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TENSILE YIELD AND FRACTURE BEHAVIOUR OF BETA-III TITANIUM

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INTRODUCTION

Recent investigations of the yield and fracture toughness behaviour of unaged, metastable β titanium alloys have shown that these materials may undergo a macroscopic ductile-to-brittle transition with decreasing test temperature, [1 - 4]. This transition is accompanied on a microscale by a change in fracture mode from dimple rupture to cleavage crack propagation. These observations have however been restricted to alloy systems where the primary deformation mechanism involves slip. The deformation mode of metastable β titanium alloys need not be restricted to slip, both stress-induced martensite [5, 6] and twinning [7 - 12] having been reported in similar alloy systems. The investigation reported herein was designed as an extension of the previous studies of metastable β titanium alloys to systems where the primary deformation mode is one other than slip. A commercial alloy, designated Beta-III, was selected for this purpose. Prior examinations [9 - 12] of this alloy have shown that the deformation processes in this material involve multiple-order twinning and to a lesser extent slip and stress-induced orthorhombic martensite formation.

EXPERIMENTAL PROCEDURE

The chemical composition of the material used was (wt. pct.): 10.2 Mo, 5.8 Zr, 4.7 Sn, 0.03 Fe, 0.014 N, 0.02 C, 0.13 O, bal. Ti. It was received in the hot rolled and α-β solution treated condition. The influence of solution treatment condition was examined by re-solution treating 1.9 cm thick sections of the as-received plate for 1 hour at 1005, 1060, 1144 or 1255K followed by water quenching. Solution treatment at 1005K, i.e., below the α-β transus, yielded a three phase (α+β+ω) alloy [13], the α phase being situated either on prior β grain boundaries or within the β+ω matrix.[2] Beta solution treatment resulted in a β+ω microstructure [9, 12]. In general, the β grain structure following the low temperature β treatments was highly elongated and nonuniform. This nonuniformity decreased with increasing temperature, until at the highest solution treatment temperature considered, i.e., 1255K, the beta grain structure was equiaxed.

Tensile measurements were performed utilizing specimens prepared after completion of all heat treatments. The tests were carried out over a temperature range of 77 to 653K at a constant strain rate of 0.006/min on 0.635 cm diameter samples having a gauge length to diameter ratio of 4, the tensile axis always lying in the prior hot rolling direction.

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2. ω is a hcp transitional structure which is not resolvable by optical microscopy and develops during either cooling from the elevated temperature solution treatment or subsequent aging at low temperatures [7].

Following failure, selected samples were examined on a JEOLCO JSM-2 scanning electron microscope operated at 25KV. Additional samples were also nickel-plates, mounted and polished so that the subsequent examination plane was normal to the original fracture surface. Finally, transmission electron microscope specimens were prepared by standard techniques [9] and examined in a JEOLCO 200 Electron Microscope operated at 200 KV.

EXPERIMENTAL RESULTS

Flow Stress

Variations in solution treatment condition appear to have little effect on the yield strength-temperature behaviour of Beta-III, Figure 1. At temperatures below 223K, the yield strength increases rapidly with decreasing temperature. Between 223 and 353K the yield strength seems to be independent of temperature. Finally, at still higher temperatures, serrated yielding and a strength minima are observed. A more detailed examination of the latter phenomean [14] has shown that it is related to the isothermal $\beta \rightarrow \omega$ transformation which occurs at these test temperatures.

Figure 1 does indicate however that the absolute levels of the yield and ultimate tensile strength depend upon whether the solution treatment is carried out above or below the beta transus, the latter resulting in, at any temperature, higher strength levels. The details of the *beta* solution treatment do not seem to greatly influence the observed strengths, at least within that temperature regime where the yield and ultimate tensile strengths appear to be insensitive to test temperature. This suggests that the noted variations in grain shape and size produced by altering the solution treatment temperature have a negligible effect on the flow stress characteristics of Beta-III. It appears that for Beta-III, solution treated above the beta transus, the deformation barrier spacing is controlled by the presence of the athermal ω phase whose size and interparticle distance is approximately 25Å [9, 12], i.e., substantially less than the β grain size.

