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MECHANICAL BEHAVIOR OF TITANIUM ALLOYS (THE INFLUENCE OF ALPHA/--ETC(U)
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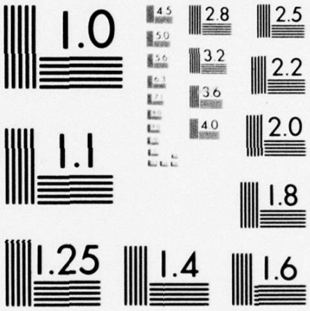
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MECHANICAL BEHAVIOR OF TITANIUM ALLOYS

TECHNICAL REPORT

Submitted to:

Office of Naval Research

By:

C. G. Rhodes and N. E. Paton

Rockwell International
Science Center

(14) SC5056.2TR

(6) MECHANICAL BEHAVIOR OF TITANIUM ALLOYS

(THE INFLUENCE OF α/β INTERFACE PHASE ON TENSILE PROPERTIES IN Ti-6Al-4V).

(Alpha/Beta

(9) Technical rept. Feb[redacted]977 ~~to~~ Feb[redacted]978,

Submitted to:

Office of Naval Research
800 N. Quincy St.
Arlington, VA 22217
Attn: Dr. Bruce MacDonald
M/F: Contract No. N00014-76-C-0598

(15)

(10) C. G./Rhodes
N. E./Paton

(11) 23 Jun[redacted]78

(12) 38p.

389 949

Rockwell International
Science Center

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Unclassified

SECURITY CLASSIFICATION OF THIS PAGE (*When Data Entered*)

REPORT DOCUMENTATION PAGE		READ INSTRUCTIONS BEFORE COMPLETING FORM
1. REPORT NUMBER	2. GOVT ACCESSION NO.	3. RECIPIENT'S CATALOG NUMBER
4. TITLE (*and Subtitle*) Mechanical Behavior of Titanium Alloys (The Influence of α/β Interface Phase on Tensile Properties in Ti-6Al-4V).		5. TYPE OF REPORT & PERIOD COVERED Technical Report February, 1977 - Feb. 1978 6. PERFORMING ORG. REPORT NUMBER SC5056.2TR
7. AUTHOR(s) C. G. Rhodes and N. E. Paton		8. CONTRACT OR GRANT NUMBER(s) N00014-76-C-0598
9. PERFORMING ORGANIZATION NAME AND ADDRESS Science Center, Rockwell International 1049 Camino Dos Rios Thousand Oaks, CA. 91360		10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS 471
11. CONTROLLING OFFICE NAME AND ADDRESS Dr. Bruce MacDonald Office of Naval Research 800 N. Quincy, Arlington, Va. 22217		12. REPORT DATE June, 1978 13. NUMBER OF PAGES 34
14. MONITORING AGENCY NAME & ADDRESS(*if different from Controlling Office*)		15. SECURITY CLASS. (*of this report*) Unclassified 15a. DECLASSIFICATION/DOWNGRADING SCHEDULE

16. DISTRIBUTION STATEMENT (*of this Report*)

Approved for Public Release; Distribution Unlimited.

17. DISTRIBUTION STATEMENT (*of the abstract entered in Block 20, if different from Report*)

18. SUPPLEMENTARY NOTES

ALPHA/BETA

19. KEY WORDS (*Continue on reverse side if necessary and identify by block number*)

ALPHA

Titanium, Interface Phase, Tensile Properties, Electron Microscopy.

20. ABSTRACT (*Continue on reverse side if necessary and identify by block number*)

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is not unambiguously clear, but evidence indicates that yield strength increases with increasing interface phase width when the interface phase exceeds about 2500A and elongation decreases with increasing interface phase width when the interface phase is less than about 2500Å. It is shown that the interface phase raises yield strength and lowers ductility by acting as a barrier to slip and promoting planar dislocation arrangements in the primary alpha. The concepts of a "rule of mixtures" and "continuous phase" are examined in terms of describing tensile properties of Ti-6Al-4V.

THE INFLUENCE OF α/β INTERFACE PHASE ON TENSILE PROPERTIES IN Ti-6Al-4V

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ABSTRACT

The effects of α/β interface phase on room temperature tensile properties in Ti-6Al-4V having an equiaxed primary α microstructure have been studied systematically. Due to the conditions under which it grows, manipulation of the interface phase width results in alteration of the volume fraction of primary α in the alloy. Tensile yield strength and elongation were correlated to interface phase width and volume fraction primary α. The relative individual influence of each of these microstructural features on properties is not unambigously clear, but evidence indicates that yield strength increases with increasing interface phase width when the interface phase exceeds about 2500Å and elongation decreases with increasing interface phase width when the interface phase is less than about 2500Å. It is shown that the interface phase raises yield strength and lowers elongation by acting as a barrier to slip and promoting planar dislocation arrangements in the primary alpha. The concepts of a "rule of mixtures" and "continuous phase" as applied to alpha plus beta phase microstructures are examined in terms of describing the tensile properties of Ti-6Al-4V.

INTRODUCTION

"Interface phase" and "interfacial layer" are terms used to describe the microstructural feature which is present under certain conditions in the α/β interfaces in two-phase titanium alloys.(1-4) Interface phase was previously shown to occur as an intermediate step in the $\beta \rightarrow \alpha$ transformation during slow cooling, presumably as a result of sluggish diffusion of beta stabilizers.(4) Slow cooling is an integral part of many processing techniques for titanium alloys, and consequently the potential effects of interface phase on mechanical properties in these alloys warrant investigation.

The presence of a broad layer of interface phase--it has been observed as broad as 1 micron(4)--may lead to deleterious effects on mechanical properties. For instance, the interface phase might provide an easy crack path or crack nucleation sites in a fatigue failure or a tensile overload failure. Additionally, the interfacial layer may affect ductility by inhibiting slip between the α and β phases or by providing prolific dislocation sources for both the α and β phases.

Margolin et al.,(3) on the other hand, have speculated that interface phase should have little or no effect on ductility. They suggest that slip is easily accommodated across α/β interfaces, however, they unwittingly cite results on alloys which have been heat treated to minimize interface phase.(3) Margolin et al. conclude that the possible effects of interface phase on crack propagation would be, at most, only a secondary effect.

In light of the uncertainties surrounding the influence of interface phase, a study was undertaken to examine the effects of interface phase on mechanical properties in Ti-6Al-4V. This paper reports the results of the tensile properties portion of the study.

EXPERIMENTAL

Two heats of Ti-6Al-4V were used in this work, the chemical analyses are listed in Table I. The "as-received" microstructure of both heats consisted of ~90% equiaxed primary alpha particles, having approximately a 12μm diameter in a continuous β matrix. The heat treatments were carried out with the samples either in a dynamic inert gas atmosphere or encapsulated in evacuated ampoules. A programmable controller supplying power to the furnace was used for controlled-cooling-rate treatments.

Tensile tests were conducted on an Instron testing machine with an extensometer attached to the sample. The samples were in the form of round bars with a 6.35 mm diameter and 31.75 mm gauge length. A strain rate of 2.6×10^{-4} $\sec^{-1}$ was used in all tensile tests. The starting material of both heats used in the study was in the form of plate. Tensile specimens from each heat were taken such that the tensile axis corresponded to the same orientation of the plate for all samples to reduce texture effects in the tensile results.

Thin foils for transmission electron microscopy were prepared by conventional electropolishing techniques(5) or by ion milling.(6) The foils were examined in a Philips EM-300 electron microscope equipped with a double tilt goniometer stage.

RESULTS

Starting Conditions

In order to determine the influence of interface phase on tensile properties in Ti-6Al-4V, the other microstructural variables must be held constant, or at least allowed to vary within a minimal range. The major microstructural feature to be controlled in α/β titanium alloys is the primary alpha, because its morphology and volume fraction can affect tensile properties.(7) A fixed morphology of primary alpha was, therefore, considered necessary for this study and, inasmuch as Ti-6Al-4V is most frequently used in an α/β processed condition, an equiaxed primary alpha morphology was selected. The volume fraction of primary alpha was less easily controlled because of the condition required to vary the interface phase. It was shown previously(4) that interface phase can only be altered by altering the cooling rate from temperatures high in the $\alpha+\beta$ phase field or in the β-phase field. Since the amount of primary alpha phase will be affected by cooling rates and quenching temperatures, the variety of treatments used to produce various interface phase widths resulted in volume fractions of primary alpha ranging from .73 to .90 in tensile samples.

Previous work(4) showed that the interface phase width in Ti-6Al-4V can be varied from 500Å to 4500Å. The constraint of retaining a constant volume fraction primary alpha while developing various widths of interface phase proved virtually prohibitive in the formulation of heat treatments. The heat treatments selected for tensile specimens were either single-stage (controlled-cool treatments) or two-stage (controlled-cool followed by isothermal treatments). The isothermal step in the two-stage treatments was intended to reduce the width of the interface phase which had been formed during the slow cooling step. This was based on the concept that the lower amount of equilibrium alpha phase at the isothermal step temperature would promote transformation of $\alpha \rightarrow \beta$ which would initiate at α/β boundaries and consume interface phase. These treatments were only moderately successful in significantly altering interface phase widths at isothermal-treatment temperatures of 760°C (1400°F) and below, apparently because diffusion (transformation) rates are low at those temperatures and the maximum isothermal hold time was six hours. The single-stage, controlled-cool treatments were more effective in altering the interface phase width, but at the same time these treatments produced a wider variation in volume fraction primary alpha phase.

