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# Superplastic deformation of strongly textured Ti–6Al–4V

## Part 2 *Changes in microstructure and texture*

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Changes in microstructure and texture during superplastic deformation of strongly textured Ti–6Al–4V bar have been determined in order to establish the cause of stress and strain anisotropy. The effect of strain on the microstructure of the alloy was to cause a progressive break-up, due to grain-boundary sliding, of an initially directional microstructure containing contiguous $\alpha$-phase. The texture of the $\alpha$-phase, however, changed very little with superplastic strain but that of the $\beta$-phase was randomized. Shape changes predicted by permitted deformation modes in the $\alpha$-phase did not correlate with the observed shape changes, whereas the observed anisotropy could be explained by the break-up of the contiguous $\alpha$-phase. A model to account for this anisotropy is described briefly, together with a typical microstructure which should exhibit isotropic superplastic deformation.

### 1. Introduction

In Part 1 [1] the stress and strain anisotropy observed during the superplastic deformation of a strongly textured Ti–6Al–4V alloy was described. The strain anisotropy was manifest as an elliptical cross-section after superplastic strain. The first reported observation of this phenomenon was that of Sherby and co-workers [2, 3], who found that for Zn–Al alloys the ellipticity could be correlated with crystallographic texture. This apparent importance of texture in Zn–Al alloys was then confirmed by Naziri and Pearce [4]. Later, Taplin and co-workers [5–7] showed that Al bronze could also exhibit ellipticity but they could not distinguish between texture and an aligned microstructure as the cause of this type of deformation. Subsequently, Edington and co-workers showed that in a Sn–Pb eutectic alloy [8] and in an Al–Cu–Zr alloy [9] it was the presence of an initially aligned microstructure that was the cause of the ellipticity.

In the present paper we give evidence of microstructural and textural changes to show that the origin of the observed ellipticity in Ti–6Al–4V alloy is also microstructural, i.e. it is due to the presence, in the initial microstructure, of directionality consisting of contiguous $\alpha$-grains.

### 2. Material examination

The alloy composition, test piece configuration and test procedure used for superplastic deformation have been described in Part 1 [1]. To determine the origins of the observed anisotropic deformation, changes in microstructure and $\alpha$-phase texture were determined as described below.

1. The microstructures actually present at the deformation temperature were investigated by re-heating superplastically deformed material to the deformation temperature under argon, holding for 2 to 3 min, and water quenching. The samples were then etch-polished on a 0.5 $\mu$m alumina wheel onto which a solution of 0.5% HF, 6% $HNO_3$ and distilled water was dripped. They were etched using Kroll's reagent. Detailed metallographic analysis of the primary $\alpha$-phase was undertaken in the L-ST plane of the bar (where L is the longitudinal direction and ST the short transverse direction of the bar). The primary $\alpha$-phase grain size in the L and ST directions was determined from

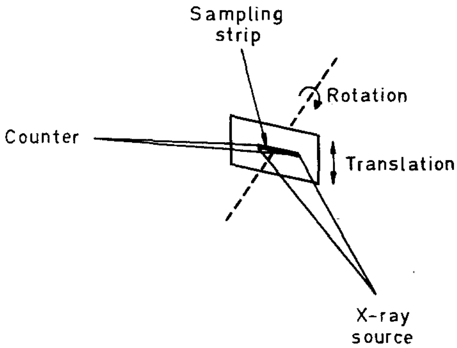


*Figure 1* Specimen movements used to assess the degree of microstructural alignment.

individual measurements of between 350 and 680. grains and the mean value reported as $d_{\text{MLI}}$ (MLI = mean linear intercept). The aspect ratio of the contiguous α-phase was determined from a minimum of 200 separate linear intercept measurements in the L and ST directions. All measurements were made at a magnification of ×750.

2. X-ray diffraction analyses were carried out on electropolished sections cut from as-received material and from the gauge lengths of superplastically deformed test pieces. X-ray intensity measurements were made on a Philips diffractometer fitted with a curved graphite monochromator in the diffracted beam, each specimen being rotated in its own plane. (0 0 0 2) and $(1\,0\,\bar{1}\,0)$ pole figures for the α-phase were obtained on a Philips texture goniometer, the intensity levels being normalized to a randomly textured specimen of similar cross-sectional area. The geometry of the Schulz method [10] used for pole figure measurement is that of a line focus on the specimen (Fig. 1). Use was made of this configuration on the texture goniometer to detect changes in the degree of alignment of the microstructure. Specimens were moved progressively under the X-ray beam (i) by rotation but with no translation, and (ii) by translation but with no rotation (Fig. 1). In both cases the specimens were at the exact Bragg angle for the (0 0 0 2) α-phase reflection i.e. at zero tilt, which corresponds to the centre of the (0 0 0 2) pole figure.

