

Effect of hydrogen on the tensile ductility of Ti6Al4V

Part I Orientation effects in slow straining

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Smooth tensile specimens were machined from a round Ti6Al4V bar, the crystallographic orientation of which was determined by X-ray diffraction analysis. The preferred orientation was mainly $\langle 10\bar{1}0 \rangle$ texture. When the specimens of various hydrogen contents were slowly strained, those having more than 2000 p.p.m. hydrogen exhibited slow-strain embrittlement in both rolling and transverse directions, whereas below 2000 p.p.m. hydrogen, the effect of hydrogen on the tensile properties was different in the two directions. Hydrogen decreased both the proportional limit, σ_0 and the 0.2% flow stress, $\sigma_{0.2}$, in specimens parallel to the rolling direction but there was no change in either σ_0 or $\sigma_{0.2}$ with increasing hydrogen content in the transverse specimens. The strain hardening behaviour of all specimens could be divided into two stages, whatever the orientation of specimen or hydrogen content. The first-stage strain-hardening exponent, n_1 , increased with increasing hydrogen content, but n_1 for the transverse specimens was smaller than for specimens parallel to the rolling direction, and the variation of n_1 with hydrogen content was also lower in the transverse specimens. These differences were attributed to the differing deformation modes for specimens of different orientation. There was, however, no effect of orientation on the microfractography.

1. Introduction

The fabrication of round bars or sheets in titanium alloys introduces varying degrees of crystallographic texture which naturally produces anisotropy of the mechanical properties [1–5] and may cause problems in subsequent use of the materials. There have been a number of investigations of the relationship between the anisotropy of mechanical properties and the crystallographic preferred orientation [6–8] and some techniques have been suggested for reducing the anisotropy [9, 10]. The different textures observed are dependent on the alloy composition, both the schedule and the temperature involved in fabrication and the subsequent heat treatment. The most prominent texture involves orientation of the basal planes parallel to the rolling plane and this is produced by cross-rolling at low temperature. Rolling at higher temperatures produces a transverse texture, where the basal planes are normal to the rolling plane. In addition, mixed textures having a basal as well as a transverse component or a random orientation are often observed [7, 9]. For sheet or rectangular bar, the plastic strain ratio, R , of width to thickness strain values in a uniaxial tensile test, may be employed to describe the plastic anisotropy, and R -values for titanium alloys vary over a very wide range of (0.5–12). The R value

for basal texture is much higher than for the mixed texture [6]. When the texture-strengthening effect was studied using a yield-locus diagram, it was found to be as large as 60% for the material with a basal texture, but only 15% for the transverse [7]. However, the R parameter and yield-locus diagram are unsuitable for studying the anisotropy of round bars. Although the effect of hydrogen on the mechanical properties of titanium alloys has been examined [11–13], there is only a limited amount of work concerned with the additional effect of different crystallographic orientation. The present work compares the effect of hydrogen on the slow strain tensile properties of smooth specimens in both the rolling and a transverse direction of a strongly textured Ti6Al4V round bar.

2. Experimental procedure and results

The 110 mm diameter round bar of Ti6Al4V used in this study contained (wt %) 6.55 Al, 4.19 V, 0.19 Fe, 0.185 O and 0.001 H and was received in the hot-rolled and mill-annealed condition. The microstructure in both rolling and transverse directions showed distinct orientation (Fig. 1) that consisted essentially of α grains separated by grain-boundary β phase and α/β eutectoid. The α grains are extended parallel to the

