JOURNAL OF MATERIALS SCIENCE 21 (1986) 4071–4080

Superplastic deformation of sheet test pieces machined from Ti–6Al–4V bar

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Sheet tensile test pieces were machined in three orientations from edge textured Ti–6Al–4V bar and tested at temperatures in the range 800 to 975° C and at strain rates of 3×10^{-4} and $1.5 \times 10^{-3}\,sec^{-1}$. Bands of contiguous alpha grains aligned in the rolling direction caused local variations in the flow stress, strain to necking, strain rate sensitivity, plastic strain ratio values and surface roughness. Texture effects were only detected at the lowest test temperature (800° C) and highest strain rate ($1.5 \times 10^{-3}\,sec^{-1}$).

1. Introduction

Studies of the influence of texture and microstructure on the superplasticity of round test pieces machined from Ti–6Al–4V bar [1–4] revealed that stress and strain anisotropy was caused by directionality in the starting microstructure which contained contiguous alpha grains aligned in the rolling direction. There was little effect of texture. Round test pieces machined in the three principal directions of the bar became oval in cross-section during superplastic straining [1]. However, the superplastic strain anisotropy was found to be different to that developed during room temperature deformation, which was related to the crystallographic texture of the alpha phase [5]. The room temperature alpha texture was retained after 300% superplastic strain, although with a reduced intensity [2, 6]. Consequently, the sense of the room temperature anisotropy was unchanged by superplastic strain [7]. Superplastic deformation caused the alpha phase banding to break up and resulted in a more homogeneous microstructure and increasingly isotropic behaviour [2].

In order to relate the results on the round test pieces to observations of anisotropic superplastic deformation in commercial rolled sheet [8], 2 mm thick sheet test pieces were machined from the edge textured bar used in the original work [1, 2]. This paper describes the influence of microstructure on the superplastic deformation of the sheet test pieces at 800 to 975° C. The anisotropy is measured in terms of the plastic strain ratio, R.

2. Experimental procedure

The Ti–6Al–4V bar was supplied by IMI Titanium in the form of rectangular bar of cross-section 160 mm × 55 mm. The chemical composition was 6.34 wt % Al, 4.35 wt % V, 0.18 wt % Fe, < 0.006 wt % H_2, < 0.01 wt % N_2, < 0.15 wt % O_2, balance Ti. Throughout this paper L, T and S are, respectively, the longitudinal, transverse and short transverse directions of the original bar. The three test piece orientations are referred to as TL, ST and LT. The first letter in each case refers to the tensile axis of the test piece and the second letter refers to the orthogonal direction in the plane of the test piece (Fig. 1). Test pieces 2 mm thick were machined from the bar in the TL and LT orientations. However, the bar thickness was insufficient for the extraction of the ST orientation test pieces. Therefore, 4 mm thick blanks for the gauge length were machined from the bar in the ST orientation and extended by electron beam welding pieces of 4 mm thick Ti–6Al–4V sheet to each end. Final machining of the test pieces to 2 mm thickness was then carried out. All the test piece heads were then reinforced by plates of 2 mm thick Ti–6Al– 4V sheet spot welded onto each side in order to minimize deformation of the test piece heads during testing.

Uniaxial superplastic straining up to 300% was carried out in a deoxygenated argon atmosphere with a vertical three zone furnace controlled to $\pm 2°$ C over a length of 100 mm. Four test temperatures of 800, 875, 925 and 975° C and two strain rates of 3×10^{-4} and $1.5 \times 10^{-3}\,sec^{-1}$ were used. The strain rate was maintained nominally constant by increasing the cross-head speed after every 50% strain increment. The alpha phase proportions estimated from the data in [4] were, at 800, 875, 925 and 975° C, 81, 68, 50 and 10% respectively. The R value, defined as the ratio of width strain to thickness strain; $R = \varepsilon_w/\varepsilon_t = \ln (w/w_0)/\ln (t/t_0)$, where w_0 and w are the initial and final widths and t_0 and t are the initial and final thicknesses was calculated from micrometer measurements made before and after straining.

3. Results

3.1. Test piece shape after superplastic strain

Fig. 2 shows the effect of superplastic strain at $3 \times 10^{-4}\,sec^{-1}$ on test pieces of each orientation after testing at temperatures in the range 800 to 975° C. The test pieces pulled at the higher strain rate of $1.5 \times 10^{-3}\,sec^{-1}$ were similar in appearance to those shown in Fig. 2.

The TL orientation test pieces after testing were characterized by a very irregular strain distribution

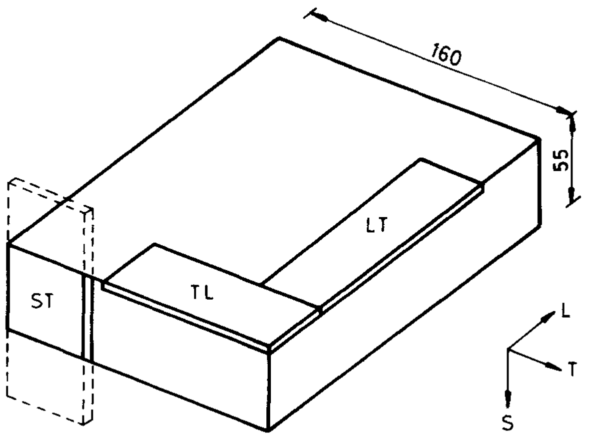


Figure 1 Test piece orientation diagram. Dimensions in mm.

along the gauge length with bands of relatively undeformed material running across the test piece. This behaviour was most marked for the 875 and 925° C test temperatures and the banding is shown at A in Fig. 2. The TL orientation test pieces pulled at 975° C failed at a relatively low strain with no evidence of banding across the gauge width.

For the ST orientation the banding direction was through the thickness of the test pieces and small blocks of relatively undeformed material were left protruding out of the surface, e.g. Fig. 3. This effect was more clearly visible near the ends of the test piece gauge lengths, i.e. in the low strain regions than in the centre of the gauge lengths or in the necked regions, where the test piece surfaces were relatively smooth. The ST orientation test pieces pulled at 975° C exhibited relatively low total elongations and, for both strain rates, showed shallow trenches in the test piece surface running roughly perpendicular to the tensile axis (Fig. 4a). Each of these features was an incipient neck as shown schematically in Fig. 4b. This behaviour was peculiar to the ST test piece orientation. Final necking at 975° C occurred along a line at an angle to the tensile axis. The 800° C ST test pieces also showed this slant type local necking (Fig. 2a).

By contrast, the LT orientation test pieces pulled at temperatures up to 925° C showed more uniform deformation with relatively smooth surfaces and parallel gauge lengths (Figs 2a to c).