## 3. Microstructural analysis

Typical room temperature microstructures before and after superplastic deformation at 928° C are shown in Fig. 2. The ratio of the volume fraction of the α ($V_\alpha$) and β ($V_\beta$) phases was $V_\alpha/V_\beta$ = 90/10 in both microstructures, but after superplastic deformation the distribution of the α- and β-phases was more uniform and the α-grain size was larger. The α-grain size was dependent on the strain rate, strain and test temperature [11]. However, the microstructure at the test temperature, which controls the hot deformation behaviour, cannot be deduced from the slow furnace-cooled microstructure*. To obtain information on the α-phase grain size at the test temperature, samples of the superplastically deformed material were therefore re-heated to the test temperature and water quenched. This caused the β-phase to transform to acicular secondary α-phase, which can be readily distinguished from the primary α-phase. The short re-heat times did not affect the grain sizes because grain growth is slow under static conditions [11].

In material quenched prior to superplastic deformation, two different α/β microstructures were apparent: an aligned microstructure with the α-phase aligned and contiguous over several grains in the rolling direction of the bar; and a non-aligned microstructure with a uniform distribution of nominally equiaxed α and transformed β (acicular α) grains. We define these two different types of microstructure as bands A and B, respectively. The aligned microstructure etched more rapidly than the non-aligned microstructure to give a banded appearance at low magnifications (Fig. 2). The bands, which were variable in width and length and present throughout the section, may reflect compositional variations. Electron probe micro-analysis did not reveal any differences in chemical composition but X-ray diffraction in the L-ST plane by a back-reflection micro-beam technique indicated that the (0 0 0 2) planes in band A microstructure were more strongly aligned than in band B microstructure.

The effect of superplastic deformation at 982° C and an initial strain rate $\dot{\epsilon}_{\text{I}} = 4.2 \times 10^{-4}\ \text{sec}^{-1}$ on the aligned and non-aligned microstructures in the L-ST plane is shown in Figs. 2 and 3. After a strain of 1.33 (278%) the banding is still visible (Fig. 3c) even though the aligned and contiguous α-phase grains

*The $V_\alpha/V_\beta$ ratio is dependent on the test temperature with $V_\alpha$ decreasing to zero at the β-transus temperature. During slow cooling β-phase transforms to secondary α-phase and some secondary α-phase is also precipitated on the primary α-phase. Primary and secondary α-phase are often indistinguishable. The α-phase grain size after the slow cool to room temperature does not, therefore, indicate the primary α-grain size, volume fraction or distribution at the test temperature.