Additional observations confirmed that deformation at the more elevated temperatures, e.g., 298K, involved both multiple-order twinning and sub-grain formation, Figure 2. The globular α particles, e.g., point 1 in figure 2, did not appear to materially influence the deformation processes in the α-β solution treated condition. Decreasing temperature tended to restrict the extensive dislocation sub-boundary formation, the primary deformation mode at 77K being twinning, Figure 3. Dark field electron microscopy also indicated a lower density of athermal ω within the twinned regions. The latter observation is consistent with those of earlier investigators [12] who reported the destruction of athermal ω during twin deformation of Beta-III.

Fracture

High temperature beta solution treatment tends to increase the room temperature tensile ductility over that obtained by sub-transus treatment, Figure 1. Furthermore, while the tensile ductility following the former treatments appears to vary smoothly with temperature and seems again to be rather insensitive to grain size and shape, the behaviour of sub-transus heat treated Beta-III is quite complex. Indeed, between 163 and 353K the reduction in area values can best be divided into two distinct groups. Nevertheless, when the test temperature decreases below 163K, all of the solution treatment conditions displayed a similar decrease in tensile ductility.

Fractographic examinations showed however, that this decrease of tensile ductility was *not* due to the presence of increasing amounts of cleavage failure with decreasing temperature. Rather it is due to the introduction of a *macroscopic* low energy "shear" fracture mode, Figure 4. *Microscopic* examination of these shear failures revealed however, that the dimples were rather more equiaxed in appearance than might have been anticipated from the macroscopic fracture appearance, Figure 5.

Further examination of the fracture behaviour in Beta-III showed that the maxima in tensile ductility observed in all solution treated conditions could be related to the presence of secondary, principally intergranular cracks propagating normal to the predominant fracture direction, i.e., parallel to the tensile axis, Figure 6. Indeed, the tensile ductility behaviour exhibited by sub-transus solution treated Beta-III could be traced to a similar phenomean: that group of samples exhibiting cup-cone fracture coincident with secondary cracking displayed high ductilities, the "shear" failure mode being associated with decreased ductility.

The relationship between deformation twinning and the crack nucleation process is illustrated by Figure 7. Crack initiation has occurred at the intersection of (i) two twins, A and (ii) a twin and grain boundary, B. Some indication of crack propagation along the twin-matrix interface is also seen in Figure 7.

DISCUSSION

The present results confirm that the primary deformation mode in Beta-III titanium is twinning. Roberson, et al [11] have shown that the operative twin system depends upon crystal orientation, $\{332\}$, $\langle 113 \rangle$, $\{112\}$, $\langle 111 \rangle$ and an irrational mode $\{2, 4.8, 4.8\}$ all having been observed. It is clear that in a polycrystalline sample, all of these systems could contribute to the overall strain, the relative importance of each depending upon the occurrence of a distince crystallographic texture and the direction of load application relative to this texture. Although twinning is the predominant deformation mechanism throughout the temperature range considered, decreasing test temperature does have an important influence on the fine details of the deformation. At the more elevated temperatures, multiple-order twinning and sub-grain formation is observed. Reductions in temperature appear to minimize the degree of slip and sub-boundary formation. It is proposed that it is this variation in the ability of Beta-III to relieve internally imposed stress concentrations that directly controls its tensile fracture behaviour. The stress concentration which arises when a twin band impinges upon another twin/grain boundary/alpha particle can be relieved by either the emission of new dislocations and/or additional twins or by cracking [15]. The transmission electron microscopy observations, Figures 2 and 3, suggest that as the test temperature is reduced, the relaxation of such internal stress concentrations by slip becomes more difficult and cracking becomes an attractive relaxation mechanism.

The above re-emphasizes the important role that twin deformation plays in the fracture initiation behaviour of unaged Beta-III. Figure 7 shows that twin-twin and twin-grain boundary intersections may be effective crack nucleation sites in the β-solution treated alloy. Another possible initiation site, i.e., twin-particle interfaces, although not directly observed in the present investigation, may be of importance in the sub-transus treated condition.

