LA-UR -92-659

Conf-920604--1

LA-UR--92-659

DE92 008490

TITLE Influence of Strain Rate on the Structure/Property Behavior of the Alpha-2 Alloy $Ti-24.5Al-10.5Nb-1.5Mo$

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SUBMITTED TO 7th World Conference on Titanium
June 28 - July 2, 1992
San Diego, CA

MAR 0 4 1992

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INFLUENCE OF STRAIN RATE ON THE STRUCTURE/PROPERTY

BEHAVIOR OF THE ALPHA-2 ALLOY Ti-24.5Al-10.5Nb-1.5Mo*

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Abstract

While the influence of strain rate and temperature on response of many titanium alloys has been extensively studied, the effect of strain rate on the mechanical and deformation response of titanium aluminides has received limited attention. In this paper, a preliminary study of the effect of strain rate and temperature on the substructure evolution and mechanical response of a (Ti-24.5Al-10.5Nb-1.5Mo) Ti_3Al alloy is presented. The compressive true stress-true strain response of Ti-24.5Al-10.5Nb-1.5Mo was found to depend on both the applied strain rate, varied between 0.001 and 6000 s^{-1}, and the test temperature, varied between 25°C and 700°C. The strain-rate sensitivity of the flow stress of Ti-24.5Al-10.5Nb-1.5Mo at 2% strain is 0.008. The rate of strain hardening in Ti-24.5Al-10.5Nb-1.5Mo is seen to increase slightly with increasing strain rate. The calculated strain-hardening rate for the quasi-static (0.001 s^{-1}) data at 25° C is 3300 MPa / unit strain. The hardening rate at 6000 s^{-1} at 25° C is 4200 MPa / unit strain. Calculation of shear modulus normalized hardening rates yields μ / 165 for the quasi-static rate data and μ / 130 for the high-rate data assuming a shear modulus of 57 GPa for Ti_3Al at room temperature. Increasing temperature at high strain rates (~2000 s^{-1}) is observed to decrease the flow stress at 2% strain from 1110 MPa at 25°C to 1000 MPa at 200°C and to 800 MPa at 700°C. Increasing temperature is seen to have no effect on the rate of strain hardening; the calculated hardening rate at 700° C at 4000 s^{-1} is 4178 units / unit strain. The substructure evolution of Ti-24.5Al-10.5Nb-1.5Mo was observed to depend on the applied strain rate and temperature of deformation. Preliminary dislocation g•b analysis revealed that following room temperature deformation at low strain rate the majority of the dislocations are a-dislocations lying on basal planes, 2nd order pyramidal (a/2 + c) slip on $\{1\bar{2}11\}$, and 1st order pyramidal a-slip on $\{1011\}$. Increasing the rate of deformation at room temperature to 6000 s^{-1} is seen to result in increased a-slip on prism planes and a decreased amount of basal slip. At high-strain-rates and elevated temperatures the substructure was seen to be generally similar to that observed following high-rate deformation at room temperature except for an increased amount of basal slip and a somewhat higher incidence of 2nd order pyramidal slip. The defect generation and the rate sensitivity of Ti-24.5Al-10.5Nb-1.5Mo are discussed as a function of strain rate and temperature and contrasted to that observed in conventional titanium alloys and TiAl.

* Work performed under the auspices of the U.S. Department of Energy

Introduction

The alpha-2 titanium aluminides are receiving increasing attention as potential high temperature structural materials to provide propulsion enhancement in aerospace applications, principally gas turbine engines.[1,2] Although the influence of strain rate on the microstructure / property relationships of pure titanium and a variety of titanium alloys has been extensively studied[4], the effect of strain rate on the stress-strain and deformation response of titanium aluminides has received limited attention[3-6]. Dynamic constitutive behavior is however crucial to understanding the high-speed deformation and impact performance of titanium aluminides under high-rate loading histories such as might be encountered during foreign object damage, high-speed forging, or machining. To date, only two studies have investigated the mechanical response of α_2-based alloys to high-rate deformation[3,5]. Results of strain-rate change tests on Ti-25Al-9Nb and Ti-24Al-11Nb revealed that the strain hardening and strain-rate hardening responses of the two alloys were very similar[3]. The strain-rate hardening exponents were measured to be 0.008 and 0.010 at 27^{0}C for the Ti-25Al-9Nb and Ti-24Al-11Nb, respectively. The rate sensitivity of the Ti-24Al-11Nb alloy was found to be reasonable constant with increasing temperature; an "m" value of 0.007 was measured at 650^{0}C[3].