3.2. Microstructure

The microstructure of the mill annealed (as-received) bar contained bands of heavily deformed material, e.g. at A in Fig. 5 and more homogeneous, non-directional microstructure such as at B in Fig. 5. There is some evidence [9] that this banding is associated with small variations in composition. Examination of the microstructure annealed at 925° C and quenched into water revealed [8] that the regions of strongly

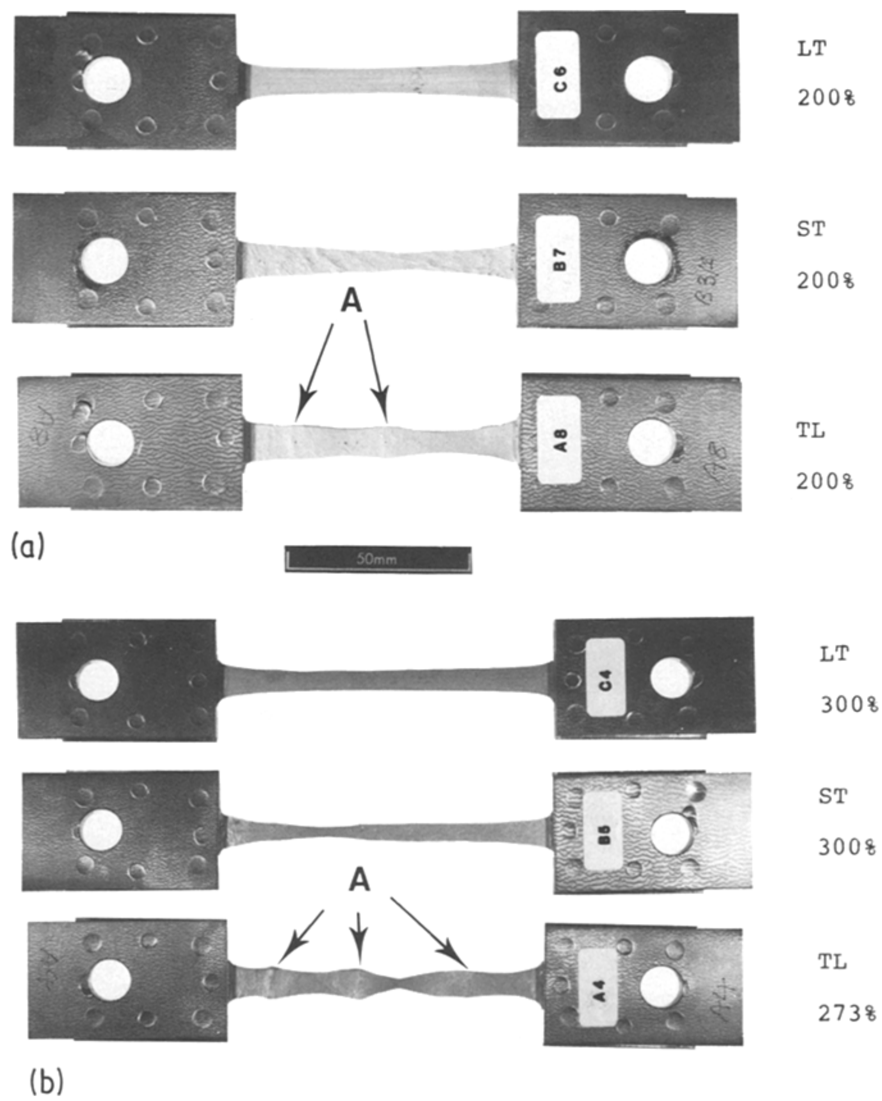
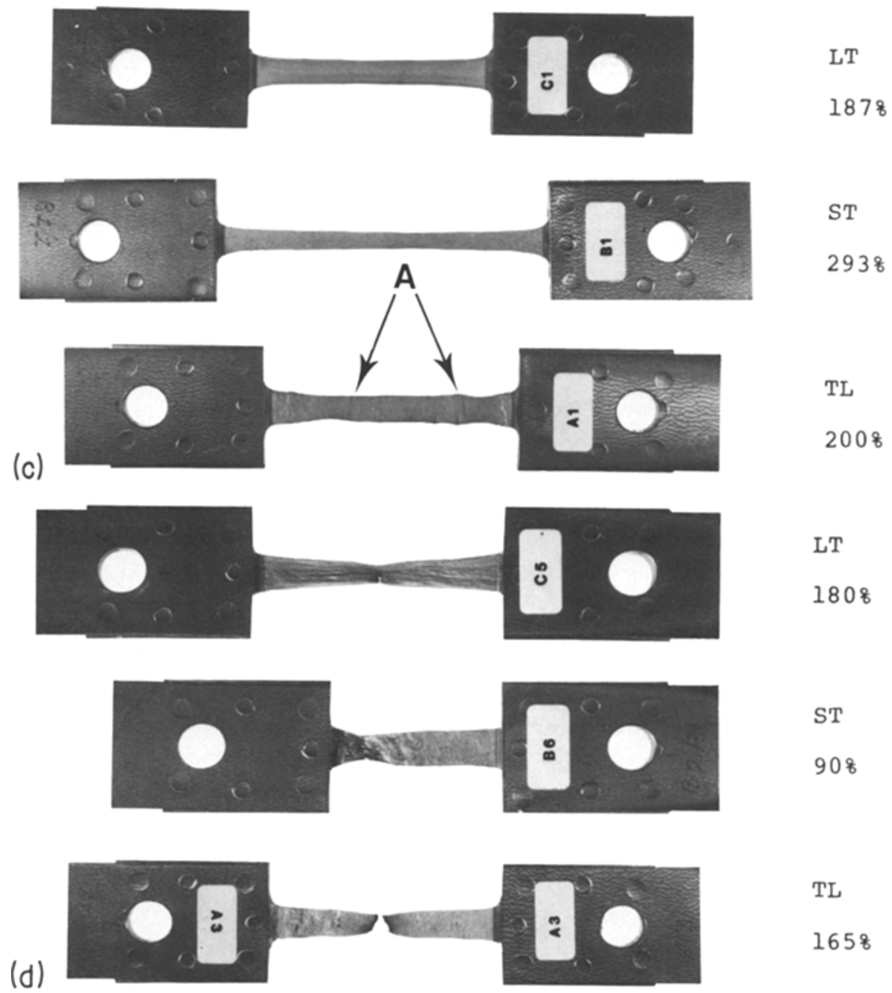


Figure 2 Test piece shape after strain at 3 × 10^{-4} sec^{-1} (a) 800° C, (b) 875° C, (c) 925° C, (d) 975° C.