There are other microstructural features which may vary as a result of the manipulation of the interface phase width. Two of these features which may influence properties are beta phase decomposition and Ti_3Al formation in the primary alpha. During continuous slow cooling from high in the $\alpha+\beta$ phase field, there is little likelihood of α-phase precipitation within the β-phase, but upon quenching from a predetermined temperature at the end of the controlled-cooling step, the beta phase may decompose, depending upon its composition. There may also be decomposition of the beta phase during the isothermal step of the two-stage heat treatments, again depending upon the β-phase composition. The presence of transformed β in the microstructure can have an effect on properties.(7)

The formation of Ti_3Al is a sluggish reaction, but is known to occur in Ti-6Al-4V.(8) The very slow cooling rates required to promote large interface layer widths are conducive to α_2 (Ti_3Al) formation. The presence of α_2 particles in α-phase has been shown to influence tensile properties.(8)

All of the microstructural features described here can have an effect on mechanical properties. Each has a different degree of influence on any particular property and the relative contribution of each acting singly or in concert with others can only be analyzed when all variables are controlled. In order to isolate the effects of interface phase on tensile properties, therefore, the other microstructural variables must be held reasonably constant.

Tensile Results

Room temperature tensile tests were performed on fourteen specimens given a variety of subtransus heat treatments aimed at producing a variety of interfacial layer widths. In all cases, however, the primary α particles were equiaxed. The tensile results, presented in Table II, reveal that a wide range in tensile yield strengths (from 683 to 903 MN/m^2) can be developed by altering the heat treatment of α/β processed Ti-6Al-4V. The ultimate tensile strength is influenced considerably less by heat treatment within each alloy heat, although there is a systemic difference in ultimate strength between the two alloy heats.

Tensile elongation varies from 12.6 to 18.3% among the samples with no systematic difference between the two alloy heats. Generally, the tensile ductility increases as the yield strength decreases. Table II also lists the volume fraction of primary alpha phase for each tensile test specimen. These values are seen to range from 0.73 to 0.90.

In order to determine the inherent scatter in identical samples, five tensile specimens from heat #2 were given the identical treatment of 980^{0}C/2 hrs, cooled at 56^{0}C/hr to 760^{0}C/WQ, which should produce a microstructure containing a narrow interface phase. These tensile results showed an average yield strength of 700.5 ±15.8 MN/m^2 and an average elongation of 15.3 ±1.1%, where the limits are 95% confidence intervals. Although this small spread may result from variations in microstructure, it is assumed that the microstructure is constant among the samples and the scatter is the statistical variation for identical samples. This means that correlations drawn between properties and microstructure must be made among samples in which yield strength varies by more than 31.6 MN/m^2 and elongation by more than 2%.

Microstructures

A typical example of the microstructure of the tensile samples is shown in Fig. 1, where the equiaxed alpha particles as mentioned earlier, are on the order of 12 microns in diameter. The effects of volume fraction primary alpha on tensile yield strength and tensile elongation are illustrated in Figs. 2 and 3. These plots indicate that there is a correlation of the amount of primary alpha in the microstructure with these two tensile properties, although there is scatter in the data. It would be logical to assume that other microstructural features mentioned earlier would also be contributing to the yield strength and elongation of each of these samples and may, in fact, account for some of the scatter observed in the plots of Figs. 2 and 3.

Transmission electron microscopy is required for characterization of interface phase, transformed β, and α_2 particle precipitation in the 6-4 alloy specimens. Although the heat treatments were designed to preclude the precipitation of α-phase particles within the β-phase matrix, test specimens 5 and 6 were found to contain some transformed β regions. However, since α precipitation in the β-phase generally acts to strengthen the alloy(7) and no apparent strengthening has occurred in tests 5 and 6 due to α-phase particle precipitation (compare tests 5 and 7 which are within scatter, or tests 6 and 3), transformed β can be ignored as a significant microstructural feature in the tensile tests.

Each of the fourteen tensile samples was examined for α_2 formation and test specimens 1, 9, and 10 were found to contain α_2. Figure 4 illustrates a typical example of the distribution of the fine α_2 particles. Inspection of Table II reveals that test specimen 1 had the greatest yield strength of both alloy heats and the lowest elongation of heat 1. Similarly, test specimens 9 and 10 had the highest yield strength and lowest elongation of heat 2. It is quite likely that the presence of α_2 in the primary α has influenced the tensile properties.

In the course of this investigation, we had occasion to re-heat samples which contained α_2 particles. It was noted that reheating to temperatures as high as 760^{0}C does not dissolve the α_2 particles, whereas 816^{0}C is sufficient for re-solution of the particles. This behavior is illustrated in Fig. 5.

The final microstructural feature analyzed in this study is the α/β interface phase. Before an examination of the influence of interface phase on tensile properties is made, however, the relationship between interface phase width and volume fraction primary alpha needs to be established. This relationship is presented in Fig. 6, where a least squares fit of the data indicates a positive correlation of layer width and volume fraction primary alpha. This relationship clearly arises because of the mechanism by which interface phase forms.(4)

Figures 2 and 3 indicated a dependence of tensile properties on volume fraction primary alpha. However, the curve in Fig. 6 leads to the conclusion that the tensile data shown in Figs. 2 and 3 might alternatively be interpreted in terms of interface phase width rather than, or in addition to, volume fraction primary alpha. If interface phase does indeed influence tensile properties, its relative effect on those properties might be dominant or it might be relatively minor when compared to the effect of volume fraction primary alpha on properties. In the following paragraphs, the tensile data will be presented alternatively in terms of, first, the assumption that the interface phase exerts only a minor influence on properties, and, second, the assumption that interface phase has a dominant effect on properties. This treatment of the data is made in order to assess the relative effects of interface phase and primary alpha in influencing tensile properties.

If the interface phase has only a minor effect on tensile properties, then one might expect to find that the scatter in the data points in Figs. 2 and 3 could be accounted for by the presence of interface phase. In order to examine the influence of interface phase on the scatter of the Fig. 2 and 3 data, tensile properties of test specimens having nearly identical volume fraction primary alpha (v.f.α_p) should be compared. Inspection of Fig. 2 reveals six test results within the narrow range of 0.88 $\pm$.02 v.f.α_p. Of these six, three are samples which contained α_2 particles and, therefore, are eliminated from consideration of interface phase effects. The other three show a variation in yield strength of 83 MN/m^2, or about a 10% variation. These data are plotted in Fig. 7 as a function of interface phase width. A least squares fit of the three data points from samples with no α_2 in the microstructure is included to show the trend of the data, not necessarily to imply a linear relationship. The plot indicates a positive correlation of tensile yield strength with interface phase width, and that, at least for interface phase widths greater than 2700Å, interface phase can account for the scatter in the data points of Fig. 2.

In a similar manner, inspection of Fig. 3 shows four test results within the narrow range of 0.80 $\pm$.01 v.f.α_p. These data which show a wide variation in elongation are plotted in Fig. 8 as a function of interface phase width. Again, a least squares fit of the data is drawn only to indicate the trend which is a negative correlation of tensile elongation with interface phase width. It can be seen that at least for interface layer widths less than 2800Å, interface phase can account for the scatter in the data points of Fig. 3.

The approach used in the generation of Figs. 7 and 8 was one that assumed the influence of interface phase on tensile properties to be minor compared to the influence of v.f.α_p. If, on the other hand, the effect of interface phase were dominant over v.f.α_p, the tensile data could be plotted as in Figs. 9 and 10, in which the influence of v.f.α_p is ignored and all data are plotted on a single graph. Figure 9 reveals that interface phase has little influence on tensile yield strength until it attains a width of greater than 2500Å, beyond which it has a strong influence on yield strength.

The relationship between elongation and interface phase width is shown in Fig. 10. In this case, the tensile elongation decreases with increasing interface phase width until the layer reaches ~2500Å, at which point the elongation appears to have attained a minimum value. Continued increase in interface phase width beyond ~2500Å has little influence on elongation.

The dependence of yield strength or elongation on interfacial layer width is different from the dependence of these properties on volume fraction primary alpha (compare Figs. 2 and 9,or 3 and 10). Although this observation is not evidence that one microstructural feature is dominant over the other as regards tensile properties, it seems to indicate that each has some effect on the properties. More will be said on this in the Discussion Section.

The gauge section of each of the tensile samples was examined by thin foil transmission electron microscopy in order to determine the influence of interface phase on dislocation arrangements. It was found that, in general, samples with broad interface phase exhibited more planar slip in the primary α particles whereas those with narrow interface phase contained a more tangled dislocation arrangement. These are illustrated in Figs. 11 and 12. The prominent slip bands in Figs. 11(a) and 12(a) are seen to contain a high density of planar dislocations; trace analyses indicated that these bands lie on $\{10\bar{1}0\}$ α planes. The samples with narrow interface phase, Figs. 11(b) and 12(b), show little, or no, indication of planar slip.

DISCUSSION

This work has shown that there is a correlation between tensile properties and either volume fraction primary alpha or interface phase width or both. In order to resolve the problem as to whether the properties are influenced by primary alpha or interface phase, the expected behavior will be examined and compared to the experimentally observed results.

In examining the influence of v.f.α_p on yield strength, Fig. 2, one must first consider the strengths of both the alpha and beta phases. It is assumed that the vanadium content of the primary alpha is unaltered from a maximum value of about 1.2 wt%(9) as v.f.α_p changes, and therefore, the yield strength of the alpha phase will not be affected. The vanadium content of the β phase, however, changes as the volume fraction β-phase changes. Calculations show that the composition of the β phase is ~23.5 wt% V at 90%α_p and 13.8 wt% V at 80%α_p. Extrapolation of the data of

Ling et al.(10) shows that the tensile yield strength of Ti-V (β-phase) will increase by ~ 40% as the vanadium content increases from 13.8 to 23.5 wt%.