Tests on the β-phase alloy Ti-11Al-23Nb found a rate sensitivity of 0.020 at room temperature, strain hardening in the β-phase to be significantly lower than in α_2, and a strong dependence of the yield strength on temperature[3]. Based on these α_2 and β-phase results, Gittis and Koss[3] concluded that the macroscopic deformation behavior of Ti-24Al-11Nb is controlled by the α_2 phase when the alloy is in a basketweave microstructural condition. High-rate compression tests conducted on Ti-24Al-11Nb studied with an equiaxed α_2 plus discontinuous β-phase microstructure exhibited somewhat different results[5]. While the rate sensitivity in the equiaxed Ti-24Al-11Nb alloy was also relatively low (m <0.01), the quasi-static work-hardening of the equiaxed alloy was approximately a factor of two higher than when the alloy was heat-treated to a basketweave structure[5]. Under dynamic-loading conditions, strain rate = 10^2 s^{-1}, the equiaxed Ti-24Al-11Nb alloy was also seen to show a pronounced decrease in strain hardening compared to quasi-static loading and the occurrence of localized plastic flow[5]. The purpose of this paper is to report the results of a preliminary study of the effect of strain rate and temperature on the substructure evolution and mechanical response of a Ti_3Al-based alloy; Ti-24.5 Al-10.5 Nb-1.5 Mo.

Experimental Procedures

The material used for this investigation was Ti-24.5Al-10.5Nb-1.5Mo produced by TIMET (composition in at.% of 23.8 Al, 10.6 Nb, 1.43 Mo, 0.19 O, 0.05 Fe and bal. Ti). The Ti-24.5Al-10.5Nb-1.5Mo alloy(referred to hereafter in atomic %) studied was produced by triple vacuum-arc-remelting a 410 kg, 35.5 cm-dia. ingot. The processing was started with upsetting and drawing operations which were conducted above the β-transus temperature Processing below the β-transus was initiated at a diameter of approximately 23 cm and continued until the diameter of the bar had been reduced to 10 cm. The alloy was heat-treated at 1085^{0}C for 1 hour and then in a salt bath at 815^{0}C for 30 minutes followed by aging at 704^{0}C for 8 hours. Compression samples were electro-discharge machined (EDM) from the forged bar section. The in-plane microstructure of the heat-treated Ti-24.5Al-10.5Nb-1.5Mo alloy, as shown in Figure 1a and1b, is seen to consist of equiaxed primary α_2 and fine secondary α_2 plates and ordered B2(β_0) forming a basketweave structure. The mechanical response of Ti-24.5Al-10.5Nb-1.5Mo was measured in compression using solid-cylindrical samples 7.5 mm in dia. by 4.6 mm long. Quasi-static compression tests were conducted at strain rates of 0.001 and 0.1s^{-1}. Dynamic tests, strain rates of 1000-6000 s^{-1}, were conducted as a function of strain rate and temperature utilizing a Split Hopkinson Pressure Bar. The inherent oscillations in the dynamic stress strain curves and the lack of stress equilibrium in the specimens at low strains make the determination of yield inaccurate at high strain rates. Specimens for optical metallography and TEM were sectioned from the heat treated and deformed samples. TEM foils were prepared using conventional jet polishing and observed using a Phillips CM-30.

