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The Variation of the Yield Stress of Ti Alloys with Strain Rate at High Temperatures

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This study extended investigation on the elevated-temperature yield-strength dependence of beta-phase titanium alloys on strain rate and temperature. Yield stresses were found to increase substantially with increasing strain rate at elevated temperatures due to the high strain-rate sensitivity of titanium at high temperatures. Above 1000 °C, the strain-rate sensitivities were found to increase substantially with increasing temperature and/or decreasing strain rate. The six alloys examined were TIMETAL 21S, Ti-15-3, Ti-6-4, Ti-13-11-3, Beta C, and Beta III. There was particular interest in determining the strain-rate sensitivity of these alloys through strain-rate change tests above 1000 °C. The yield stresses of all the titanium alloys at temperatures above 1093 °C were less than 1% of their ambient temperature values. Strain hardening was negligible in the alloys tested at these high temperatures. Extended tensile ductilities of 100 to 200% were observed due to the pronounced strain-rate sensitivity. The rate controlling mechanism for plasticity, based on activation energy and the strain-rate sensitivity measurements, is discussed.

Keywords high-temperature strength, strain-rate sensitivity, titanium alloys

1. Introduction

The purpose of this study was to extend the existing literature of the elevated-temperature yield-strength dependence of beta-phase titanium alloys on strain-rate ($\dot{\varepsilon}$) and temperature. The six alloys examined were TIMETAL 21S, Ti-15-3, Ti-6-4, Ti-13-11-3, Beta C, and Beta III. Determining the strain-rate sensitivity of these alloys through strain-rate change tests above 1000 °C was emphasized. This helped allow the development of constitutive equations for these alloys.

2. Experimental Procedure

The compositions of TIMETAL 21S, Ti-15-3, Ti-13-11-3, Beta C, Beta III, and Ti-6-4 are listed in Table 1. The first five alloys are beta stabilized at ambient temperature, exhibiting high strengths and good cold formability. Ti-6-4 has a two-phase alpha + beta structure at ambient temperature but is completely beta phase at the testing temperatures above 1000 °C (Ref 1). Tensile specimens were machined from sheet supplied from United Defense of Fridley, MN (TIMETAL 21S); TIMET of Denver, CO (Ti-15-3); Crucible Materials of Pittsburgh, PA (Ti-13-11-3 and Beta III); RMI Titanium of Niles, OH (Beta C); and Metals Unlimited, Inc. of Deer Park, NY (Ti-6-4). The heat treatments are described in Table 2.

The test specimens had gage dimensions of 10.2 mm length, 4.8 mm width, and 1.52 mm thickness. Machining tolerances of the finished specimen were ±25 μm in all dimensions. Specimens were cut in the plane of the sheet and surface finished on a milling machine. The specimens were subsequently placed in an evacuated quartz chamber that was purged with argon prior to heat treatment. All of the specimens were solution treated and aged for maximum ambient-temperature tensile strength as suggested (Ref 1, 2).

Tensile tests were performed on an Instron 8521 servohydraulic biaxial testing machine with computerized data acquisition. Instron universal joints were used on each side of the TZM (0.5%Ti, 0.8%Zr, bal Mo) grips to eliminate any bending moments applied to the specimen from the loading fixtures. A ten-filament (Research Inc., Minneapolis, MN) radiant heat furnace was utilized to rapidly heat the specimens. Each of the

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Table 1 Composition of the alloy specimens tested

	Composition, wt%													
Alloy	**Nb**	**H**	**C**	**Al**	**Sn**	**Zr**	**Mo**	**Si**	**N**	**Cr**	**V**	**Ni**	**Fe**	**O**
TIMETAL 21s(a)	2.4-3.0	0-0.015	0-0.05	2.5-3.5	…	…	14-16	0.15-0.25	0-0.5	…	…	…	0.2-0.4	0.11-0.15
Ti-15-3(a)	…	0.015	…	2.5-3.5	2.5-3.5	…	…	…	…	0.05	2.5-3.5	14-16	…	0.25
Ti-6-4	…	0.005	0.02	6.1	…	…	…	…	0.019	…	4.0	…	0.16	0.121
Ti-13-11-3	…	0.032	0.02	3.22	…	…	…	…	0.018	11.12	13.68	…	0.2	0.122
Beta C(a)	…	0.02	0.05	3-4	…	3.5-4.5	3.5-4.5	…	0.03	5.5-6.5	7.5-8.5	…	0.03	0.014
Beta III	…	0.008	0.18	…	4.48	5.4	11.7	…	…	…	…	…	0.04	0.126