Chakrabarti and Spretnak [16] have recently proposed a general macroscopic criteria for the cup-cone to shear tensile instability observed in the present investigation. They propose that a necessary condition for the onset of a persistent shear instability is:

$$\frac{\partial\sigma}{\partial\varepsilon}\,\delta\varepsilon + \frac{\partial\sigma}{\partial\dot{\varepsilon}}\,\delta\dot{\varepsilon} + \frac{\partial\sigma}{\partial T}\,\delta T + \frac{\partial\sigma}{\partial S}\,\delta S \leq 0 \qquad (1)$$

where σ is the true stress, ε the true strain, $\dot{\varepsilon}$ the true strain rate, T the absolute temperature and S is a parameter which characterizes the microstructure. Qualitatively, equation (1) indicates that those materials possessing a low strain rate sensitivity, a high dependence of flow stress on temperature, a low strain hardening rate and which, when deformed, yield a metallurgical structure that is weaker than the original material should exhibit localized plastic shear instability. Figures 2 and 8 show that the flow stress of unaged Beta-III although quite sensitive to temperature changes below 223K is rather insensitive to strain rate changes. Furthermore, the strain hardening behaviour is low, i.e., $n = 0.05 \pm 0.01$ at temperatures below ambient. Finally, Feeney and Blackburn [12] have shown that deformation twinning in Beta-III results in the local destruction of the athermal ω phase contained in the solution treated alloy. This suggests that the material within the as-deformed, twinned region is weaker than that within the untwinned matrix. Indeed, Rack [18] has recently shown that the general destruction of athermal ω by the passage of a shock wave does result in a decrease in the yield strength. These observations all support the contention that, at low temperatures, macroscopic shear instability may be an important crack propagation mode; the present results indicate it is the predominant mode at 77K.

In summary, it appears that the low temperature fracture process in unaged Beta-III involves, on a *microscale*, at least three stages: first, mechanical twinning accompanied by the destruction of athermal ω within the twin; second, crack initiation at twin-twin, twin-grain boundary and possibly twin-α particle interfaces and third, crack propagation involving a localized plastic shear instability. At elevated temperatures, the secondary deformation processes, i.e., slip and sub-boundary formation, tend both to strengthen the initially weakened, twinned regions by further strain hardening and relieve the stress concentrations associated with twin/obstacle impingement. Highly ductile void initiation/growth cup-cone fracture therefore ensues.

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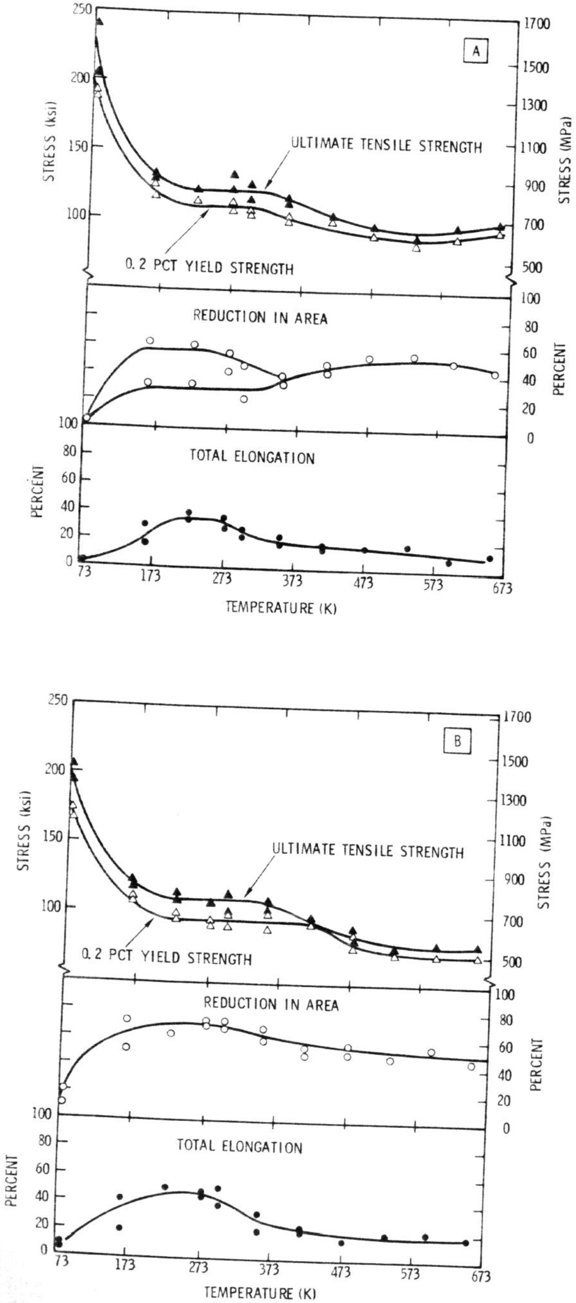

Figure 1 Tensile - Temperature Behaviour of Beta-III Solution Treated At (A) 1005, (B) 1060