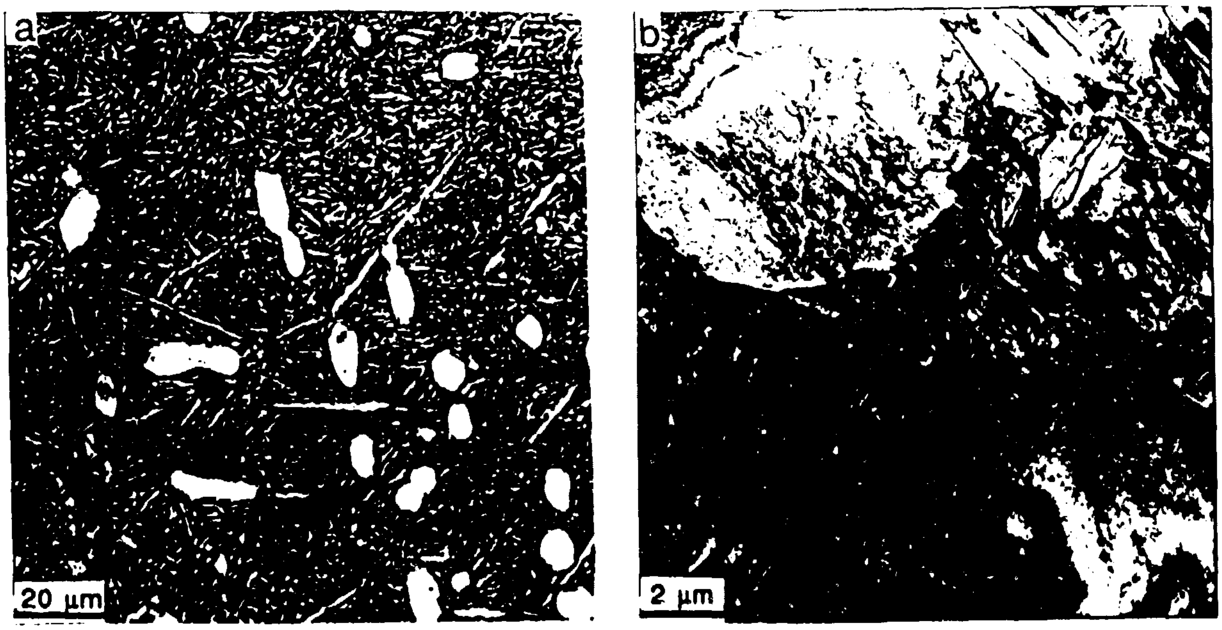

Figure 1 - a) optical micrograph, and b) TEM brightfield of the microstructure of Ti-24.5 Al-10.5 Nb-1.5 Mo showing equiaxed primary α_2 in a matrix of fine secondary α_2 plates and β_0 phase (basketweave).

Results and Discussion

Mechanical Response

The compressive true stress-true strain response of Ti-24.5Al-10.5Nb-1.5Mo was found to depend on both the applied strain rate, which ranged from 0.001 to 6000 s^{-1}, and the test temperature, which was varied between 25 and 700°C (Figures 2 and 3). The flow stress of Ti-24.5Al-10.5Nb-1.5Mo is seen to increase with increasing strain rate. The strain-rate sensitivity, "m" ($= \delta \ln\sigma / \delta \ln\dot{\varepsilon}$), of Ti-24.5Al-10.5Nb-1.5Mo at 2% strain is 0.008. This rate sensitivity is less than one third that measured in Ti-48Al-1V[6] but higher than Ni_3Al[7] which displays no strain-rate sensitivity of the yield stress at 25°C.

The quasi-static mechanical behavior of Ti-24.5Al-10.5Nb-1.5Mo (to true strains of 0.20 in compression) exhibits stress-strain curves with falling work-hardening rates approaching saturation at approximately 1700 MPa (μ/100) similar to Ni_3Al[7] which saturates at a stress level of ~μ/105 but contrary to the almost linear stress-strain sustained hardening response of Ti-48Al-1V[6] to similar stress levels (assuming a polycrystalline shear modulus). Based on this assumption, Ti-24.5Al-10.5Nb-1.5Mo , Ti-48Al-1V, and Ni_3Al all harden to stress levels roughly a factor of three higher than pure metals such as copper and tantalum which saturate at stress levels of ~μ/333 and μ/450 at low strain rates. Given the known anisotropy in the intermetallics mentioned it is likely that a portion of the ~3x factor in the saturation stress levels is due to anisotropy considerations, such as a higher Taylor factor in these low symmetry systems. The majority of the sustained hardening, however, is believed to result from a lack of dynamic recovery. The sustained plasticity exhibited by this alloy in compression is similar to the crack-free deformation in Ti-48Al-1V tested in compression to 20% strain[6]. These results clearly suggest the limited tensile ductility exhibited in some Ti-aluminides is due to the intervention of cleavage or intergranular fracture at stresses lower than that required to activate slip rather than solely an inherent lack of sufficient slip systems.

In disordered metals the decrease in work hardening at high stresses and strains is generally linked to dynamic recovery. This is usually associated with the loss of dislocations at cell walls, tangles, etc. It is worth noting in the case of intermetallics, however, that the absolute stress levels are unusually high, i.e., ~ $\mu/100$ compared with ~$\mu/300$, suggesting that: 1) dynamic recovery mechanisms in intermetallics are physically different than those in disordered metals, and 2) it is possible that other mechanisms are limiting (saturating) the flow stress such as mutual annihilation by mechanical recombination of dislocations. These results point demonstrate that contrary to disordered metals Stage II work-hardening in some ordered systems continues to very high stress levels. The details of this process in ordered intermetallics will most probably depend on the local dislocation configuration. Independent of the controlling mechanisms these results clearly suggest that radically different dynamic recovery processes appear to be operative between α_2 and γ-based titanium alumindes.