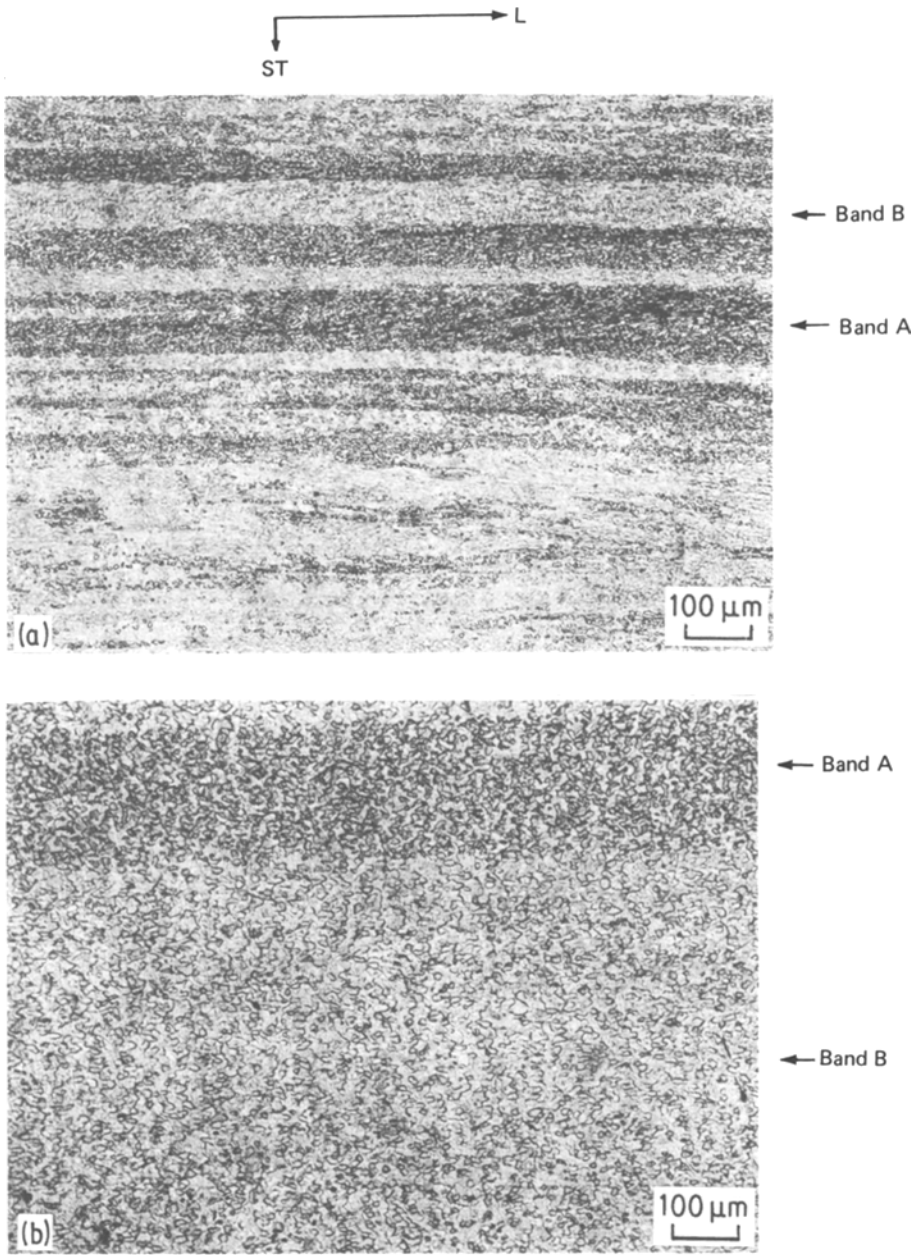


*Figure 2* Typical water-quenched microstructures. (a) Initial microstructure after 0.5 h at 928° C, (b) after a superplastic strain of 1.33 (278%) at 928° C. $\dot{\epsilon}_I = 4.2 \times 10^{-4}\ \text{sec}^{-1}$.

were no longer present in the rapidly etching regions (band A) and the non-aligned microstructure (band B) was unchanged (Fig. 3c). Grain growth occurred in both types of microstructure (Figs. 3a to c).

The effect of superplastic strain on the mean $\alpha$-phase grain size ($d_{\mathrm{MLI}}$) after deformation at, and a water quench from, 928° C is shown in Fig. 4 for both bands A and B microstructures. The ratio $V_\alpha/V_\beta$ at 928° C was 50/50. At 928° C the $\alpha$-grain growth rate/unit strain was similar in bands A and B (Fig. 4). The mean $\alpha$-grain size in the L and ST directions following the interrupted test sequences [1] were similar to those for the continuous tests. The $\alpha$-phase grain size became more equiaxed with increasing superplastic strain at 928° C (Fig. 5).