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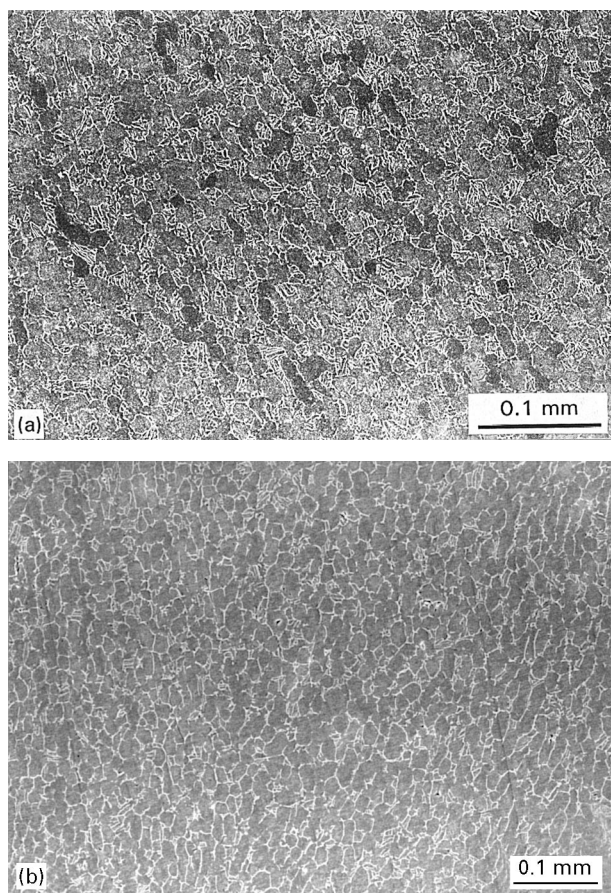


Figure 1 Microstructure of sections from the as-received Ti6Al4V round bar: (a) longitudinal; (b) transverse.

rolling direction, but also, to some extent, in various radial directions. In order to identify the preferred orientation in the as-received bar, inverse pole figures were produced from X-ray data obtained from a Philips X-ray diffractometer employing copper K_α radiation and a scanning speed of $0.3^\circ \text{ min}^{-1}$. Specimens were in the form of round discs having diameter 20 mm and thickness 5 mm, with the surface to be irradiated finely ground and chemically polished, and were of two orientations having the surface irradiated parallel to either the rolling (L) or a chosen transverse direction (T) that was parallel to the longitudinal axis of the relevant (transverse) tensile specimens. Filings of the bar annealed at 900°C for 1 h were employed as standards to represent random orientation. The texture coefficient was determined by the normal method [14] and the $\langle 10\bar{1}0 \rangle$ texture coefficient in the T specimen was greatest (Fig. 2), which means that the preferred orientation of the rolling direction was essentially $\langle 10\bar{1}0 \rangle$.

Smooth tensile specimens having a 11.35 mm gauge length of 3.24 mm diameter were machined with their major axes parallel to either the rolling (L) or the specific transverse direction (T). By invariably cutting blanks along fixed directions with reference to a mark on the original bar, the orientation of the tensile and X-ray diffraction specimens could be identified with respect to the dominant texture. The slow straining of smooth tensile specimens was performed at a cross-head speed of $2.8 \times 10^{-5} \text{ mm s}^{-1}$ giving an initial

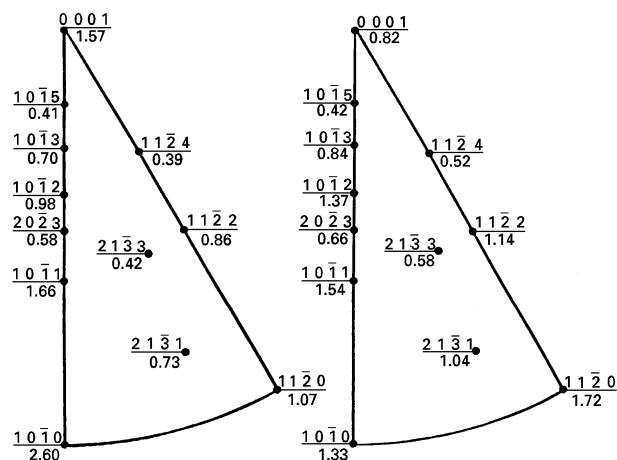


Figure 2 Inverse pole figures for the Ti6Al4V with irradiated surfaces normal (T) and parallel (L) to the rolling direction.

strain rate of $2.5 \times 10^{-6} \text{ s}^{-1}$ and curves of true stress versus true strain were evaluated from the recorded curves relating load and extension. Strain-hardening exponents were calculated according to the formula proposed by Ludwik [15]

$$\sigma = \sigma_0 + K\varepsilon_p^n \quad (1)$$

where σ is the true stress, σ_0 is the true stress for initiation of plastic deformation, ε_p the true plastic strain and K and n are, respectively, termed the strength coefficient and the strain-hardening exponent. The reduction in area at fracture was assessed from a mean of diameters obtained on a measuring microscope and the final area of cross-section was also employed to calculate the true fracture stress, σ_f , from the breaking load.