(a) Alloy compositions given are typical ones per Ref 1.

titanium alloys was tested at 20, 1093, and 1316 °C. The (elevated) test temperatures of the specimens were maintained to within ±10 °C of the set temperature at the yield stress and ultimate tensile strength (UTS) and to within ±15 °C at the termination of the test (typically, after about 30% elongation). Only 3.0 to 3.5 min were required to heat the specimens to test temperature for tests performed at 1093 and 1316 °C. This rapid heating was intended to minimize thermal effects on the microstructure. Temperatures were measured on the specimen surface using Pt/Pt-13Rh type thermocouples. The temperature gradient from the surface to the center of the specimen was calculated to be less than 1 °C, primarily due to the small thickness of the specimens. Temperatures were relatively uniform along the gage length (within 5 °C).

Strain-rate change tests were used to determine the strain-rate sensitivity of the alloys at each test temperature. The strain rate sensitivity:

$$m = \partial[\ln(\sigma)]/\partial[\ln(\dot{\varepsilon})]|_{s,T} \qquad \text{(Eq 1)}$$

was determined at an approximately constant structure (i.e., a particular microstructure, s) by measuring the change in the yield stress, σ, with an "instantaneous" change in the applied strain rate, $\dot{\varepsilon}$. The changes in strain rate were from 10^{-4}/s to 10^{-2}/s at 20 °C, and 10^{-3}/s to 10^{-2}/s at 1093 and 1316 °C. It will be shown that there is a relatively small amount of hardening (actually, a few percent softening, typically) at the elevated temperatures tested so that m is approximately the inverse of the steady-state strain-rate sensitivity exponent:

$$n = (\partial \ln \dot{\varepsilon}/\partial \ln \sigma) \qquad \text{(Eq 2)}$$

An offset of 0.006 was used to determine the onset of subsequent plastic flow at ambient temperature, and a 0.0006 strain offset was used at the elevated temperatures. High purity (grade 5) argon was used to purge a quartz chamber surrounding the titanium specimens during the high-temperature tests to ensure that the tests would not be influenced by high-temperature oxide embrittlement of the titanium alloys (Ref 3). Stress-

Table 2 Heat treatments of the tensile specimens

	Solution anneal		Age	
Specimen	Temperature, °C	Time, min	Temperature, °C	Time, h
TIMETAL 21S	815	15	480	20
Ti-15-3	815	15	510	14
Ti-6-4	815	15	540	4
Ti-13-11-3	815	15	480	72
Beta C	815	15	480	16
Beta III	815	15	480	8

Table 3 Ambient temperature tensile properties, 20 °C, 10^{-4}/s

	Yield strength, MPa	Ultimate tensile strength, MPa	TFS (a), MPa	Reduction in area, %	Elongation, %	Strain-rate sensitivity factor, *m*
TIMETAL 21S	1406	1420	1544	9.2	8.3	0.008
Ti-6-4	1007	1055	1227	14.6	13.5	0.011
Ti-15-3	1193	1269	1454	19.9	13.8	0.008
Ti-13-11-3	955	972	1013	4.4	4.0	0.009
Beta C	1027	1103	1275	14.4	11.6	0.008
Beta III	1213	…	…	…	…	0.016

Figure 1 values for 10^{-3}/s. (a)TFS, true fracture stress (σ_f)