There is some question as to whether a rule of mixtures predicts the strength of the α/β alloys or whether the strength more closely follows that of the continuous phase.(7) A rule of mixtures has likely been found lacking in predicting mechanical properties in α/β alloys because of the duplex (or triplex) distribution of α-phase, for these alloys are frequently treated to produce primary alpha, secondary alpha, and finely dispersed alpha in a transformed β matrix. The (normalized) contribution of each of these to the yield strength of the alloy would not be equal because each provides a different strengthening mechanism. However, in the present study, a rule of mixtures could be applied because the microstructure consists of coarse, equiaxed primary α particles in a nontransformed β matrix; that is, the alloy is composed of a mixture of two phases in which the presence of each phase only minimally influences the properties of the other. If one assumes a rule of mixtures to account for the two-phase alloy yield strength (Fig. 2), then the increase in β-phase yield strength (based on the extrapolation of Ling et al. data) only accounts for a 20.6 MN/m^2 increase in the alloy yield strength for an increase in primary alpha from .8 to .9 volume fraction. Clearly, a rule of mixtures analysis does not account for the large increase in yield strength with v.f.α_p.

If one assumes that the yield strength of the two-phase alloy is governed by that of the continuous phase, in this case the β-phase, it would be expected that the alloy yield strength should increase by the 40% which was shown to be the β-phase increase. The observed increase, Fig. 2, is

124 MN/m^2 which corresponds to about a 17% increase. In this instance, the increase in alloy yield strength is significantly overestimated by the assumption that the continuous phase dominates the tensile properties. It appears, then, that the rule of mixtures does not apply in a system where it logically could and that the continuous-phase theory does not apply, but rather the observed behavior lies somewhere between these two extremes. The observed behavior could result from β-phase solid solution strengthening which contributes to the yield strength in some modification of the rule of mixtures or it could result from the presence of an additional microstructural feature, namely interface phase.

Further evidence that the behavior observed in Fig. 2 is not necessarily the result of β-phase strengthening is seen in the data of Holden et al.(11) for β-phase alloys in the Ti-Mn system. These data, shown in Fig. 13, indicate that the yield strength of the two-phase alloy is reduced by an increase in primary alpha phase, contrary to the data in Fig. 2. The samples from which Holden et al.'s data were obtained were treated so that interface phase would be minimized or eliminated.

The rule-of-mixtures calculation predicts a very slight positive correlation of tensile yield strength with increasing v.f.α_p and the Holden et al. data predict a negative correlation of tensile yield strength with increasing v.f.α_p. Neither of these predicts the strong positive correlation observed in Fig. 2. The "continuous-phase concept" predicts the behavior of Fig. 2, but it is based on experimental observation rather than developed theory. If one rejects the "continuous-phase concept" then the shape of the curve in Fig. 2 must be addressed. If one accepts the "continuous-phase concept" as explaining the positive correlation of yield strength with volume fraction α_p, the wide spread in yield strength values for test samples having virtually identical v.f.α_p needs to be examined. Each of these will be considered.

If one assumes that a rule-of-mixtures concept should apply, then the curve in Fig. 2, which does not follow a rule-of-mixtures prediction based on two constituents, can be explained in terms of interface phase. The data of Fig. 9 reveal that, when the interface phase width exceeds about 2500Å, it begins to exert a positive influence on tensile yield strength. It would be reasonable to assume, under these conditions, that the alloy yield strength would be predicted by a rule of mixtures including the three microstructural features: v.f.α_p, v.f.β, and interface phase width. This explains why applications of the rule of mixtures using only two microstructural constituents have failed in the past.

If, on the other hand, one accepts the "continuous-phase concept," then interface phase can account for the scatter in the data of Fig. 2. This was demonstrated in Fig. 7 where a positive correlation of tensile yield strength with interface phase width for a constant v.f.α_p (and a constant β-phase composition) is apparent. Although, under these conditions, interface phase effects were considered secondary to β-phase solid solution strengthening, the data of Fig. 7 show a significant increase in yield strength due to interface phase over the range examined. Hence, whether or not one accepts the "continuous-phase concept," interface phase is seen to exert a significant influence on tensile yield strength.

Although the preceding discussion was limited to yield strength, a similar one can be made for tensile elongation. Analyses of Figs. 3, 8, and 10 lead to the conclusion that interface phase has a negative correlation with tensile elongation, by means of arguments similar to those applied in the yield strength discussion.

A mechanism by which interface phase increases tensile yield strength and reduces tensile elongation is indicated from the transmission electron microscopy results. The planar dislocations observed in samples with broad interface phase (and corresponding higher yield strength and lower elongation) indicate that the interface phase acts as a barrier to slip transfer from the α to β phase. The back stresses built up by the planar dislocations will induce more rapid strain hardening and increase the yield strength. A lack of planar slip in the samples with narrow interface phase indicates an easier transfer of slip from the α phase, resulting in decreased strain hardening and a lower yield strength. Elongation is affected in a similar manner. The ease of slip transfer results in higher elongation in those samples with narrow interface phase widths whereas the reduced dislocation generation and motion resulting from the planar dislocation arrangement in samples with broad layers will cause a decrease in total strain.

The relationship between interface phase width and volume percent primary alpha illustrated in Fig. 6 is to be expected because of the interface phase formation mechanism.(4) The interface phase forms during slow cooling simultaneously with the transformation of α to β, i.e., as the volume fraction of primary α is increasing. Although this fact makes it difficult to produce a wide range of interface phase widths for a constant volume fraction primary alpha, several of the samples did have a reasonable range of interface phase widths as indicated by the scatter in the data of Fig. 6. The results demonstrate that interface phase widths can be altered even within the constraints of a fixed v.f.α_p, leading to the conclusion that tensile properties in Ti-6Al-4V can be manipulated by control of interface phase widths.

The formation of α_2 (Ti_3Al) in this alloy was not unexpected for it has been observed before.(8) However, it has generally been observed after long time isothermal treatments. The presence of α_2 after cooling at 56°F/hr to RT should serve as an alert to users of Ti-6Al-4V that processing conditions utilizing slow cooling rates may introduce this potentially detrimental microstructural feature.

CONCLUSIONS

This study has shown that tensile yield strength and elongation in Ti-6Al-4V containing equiaxed primary alpha are influenced by α/β interface phase and/or volume fraction primary alpha. The relative individual influence on properties of each of these microstructural features is not unambiguously clear, but evidence indicates that (1) yield strength increases with increasing interface phase width when the interface phase exceeds about 2500Å and (2) elongation decreases with increasing interface phase width when the interface phase is less than about 2500Å. Interface phase raises yield strength and lowers elongation by acting as a barrier to slip and promoting planar dislocation arrangements in the primary alpha.

ACKNOWLEDGEMENTS

We are pleased to acknowledge the experimental assistance of L.F. Nevarez, R.A. Spurling, E.H. Wright, and P.Q. Sauers. This work was supported by the Office of Naval Research (Contract N00014-71-C-0363).

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TABLE I

COMPOSITION OF Ti-6Al-4V, BY WEIGHT PERCENT

Heat	Ti	Al	V	Fe	C	O	H	N
#1	Bal	6.15	4.09	0.18	0.011	0.129	0.0067	0.019
#2	Bal	6.1	4.0	0.19	0.02	0.122	0.0088	0.018

TABLE II

TABLE II

ROOM TEMPERATURE TENSILE TEST RESULTS FOR TWO HEATS OF Ti-6Al-4V

Test No*	Heat Treatment	Interface Phase Width	Vol.Fract. Primary α	Y.S. MN/m^2	U.T.S. MN/m^2	Elong. (%)
1	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ RT +540°C/4 hrs/WQ	2370Å	0.88	903	972	12.9
2	925°C/2hrs $\xrightarrow{56^\circ C/hr}$ 650°C/WQ	3480	.87	876	931	14.6
3	925°C/2hrs $\xrightarrow{56^\circ C/hr}$ 705°C/WQ	3910	.86	855	903	15.2
4	925°C/2hrs $\xrightarrow{56^\circ C/hr}$ 815°C/WQ	950	.79	745	986	18.3
5	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ RT + 815°C/4hrs/WQ	0	.76	731	986	16.3
6	980°C/2hr $\xrightarrow{56^\circ C/hr}$ RT + 760°C/6hrs/WQ	2760	.81	724	972	14.4
7	870°C/2hrs $\xrightarrow{28^\circ C/hr}$ 805°C/WQ	0	.77	710	958	16.4
8	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ 760°C/WQ	Not Meas	.73	683	965	16.2
9	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ RT +540°C/30min/WQ	2970	.90	834	869	12.6
10	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ RT +540°C/30min/AC	2420	.87	820	848	15.2
11	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ RT +540°C/30min/WQ	2730	.89	793	827	15.2
12	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ RT +730°C/30min/WQ	600	.76	710	910	16.2
13	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ 815°C/WQ	640	.79	703	910	16.6
14	980°C/2hrs $\xrightarrow{56^\circ C/hr}$ RT $\xrightarrow{25^\circ C/hr}$ 815°C/WQ	710	.80	689	903	17.0

* Tests 1-8 are Heat #1, Tests 9-14 are Heat #2.