The low-pressure high-temperature method of charging with hydrogen was used. The purified hydrogen was introduced to the specimen at 600°C over a period of 2 h, after evacuating to 1 Pa, and then the specimens were cooled in the vacuum furnace. The hydrogen content was controlled by adding measured amounts of hydrogen from a gas burette [16] and this was believed to give a more dependable hydrogen concentration than subsequent analysis.

Slow straining smooth specimens in tension revealed a hydrogen-induced embrittlement above 2000 p.p.m. hydrogen in specimens parallel to both rolling and transverse directions, and the effect of hydrogen on the tensile properties below 2000 p.p.m. hydrogen is different for the two orientations (Figs 3 and 4). Hydrogen produces an obvious reduction in σ_0 and in the stress for 0.2% plastic strain, $\sigma_{0.2}$, for specimens parallel to the rolling direction, but there is no change with increasing hydrogen for the transverse specimens. It is also notable that the fracture stress and the ductility parallel to the rolling direction for a particular hydrogen content are better than for transverse specimens.

A logarithmic plot of $\Delta\sigma (= \sigma - \sigma_0)$ against plastic strain (Fig. 5) allows determination of the strength coefficient and the strain-hardening exponent. The strain-hardening behaviour of all specimens may be divided into two stages, whatever the orientation or

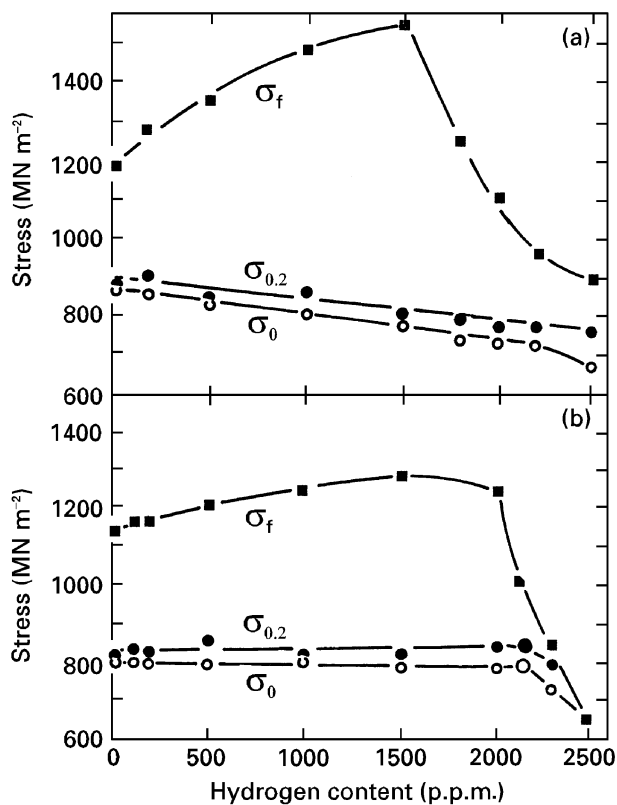


Figure 3 Effect of hydrogen content on the strength of the Ti6Al4V bar with the stressing direction (a) parallel and (b) normal to the rolling direction.

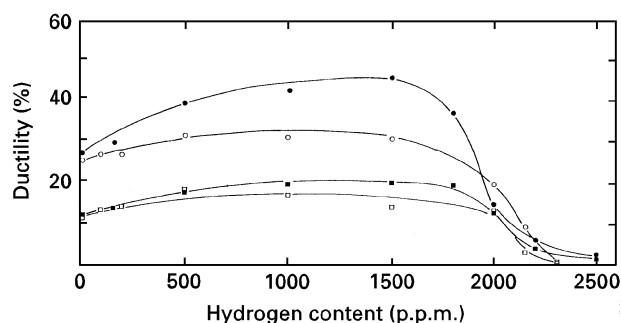


Figure 4 Effect of hydrogen content on the ductility of the Ti6Al4V bar with the stressing direction (●, ■) parallel and (○, □) normal to the rolling direction, in terms of (○, ●) percentage reduction in area and (□, ■) elongation.

hydrogen content of the specimen. The strain-hardening exponent in the first stage, n_1 , for both orientations increases with increasing hydrogen, but n_1 for the transverse direction specimens is smaller than for the rolling direction and, in addition, is less affected by hydrogen content (Fig. 6). The effect of hydrogen on tensile properties at slow strain rates is similar to that previously reported for similar material but with a slightly different microstructure and preferred orientation [17].