The effect of superplastic strain on the

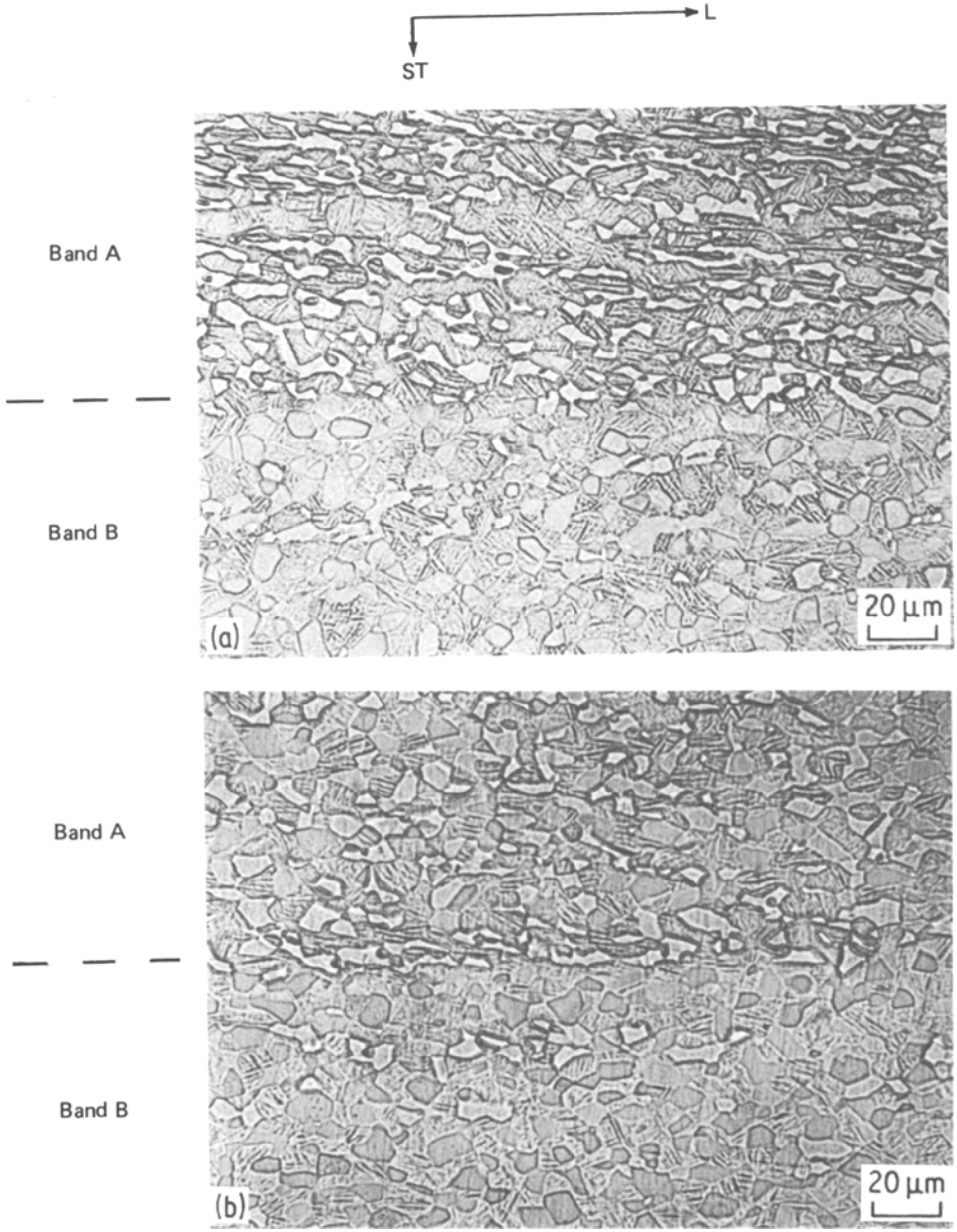


*Figure 3* Effect of superplastic strain at 928° C on the microstructure in the L-ST plane (a) $e = 0$ (0%) (b) $e = 0.68$ (97%) (c) $e = 1.33$ (278%). $\dot{\epsilon}_I = 4.2 \times 10^{-4}\ \mathrm{sec}^{-1}$.

contiguity of the $\alpha$-phase bands A and B is best measured in terms of the contiguous $\alpha$-phase aspect ratio i.e. the ratio of the mean linear intercepts in contiguous $\alpha$-phase *without* regard to $\alpha/\alpha$ grain boundaries. This ratio is plotted against superplastic strain in Fig. 6. The curves show clearly that the contiguous $\alpha$-phase in the aligned microstructure (band A) is broken up after $\approx 0.9$ strain (150%); sufficiently so that at greater strains the ratios for bands A *and* B are similar.

The effect of non-superplastic deformation at 928° C on the microstructure is shown in Fig. 7. The contiguous $\alpha$-phase in band A was unchanged at the higher strain rate and the resultant $\alpha$-grain size after a strain of 1.38 (300%) was less than that after an equivalent amount of superplastic strain (cf. Fig. 7 and Fig. 3c).

## 4. X-ray diffraction analysis

The effect of superplastic strain on the texture of

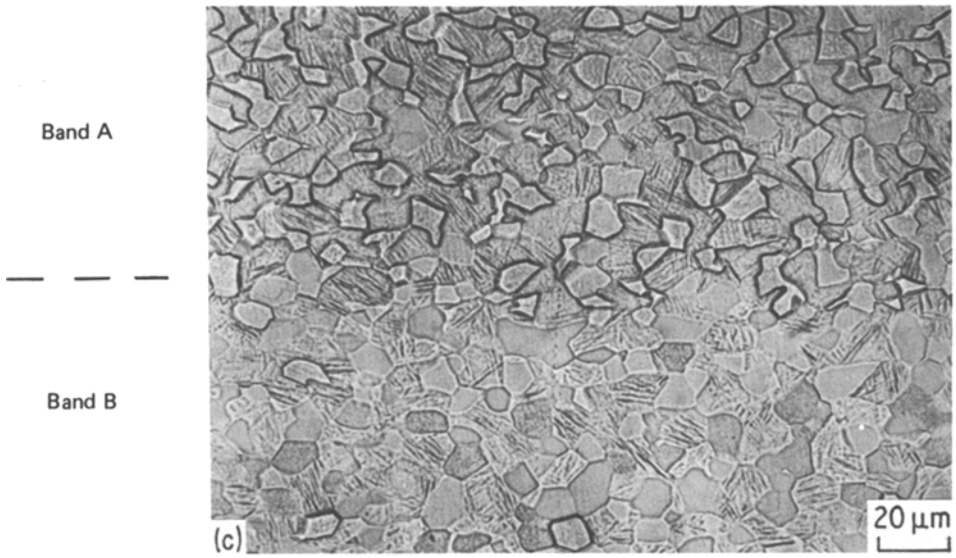


*Figure 3* Continued .

