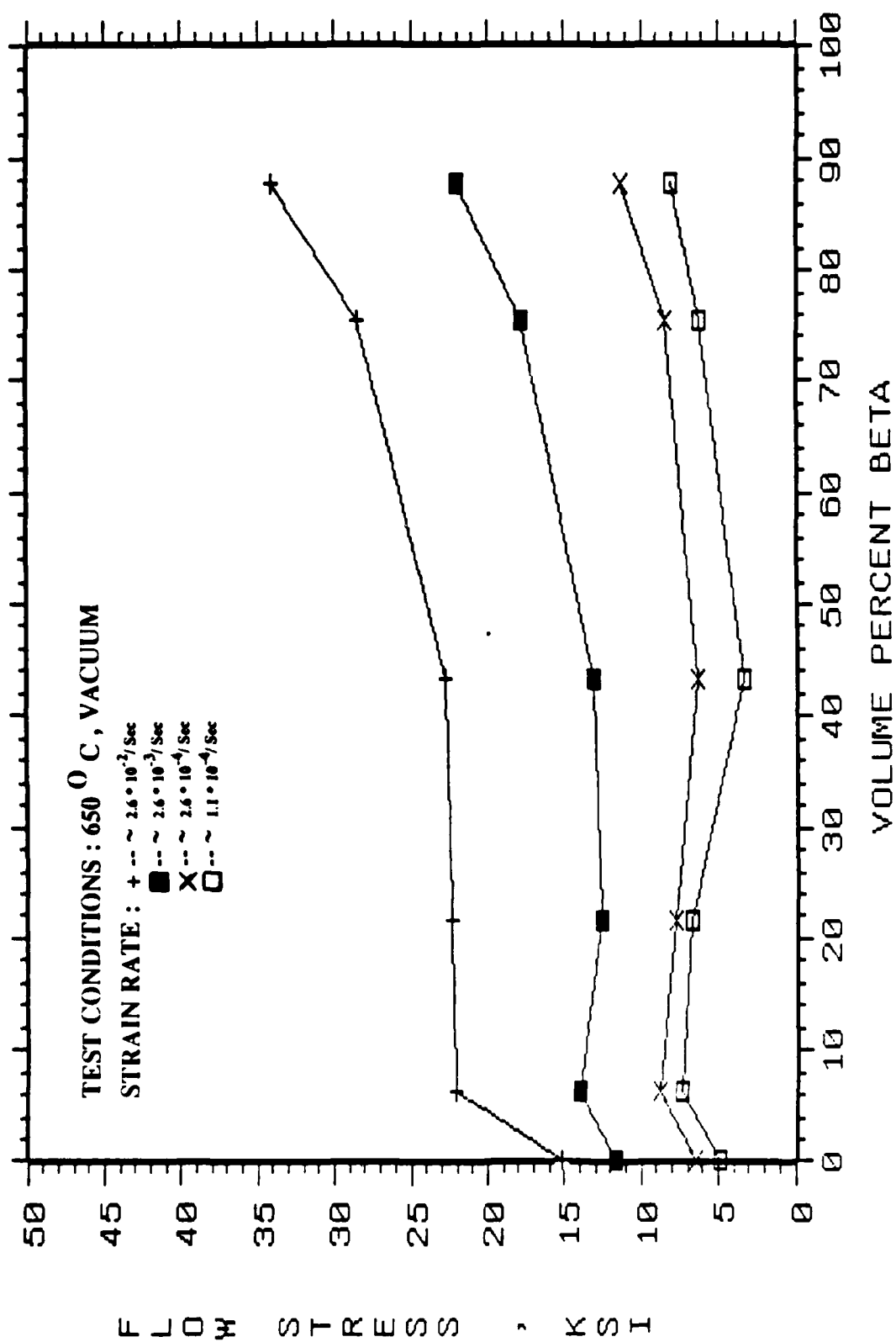


FIG 22 : FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 0.5 % PLASTIC STRAIN