There is a slight difference in the macrofractography of the two orientations, which may be because of the different alignment of α in the microstructure, but the macrofractography observed by the SEM is similar for both orientations. The effect of hydrogen content on the macrofractography is, however, obvious. The frac-

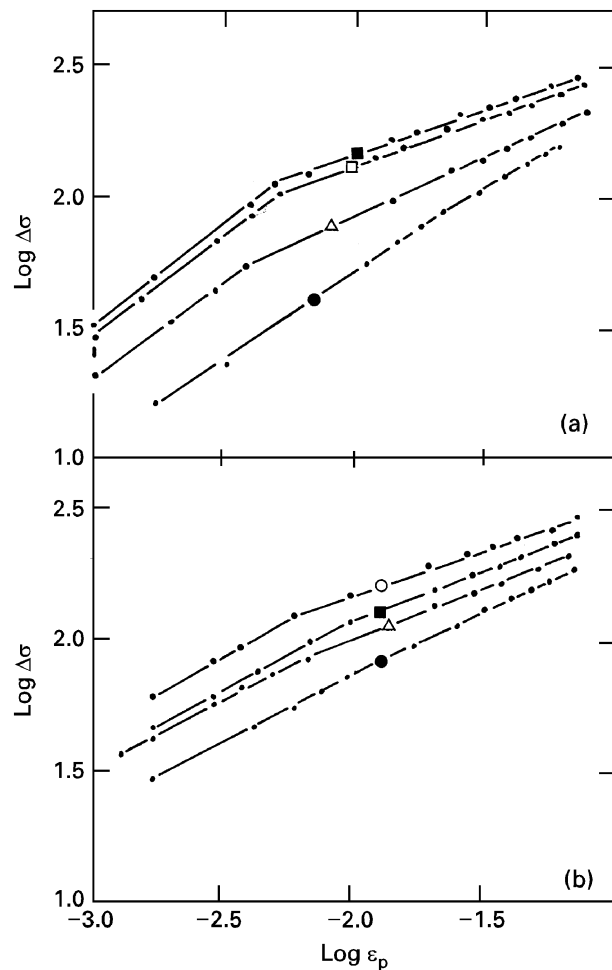


Figure 5 Logarithmic plot of $\Delta\sigma$ versus plastic strain for Ti6Al4V bar containing various amounts of hydrogen (p.p.m. by weight): (●) 10, (Δ) 500, (□) 1010, (■) 1500, (○) 2000; (a) longitudinal and (b) transverse specimens.

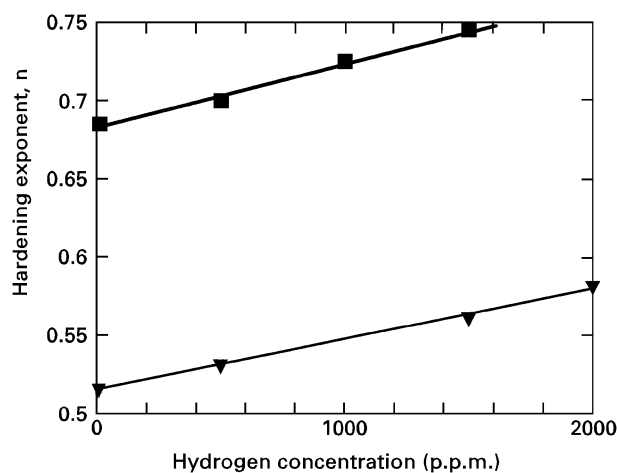


Figure 6 Effect of hydrogen and orientation on the hardening exponent, n_1 , in the first stage of deformation. Orientation: (▼) n -trans., (■) n -long.

ture surface consists of deep equiaxed dimples, shear rupture and transgranular cleavage at the lower hydrogen content (Fig. 7) and a fine dimpled region appears in the range 500–1500 p.p.m. (Fig. 8). When the hydrogen content exceeds about 2000 p.p.m., there appears, in addition to the fine dimpled region, some interfacial separation of the α - and β -phases in the α/β