Figure 2 Continued.



directional microstructure consisted largely of equiaxed, but contiguous alpha grains aligned in the rolling direction. This microstructural banding is discussed in [1, 10] and shown schematically in Fig. 6.

Sections through the ST orientation test pieces after superplastic strain showed that the protrusions on the surfaces (Fig. 3) corresponded to areas of aligned microstructure, such as at A in Fig. 7 and that depressions in the surface were associated with more equiaxed microstructure, e.g. at B in Fig. 7. Similarly, the banded regions of the TL test pieces, such as at A in Fig. 2, contained the aligned microstructure, running across the width of the test piece.

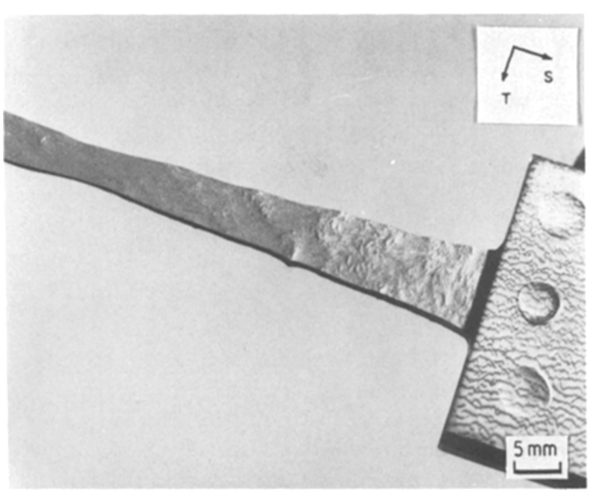


Figure 3 Surface of ST orientation test piece after strain at 800° C and 3×10^{-4} sec^{-1}.

The effect of strain at 875° C on the microstructure of an ST test piece is shown for the LS section in Fig. 8; the microstructures after testing have been reheated to 875° C and quenched into water. Superplastic strain increased the average grain size and broke up the banded microstructure.

A section through an ST test piece pulled at 975° C revealed that the microstructure in the necked regions contained practically no primary alpha phase and showed relatively large alpha plate colonies indicative of a large prior beta grain size (Fig. 9). The neck which propagated to failure contained only a few isolated primary alpha grains indicating that, at the forming temperature, this region was composed almost entirely of beta phase (Fig. 10a). By contrast, the microstructure of a 975° C TL test piece showed a uniform distribution of primary alpha phase, e.g. Fig. 10b which shows the microstructure close to the point of failure. These results suggest that local composition variations may lead to regions rich in beta phase with less resistance to necking.

The test pieces which had almost necked to failure showed cavitation in the necked region, e.g. at A in Fig. 11. The cavitation seemed to originate either in the beta phase or at the alpha/beta phase boundaries and the cavities appeared to elongate or coalesce parallel, rather than perpendicular to the tensile axis.

A scanning electron micrograph of the edge of a test

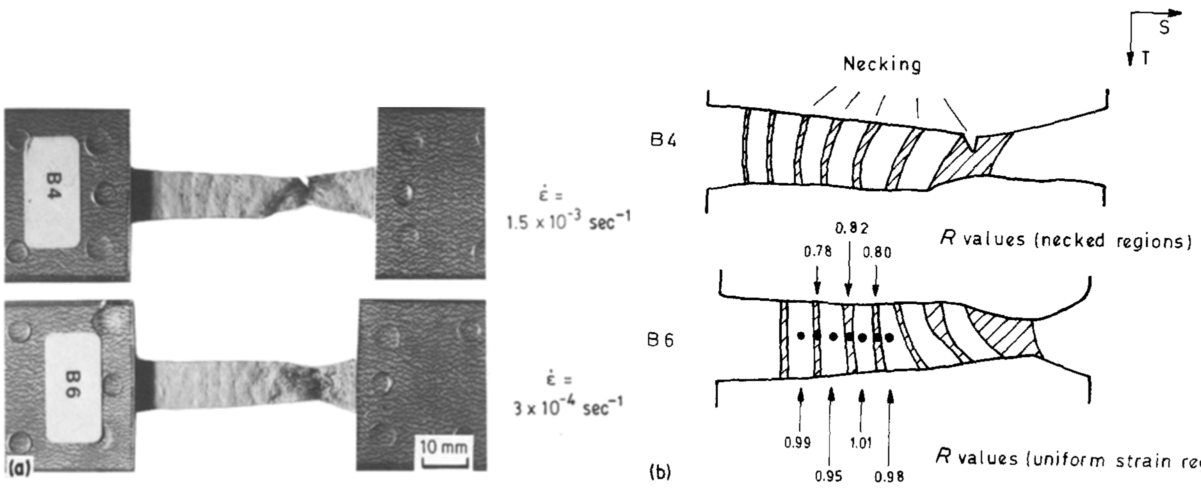


Figure 4 Deformation of ST orientation test pieces pulled at 975° C (a) appearance of necks, (b) schematic diagram of position of necks and corresponding R values.

piece after superplastic strain is shown in Fig. 12. The surfaces, which had a machined finish prior to testing, have been roughened by grain boundary sliding and the emergence of new surface grains.

3.3. Flow stress

The superplastic flow stresses are shown in Fig. 13. The flow stress values were calculated from the maximum loads, which occurred in each test at 15% strain or less, and the initial cross-section area. At such low strains the test piece gauge lengths would still be approximately of uniform cross-section and parallel sided. Flow stresses were higher for the high strain rate and decreased with increasing temperature. For each strain rate the flow stresses were highest for the LT orientation and lowest for the ST orientation. Similar results were reported for the round test pieces [1, 10].

3.4. Strain rate sensitivity

The m values are plotted against temperature in Fig. 14. For each temperature the m values were higher for the low strain rate tests and for each orientation and strain rate reached a maximum at 925° C. With the exceptions of the high strain rate tests at 800° C and the low strain rate tests at 975° C, the strain rate sensitivity values were in the order $m_{ST} > m_{TL} > m_{LT}$.

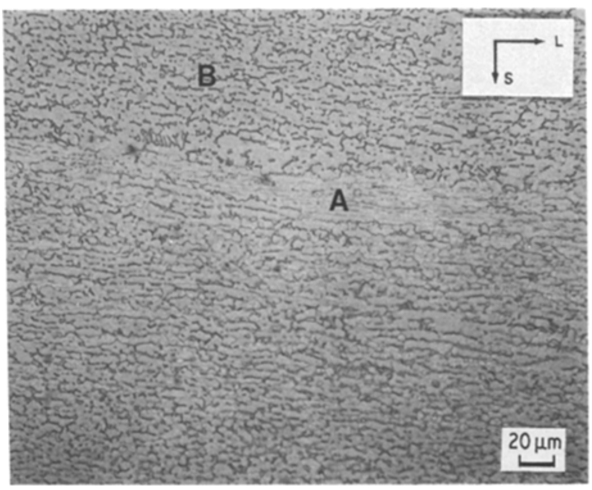


Figure 5 Microstructure of mill annealed (as-recevied) Ti–6Al–4V bar, LS section.