FIGURE CAPTIONS

Fig. 1	Scanning electron micrograph of Ti-6Al-4V depicting typical microstructure of samples used in this study.
Fig. 2	Tensile yield strength as a function of volume percent primary alpha.
Fig. 3	Tensile elongation as a function of volume percent primary alpha.
Fig. 4	Thin foil transmission electron micrograph of tensile test #1 specimen revealing Ti_3Al particle formation within primary alpha. Dark field, g $(1011)\alpha_2$.
Fig. 5	Selected area electron diffraction patterns demonstrating the stability of Ti_3Al. (a) as slow cooled, $980^oC \xrightarrow{56^oC/hr}$ room temperature; (b) as in (a) plus 760^oC/30min/WQ; (c) as in (a) plus 816^oC/5min/WQ.
Fig. 6	Interface phase width as a function of volume percent primary alpha.
Fig. 7	Tensile yield strength as a function of interface phase width for samples containing 88±2 volume percent primary alpha.
Fig. 8	Tensile elongation as a function of interface phase width for samples containing 80±1 volume percent primary alpha.
Fig. 9	Tensile yield strength as a function of interface phase width.
Fig. 10	Tensile elongation as a function of interface phase width.
Fig. 11	Thin foil transmission electron micrographs of two Ti-6Al-4V tensile tested specimens. $[11\bar{2}0]$ α zone normal in both micrographs. (a) test specimen #11, σ_{ys} = 115 ksi, ϵ = 15.2%, $\vec{g} = 1\bar{1}01$; (b) test specimen #12, σ_{ys} = 103 ksi, ϵ = 16.2%, $\vec{g} = 1\bar{1}01$.
Fig. 12	Thin foil transmission electron micrographs of two Ti-6Al-4V tensile tested specimens, $[11\bar{2}3]$ α zone normal in both micrographs. (a) test specimen #11, σ_{ys} = 115 ksi, ϵ = 15.2%, $\vec{g} = 10\bar{1}1$; (b) test specimen #12, σ_{ys} = 103 ksi, ϵ = 16.2%, $\vec{g} = 10\bar{1}1$.
Fig. 13	Tensile yield strength as a function of volume percent primary alpha in Ti-Mn alloys. Data from Holden et al.(11).

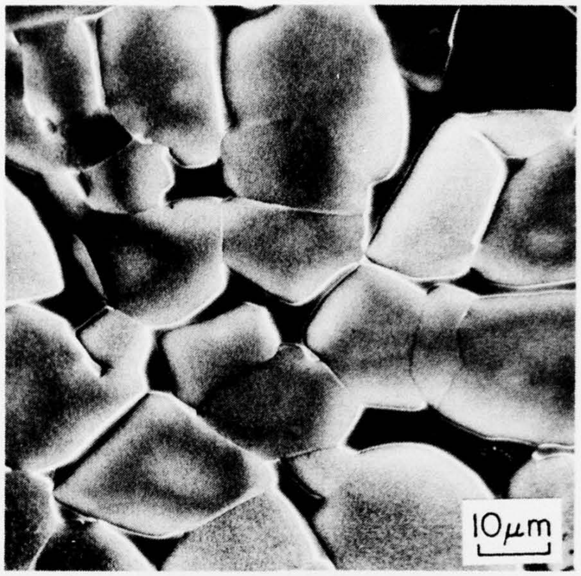

Fig. 1 Scanning electron micrograph of Ti-6Al-4V depicting typical microstructure of samples used in this study.

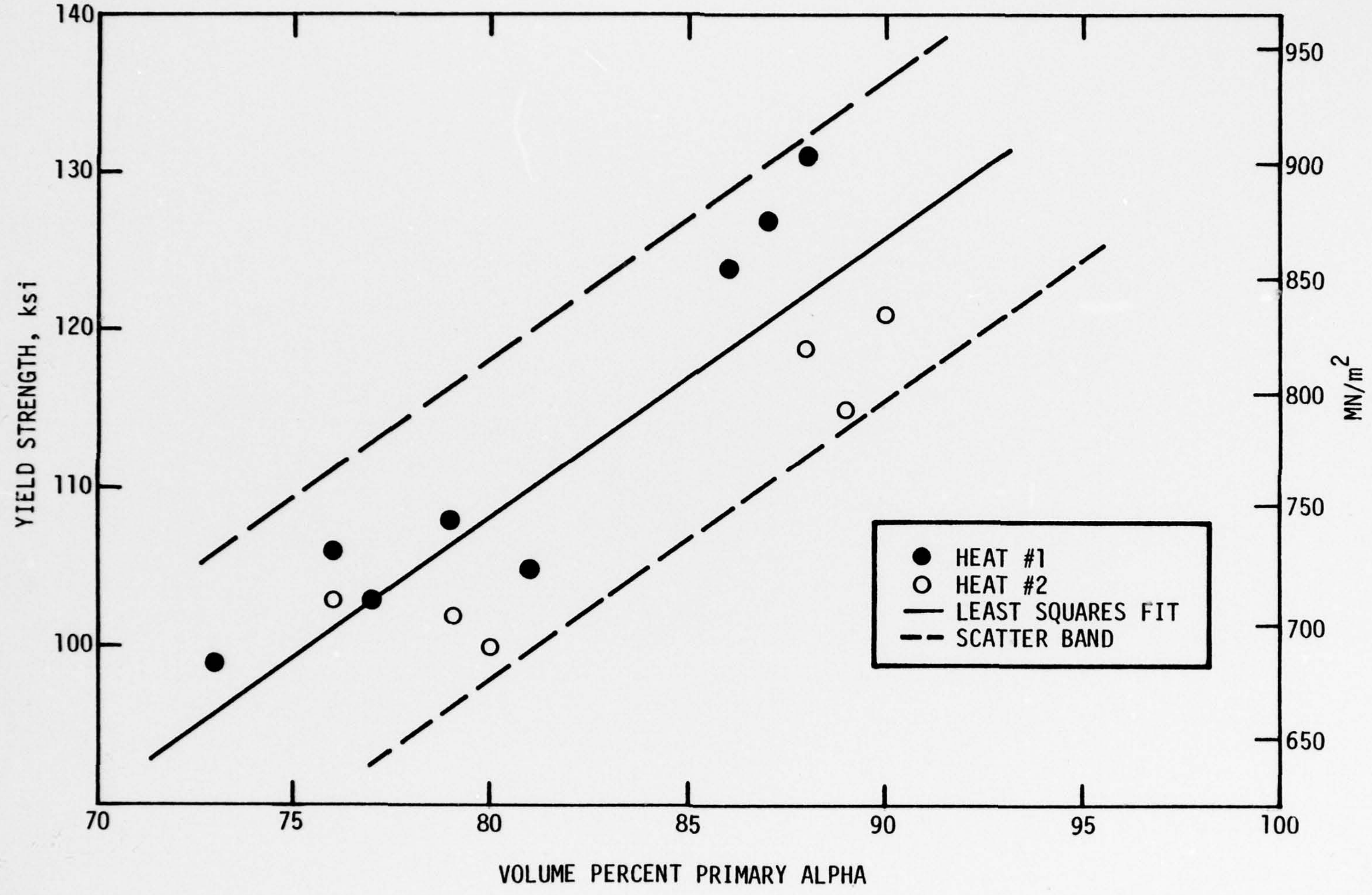

Fig. 2 Tensile yield strength as a function of volume percent primary alpha.