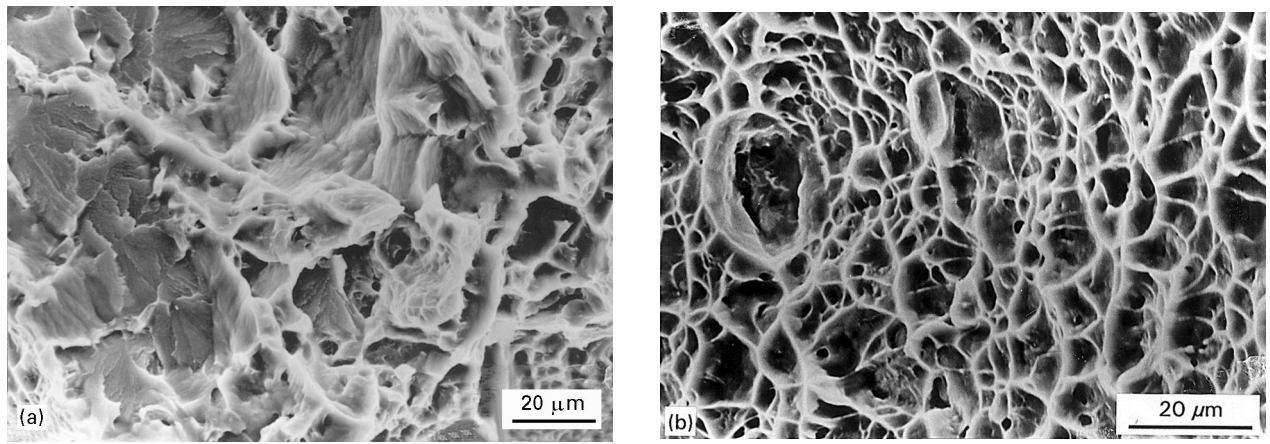


Figure 7 Fracture surfaces of specimens of the as-received Ti6Al4V: (a) longitudinal, and (b) transverse.

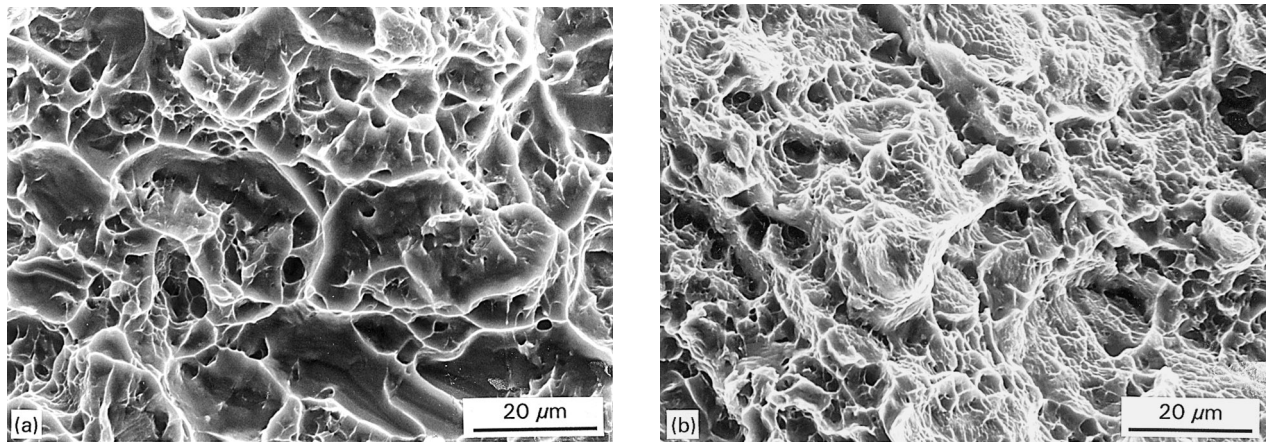


Figure 8 Fracture surfaces of Ti6Al4V specimens containing hydrogen: (a) longitudinal (1400 p.p.m.), and (b) transverse (1500 p.p.m.).