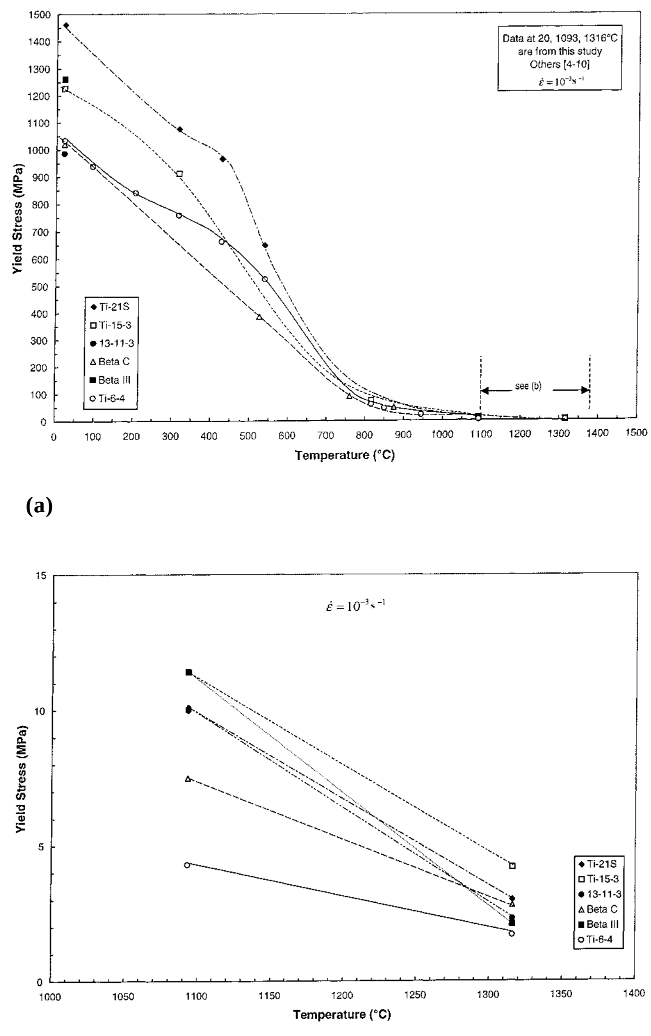
Table 4 Elevated temperatue strain-rate sensitivity, activation energy, ultimate yield, and ultimate strength values

	Testing temperature						
	1093 °C			1316 °C			
Alloy	Strain rate sensitivity factor (*m*)	YS(a), MPa	UTS(b), MPa	Strain rate sensitivity factor (*m*)	Activation energy (Q), kJ/mol	YS(a), MPa	UTS(b), MPa
TIMETAL 21S	0.34	10.1	27.3	0.44	258	3.0	13.0
Ti-6-4	0.38	4.3	10.3	0.44	198	1.7	4.7
Ti-15-3	0.39	11.4	24.3	0.60	170	4.2	10.9
Ti-13-11-3	0.41	10.0	27.6	0.33	330	2.3	10.2
Beta C	0.35	7.5	20.9	0.36	229	2.8	9.4
Beta III	0.32	11.4	29.4	0.38	399	2.1	14.2

(a) YS, yield strength 10^{-3}/s. (b) UTS, ultimate tensile strength 10^{-2}/s.

strain relationships were determined by subtracting the Instron machine compliance from the observed load elongation, based on crosshead displacement. Yield stresses and UTS were reported as engineering values based on initial cross-section area, while the true fracture stress, σ_f, was based on the final area. Yield stresses were measured using a plastic strain offset of 0.02. The total elongation and reduction in area were measured directly from the specimen at the conclusion of the tensile tests. It was found that the final cross-section (after ≅30% total strain) was typically 3% larger, based on the load-displacement curve. This might be due to slight deformation with the grips.

Optical metallography was performed on two of the titanium alloys (Ti-15-3 and Ti-6-4) to determine the effects of the high temperature on the grain size and other microstructure. The average grain size for the Ti-15-3 alloy was measured after 1, 2, 5, and 10 min at 1093 °C. The Ti-6-4 alloy specimens were examined after heat treating and in the grip section after the tensile tests at 1093 and 1316 °C (approximately 3.5 min). The initial average grain size of the Ti-15-3 alloy was about 50 μm, and the average grain size increased to about 130 μm after 1 min at 1093 °C. The grain size had stabilized at about 240 μm after 5 min at 1093 °C. The initial grain size of the heat treated Ti-6-4 alloy was small, at approximately 10 μm, and grain size increased to about 260 μm after 3.5 min at 1093 °C and to about 320 μm after 3.5 min at 1316 °C.