This was the reverse of the order of flow stresses (Fig. 13).

3.5. *R* values

R values, measured at various points along the gauge length of each of the test pieces, are plotted in Fig. 15 against the local axial true strain ε_l for the strain rate $3 \times 10^{-4}\,sec^{-1}$. This strain rate is only a nominal value as the irregular strain distributions in the TL and ST test pieces imply a wide variation in local strain rate.

The most significant feature of the R values for the ST test pieces was that for test temperatures of 800 to 925° C all the R values measured were greater than 1.0, although a large amount of scatter was recorded. For the 975° C test pieces R values ranged from 0.77 inside the necked regions to about 1.0 in the adjacent unnecked area (see Fig. 4b).

For the LT orientation there was much less variability in R than for the other orientations. All the R values measured for this orientation were less than 0.9.

The width and thickness strains ε_w and ε_t have been plotted against ε_l for the TL orientation 875° C low strain rate test piece in Fig. 16a and the measurement locations on the test pieces identified in Fig. 16b. The banded regions apparently deformed more isotropically than the non-banded regions. This is also shown in Fig. 15, in which R values measured on the banded regions of the TL orientation test pieces, such as at A in Fig. 2, and in the regions between the bands have been plotted separately as closed and open symbols respectively. No trend of R value with strain was apparent.

The R values measured on the test pieces pulled at $1.5 \times 10^{-3}\,sec^{-1}$ showed similar trends [8], except that, for the ST orientation tests at 800° C, R values less than 1.0 were measured and R ranged from 0.9 to 1.1.

4. Discussion

4.1. Microstructure and flow stress

The results have shown that large variations in width and thickness strains in superplastically deformed sheet test pieces were related to aligned, contiguous alpha phase bands. The orientation of the banding

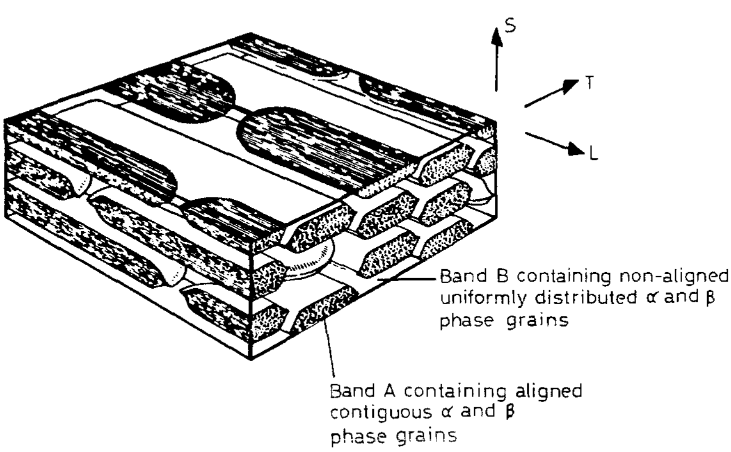


Figure 6 Schematic distribution of aligned (band A) and non-aligned (band B) microstructure.

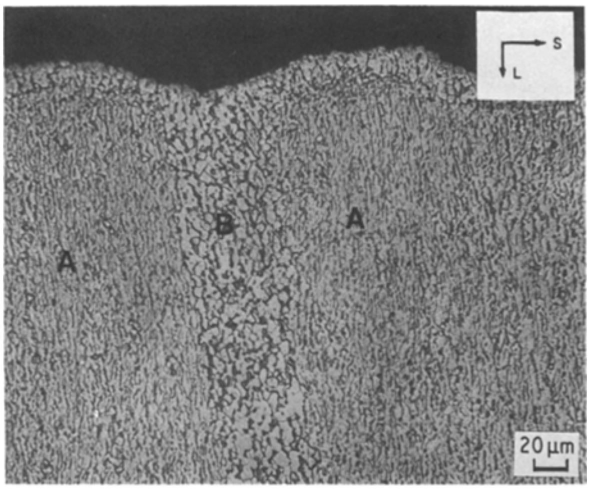


Figure 7 Effect of strain at 875° C and $3 \times 10^{-4} sec^{-1}$ on an ST orientation test piece. Tensile axis S.

relative to the sheet plane affected the strain to the onset of necking, R values and surface roughness.

In general, the superplastic flow models for single phase materials [11–13] apply to Ti–6Al–4V in that large grain offsets occurred (Fig. 12) and cavitation was only detected in the necked regions. In the areas of aligned microstructure more alpha/alpha boundaries must be sheared than in the equiaxed areas, where more alpha/beta and beta/beta boundaries exist. Boundary sliding measurements on a Pb–62% Sn alloy [14] and on Zn–22% Al [15] revealed that the sliding rate for each type of interface depended on the value of δD_g, where δ is the boundary width and D_g is the coefficient of boundary diffusion. This was the basis for a deformation model [10] for anisotropic superplastic deformation in two phase alloys, which accounted for the break-up of banded microstructures and predicted a flow softening by the progressive and irreversible conversion of alpha/alpha interfaces into alpha/beta interfaces. The bulk diffusion coefficient for the alpha phase is 100 to 1000 times that of the beta phase in the temperature range investigated [16]. Thus, it is expected that the flow stress of the microstructure containing aligned alpha phase would be higher than that of the non-aligned microstructure. A further factor, tending to increase the flow stress in the aligned regions, is that the aligned grains are likely to be of a similar orientation within each string with