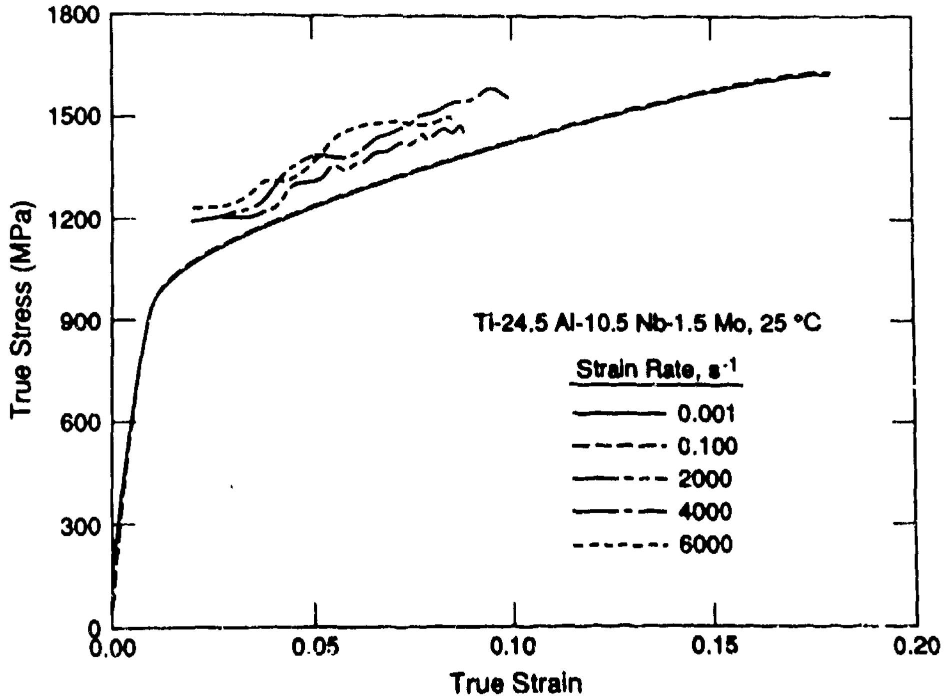

Figure 2 - Compressive Stress-Strain response of Ti-24.5Al-10.5Nb-1.5Mo as a function of the applied strain rate at 25° C.

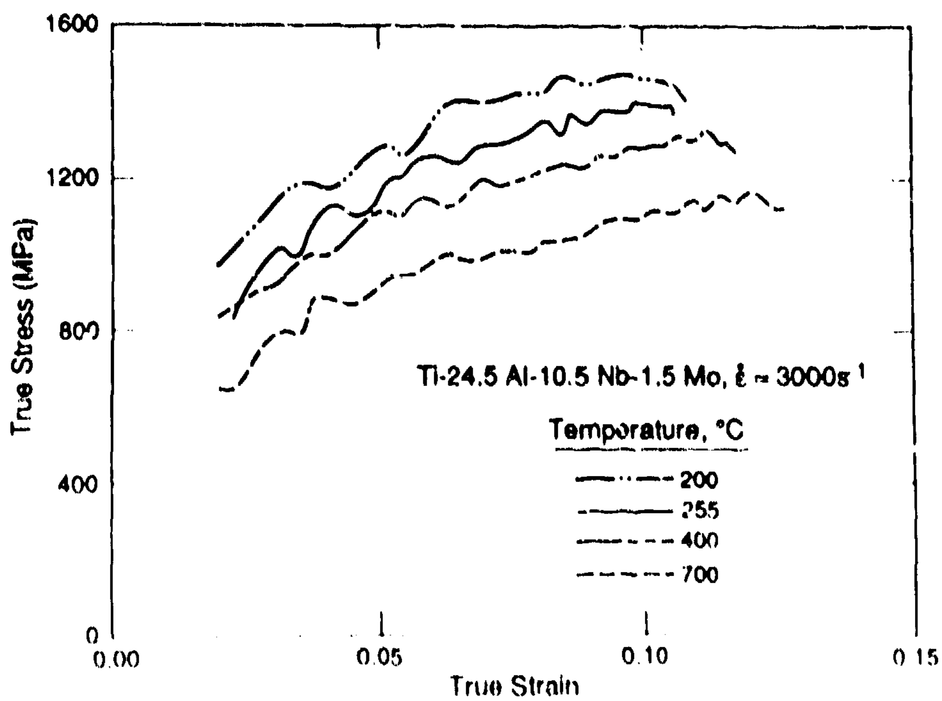

Figure 3 - Compressive stress strain response of Ti-24.5Al-10.5Nb-1.5Mo as a function of test temperature at a strain rate of ~3000 s⁻¹.