(a)

(b)

Fig. 1 Yield stress variation of titanium alloys (a) with temperature at a strain rate of 10^{-3}/s (Ti-6-4 has $T_\beta \cong 960$ °C) and (b) at very high temperatures

3. Results and Discussion

Figure 1(a, b) illustrates the results of titanium alloy tensile tests at 20, 1093, and 1316 °C from this study and from other investigations at various strain rates and temperatures (Ref 4-11). These data are also available in Table 3 and 4 as well as in Ref 12 by the authors. Figure 1 shows the variation in yield stress of the titanium alloys with temperature at a strain rate, $\dot{\varepsilon}$, of 10^{-3}/s. It can be observed in Fig. 1(a) that the yield stress decreases dramatically with increasing temperature up to 538 °C for all six alloys, and the values at this temperature are about half those at ambient temperature. The yield stress decreases to less than 10% of that at ambient temperature values above 760 °C. TIMETAL 21S exhibits the highest yield stress (1406 MPa at ambient temperature) of the alloys tested at temperatures from ambient to 538 °C. Ti-6-4 generally exhibits a relatively low yield stress (1007 MPa at ambient temperature) at all test temperatures compared to the other (beta and/or near beta alloys at ambient temperature) alloys. The yield stress of Ti-6-4 is less than 11 MPa at 1093 °C and less than 5 MPa at 1316 °C. The strength of the titanium alloys is principally provided by solid solution hardening.

Figure 2 shows the effect of strain rate on the yield stress of Ti-6-4 alloy based on data over a range of 5 orders of magnitude in strain rate at a variety of temperatures. Increasing strain rate results in increasing yield (or flow) stress. Strain-rate sensitivity appears to change with strain rate and with temperature. Paton and Hamilton concluded that some of the data of Fig. 2 reflect superplastic behavior (Ref 13) where a two-phase mixture is apparently present and *m* approaches 1.0. However, the

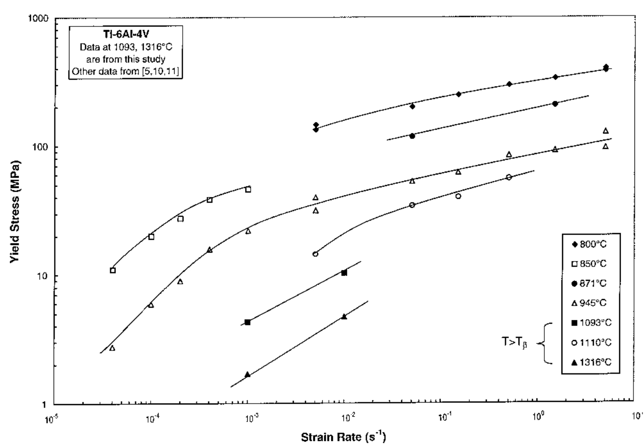


Fig. 2 Variation in yield stress of Ti-6-4 with strain rate at a variety of temperatures

exponent typically drops (Ref 14), and grain sizes can be large in beta-titanium alloys above 1000 °C.

Figure 3(a, b) shows the effects of strain rate and temperature, respectively, on the strain-rate sensitivity of Ti-6-4 alloy (at an approximately constant structure as determined at a constant plastic strain of 0.02). The strain-rate sensitivity is shown to increase with increasing temperature and decreasing strain rate. Figure 3 shows that strain-rate sensitivities vary with temperature and strain rate from values at ambient temperatures of about 0.011 for Ti-6-4 (10^{-3}/s) in this study to as high as about 0.80 (10^{-4}/s) at 945 °C (below T_β, the beta transus), as reported by Hamilton (Ref 5), for superplasticity. The *m* values of this study at 1093 and 1316 °C (above T_β) fall typically in the range 0.3 to 0.6 (at 10^{-2}/s to 10^{-3}/s strain rates) for all of the alloys tested. (For Ti-6-4, $m = 0.4$, which is within the same range and consistent with other work in Fig. 3a and b.)