the α-phase in this strongly textured Ti–6Al–4V alloy is shown in Fig. 8. After strains of up to 1.33 (278%) the (0 0 0 2) and ($1 0 \bar{1} 0$) pole figures became more diffuse but the preferred orientation remained essentially unchanged. A more detailed study of the relationship between the texture in the α-phase and the microstructure was carried out using a diffractometer by measuring variations in X-ray intensity ratios with superplastic strain in the L–ST plane. The effect of strain on the intensity of the ($1 0 \bar{1} 0$) reflection (normalized to the ($1 0 \bar{1} 1$) reflection (the strongest reflection in a random sample [12]) is shown in Fig. 9. A similar plot for the ratio of the intensities of the ($1 0 \bar{1} 0$) to the (0 0 0 2) reflection is shown in Fig. 10†. The ratios decrease with increasing superplastic strain

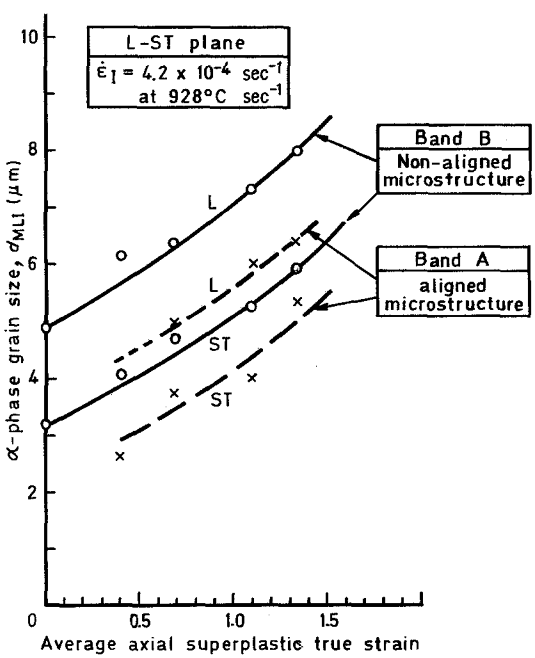


*Figure 4* Effect of superplastic strain on the α-phase grain size ($d_{MLI}$) at 928° C. $\dot{\epsilon}_I = 4.2 \times 10^{-4}$ sec$^{-1}$.

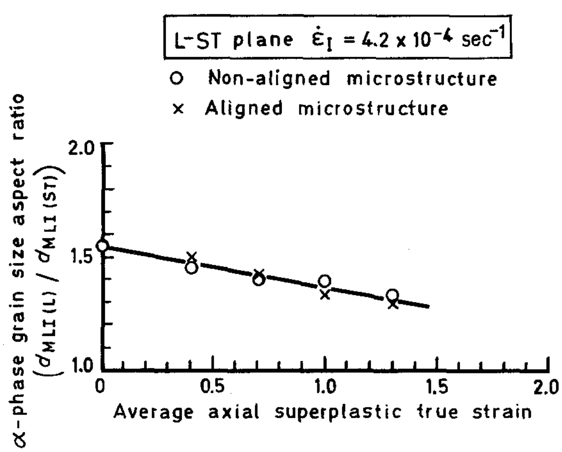


*Figure 5* Effect of superplastic strain on the aspect ratio of the α-phase grain size.

†These curves indicate a more rapid decrease in texture than that shown by the pole figures (Fig. 8). This difference, however, can be attributed to differences in the focusing conditions between the diffractometer and the texture goniometer. Diffractometer analyses correspond only to one region of the pole figure, namely the specimen normal position. Thus small changes in intensity at this position need not necessarily indicate any changes in the overall positions or intensities of poles.

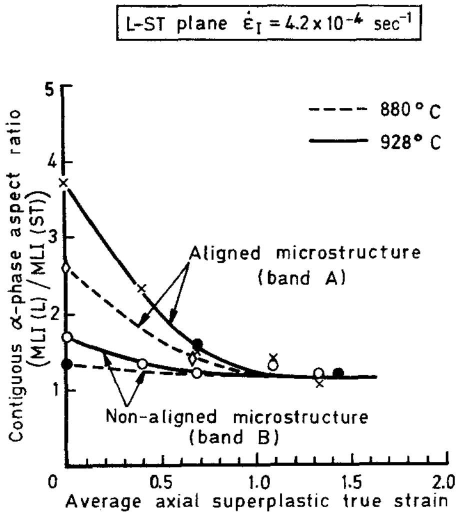


*Figure 6* Effect of superplastic strain and deformation temperature on the contiguous α-phase aspect ratio.

indicating a decrease in the preferred orientation, but the important point to note is that the ratios do *not* decrease to values for a random sample. This is consistent with the pole figure data.

An attempt was made to quantify the effect of increasing superplastic strain on the degree of alignment in band A by monitoring the intensity of the (0002) reflection, $I_{0002}$, from the L-ST plane as the specimen was moved under the line-focus of the X-ray beam when it was mounted on the texture gonimeter (Fig. 1). The variation in the X-ray intensity ($\Delta I$) (Fig. 11) is expressed as $\Delta I = I_{max} - I_{min}$, where $I_{max}$ is the strongest and $I_{min}$ the weakest intensity recorded during a scan (cf. insets, Fig. 11) and indicates that either the α-phase orientation becomes more random or the distribution of the α-phase in the plane of the section becomes more uniform. The latter explanation is consistent with the metallographic and other X-ray diffraction data.