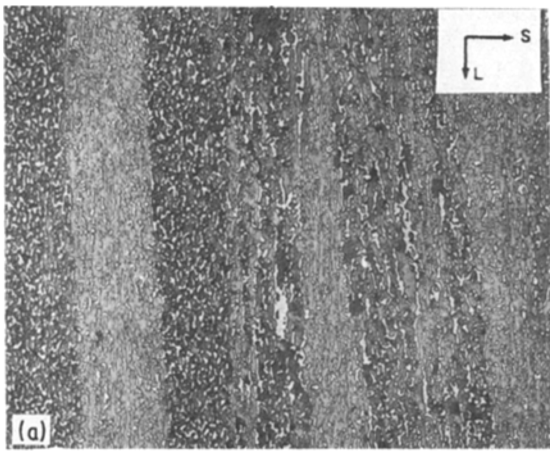
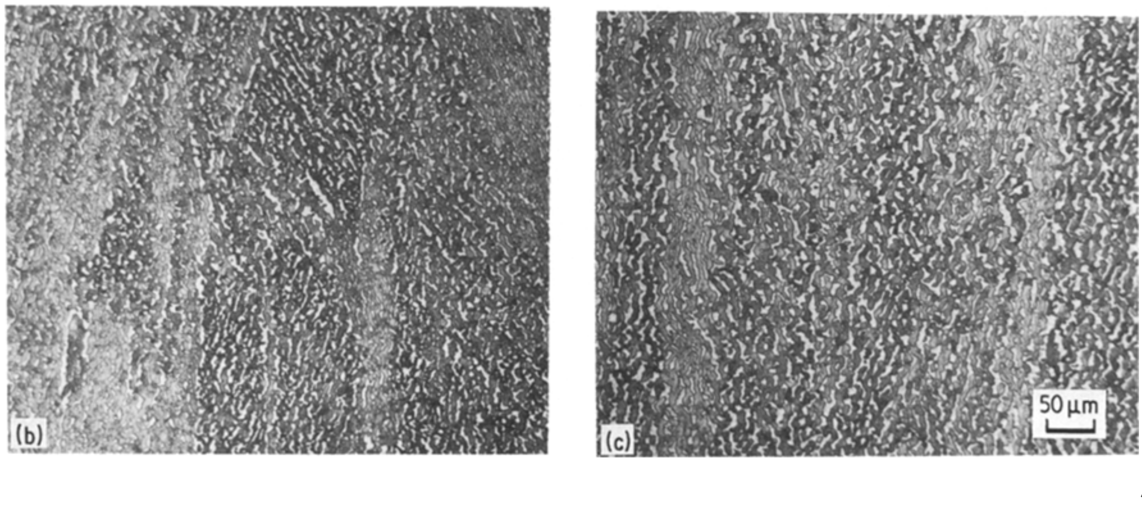


Figure 8 Effect of strain at 875° C and $3 \times 10^{-4} sec^{-1}$ on LS section microstructure. Tensile axis S. Material reheated to 875° C and quenched into water (a) annealed at 875° C (unstrained), (b) 24% strain, (c) 169% strain.



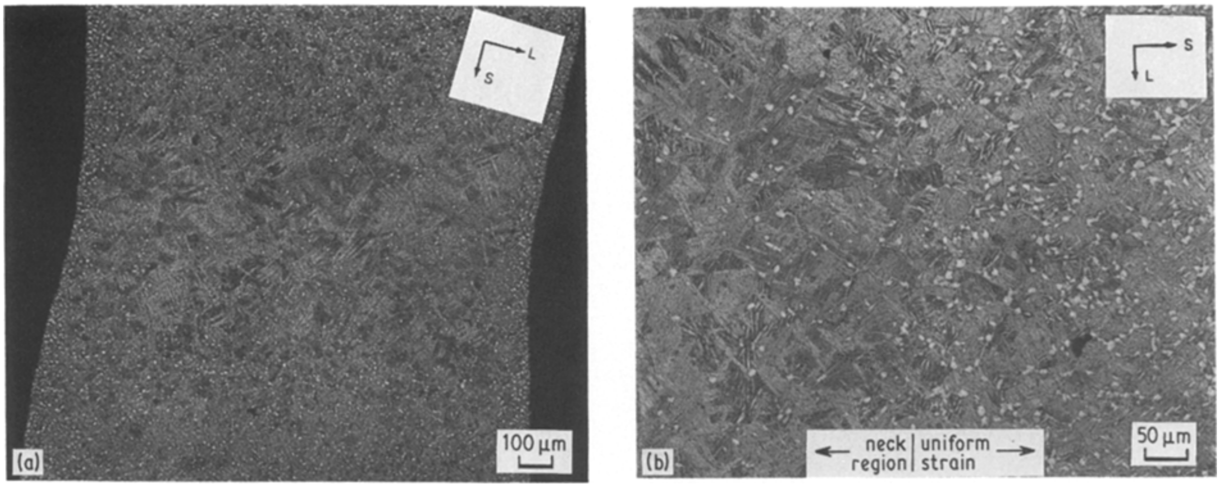


Figure 9 Microstructure inside neck and in adjacent region after strain at 975° C and $1.5 \times 10^{-3}\,sec^{-1}$. Tensile axis S.

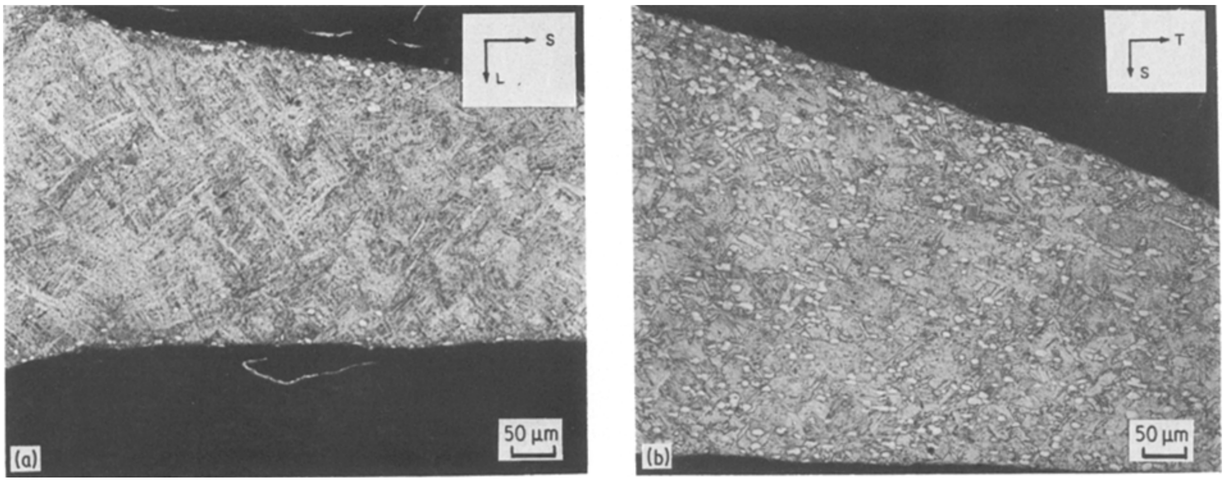


Figure 10 Microstructure inside necked regions after strain at 975° C (a) ST orientation test piece. Strain rate $3 \times 10^{-4}\,sec^{-1}$, (b) TL orientation test piece. Strain rate $1.5 \times 10^{-3}\,sec^{-1}$.

relatively narrow grain boundaries and possibly even some low angle boundaries. Evidence for this is that the alpha phase texture is much more intense in the aligned regions than in the non-aligned regions [2] and the boundaries between the contiguous, aligned grains etch very slowly and are difficult to detect optically.