As these *m* values in the Ti alloys vary substantially from ambient temperature to 1316 °C, there can be an associated change in the rate-controlling mechanisms for plasticity.

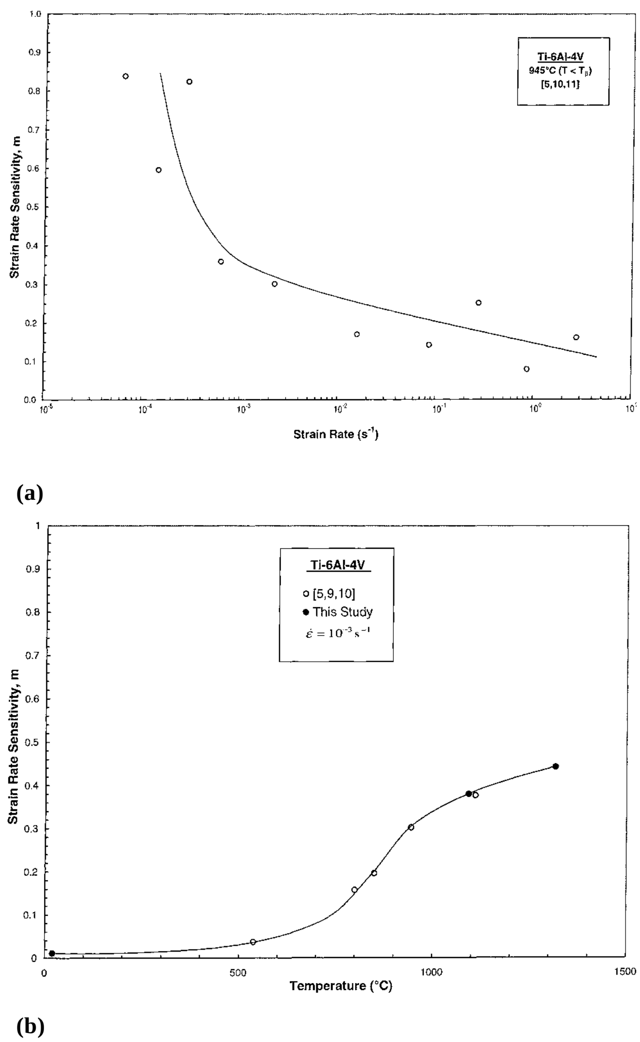


Fig. 3 Variation in strain-rate sensitivity of Ti-6-4 (a) with strain-rate at 945 °C ($T_\beta \cong 980$ °C) and (b) with temperature ($T_\beta \cong 960$ °C)