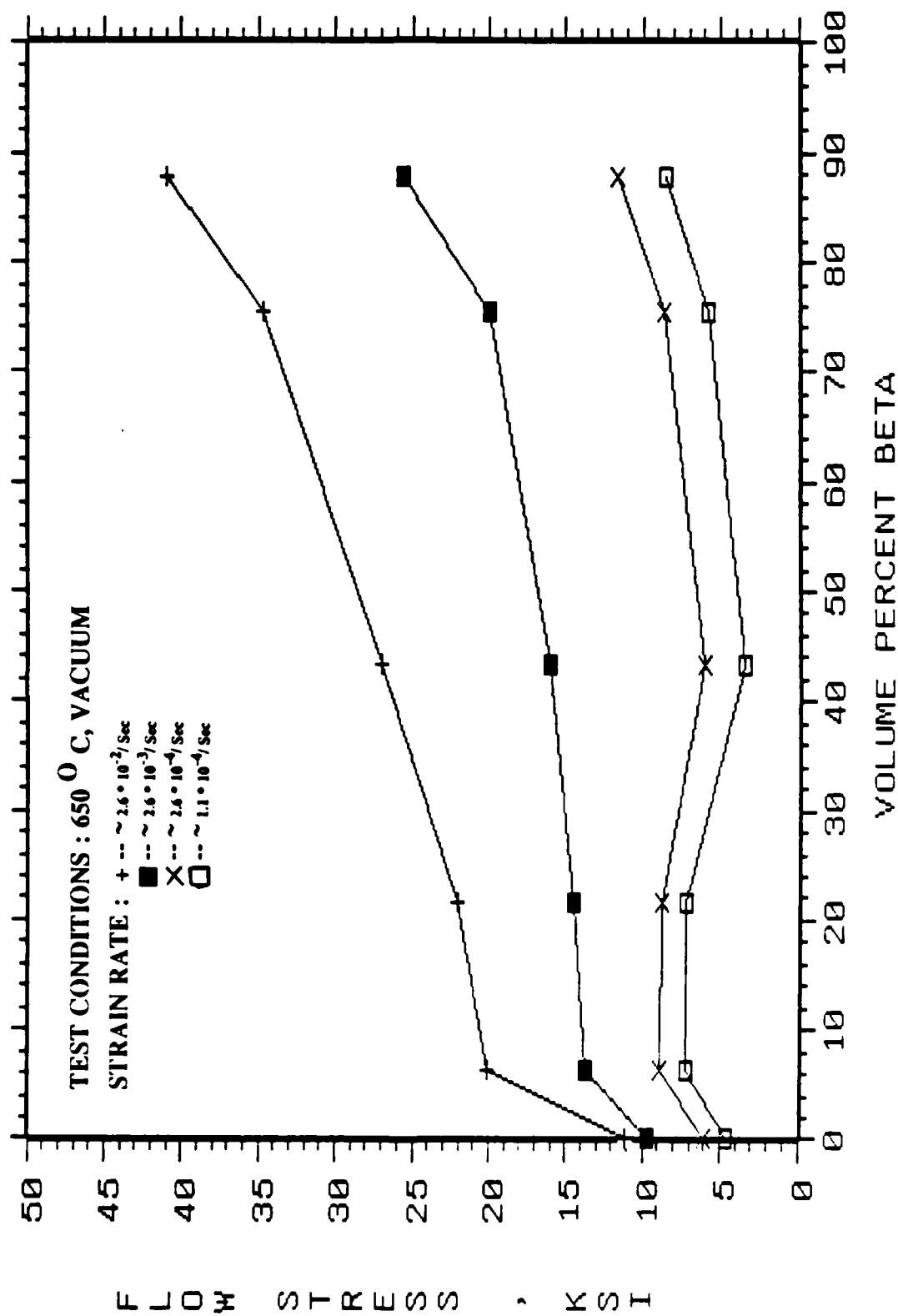


FIG 23: FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 10 % PLASTIC STRAIN