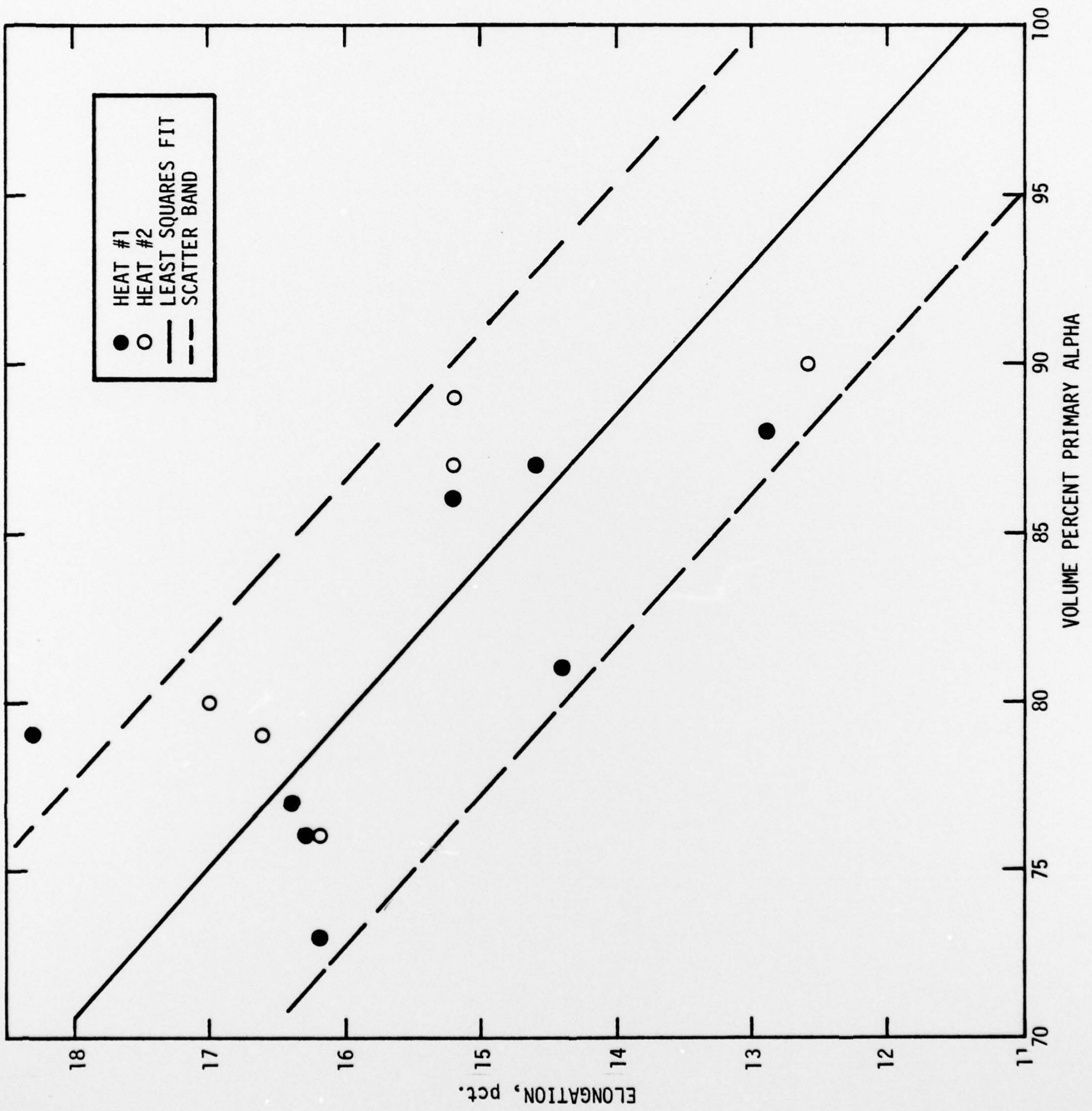

Fig. 3 Tensile elongation as a function of volume percent primary alpha.

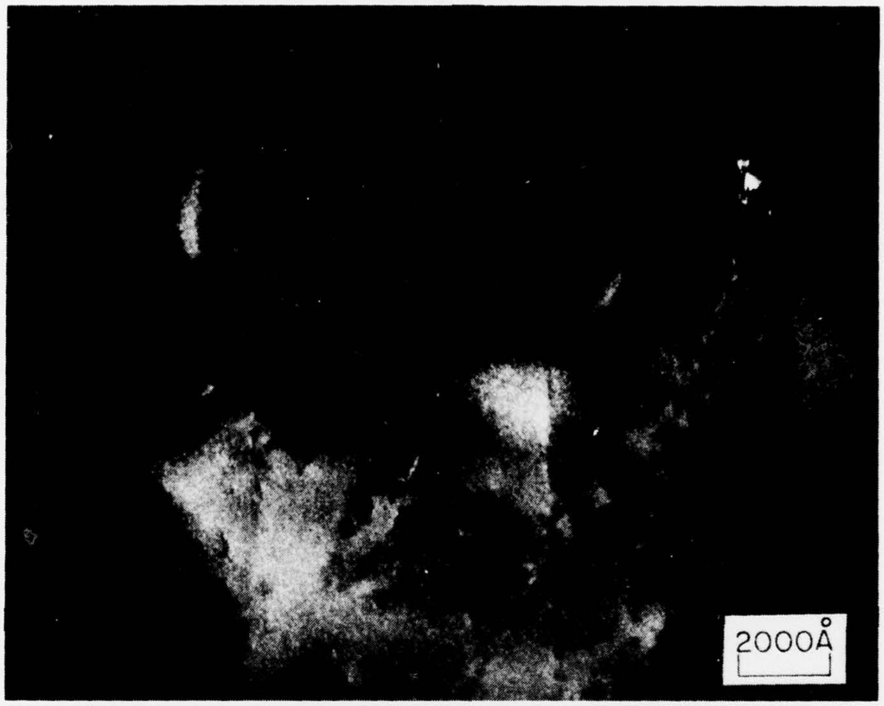

Fig. 4 Thin foil transmission electron micrograph of tensile test #1 specimen revealing Ti_3Al particle formation within primary alpha. Dark field, g (1011)α_2.

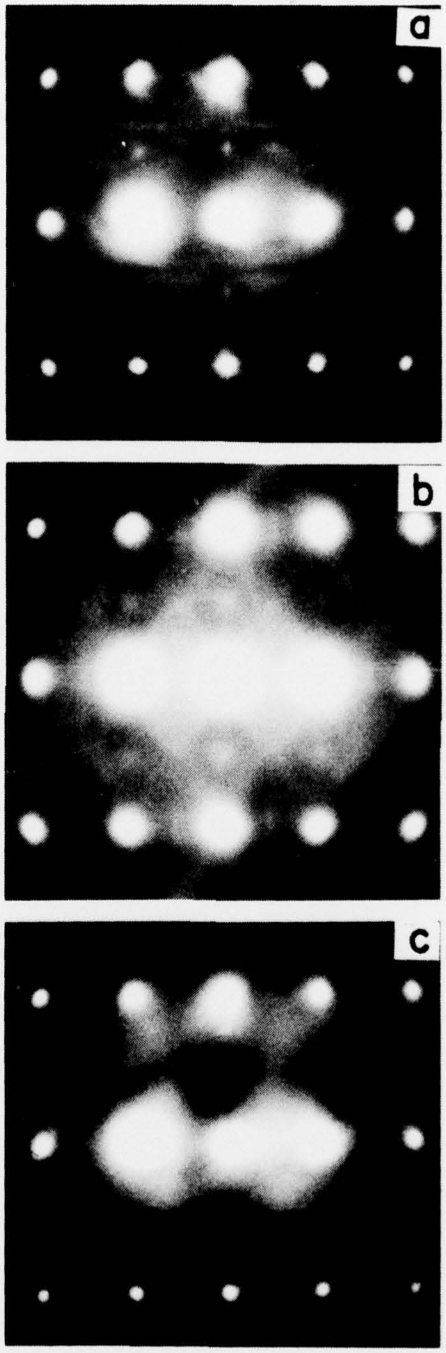

Fig. 5 Selected area electron diffraction patterns demonstrating the stability of Ti_3Al. (a) as slow cooled, $980^{\circ}C \xrightarrow{56^{\circ}C/hr}$ room temperature; (b) as in (a) plus $760^{\circ}C$/30min/WQ; (c) as in (a) plus $816^{\circ}C$/5min/WQ.

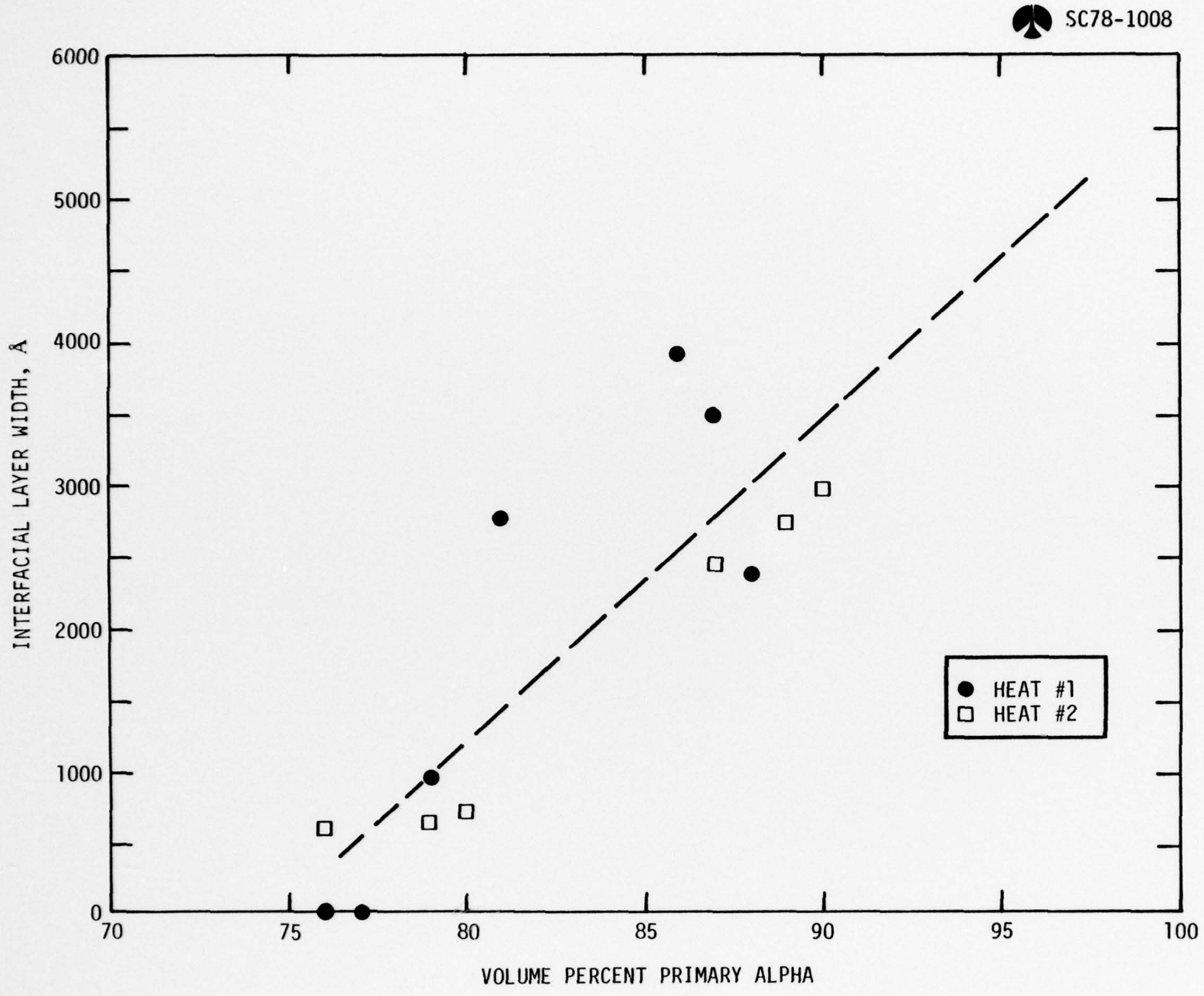

Fig. 6 Interface phase width as a function of volume percent primary alpha.