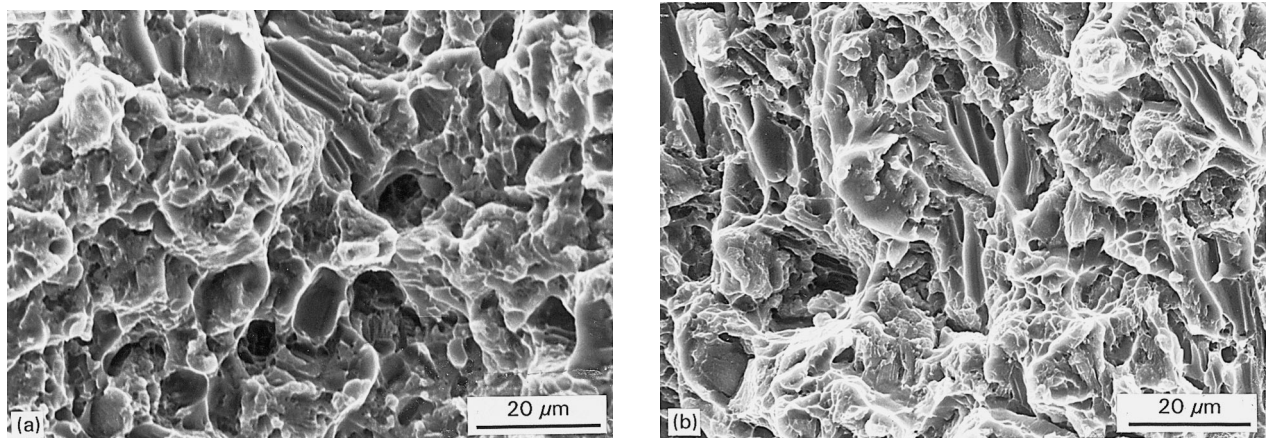


Figure 9 Fracture surfaces of Ti6Al4V specimens containing high levels of hydrogen: (a) longitudinal (2500 p.p.m.) and (b) transverse (2000 p.p.m.).

eutectoid and some secondary cracking along the α grain boundaries (Fig. 9).

3. Discussion

There seems to be two contributions to anisotropy in the as-received Ti6Al4V bar: an inhomogeneous microstructure (Fig. 1) and preferred orientation of the α -phase. Although there are α grains elongated in

different directions in both specimen orientations, this is not very severe, and the principal reason for anisotropy is thought to be the preferred orientation. The inverse pole figure for the T specimen obtained from X-ray diffraction indicates that the preferred orientation in the rolling direction is mainly $\langle 10\bar{1}0 \rangle$ whereas the L specimen indicates that in the normal direction to be mainly $\langle 11\bar{2}0 \rangle$. This latter direction is normal to the axis of the transverse tensile specimens

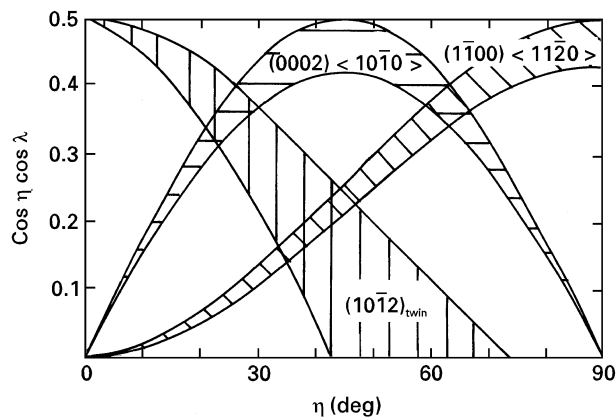


Figure 10 Orientation dependence of the resolved shear stress factor for slip and twinning in α titanium (after Partridge [18]).

and therefore the axial direction is $[0001]$, i.e. normal to both $\langle 10\bar{1}0 \rangle$ and $\langle 11\bar{2}0 \rangle$.

Partridge pointed out [18] that the principal mode in the early stages of plastic deformation is related to η , the angle between the stress axis and the $[0001]$ direction, as shown in Fig. 10 where $\cos \eta$ is the orientation factor for the resolved shear stress for slip or twinning, and λ indicates the angle between the stress axis and the principal directions for slip or twinning. When η is zero, the initial plastic deformation mode is $\langle 10\bar{1}2 \rangle$ twinning, which is true for the transverse specimens. Even though the material studied is $(\alpha + \beta)$ Ti6Al4V, as opposed to single-phase α , α is the principal phase and may be considered mainly in interpreting the deformation [19]. The applicability of Partridge's conclusion to Ti6Al4V, is supported by the observation that twinning can be seen in the deformed transverse specimens (Fig. 11) but not in those parallel to the rolling direction. The twins observed here are thick and lenticular and this suggests that the twin plane may be $\{10\bar{1}2\}$ or $\{11\bar{2}2\}$ [20–22]. η for the rolling direction specimens is 90° and hence their principal deformation mode should be $\{1\bar{1}00\}\langle 11\bar{2}0 \rangle$ prismatic slip from Fig. 10.