The rate of strain hardening in Ti-24.5Al-10.5Nb-1.5Mo is seen to increase slightly with increasing strain rate. The calculated average strain-hardening rate, θ , for the quasi-static (0.001 s^{-1}) data at 25^0 C is 3300 MPa / unit strain. The hardening rate at 6000 s^{-1} at 25^0 C is 4200 MPa / unit strain. Normalizing the work hardening rates with the Taylor Factor for a random polycrystal, [θ / $(3.07)^2$], yields work hardening rates of μ / 165 for the quasi-static rate data and μ / 130 for the high-rate data assuming a shear modulus of 57 GPa for Ti_3Al at room temperature[8]. In comparison to a strain hardening rate of μ/175 for Ti-48Al-1V at 4500 s^{-1} at 25^0C[6], the shear modulus normalized strain hardening rate of Ti-24.5Al-10.5Nb-1.5Mo is seen to exhibit sustained work-hardening at high-strain-rate similar to Ti-48Al-1V rather than that exhibited by conventional Ti-alloys such as Ti-6Al-4V as discussed below. While the observation of stable work hardening in Ti-24.5Al-10.5Nb-1.5Mo is in contrast to previous high-rate data on Ti-24Al-11Nb[5], the present results are consistent with the fine-scaled crack-free basketweave microstructure and absence of orthorhombic phase in this study.

The rate sensitivity and strain hardening rates in Ti-24.5Al-10.5Nb-1.5Mo exhibit somewhat different trends when compared to those seen in Ti-48Al-1V [6] or a conventional disordered Ti-alloy like Ti-6Al-4V as a function of strain rate. While the rate sensitivities of the yield and flow stresses of Ti-6Al-4V in both tension and compression are also high, the rate sensitivity "m" is 0.015[4]. This value is roughly twice that of Ti-24.5Al-10.5Nb-1.5Mo while the yield stress levels are all in the range of 1000 MPa quasi-statically for Ti-6Al-4V compared to approximately 950 MPa for Ti-24.5Al-10.5Nb-1.5Mo in this study. The strain hardening of Ti-6Al-4V as a function of strain rate, using a shear modulus of 44.5 GPa at 25^0C[8] is μ / 146 at a strain rate of 0.0001 s^{-1} and μ / 436 at a strain rate of 3000 s^{-1}. The normalized quasi-static strain-hardening rate for Ti-6Al-4V is similar to the high and low-rate normalized hardening rates for Ti-24.5Al-10.5Nb-1.5Mo. Both of the normalized rates, i.e. ~ μ/120 to μ/200, are consistent with Stage II-type hardening behavior such as that seen in disordered FCC polycrystals. The drastic decrease in the strain hardening in Ti-6Al-4V at high strain rate reflects a significant influence of adiabatic heating[4]. In contrast the Ti-24.5Al-10.5Nb-1.5Mo alloy displays no temperature-induced decrement in the strain hardening but rather an increased hardening rate at high strain rates. Increasing temperature at high strain rates is observed to decrease the flow stress of Ti-24.5Al-10.5Nb-1.5Mo (Figure 3). Increasing temperature is seen to have only a small effect on the strain hardening rate; the calculated hardening rate at 700^0 C at 4000 s^{-1} is 4178 units / unit strain. This observation is similar to the lack of a temperature-sensitivity on strain hardening seen in Ti-24Al-11Nb[3] up to 650^0C. The overall decrease in flow stress with increasing temperature, although at high-strain-rate, is also consistent with several previous studies[2,3,9] on the temperature dependence of the yield stress for Ti_3Al.