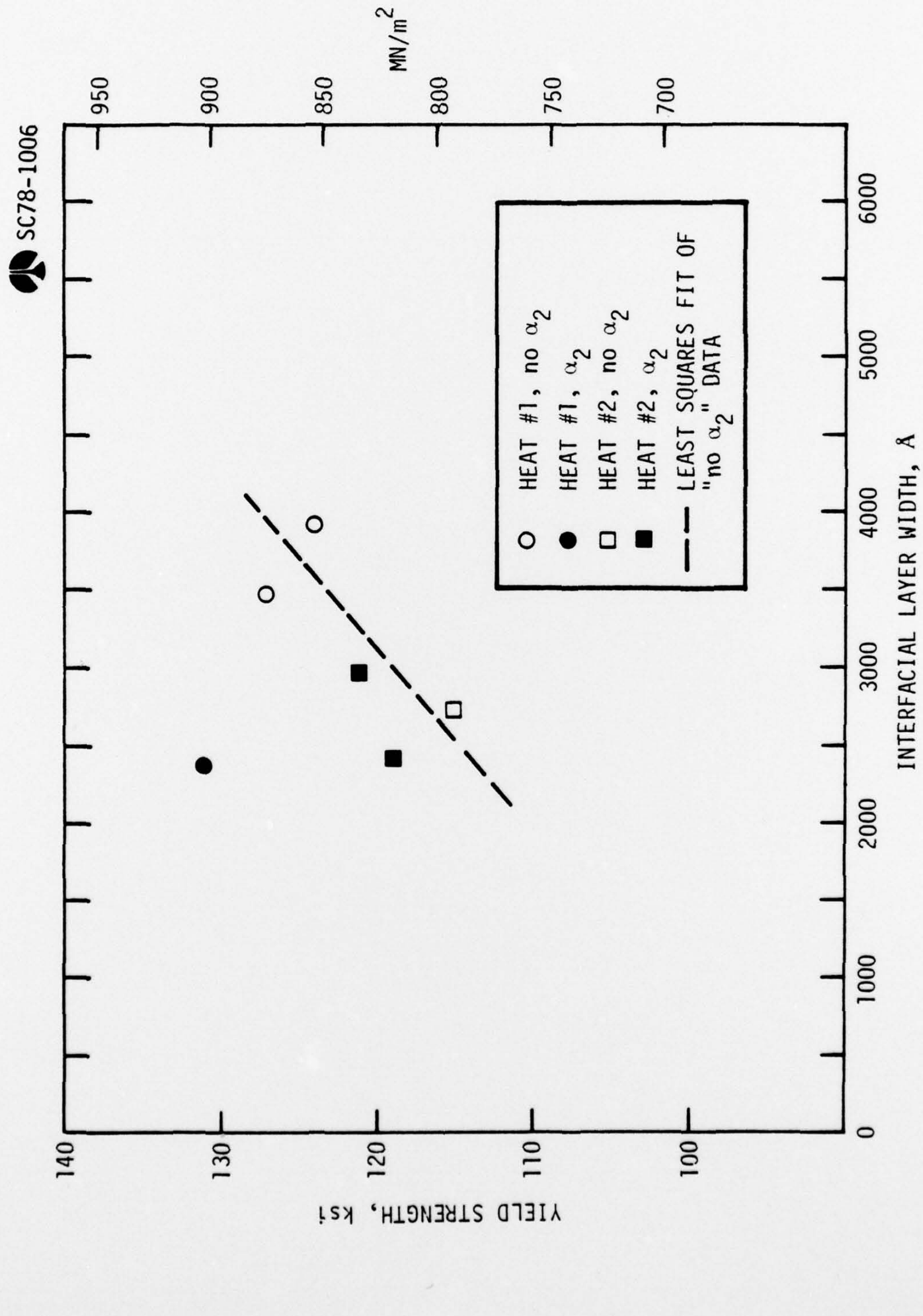

Fig. 7 Tensile yield strength as a function of interface phase width for samples containing 88±2 volume percent primary alpha.

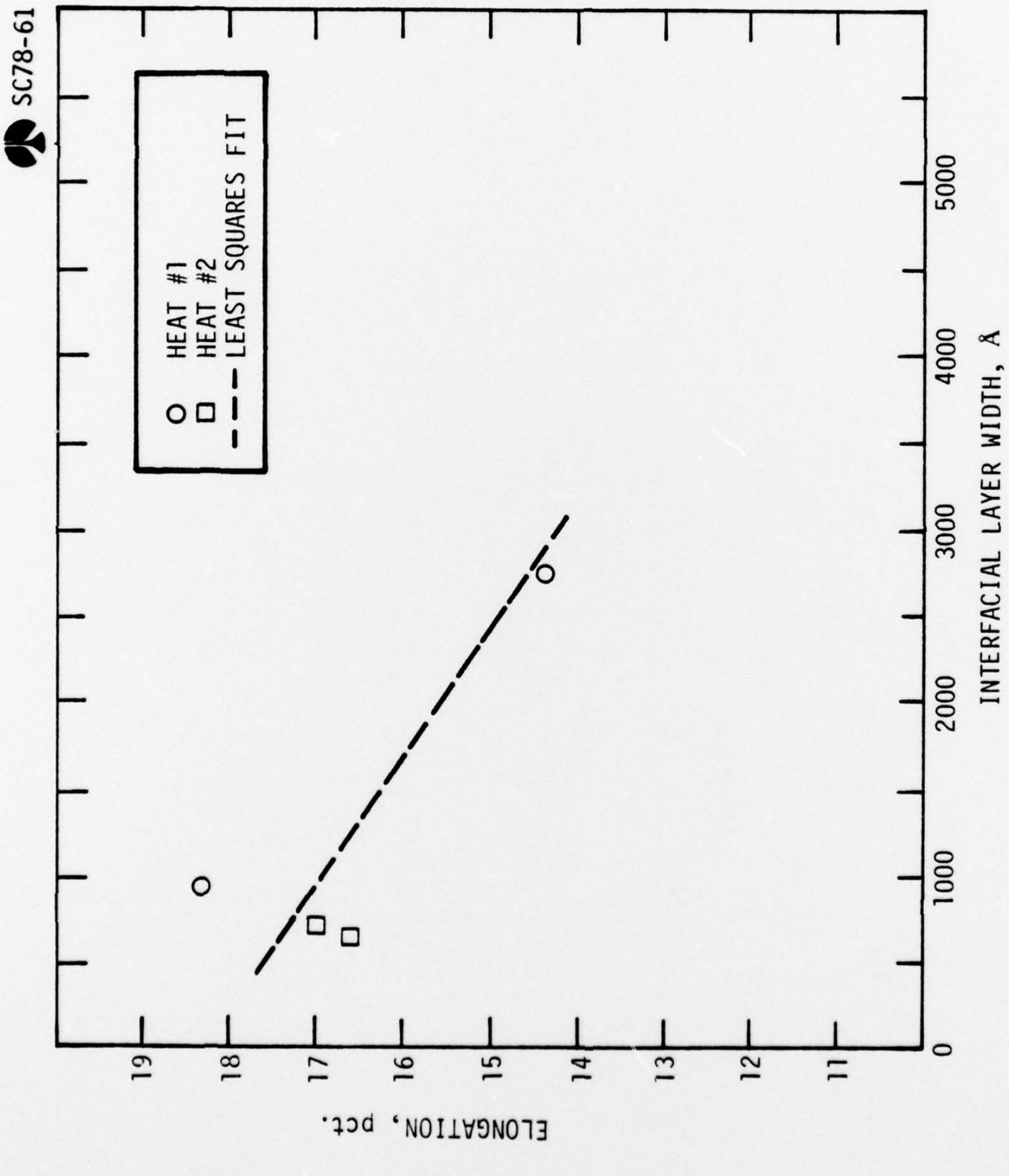

Fig. 3 Tensile elongation as a function of interface phase width for samples containing 80±1 volume percent primary alpha.