The observed reduction in the elastic limit and 0.2% flow stress induced by hydrogen for specimens in the present work parallel to the rolling direction has also been reported by Wan and Gao [23] in slow tensile straining (10^{-6} s^{-1}) of Ti6Al4V. The effect of hydrogen on the creep behaviour of Ti6Al4V at room temperature indicates that both creep strain and dislocation density increase with hydrogen content and it is suggested that dissolved hydrogen atoms soften the alloy at the beginning of deformation by increasing the dislocation mobility [24] in a similar way to the situation in steels [25, 26].

Pressouyre [27] summarized some trapping energy data for hydrogen in steels, including substitutional solute atoms, dislocations, grain boundaries and the interfaces of second phases and identified dislocations and grain boundaries as reversible traps having similar trapping energy. Because the twinning interface is a coherent interface and the distortion produced by twinning is therefore less than by an incoherent inter-

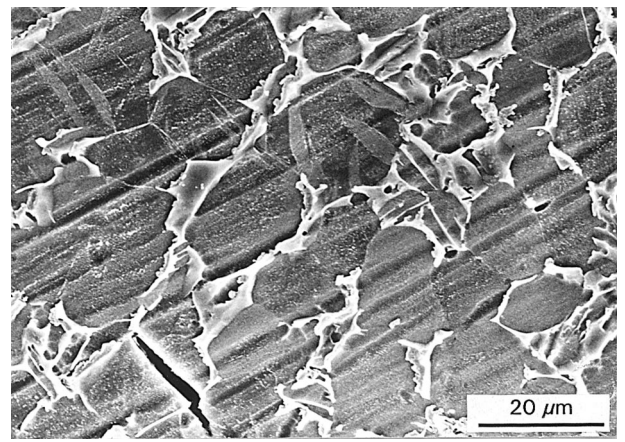


Figure 11 Twinning observed in an axial section from the necked portion of an unbroken transverse specimen of Ti6Al4V containing 10 p.p.m. hydrogen.

face, such as a grain boundary, it seems likely that the interaction of hydrogen with twinning is less than with dislocations. The absence of any effect of hydrogen on the stresses for initial plastic deformation in the transverse specimens may be related to this. However, the twinning mode of deformation becomes replaced by slip with increasing strain [28] and then the effect of hydrogen on the fracture properties such as elongation, reduction in area and fracture stress, becomes similar for both rolling and transverse specimens.

4. Conclusions

1. The preferred orientation of the mill-annealed round Ti6Al4V bar in the rolling direction is mainly $\langle 10\bar{1}0 \rangle$ in the α phase.
2. Above 2000 p.p.m. hydrogen there is hydrogen-induced slow strain embrittlement in smooth specimens cut parallel to either the rolling or transverse direction.
3. The hydrogen obviously decreases the proportional limit, σ_0 , and the 0.2% flow stress, $\sigma_{0.2}$, in the rolling direction specimens but there is no change with increasing hydrogen content in the transverse specimens.
4. The strain-hardening behaviour of all specimens could be divided into two stages. Whatever the orientation or hydrogen content of specimens, the first-stage strain-hardening exponent, n_1 , increases with increasing hydrogen content, but n_1 in the transverse direction specimens is smaller than in the rolling direction specimens and the slope of n_1 with hydrogen content is also lower in the transverse direction specimens.
5. The variable effect of hydrogen on the mechanical properties in the two orientations is related to the particular deformation mode operating. The initial deformation mode seems more likely to be twinning in the transverse direction specimens and slip in the rolling direction specimens and it is reasonable to expect that the interaction of hydrogen with twinning is different from its interaction with dislocations.
6. Although there is a distinct effect of microstructure on macro-fractography, there seems to be no

effect of orientation on the micro-fractography. The change from transgranular dimples and cleavage to fine dimples and interface separation may be attributed to increasing hydrogen content.

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