Substructure Evolution

The fine scale of the as-heat-treated basketweave microstructure in the current study is consistent with the previous observation that Mo or V additions appear to promote finer-scale microstructures through their influence on retarding the $\beta - \alpha_2$ transformation[9]. The substructure evolution of Ti-24.5Al-10.5Nb-1.5Mo was observed to depend on the applied strain rate and temperature of deformation. The substructure of Ti-24.5Al-10.5Nb-1.5Mo deformed to ε=0.10 at 0.001 s^{-1} at 25^0C was seen to consist of more long straight screw dislocations.(Figure 4) The observation of straight screws is consistent with dislocation storage observations in BCC metals, such as tantalum, and suggests a greater degree of edge mobility relative to screw dislocation motion and reduced cross-slip of screw dislocations. Preliminary dislocation g•b analysis revealed that following room temperature deformation at low strain rate the majority of the dislocations are a-dislocations lying on basal planes and 1st order pyramidal slip on $\{10\bar{1}1\}$, and 2nd order pyramidal (a/2 + c) slip on $\{1\bar{2}11\}$. Increasing the rate of deformation at room temperature to 6000 s^{-1} is seen to result in increased a-slip on prism planes, and 1st order pyramidal planes, and a lessor amount of basal slip. In Figure 4a, dislocations are seen to lie on $(1\bar{1}01)$ and (0001) traces at low strain rates whereas they lie on $(\bar{1}010)$ and $(110\bar{1})$ plane traces following high strain rate deformation (Figure 4b).

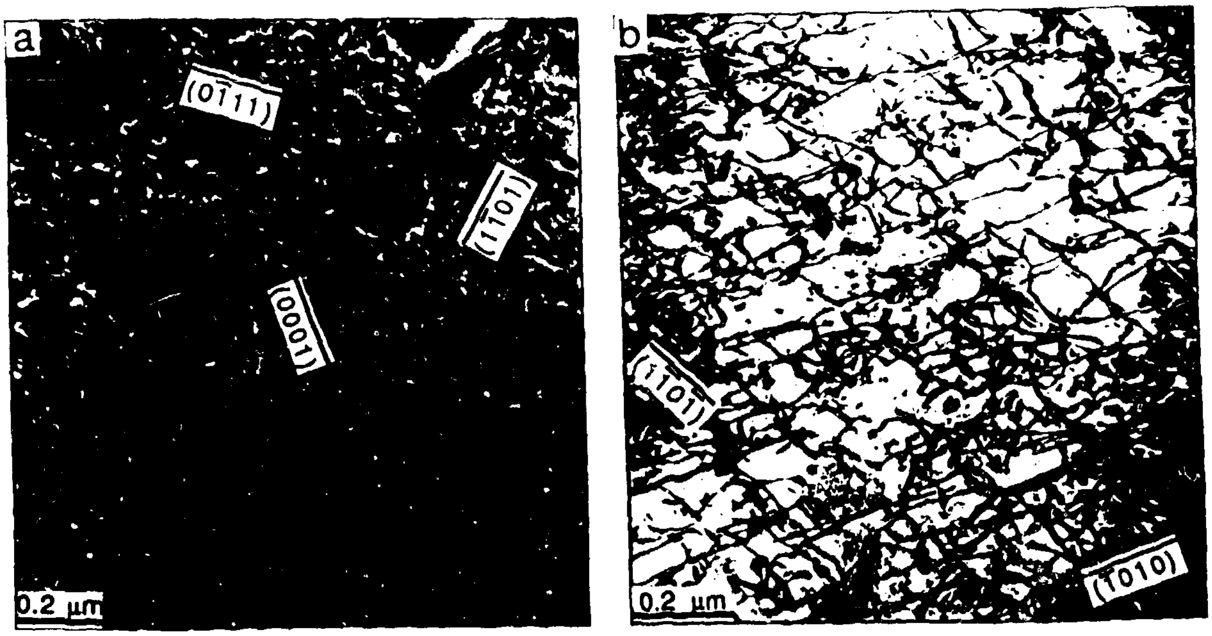

Figure 4 - Substructure of Ti-24.5Al-10.5Nb-1.5Mo following deformation at 25° C to $\varepsilon = 0.10$, a) 0.001 s^{-1} showing basal slip and 1st order pyramidal slip, [1$\bar{2}$10] zone axis, and b) 6000 s^{-1} showing prism slip and 1st order pyramidal slip, [1$\bar{2}$16] zone axis.