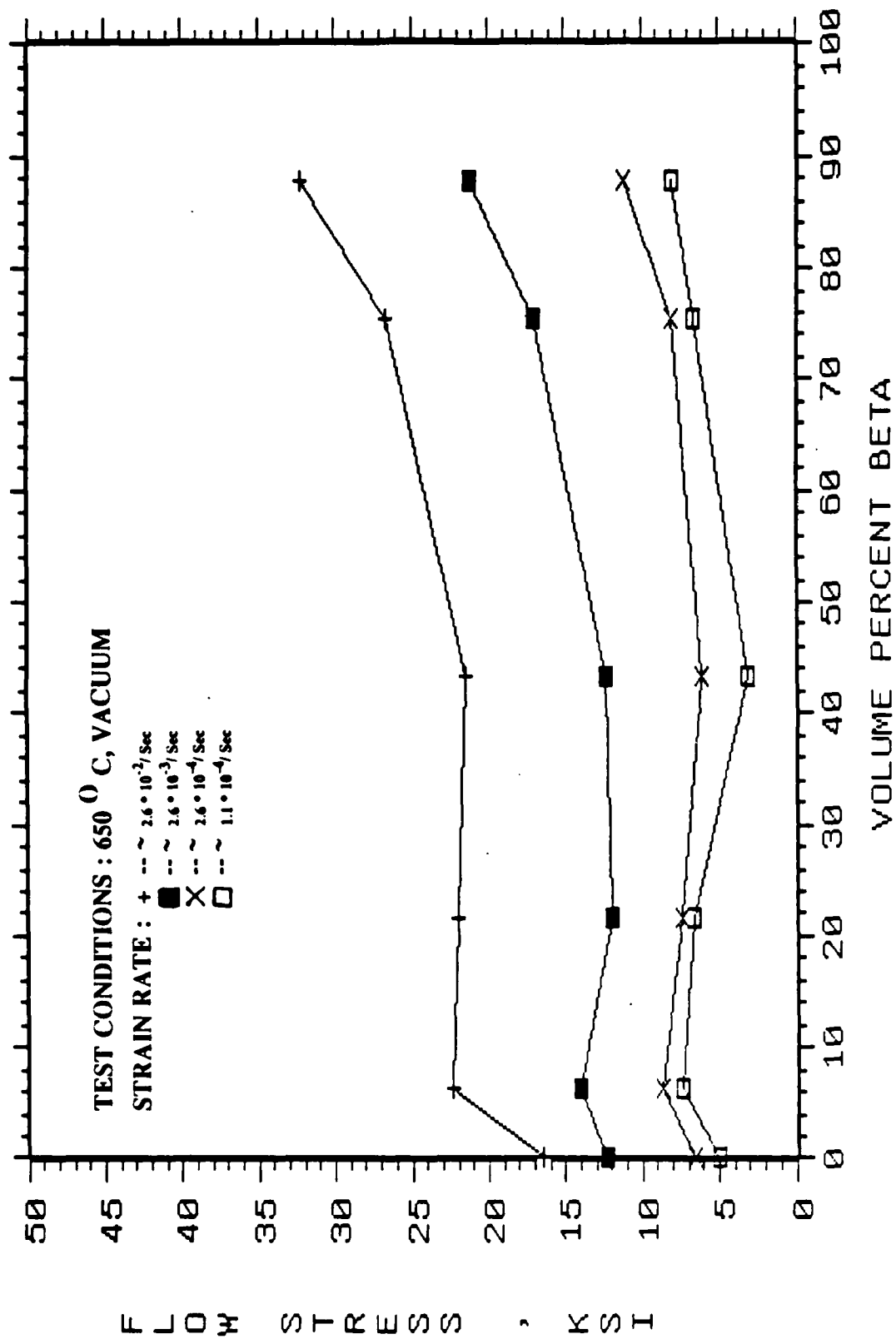


FIG 24: FLOW STRESS VS BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS
THE FLOW STRESS CORRESPONDS TO 20 % PLASTIC STRAIN