## 5. Discussion

The present results have isolated for the first time the effects of microstructure and texture on the superplastic deformation of Ti–6Al–4V. In Part 1 [1] we showed that in material with a strong basal-edge texture the sense of the observed strain anisotropy did not correlate with that predicted by the deformation modes of the α-phase. The X-ray

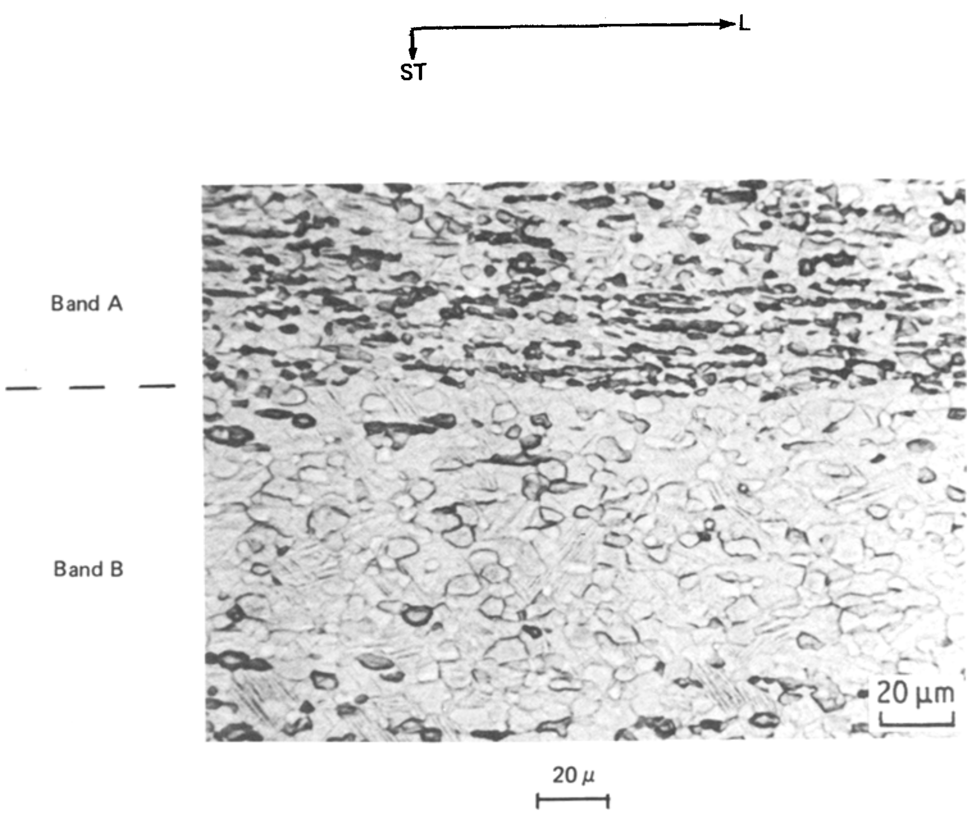


*Figure 7* Typical microstructure after a non-superplastic true strain of 1.38 (300%) at 928° C. $\dot{\varepsilon}_I = 1.05 \times 10^{-2}$ sec$^{-1}$

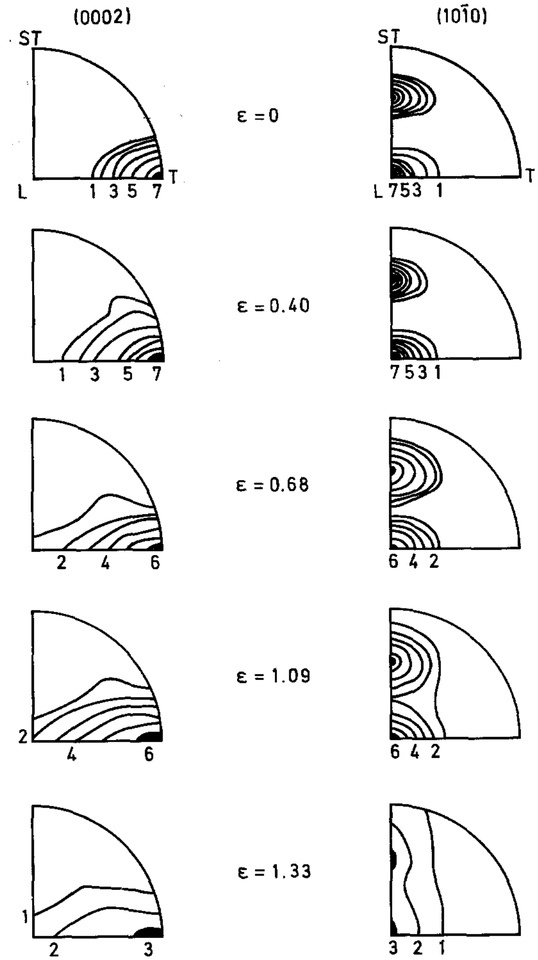


*Figure 8* (0 0 0 2) and (1 0 $\bar{1}$ 0) pole figures before and after varying degrees of superplastic strain at 928° C. $\dot{\epsilon}_I = 4.2 \times 10^{-4}$ sec$^{-1}$.