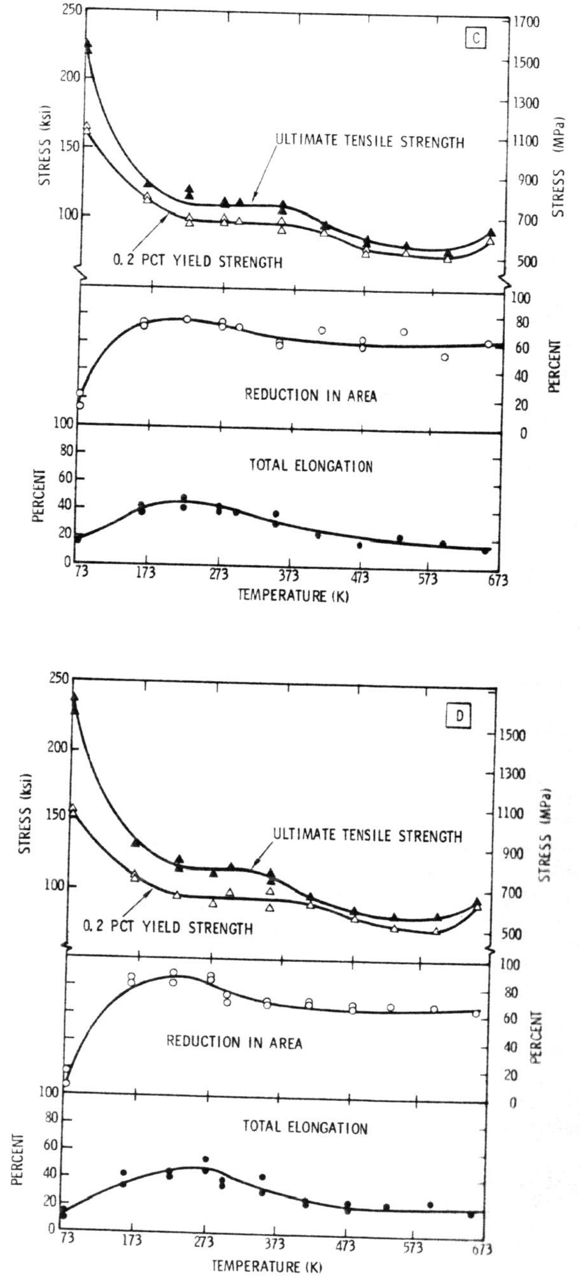

Figure 1 Tensile - Temperature Behaviour of Beta-III Solution Treated At (C) 1144, and (D) 1255K

Figure 2 Multiple-Order Twinning and Sub-Grain Formation in Beta-III Solution Treated At 1005K and Tested at 298K

Figure 3 Deformation Twinning in Beta-III Solution Treated At 1005K and Tested At 77K

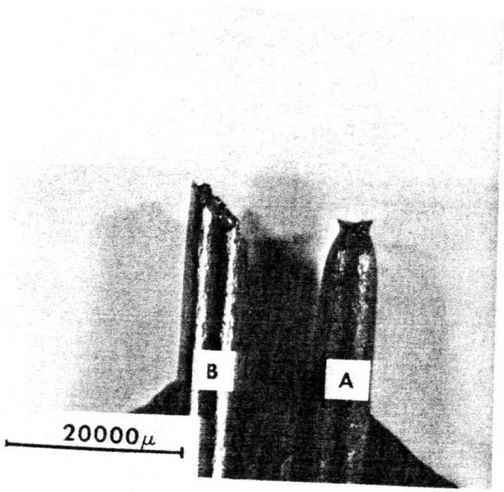

Figure 4 Macrograph of Tensile Failures in Beta-III Solution Treated at 1144K and Tested At (A) 298K and (B) 77K

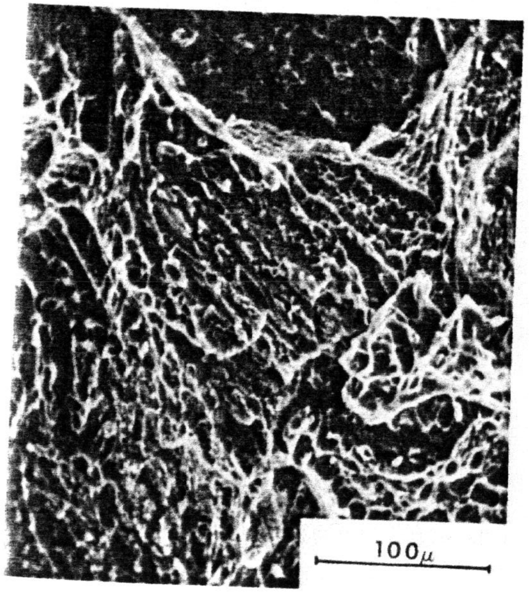

Figure 5 Scanning Electron Fractograph of Tensile Failure in Beta-III Solution Treated At 1005K and Tested At 77K