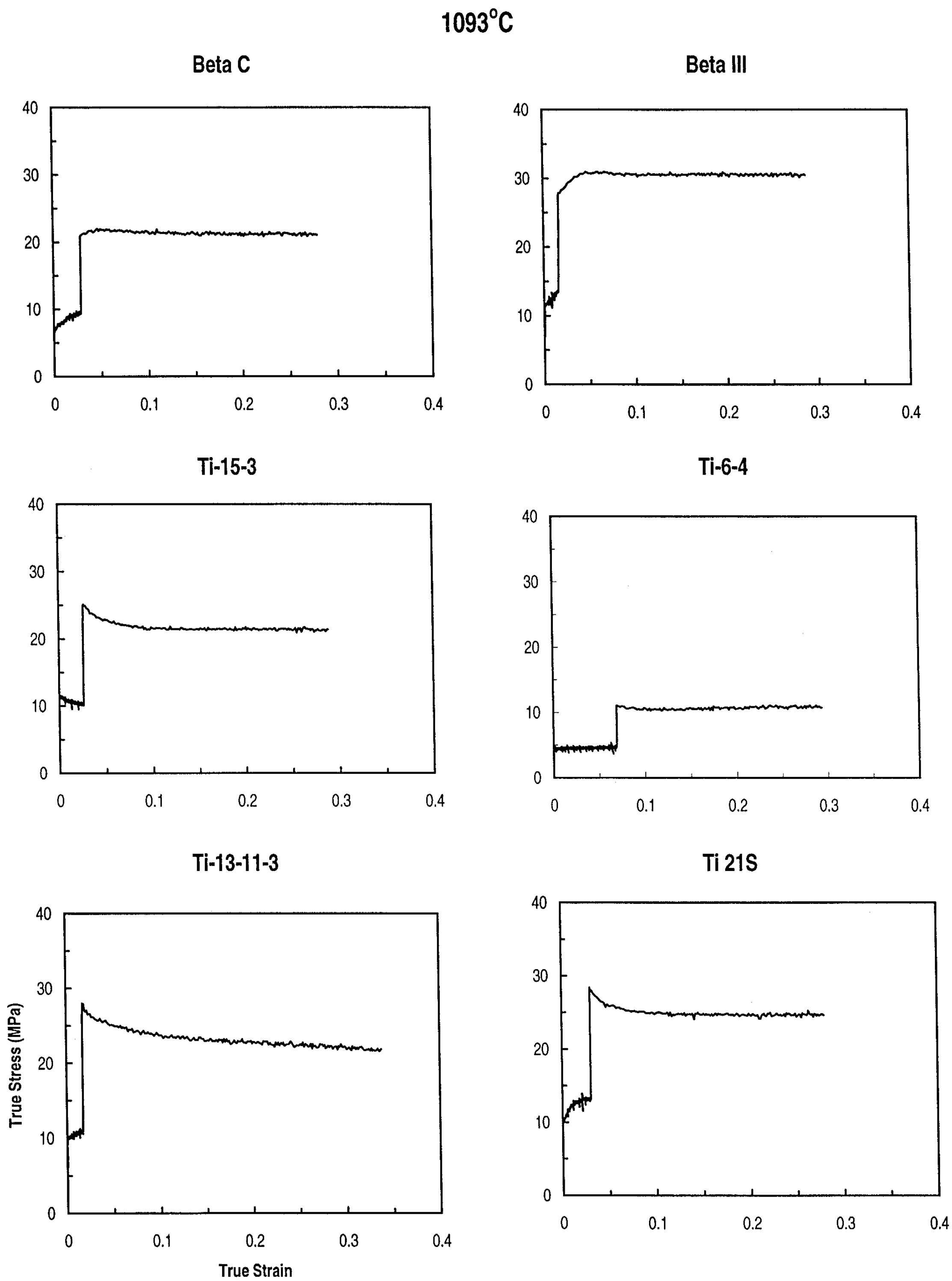
Again, some have suggested superplastic behavior for very high *m* values near 1.0 at some high temperatures, where fine grain sizes (e.g., <20 μm) are present (Ref 5, 13) in Ti-6-4 at $\mathbf{T} < \mathbf{T}_\beta$. Some of the data of Fig. 2 appear to reflect superplastic behavior (Ref 13) in Ti-6-4, where a two-phase mixture is apparently present below T_β. However, in beta alloys, the stress exponent appears somewhat lower (Ref 5,14) than in those where superplasticity occurs, and *m* decreases, although only to 0.3 to 0.6, as mentioned earlier. Elongations to fracture in the present study were less than 200%, less than expected for "conventional" superplasticity associated with grain boundary sliding. The mechanism of plastic deformation is unclear. Hamilton (Ref 5), Malcor et al. (Ref 10), and Morgan and Hammond (Ref 7) appeared to have encountered a similar dilemma (at 800 to 900 °C) as the one encountered in the present study for β-titanium alloys. Morgan and Hammond suggested that their observed "superplastic behavior" (relatively high *m* and high ductility) was provided by subgrains within the larger grains. They suggested that diffusional (Coble and/or Nabarro-Herring) creep together with, perhaps, dislocation creep (five-power-law) rather than grain-boundary sliding in "conventional superplasticity" gave rise to the observed behavior in Beta III and other β-alloys. Malcor et al. observed, from optical metallography, substantial grain distortion, which does not support classic superplasticity from the original β grains in their Ti-6-4 at $T > T_\beta$. They observed serrations of the high angle boundaries, indicative of subgrain formation (Ref 15), as Morgan and Hammond and Hamilton observed. However, they also eventually observed fine grains, which they claimed might be the result of discontinuous dynamic recrystallization or continuous reactions (they use the term "continuous recrystallization" from dynamic recovery for the latter phenomenon) (Ref 15).Whether this leads to superplasticity or diffusional creep is unclear. As discussed later, one difficulty with diffusional (Coble) creep in association with subgrains is that the activation energy, *Q*, for plasticity would be expected to be that for dislocation pipe, Q_p, or grain boundary, Q_{gb} diffusion, which, as discussed subsequently, is not the case.