diffraction and metallographic evidence presented here indicate that this anisotropy is a consequence of having regions of contiguous α-phase parallel to the rolling direction in the initial microstructure. The effect of superplastic strain is to progressively break up the contiguous α-phase but to have little effect on the α-phase texture. Thus after a strain of ≈0.9 the contiguity has been removed, i.e. bands A and B are almost the same (Fig. 6). At this point the strain anisotropy has essentially disappeared, as witnessed by the near unity value for the diameter ratio in interrupted tests [1] (i.e. material becomes more isotropic with increasing superplastic strain).

Subsequent room temperature deformation of test pieces machined from superplastically deformed material gave striking confirmation that the α-phase texture remained because the room temperature anisotropy was the same before and after a strain of 1.38 (300%) [11]. It is clear, therefore, that a small average grain size, high strain rate sensitivity (*m*-value) and any particular room temperature texture are not sufficient prerequisites to ensure isotropic superplasticity in the Ti–6Al–4V alloy.

Previous to this investigation, Paton and Hamilton [13] had concluded that the α-phase texture may not affect the ability of Ti–6Al–4V to deform superplastically, although Kaibyshev *et al.* [14] reported that the presence of a strong texture, as opposed to a random texture, enabled superplastic deformation to occur at lower temperatures and faster strain rates. The present results do not

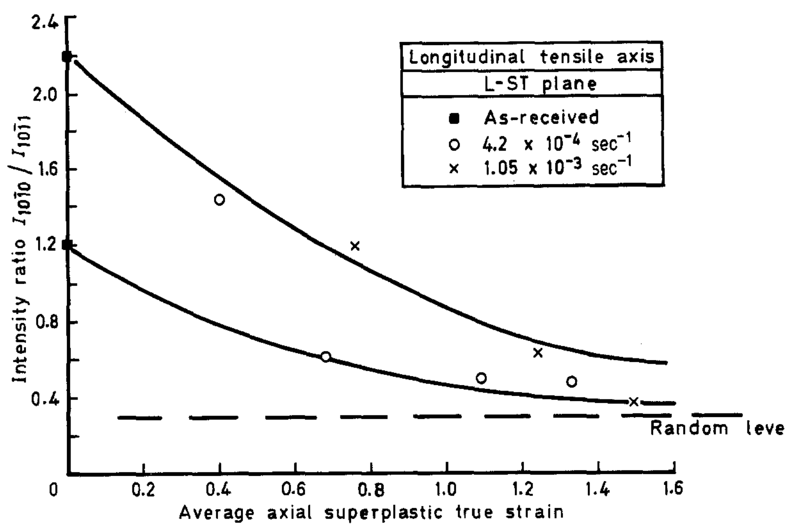


*Figure 9* Effect of superplastic strain and strain rate at 928° C on the intensity ratio $I_{10\bar{1}0}/I_{10\bar{1}1}$. (■Points indicate maxium and minium values recorded for a number of specimens in the as-received condition.)

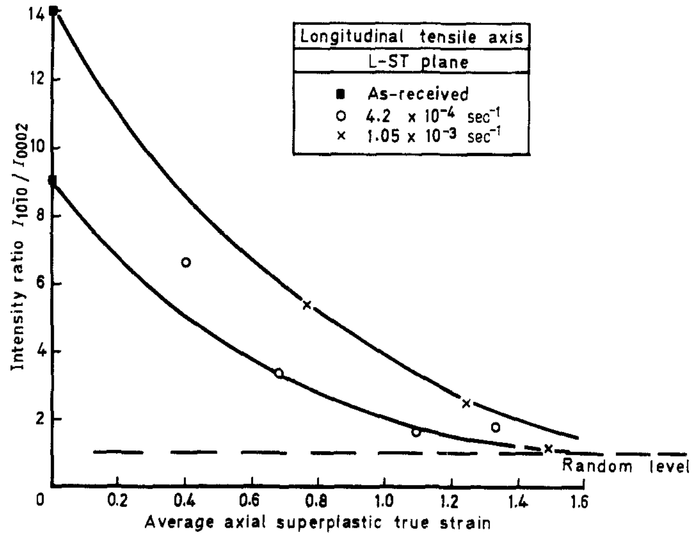


*Figure 10* Effect of superplastic strain and strain rate at 928° C on the intensity ratio $I_{10\bar{1}0}/I_{0002}$. (▪Points indicate maximum and minimum values recorded for a number of specimens in the as-received condition.)

support this latter observation; they do, however, support reports for other systems [8, 9] that anisotropic deformation may be a consequence of a directional microstructure in the initial material.