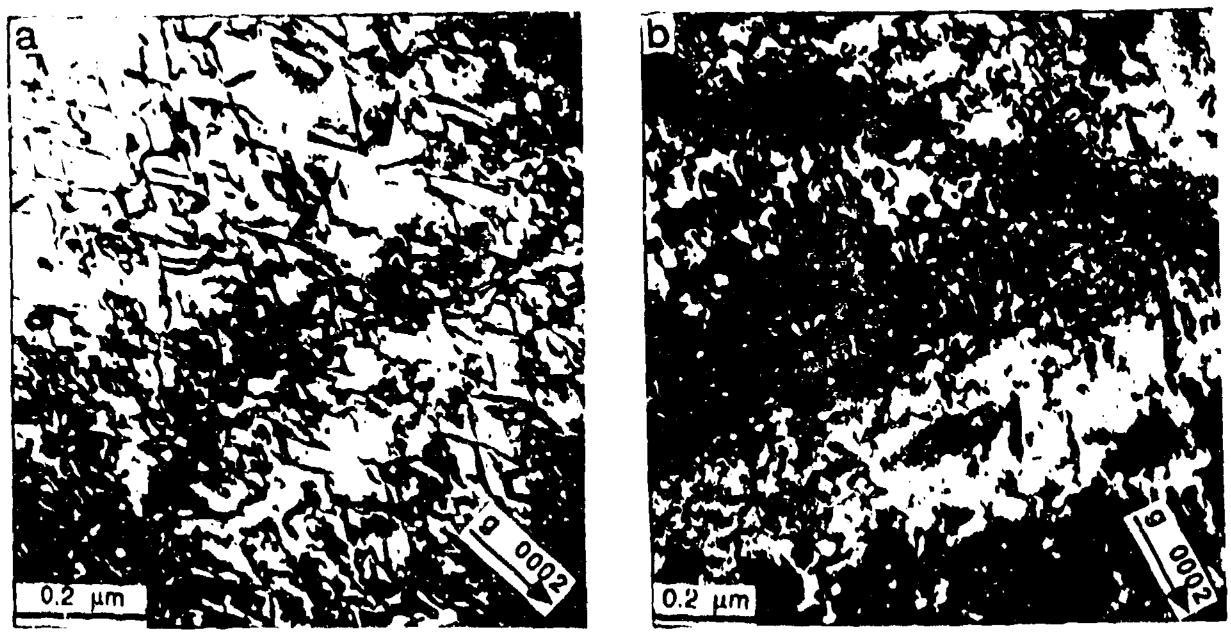

Figure 5 - Brightfield micrograph showing the presence of **a/2 + c** dislocations: a) 25°C, 6000 s^{-1} and b) 700°C, 3000 s^{-1}.

In Ti-24.5Al-10.5Nb-1.5Mo deformed simultaneously at high-strain rates and elevated temperatures the substructure was seen to be generally similar to that observed following high-rate deformation at room temperature except for an increased amount of basal slip and a somewhat higher incidence of 2nd order pyramidal slip. As shown in Figures 5a and 5b, a/2 + c dislocations were observed following high rate deformation at both 25 and 700°C. Figures 6a and 6b show the substructure of Ti-24.5Al-10.5Nb-1.5Mo following deformation at 700° C to $\varepsilon = 0.10$ at a strain rate of 3000 s^{-1}. In Figure 6a, the trace of (1011) planes is clearly seen; labeled "A". The slip traces labeled "B", are parallel to the traces of (1211) or

(0001), however the sharpness of the slip traces suggest that they are (0001) planes since the basal planes are parallel to the electron beam direction. The groups of dislocations labeled "C" are shown to lie along the traces of $(11\bar{2}1)$ or $(11\bar{2}\bar{1})$ planes. At elevated temperatures at high rate, particularly above 400^0 C, evidence of an increased amount of shearing of the β_0 laths was also seen, suggesting decreased strength of the β_0 phase at elevated temperatures. This observation is consistent with previous observations[3] that above 650^{0}C the yield strength of β_0 decreases rapidly. Finally, regardless of the temperature or strain rate of deformation, no deformation twins were seen in Ti-24.5Al-10.5Nb-1.5Mo contrary to previous observations of microtwinning at elevated temperatures[1,2]. This observation is in total contrast to the pronounced propensity for deformation twinning in Ti-48Al-1V[6] which is increases with increasing temperature and/or strain rate. Detailed dislocation analysis of the substructure evolution in Ti-24.5Al-10.5Nb-1.5Mo will be presented elsewhere.