The activation energy for deformation measurements (Table 4) for all alloys in the present study were about equal or higher than the activation energy of self-diffusion in β-titanium (Ref 16). The activation energy was calculated using:

$$Q = \frac{k}{m}\left[\frac{\partial \ln \sigma}{\partial (1/T)}\right]_{\dot{\varepsilon},\, s} \quad \text{(Eq 3)}$$

where *k* is Boltzmann's constant. The activation energy for diffusion of the principal solutes in β-titanium alloys varies from values near that of self-diffusion of β-titanium to about a factor of 2.4 higher. This might be consistent with the rate-controlling mechanism being viscous glide but less consistent with the mechanism being grain boundary or subgrain (dislocation) boundary (about half that of lattice self-diffusion) controlled deformation, although some have suggested that *Q* for superplasticity can approach that for lattice self-diffusion (Ref 14).

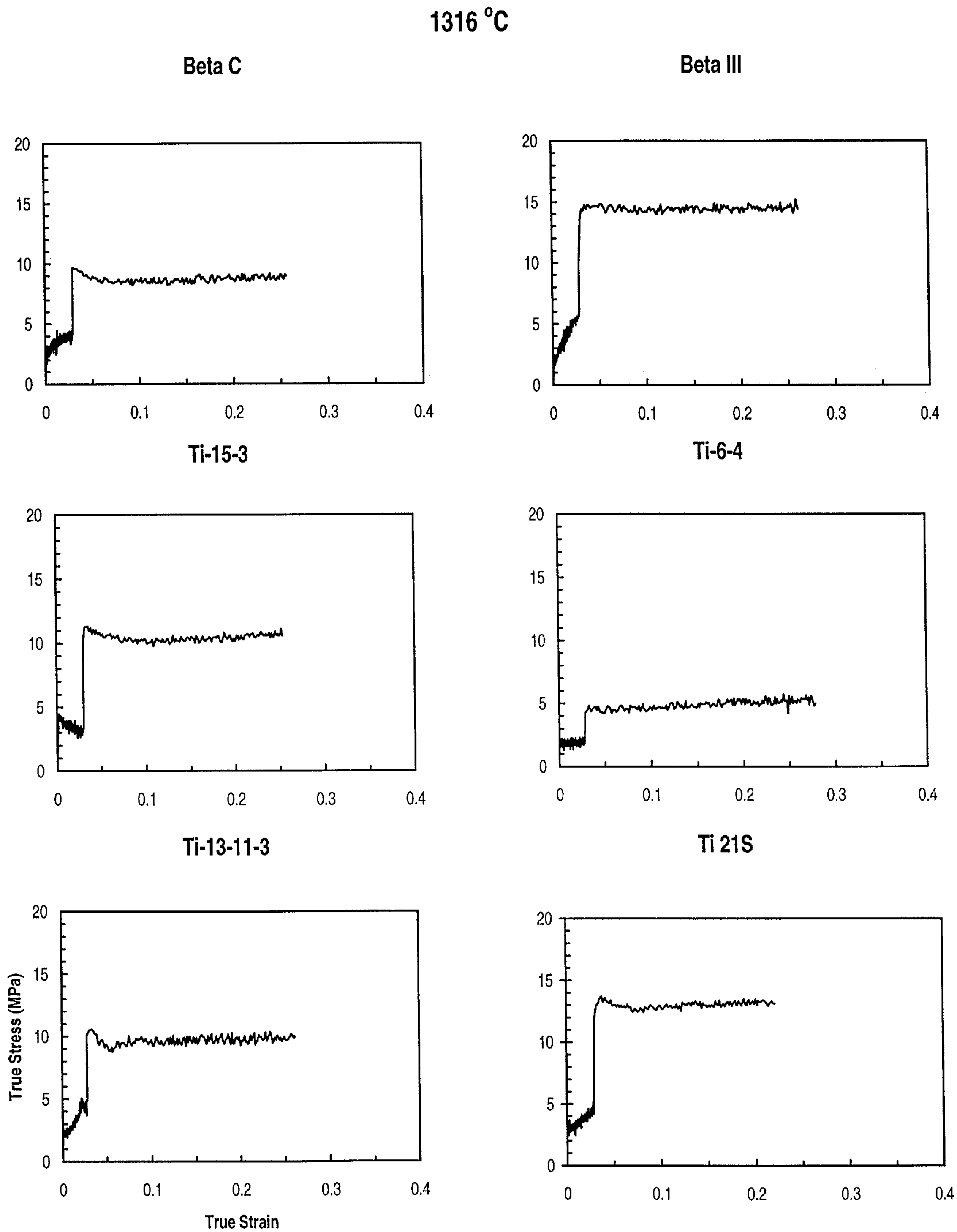
One mechanism considered for the temperature-strain rate regime of the present study was viscous drag (Ref 17) where $m \cong 0.3$. This mechanism is expected to have inverted primary creep and creep transients in strain-rate change tests (σ decreases with strain subsequent to the strain-rate changes) and an activation energy that corresponds to solute diffusion in β-titanium. Ti-15-3 shows inverted primary and transient (subsequent to strain-rate change) as expected for three-power-law, viscous glide in Fig. 4. Two other alloys (Ti-21S and Ti-13-11-3) show 10 to 20% softening during the strain-rate change transient, possibly indicative of solute drag, but normal recovery creep hardening on initial loading. These same type of "softening"