For the TL and ST orientations the areas of aligned microstructure remained relatively undeformed during testing (Fig. 2) compared with those areas where the two phases were homogeneously distributed. Therefore, for these two orientations the measured flow stress corresponded roughly to that of the flow stress of the equiaxed areas. For the LT orientation, where the tensile axis was parallel to the banding direction, the bands of aligned microstructure were constrained to deform at the same rate as the surrounding areas

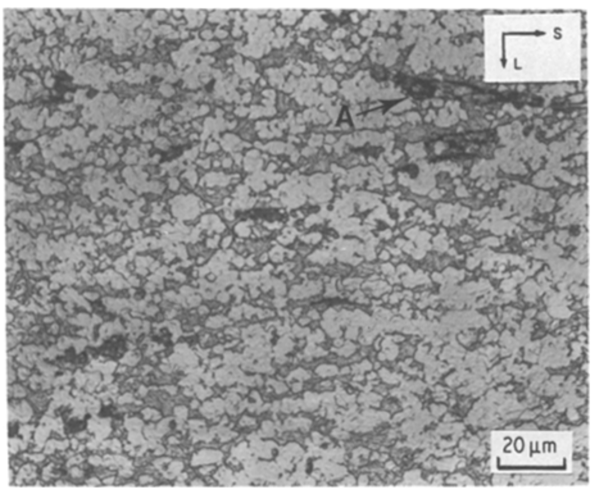


Figure 11 Cavitation inside necked region after strain at 875° C and $1.5 \times 10^{-3}\,sec^{-1}$. ST orientation test piece. Tensile axis S.

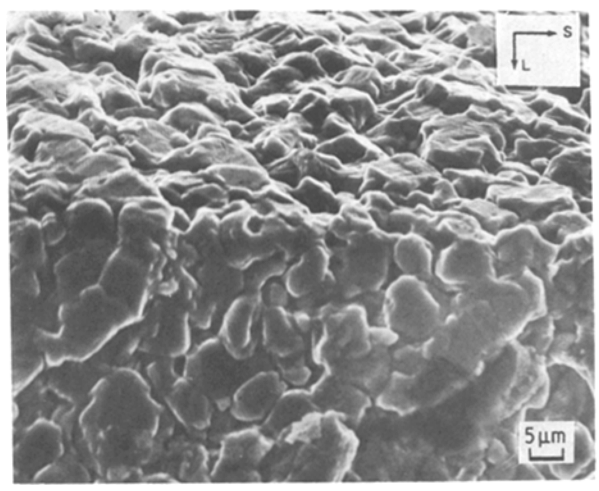


Figure 12 ST orientation test piece after strain at 875° C and $3 \times 10^{-4}\,sec^{-1}$ showing test piece edge. Tensile axis S. Scanning electron micrograph.

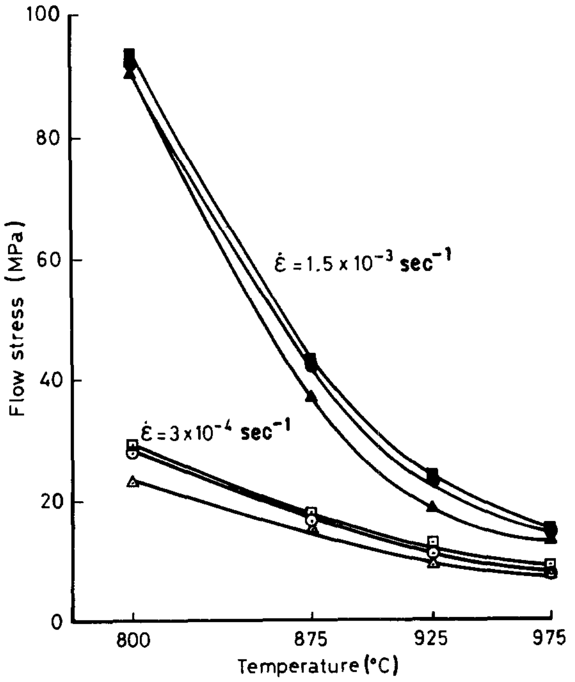


Figure 13 Flow stress at maximum load for each orientation, strain rate and test temperature. (⊙, ●) TL; (△, ▲) ST; (□, ■) LT.

and a relatively high flow stress was measured (Fig. 13). The orientation with the lowest flow stress was consistently the ST orientation. This can be explained by the observation that, although the alpha phase was most strongly aligned in the L direction, there was also some alignment in the T direction [8], which led to flow stresses in the order $\sigma_L > \sigma_T > \sigma_S$.

Quantitative analysis of the break-up of the contiguous alpha phase is not easy because measurements have to be taken from several areas that may not have been similar before straining. However, the measurements of contiguous alpha phase aspect ratios in [2] and the microstructures in Fig. 8 show that the aligned microstructure does break up during straining and it

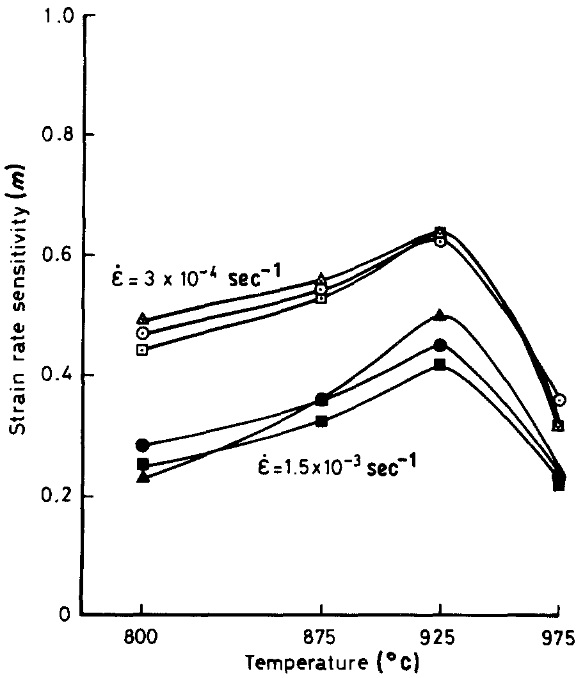


Figure 14 Strain rate sensitivity (*m*) at maximum load for each orientation, strain rate and test temperature. (⊙, ●) TL; (△, ▲) ST; (□, ■) LT.

is assumed that those areas most strongly aligned before deformation are those which deform slowest.