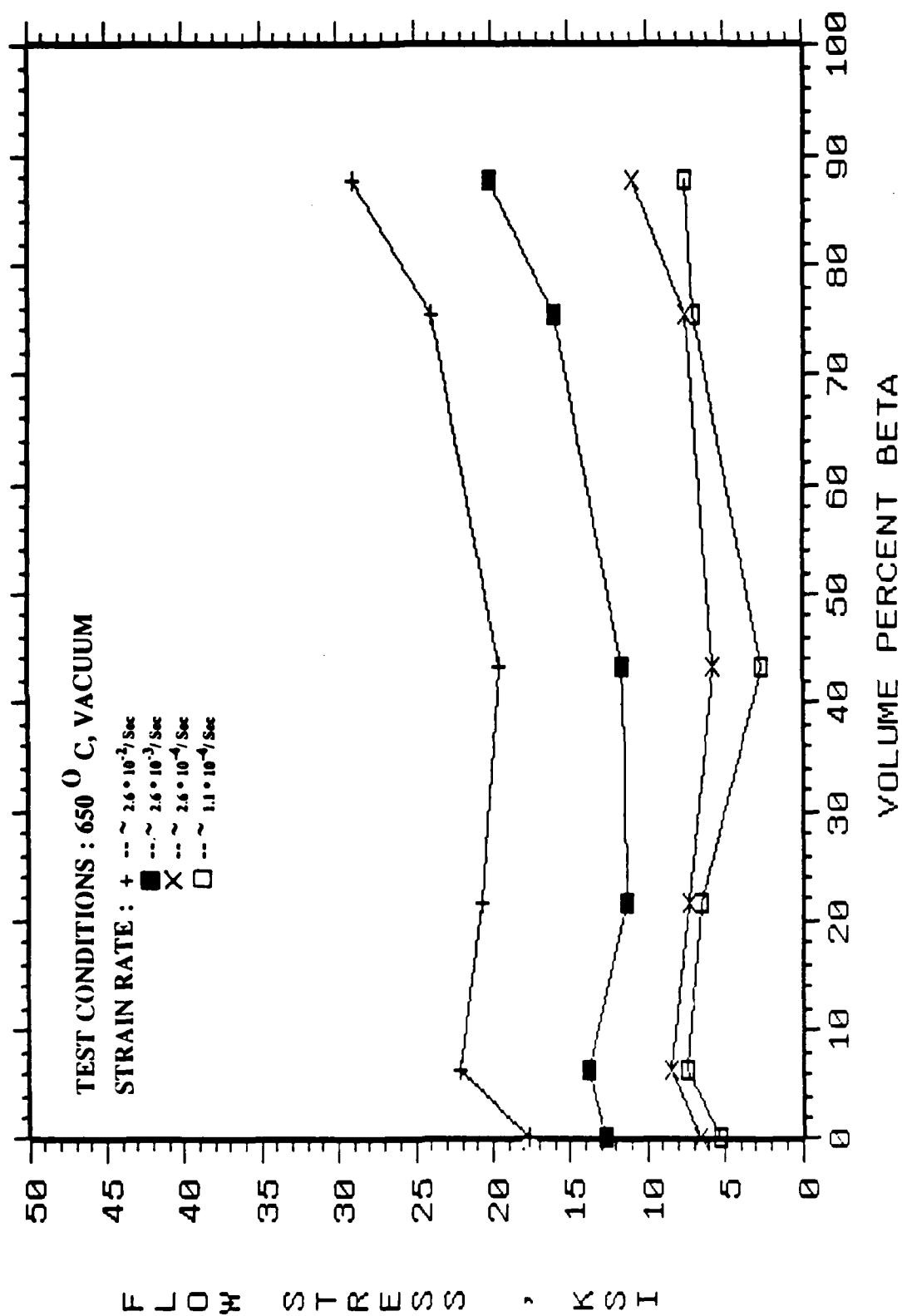
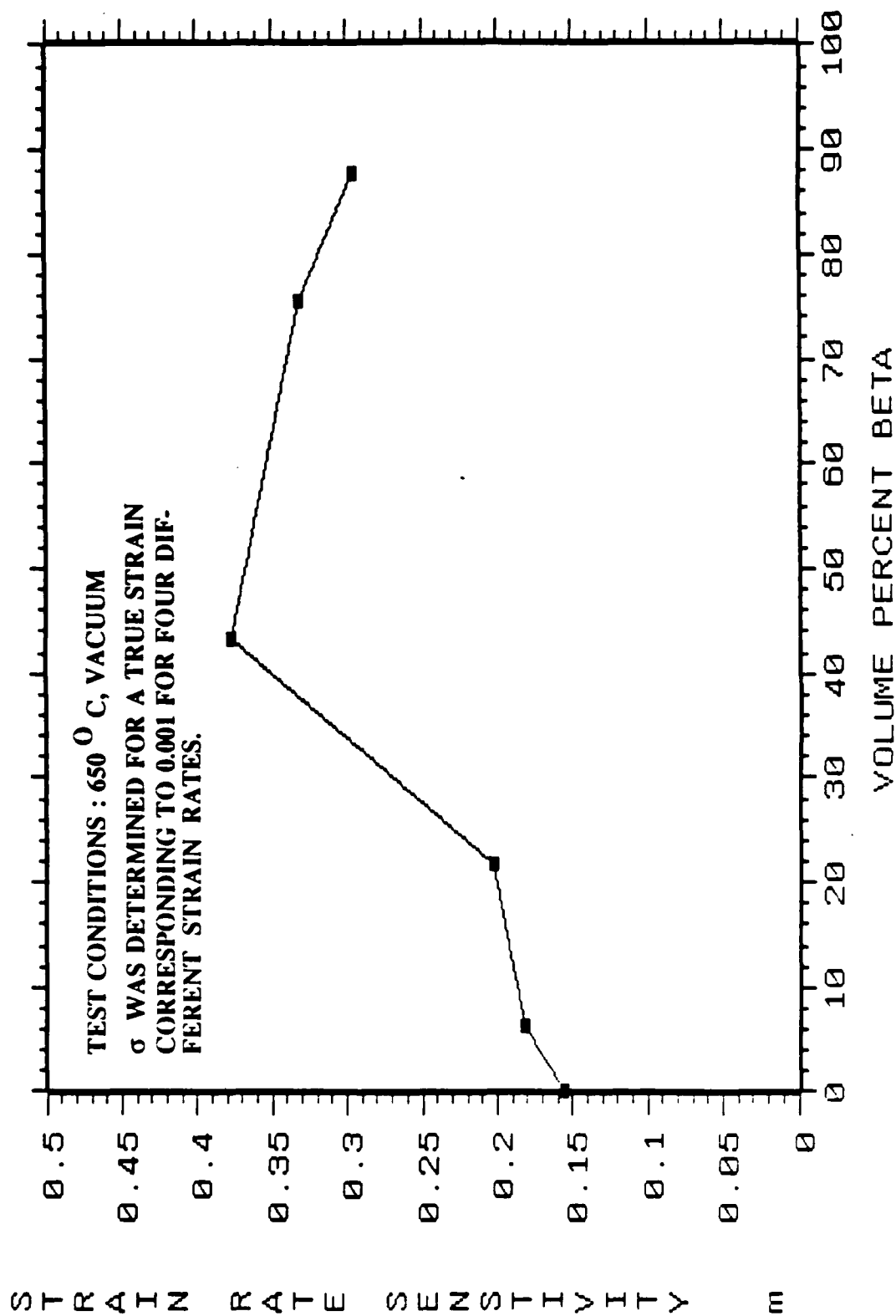