(a)

Fig. 4 Stress versus strain behavior of the various titanium alloys with strain-rate change tests at (a) 1093 °C and (b) 1316 °C. (continued) behaviors were observed by Hamilton in Ti-15-3 at 815 °C and apparently by Malcor et al. (Ref 10) in Ti-6-4 above T_β. These investigators appear to have associated this drop with subgrain formation (Ref 5). Hamilton does not appear to suggest conventional superplastic deformation involving grain boundary sliding but does not appear to propose any specific plasticity mechanism.

Of course, m is somewhat high for a recovery controlled or five-power creep ($m \cong 0.2$ or $n \cong 5$), but it has been argued (Ref 18) that $n \cong 3$ is the natural exponent at low stresses in bcc met-



(b)

Fig. 4 cont. Stress versus strain behavior of the various titanium alloys with strain-rate change tests at (a) 1093 °C and (b) 1316 °C als, such as is relevant here. This mechanism is also consistent with the activation energy being about equal to lattice self-diffusion. It has been reported that the steady-state stress exponent is 4.3 with $Q \cong 131$ kJ/mol for pure β-titanium, although this exponent might be relevant to power-law breakdown where exponents are higher than those for power-law behavior (Ref 1, 19), such as $n = 3$.

Thus, the mechanism of plasticity is unclear with the mechanisms of recovery (five-power-law) creep, classic superplasticity, or solute drag not clearly being evident. The ductilities of approximately 160 to 200% can be consistent with extended ductility in association with the relatively high m values that protract a neck (Ref 20).

4. Conclusions

- Yield stresses of all the titanium alloys at temperatures above 1093 °C were determined to be less than 1% of their ambient temperature values. Strain hardening is negligible in the alloys tested at these high temperatures.
- Yield stresses were found to increase substantially with increasing strain rate at elevated temperatures due to the high strain-rate sensitivity of titanium at high temperatures, $m \cong 0.3$ to 0.6. This contrasts with ambient temperature mechanical properties, where the titanium alloys are not particularly sensitive to strain rate.
- Above 1000 °C, the strain-rate sensitivities were found to increase substantially with increasing temperature and/or decreasing strain rate. The rate controlling mechanism for plasticity is unclear.

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