The effect of superplastic deformation on the crystallographic texture of the Ti–6Al–4V alloy has not been investigated previously. The present results are surprising since they reveal that even after a strain of 1.38 (300%) at 928° C there is only a slight spread in the α-phase poles although their diffracted intensities are reduced. The fact that the room temperature anisotropy was the same before and after a strain of 1.38 (300%) is striking confirmation of the essentially unchanged nature of the α-phase texture. Quotes for only one phase, however, may be misleading as the microstructure contains two phases and these can behave differently. Preliminary results on texture changes in the β-phase reveal that following superplastic strain its texture is randomized whereas following high temperature non-superplastic deformation a strong (1 1 0) fibre texture is developed [15]. Such differences between the texture behaviour of the two phases as a result of superplastic deformation, have also been reported for Zn–22% Al [16] and Zn–40% Al [17] alloys and an Al–33% Cu alloy [18].

From the present results it is concluded that the anisotropic superplasticity in Ti–6Al–4V was caused by the presence of contiguous α-phase. A macroscopic deformation model is described elsewhere [19] which provides a qualitative

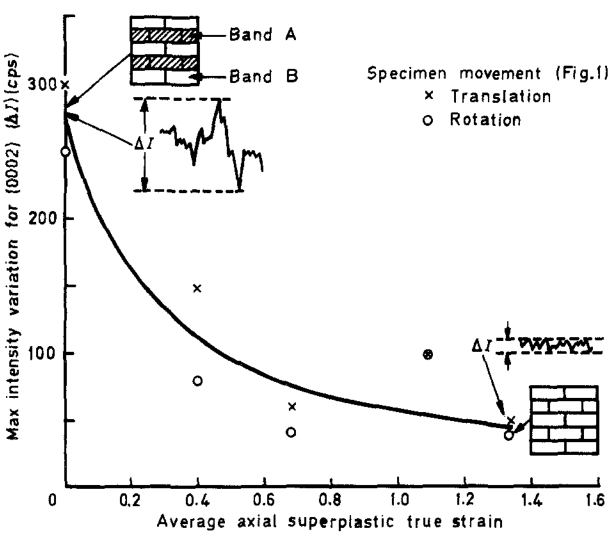


*Figure 11* Effect of superplastic strain at 928° C and $\dot{\epsilon}_I = 4.2 \times 10^{-4}\ \text{sec}^{-1}$ on $\Delta I_{0002}$ in the L-ST plane.

explanation for this anisotropy, both for the present results and for results obtained on other two-phase systems. It depends on the resistance to grain-boundary sliding being inversely proportional to the grain-boundary diffusion rate and predicts that the resistance will increase in the phase interfaces in the order: $\beta/\beta > \alpha/\beta > \alpha/\alpha$. Contiguous $\alpha$-phase, therefore, acts as a barrier to shear and in any particular direction the shear stress will be approximately proportional, and the shear strains approximately inversely proportional, to the length of the contiguous $\alpha$-phase in that direction. The anisotropy will therefore be dependent on the aspect ratio of the contiguous phase as found for Ti–6Al–4V, i.e. $\sigma_L > \sigma_{LT} > \sigma_{ST}$ and $\epsilon_L < \epsilon_{LT} < \epsilon_{ST}$. The model also predicts a Bauschinger effect and flow softening for aligned microstructures under superplastic conditions. Flow softening was not observed in the present tests but has been reported in Ti–6Al–4V by Sastry *et al.* [20].

For isotropic superplastic deformation of thin titanium alloy sheet the microstructure should ideally contain a random distribution of small equiaxed grains. Such a homogeneous microstructure is likely to be difficult to achieve owing to slight compositional variations which occur in the original ingot and also to manufacturing constraints. Nevertheless, the model [19] does suggest that isotropic deformation could be achieved even in an inhomogeneous mixture of two phases if thin overlapping regions containing aligned band A microstructure are arranged parallel to the plane of the sheet. Suitable cross-rolling combined with recrystallization anneals might produce a sheet microstructure which approximates to this sufficiently to ensure uniform thinning during superplastic deformation.

## Conclusions

1. Deformation of the strongly textured Ti–6Al–4V bar under superplastic conditions is anisotropic because of the presence, in the initial microstructure, of an aligned contiguous $\alpha$-phase. It was not dependent on the $\alpha$-phase crystallographic texture.

2. The effect of superplastic strain was progressively to break up the contiguous $\alpha$-phase. At strains >0.9 the banded structure had been removed and deformation became essentially isotropic, in spite of the $\alpha$-phase texture having changed little.

3. X-ray and metallographic data indicated that the mechanism responsible for the break-up of contiguous $\alpha$-phase was grain-boundary sliding.

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