The multiple necking occurring in the ST orientation test pieces pulled at 975° C (Fig. 4) appears to be the result of strain localization in bands where the alpha phase proportion was low at 975° C. Deformation of these areas would be favoured because sliding of beta/beta interfaces only would be required. The final beta grain size in these areas was much larger than in adjacent regions where the alpha phase retarded the beta grain growth. Presumably, this type of necking did not occur in test pieces of other orientations because the beta phase bands were not suitably oriented in relation to the tensile axis. It is shown schematically in Fig. 6 that the regions of aligned and non-aligned microstructure were roughly planar in the LT plane. Thus, the beta rich regions observed after testing at 975° C would lie in planes perpendicular to the S direction.

It is to be expected that the beta rich areas would exhibit lower flow stresses than those areas lean in beta phase for forming temperatures lower than 975° C. Thus, as well as grain alignment, a second factor that may contribute to the irregular superplastic strain distribution is a variation in phase proportion.

4.2. Strain rate sensitivity

The strain rate sensitivity (*m*) increased with test temperature and reached a maximum at 925° C, when the phase proportions were roughly equal. Above this temperature rapid beta grain growth caused a sharp reduction in *m* value.

For the TL and ST orientation tests the strain tended to be localized in the areas of non-aligned microstructure. Therefore, the *m* values measured on the TL and ST orientation test pieces corresponded approximately to the *m* values of the non-aligned microstructure, whereas for the LT orientation tests each type of microstructure deformed at the same rate and an "average" *m* value was recorded. Thus, it appears that the strain rate sensitivity of the aligned microstructure was lower than that of the non-aligned microstructure.

4.3. *R* values

In the TL orientation the banding direction was across the gauge width and *R* values measured on the bands, such as points A1–A3 in Fig. 16 were less than 1.0. It might be anticipated that the *R* values measured at other points along the gauge length, e.g. B1–B4 in Fig. 16 would be closer to 1.0 reflecting the more isotropic nature of the microstructure in these areas. However, this was not the case and for test temperatures in the range 800 to 925° C lower *R* values were measured in the non-banded areas. Work on rolled Ti–6Al–4V sheet [8] showed that *R* values measured after superplastic strain were influenced by the initial test piece shape. If a test piece of low gauge length to width ratio was used then constraint by the test piece heads restricted the strain in the width direction and low *R* values were measured. Increasing the gauge length or reducing the width minimized this geometrical effect. In the same way, *R* values measured in the non-

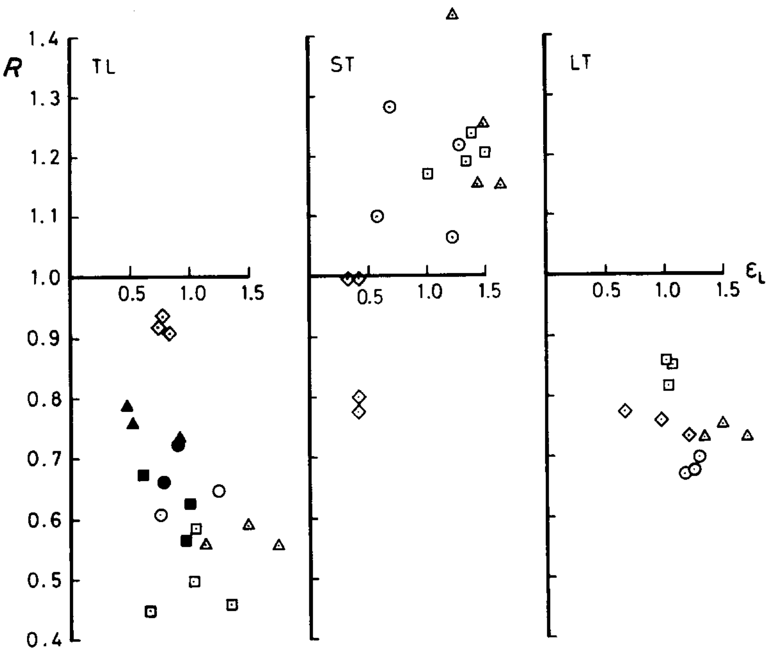


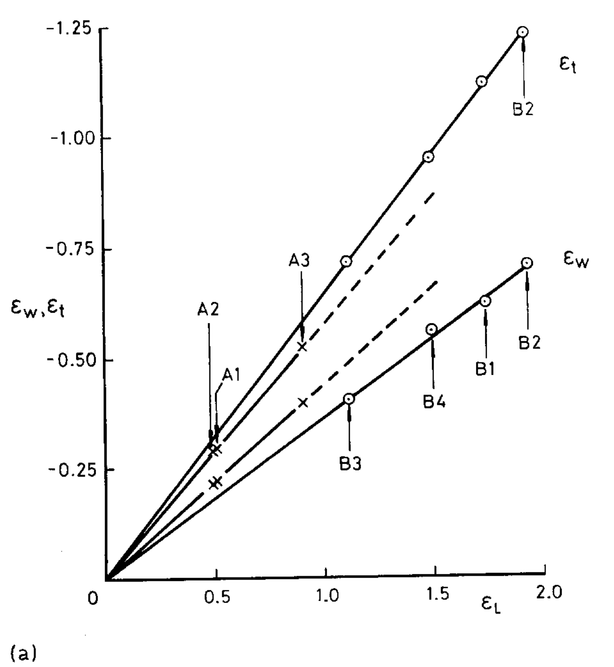
Figure 15 Effect of strain at 3 × 10^{-4} sec^{-1} on R values for each test piece orientation and test temperature . Open symbols: non-banded, closed symbols: banded. (○, ●) 800; (△, ▲) 875; (□, ■) 925; (◇) 975° C.

banded areas of the Ti–6Al–4V used here were influenced by the adjacent, less deformed areas.

The TL test pieces pulled at 975° C showed higher R values, reflecting the low alpha phase proportion and much less directional microstructure at this temperature.

For the ST orientation the alpha grain alignment acted to reduce the thickness strain during superplastic deformation and for test temperatures of 875 and 925° C all the R values were greater than 1.0. However, the test piece surfaces were very irregular, as shown in Fig. 3. Thickness measurements were taken on the surface protrusions and in the depressions at random and, therefore, a wide variation in R was recorded and any trend of R with strain ε_l was masked.