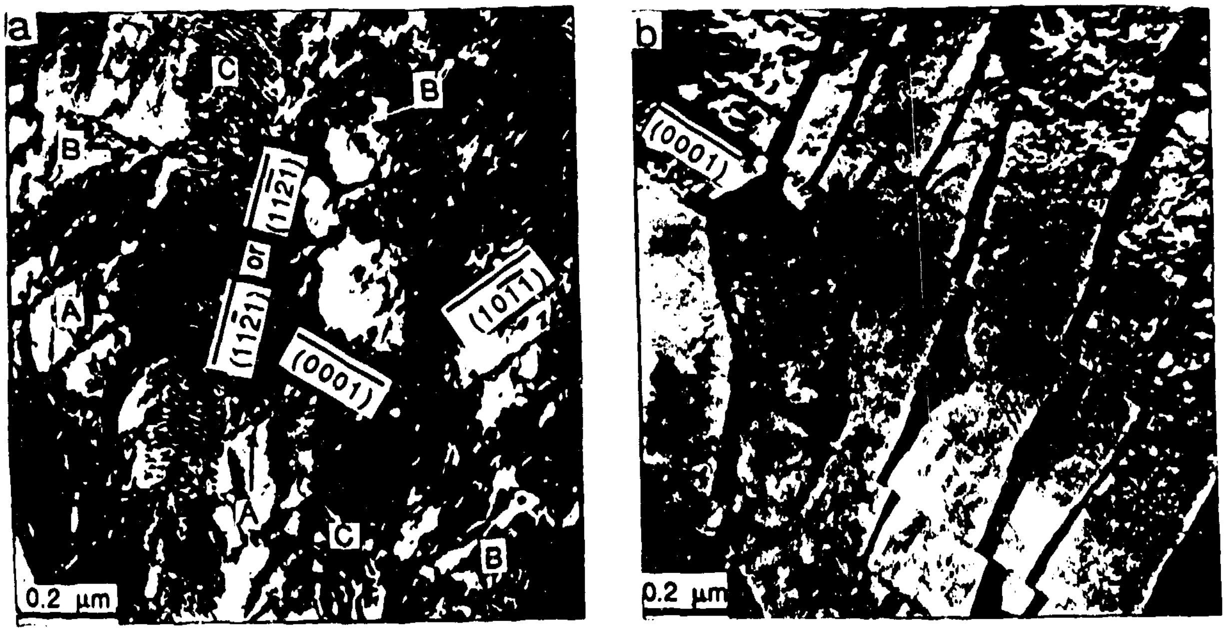

Figure 6 -Brightfield micrograph of dislocation debris in Ti-24.5Al-10.5Nb-1.5Mo following deformation to ε=0.10 at 3000 s^{-1} at 700^{0}C showing: a) pyramidal and basal slip, [1210] zone axis and b) increased incidence of shearing of β_0 lath at high temperature.

Summary and Conclusions

Based on a study of the influence of strain rate and temperature on the mechanical response and substructure evolution of Ti-24.5Al-10.5Nb-1.5Mo, the following conclusions can be drawn:

1. The compressive true stress-true strain response of Ti-24.5Al-10.5Nb-1.5Mo was found to depend on both the applied strain rate and the test temperature. The strain-rate sensitivity, m, of Ti-24.5Al-10.5Nb-1.5Mo at 2% strain was calculated to be 0.008.

2. The rate of strain hardening in Ti-24.5Al-10.5Nb-1.5Mo is seen to increase slightly with increasing strain rate. Stage II work-hardening behavior in this ordered alloy is shown to continue to very high stress levels, ~μ/100, contrary to disordered metals. Increasing temperature is seen to have no effect on the rate of strain hardening; the calculated hardening rate at 700^0 C at 4000 s^{-1} is 4178 units / unit strain.

3. The substructure evolution of Ti 24.5Al 10.5Nb 1.5Mo was observed to depend on the strain rate and temperature of deformation. Following room temperature deformation at low strain rate the majority of the dislocations are a dislocations lying on basal planes and 1st order pyramidal planes, and 2nd order pyramidal (a/2 + c) slip on {1211}. Increasing the rate of

deformation at room temperature to 6000 s^{-1} is seen to result in increased **a**-slip on prism planes and a lessor amount of basal slip. At high-strain-rates and elevated temperatures the substructure was seen to be similar to that following high-rate deformation at room temperature except for an increased amount of basal slip and a somewhat higher incidence of 2nd order pyramidal slip. At elevated temperatures at high rate evidence of an increased amount of shearing of the β_0 laths was also seen, suggesting decreased strength of the β_0 phase.

Acknowledgments

The authors acknowledge the assistance of M.F. Lopez for conducting the quasi-static compression tests and W. Wright for conducting the Hopkinson-Bar tests. The authors would like to acknowledge stimulating discussions with J.D. Embury, A.D. Rollett, and M.F. Ashby. This work was performed under the auspices of the U.S. Department of Energy.

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