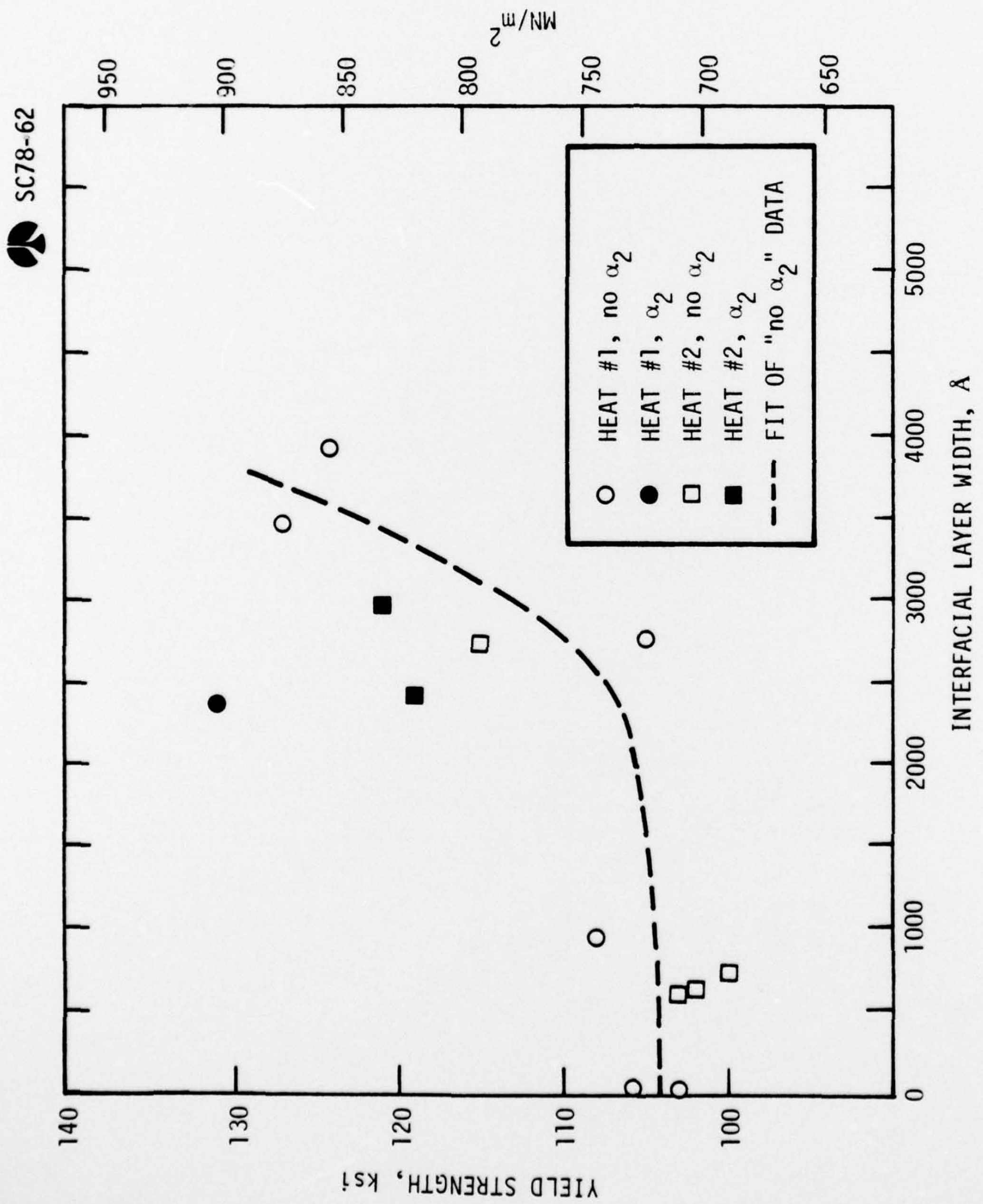

Fig. 9 Tensile yield strength as a function of interface phase width.

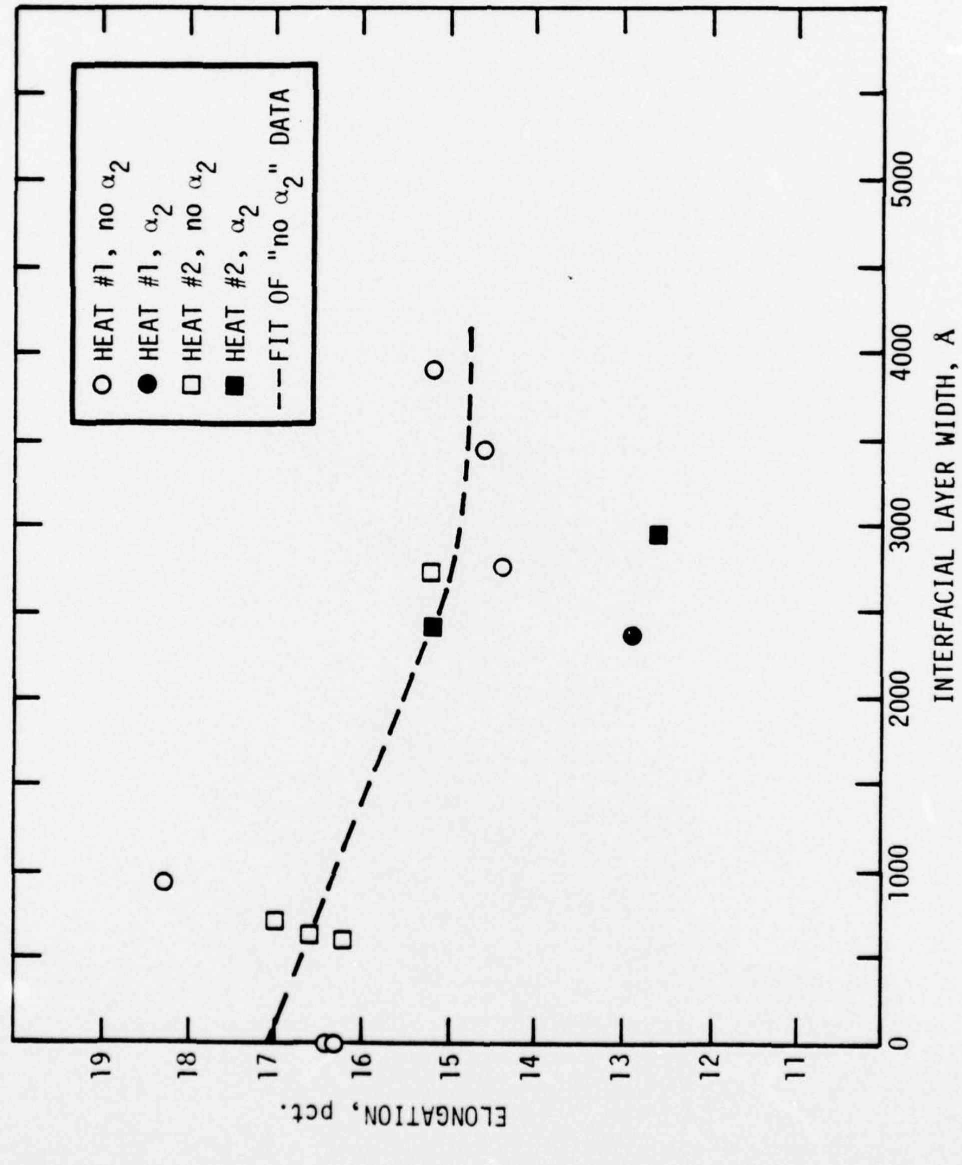

Fig. 10 Tensile elongation as a function of interface phase width.

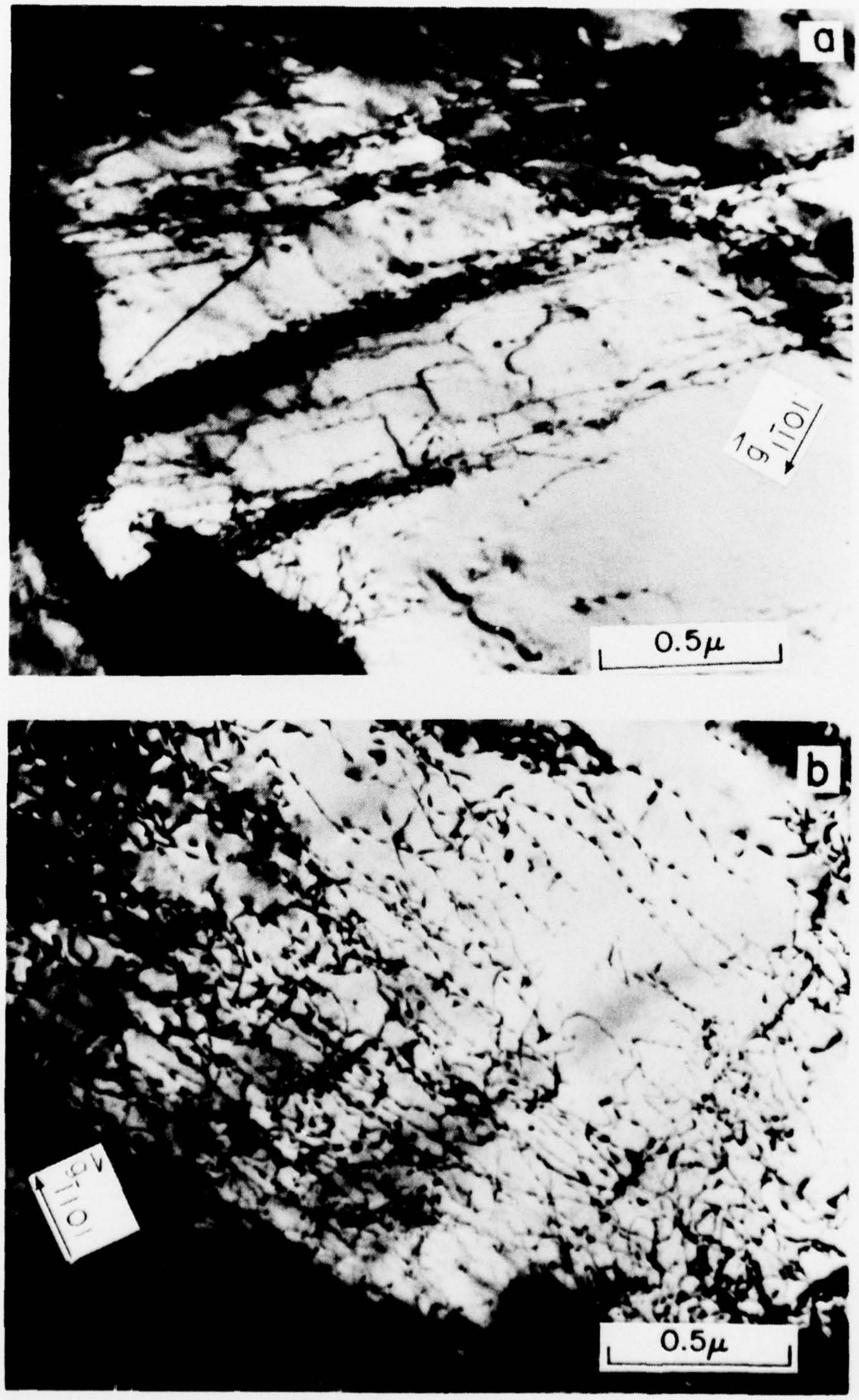

Fig. 11 Thin foil transmission electron micrographs of two Ti-6Al-4V tensile tested specimens. $[11\bar{2}0]$ α zone normal in both micrographs. (a) test specimen #11, σ_{ys} = 115 ksi, ϵ = 15.2%, $\vec{g} = 1\bar{1}01$; (b) test specimen #12, σ_{ys} = 103 ksi, ϵ = 16.2%, $\vec{g} = 1\bar{1}01$.