FIG 25: STRAIN RATE SENSITIVITY (m) Vs BETA VOLUME PERCENT FOR DIFFERENT ALPHA-BETA Ti-Mn ALLOYS DETERMINED FROM LOG σ Vs LOG ϵ PLOT. THE SLOPE OF THE PLOT IS THE STRAIN RATE SENSITIVITY (m).



4. TECHNICAL PRESENTATIONS AND PUBLICATIONS

PRESENTATIONS

1. "The Effect of Microstructure On High Temperature Tensile Properties Of Two Phase Titanium Alloys" - J.G.Shyue, S.Ankem and R.J.Arsenault - TMS - AIME Annual Meeting, Denver, Colorado, Feb. 1987.
2. "Grain Growth In Two Phase Titanium Alloys" - G.Grewal and S.Ankem - Third International Conference On Progress In Microstructure, Aachen, West Germany, May 1987.
3. "The Effect Of Temperature On The Particle Growth Of Two Phase Materials" - G.Grewal and S.Ankem ; Presented at The TMS - AIME Fall Meeting, Cincinnati, Ohio, Oct. 1987.
4. "FEM Modeling Of The Stress-Strain Behavior Of Two Phase Materials" - M.N.Vijayshankar and S.Ankem ; Presented at the TMS - AIME Fall Meeting, Cincinnati, Ohio, Oct. 1987.

PUBLICATIONS

1. "Isothermal Particle Growth In Two Phase Titanium Alloys" - G.Grewal and S.Ankem; Prepared, being submitted for publication in " Metallurgical Transactions ".
2. "Effect Of Volume Percent Of Phases On The High Temperature Tensile Deformation Of Two Phase Ti-Mn Alloys" - S.Ankem, J.G.Shyue, R.J.Arsenault and M.N.Vijayshankar; Being prepared, will be submitted for publication in " Materials Science and Engineering ".

5. KEY PERSONNEL FOR THE SECOND PERIOD

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G.Grewal - Ph.D. Candidate

M.N.Vijayshankar - Ph.D. Candidate

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