The ST orientation test pieces pulled at 975° C showed relatively low elongation and incipient necking at several locations. As each neck began to develop the deformation tended towards plane strain and the thickness strain increased relative to the width strain. This reduced the local R value, as shown in Fig. 4. However, for a neck to form across the test piece perpendicular to the tensile axis the Von Mises yield criterion predicts [17] that the axial stress required is $\sigma = 2\sigma_f/3^{1/2}$, where σ_f is the flow stress. Therefore, a neck of this geometry cannot develop because the applied stress necessary exceeds that required for general deformation along the gauge length. Under these circumstances final necking occurs along a line inclined to the tensile axis and along which the strain rate can approach zero. For isotropic material this necking line makes an angle of 54.7° with the tensile axis. In both test pieces in Fig. 4 the growth of necks running perpendicular to the tensile axis has been arrested at an early stage; neck development was favoured along a line inclined to the tensile axis. This



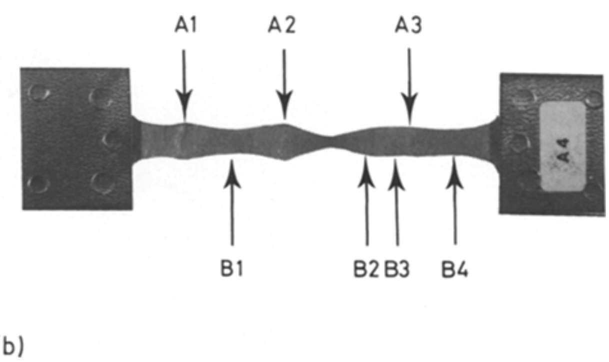


Figure 16 Effect of axial strain ε_l on ε_w and ε_t for a TL test piece pulled at 875° C and 3 × 10^{-4} sec^{-1}. (a) ε_w and ε_t, ε_l. (b) Test piece. The indices A1–A3 and B1–B4 each refer to a set of two points in (a).

type of localized necking also occurred in the 800° C ST orientation test pieces for which the strain rate sensitivity was relatively low (Fig. 14).

The LT orientation test pieces showed a more uniform strain distribution along the gauge length than the corresponding TL and ST test pieces. Consequently, there was less variation in R for each test temperature (Fig. 15). All the R values measured for this orientation were less than 1.0 as a result of the small degree of grain alignment or grain elongation in the T direction. The round test pieces pulled in the L direction at 928° C showed anisotropy in the same sense [1].

The mill annealed (as-received) bar showed an edge texture with basal poles parallel to the transverse direction [2, 18]. Pole figures indicated [18] that superplastic strain reduced both the alpha and beta phase texture intensities, although even after 350% strain in the S direction the textures were not completely randomized. Results [3] on round test pieces from the Ti–6Al–4V bar pulled in the S direction showed that for a test temperature of 800° C the sense of the strain anisotropy reversed for strain rates above about $4 \times 10^{-4}\,\text{sec}^{-1}$. For the sheet test pieces of ST orientation the texture of the bar [18] is such that $\langle 1\,1\,\bar{2}\,0 \rangle$ prism slip would increase the thickness strain at the expense of the width strain and reduce R values, i.e. the effects of texture would oppose those of the banding in the microstructure. The R values measured on the ST orientation test pieces after strain at 800° C and $1.5 \times 10^{-3}\,\text{sec}^{-1}$ ranged from 0.9 to 1.1 [18] indicating that the effects of texture were sufficient to cancel those of microstructure. Similarly, for the LT orientation texture effects tended to reduce R values, but this was only observed in the tests carried out at 800° C and $1.5 \times 10^{-3}\,\text{sec}^{-1}$. No effects of texture on R value are expected for the TL orientation because the tensile axis is parallel to the c axis and, therefore, alpha phase prism slip cannot easily occur and neither thickness strain nor width strain is preferred.

The R value results on the sheet test pieces are, for the most part, in agreement with the diametral strain ratio measurements made [1, 3] on the round test pieces machined from the same bar. An exception to this was the round test pieces pulled in the S direction; those regions of the gauge length containing aligned microstructure were found [1–3] to deform more anisotropically than the non-aligned areas. It was not possible to discriminate between aligned and non-aligned areas when the R value measurements were made on the ST orientation sheet test pieces. However, R values for the different microstructures were recorded for the TL orientation tests and showed, on average, more anisotropic deformation in the non-aligned areas (Fig. 16), i.e. the opposite of the behaviour predicted by the round test piece results. This can be explained by the constraints present in the sheet test pieces. Similarly, the results for the S orientation round test pieces showed [1–3] anisotropy at 970° C in the opposite sense to that indicated in Fig. 15 for the ST orientation sheet test pieces deformed at 975° C. The reason for this is the unusual necking behaviour related to the beta rich phase bands, which influenced the R value measurements sufficiently to reverse the apparent sense of anisotropy. This type of necking, approaching plane strain, can only occur in sheet test pieces. Thus, the use of sheet rather than round test pieces does affect the high temperature anisotropy in this material.

The anisotropy of superplastic deformation of rolled Ti–6Al–4V sheet of various thicknesses has also been investigated [8]. The sheet microstructure was much less directional than that of the bar. Nevertheless, R values as low as 0.5 were measured with the lowest R values occurring in the sheet with the highest alpha phase aspect ratio. Thus, although the rolled sheet microstructures were homogeneous, i.e. without bands of aligned and non-aligned microstructure, the superplastic anisotropy was similar in nature to that of the bar material. The results on rolled sheet [8] and on sheet test pieces machined from bar both suggest that local banding in the microstructure of Ti–6Al–4V could affect the formability under biaxial conditions [19].

5. Conclusions

1. In the sheet test pieces machined from bar the strain rate sensitivity was generally lower and the flow stress higher for the areas of microstructure containing contiguous, aligned alpha grains than for the non-aligned regions.

2. The R values were influenced by the directionality in the microstructure and hence by the test piece orientation. The sense of the superplastic strain anisotropy for each orientation was consistent with the measured order of flow stress at 875 to 975° C; $\sigma_L > \sigma_T > \sigma_S$.

3. For test temperatures above that corresponding roughly to equal phase proportions the strain rate sensitivity is sharply reduced as a result of rapid grain coarsening.

4. For ST orientation test pieces deformed at 975° C strain localization occurred in the beta rich bands perpendicular to the tensile axis. This was probably associated with variations in chemical composition throughout the bar.

5. The superplastic behaviour of the sheet test pieces machined from bar was similar to that of round test pieces from the same bar except for the slant type necking which occurred at 800 and 975° C in the sheet test pieces. Another distinction was that the R values measured on the sheet test pieces were influenced by the test piece heads and by blocks of undeformed material. This type of constraint was not apparent in the round test pieces.

6. Effects of texture on R were only detected at the lowest test temperature (800° C) and at the highest strain rate ($1.5 \times 10^{-3}\,\text{sec}^{-1}$).

Acknowledgement

This paper is published by the kind permission of the Royal Aircraft Establishment, Farnborough, Hants, GU14 6TD, UK.

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Received 20 December 1985
and accepted 11 February 1986