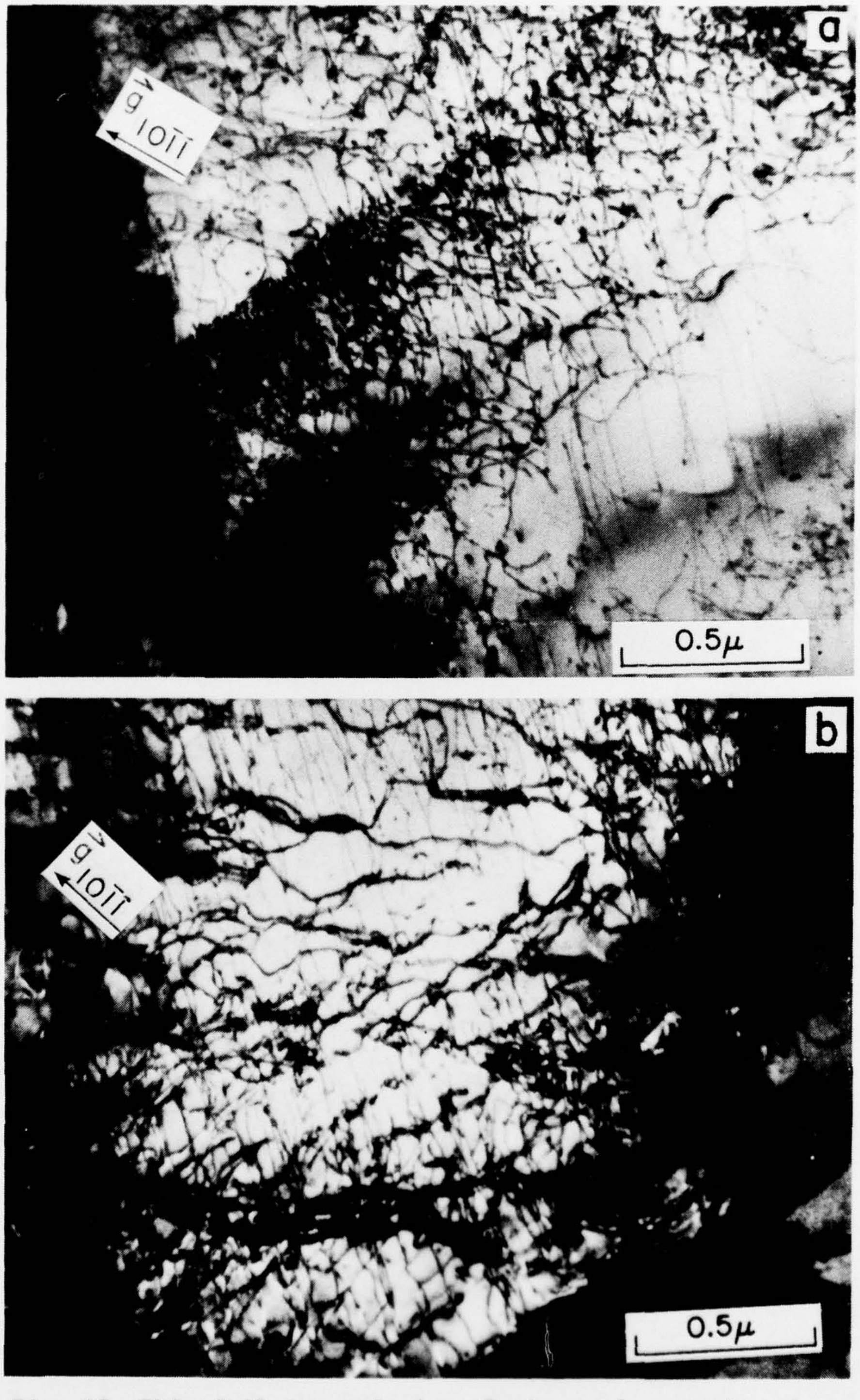

Fig. 12 Thin foil transmission electron micrographs of two Ti-6Al-4V tensile tested specimens, $[11\bar{2}3]$ α zone in both micrographs. (a) test specimen #11, σ_{ys} = 115 ksi, ϵ = 15.2%, $\vec{g} = 10\bar{1}1$; (b) test specimen #12, σ_{ys} = 103 ksi, ϵ = 16.2%, $\vec{g} = 10\bar{1}1$.

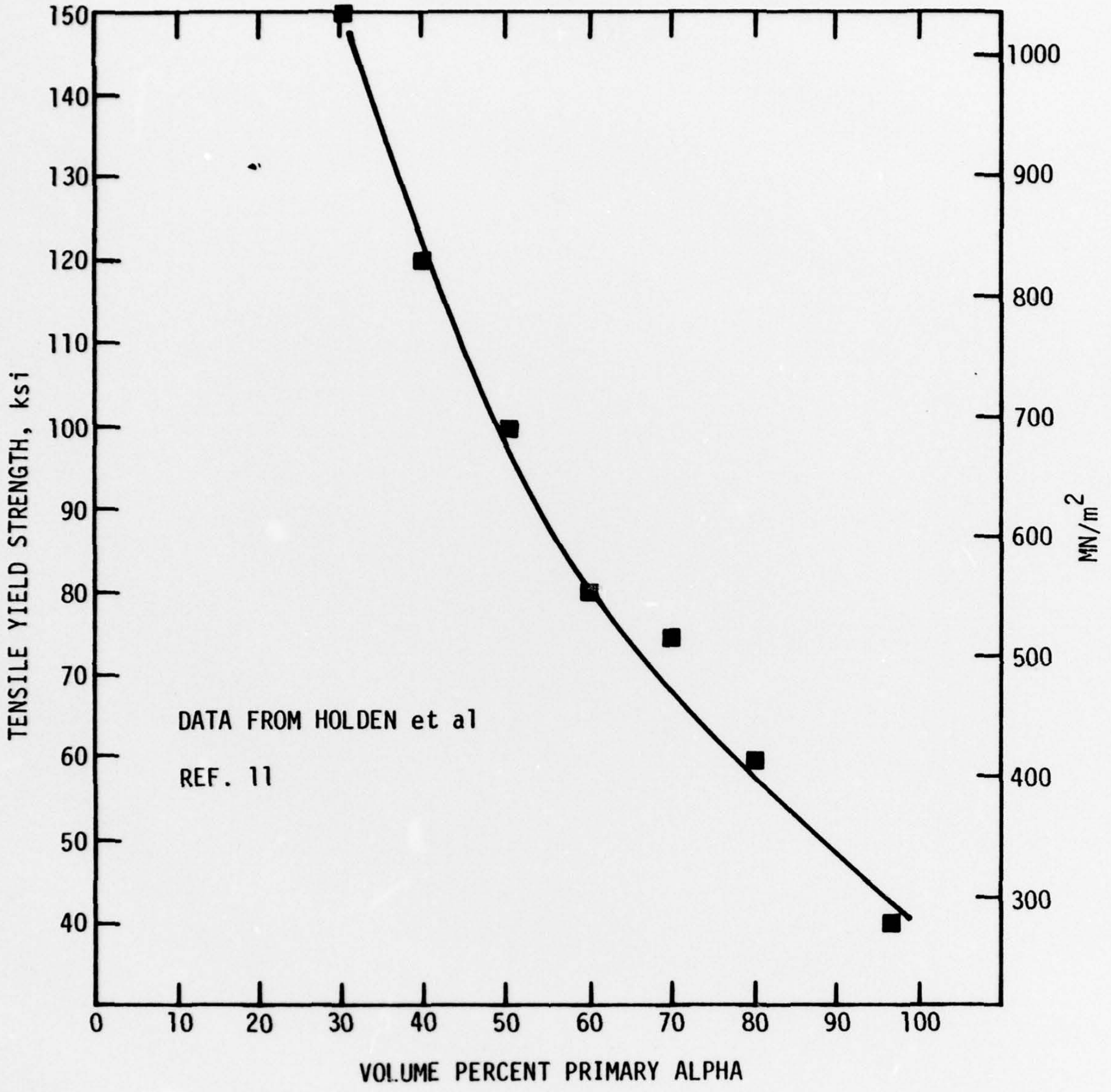

Fig. 13 Tensile yield strength as a function of volume percent primary alpha in Ti-Mn alloys. Data from Holden et al.(11).