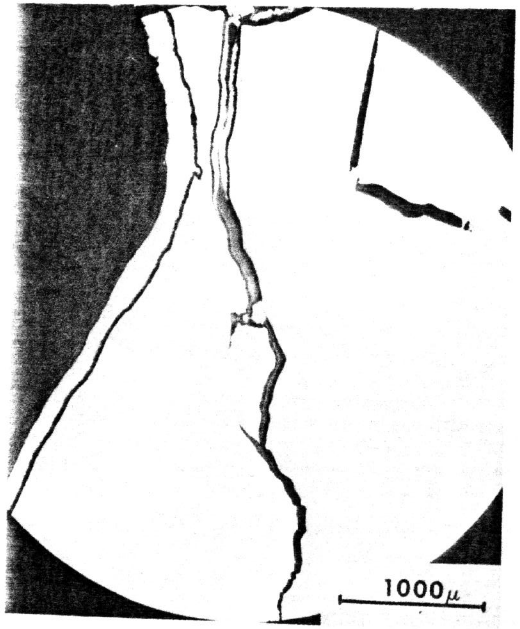

Figure 6 Illustration of Predominatly Intergranular Secondary Cracks in Beta-III Solution Treated At 1005K. RA = 60.5 pct. Test Temperature: 163K

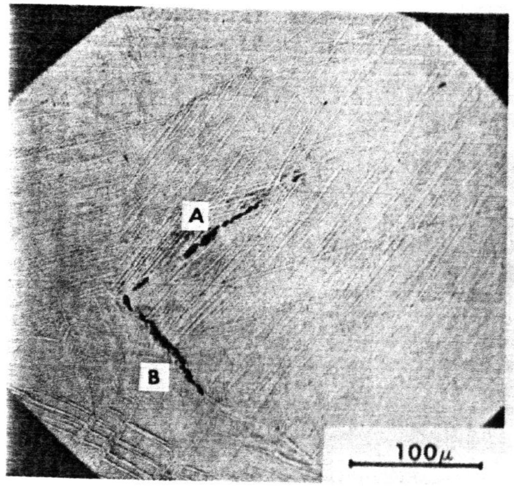

Figure 7 Twin-Twin, A and Twin-Grain Boundary, B, Crack Initiation in Beta-III Solution Treated At 1255K and Tested At 77K

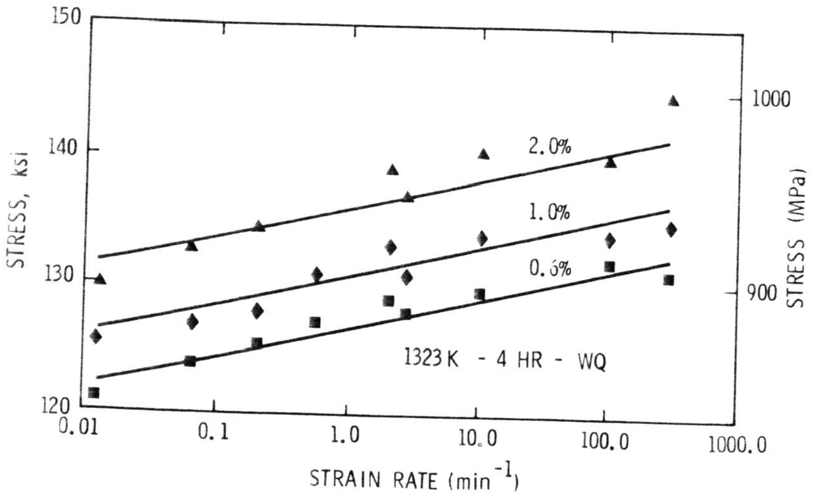

Figure 8 Strain Rate - Compressive Flow Stress Behaviour of Beta-III Solution Treated at 1323K and Tested at 298K, (17)