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# Superplastic deformation characteristics of two microduplex titanium alloys

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A comparison of the superplastic deformation behaviour of Ti–6Al–4V (wt%) between 760 and 940° C and Ti–6Al–2Sn–4Zr–2Mo between 820 and 970° C has been carried out on sheet materials possessing similar as-received microstructures. High tensile elongations were obtained with maximum values being recorded at 880° C for Ti–6Al–4V (Ti–6/4) and at 940° C for Ti–6Al–2Sn–4Zr–2Mo (Ti–6/2/4/2), under which conditions both alloys possessed a $\beta$ phase proportion of approximately 0.40. For a given deformation temperature the Ti–6/4 alloy had a slightly lower flow stress than the Ti–6/2/4/2, and this was attributed to the lower $\beta$ phase proportion in the latter alloy. However, at the respective optimum deformation temperatures the Ti–6/2/4/2 alloy had the lower flow stress. The results show that suitably processed Ti–6/2/4/2 alloy is capable of withstanding substantial superplastic strains at relatively low flow stresses, although the optimum deformation temperature is higher for this alloy than for Ti–6/4 material possessing a similar microstructure.

## 1. Introduction

Over the past decade, much interest has been shown in the phenomenon of microstructural superplasticity in titanium alloys [1–5] and in the superplastic forming of components for aerospace applications from Ti–6Al–4V (wt %) alloy (Ti–6/4) [6–8]. Although this alloy is adequate for many applications, there are other titanium alloys which exhibit superior specific mechanical properties such as creep resistance or tensile strength [9]. Several of these alloys are now available in sheet form and have the fine-grain microduplex structure necessary for superplasticity. Such alloys include Ti–6Al–2Sn–4Zr–2Mo (Ti–6/2/4/2), Ti–8Al–1Mo–1V (Ti–8/1/1) and Ti–4Al–4Mo–2Sn–0.5Si (IMI 550). Both Ti–8/1/1 and IMI 550 have higher strengths than Ti–6/4, while Ti–6/2/4/2 has a higher temperature capability and may be used up to 450° C compared with 350° C for Ti–6/4.

To date, very few detailed studies of the influence of temperature and strain rate on the superplastic tensile behaviour of titanium alloys have been published. Consequently, the present work was initiated to examine the superplastic deformation potential of Ti–6/2/4/2 sheet material and to compare this with the corresponding behaviour in Ti–6/4 sheet of similar microstructure.

## 2. Experimental procedure

Tensile specimens of 10 mm gauge length and 5 mm gauge width were machined from 0.8 mm thick Ti–6/4 and Ti–6/2/4/2 sheets with the tensile axis parallel to the rolling direction. The alloy compositions are shown in Table I. Each alloy was deformed to failure at a constant strain rate of $2.33 \times 10^{-4}\,\mathrm{sec}^{-1}$ ($1.4\%\,\mathrm{min}^{-1}$) at 30° C intervals between 760 and 940° C for Ti–6/4, and 820 and 970° C for Ti–6/2/4/2. At 940° C for both alloys and at 880° C for Ti–6/4 only, the materials were deformed to failure at constant strain rates in the range $2.33 \times 10^{-4}\,\mathrm{sec}^{-1}$ to $2.33 \times 10^{-3}\,\mathrm{sec}^{-1}$. At 940° C, strain rate change tests were performed to determine the log(stress)–log(strain) rate characteristics of the materials, and hence the variation of strain rate sensitivity of flow stress, $m$, as a function of strain rate.

All elevated-temperature tensile tests were performed in a pure, dry argon atmosphere in a split furnace with a temperature variation of $\pm 2°$ C over a length of 60 mm. Prior to deformation, the material was subject to a 30 min heating period and 10 min hold period at the test temperature.

Equilibrium $\beta$ phase proportions were determined after holding the material at temperature for 30 min followed by water quenching. The volume fraction of $\beta$ phase was measured using a Swift semi-automatic point counting device with 300 points being counted.

## 3. Results

### 3.1. Microstructure

The microstructures of the as-received materials are shown in Fig. 1. Both materials possessed plate-like $\alpha$ grains, the dimensions of which are given in Table II. After preheating to the test temperature, the clusters of $\alpha$ grains present in the as-received materials were still present in Ti–6/2/4/2 but were almost completely

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TABLE I Alloy compositions

| Alloy | Composition (wt %) | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | Al | V | Fe | O | N | C | Sn | Zr | Mo | Ti |
| Ti-6/4 | 5.94 | 3.92 | 0.20 | 0.12 | 0.002 | 0.02 | NA* | NA* | NA* | Bal |
| Ti/6/2/4/2 | 5.80 | NA* | NA* | 0.10 | 0·009 | 0·01 | 2.0 | 4.3 | 1.9 | Bal |

*NA = not analysed.

removed in Ti-6/4. Both materials exhibited very little banding, and any that did exist initially was removed during the preheat period.

The variation of $\beta$ phase proportion with temperature for the two alloys is shown in Fig. 2. It may be seen that the Ti-6/4 alloy possesses a higher $\beta$ phase proportion than Ti-6/2/4/2 at all temperatures up to at least 970° C

## 3.2. Mechanical behaviour

Elongation to failure results for materials deformed at a constant strain rate of $2.33 \times 10^{-4}\,\text{sec}^{-1}$ at various temperatures are shown in Figs 3a and b. The results shown that at this strain rate, the optimum deformation temperatures are $\sim$ 880° C and 940° C for Ti-6/4 and Ti-6/2/4/2, respectively. However, when elongation is plotted against $\beta$ phase proportion, the maximum elongation coincides with approximately 40% $\beta$ phase for each alloy (Fig. 4).

The stress-strain behaviour of the two alloys deformed at $2.33 \times 10^{-4}\,\text{sec}^{-1}$ at 30° C intervals is shown in Figs 5a and b. At each temperature, the flow stress is higher for Ti-6/2/4/2 than for Ti-6/4 although the difference diminishes as the temperature is increased. At 820 and 850° C, the Ti-6/2/4/2 alloy exhibited strain softening up to a true strain of approximately 0.2. The stress-strain rate data at 940° C also shows that the flow stress is higher for Ti-6/2/4/2, although the difference is not large, in agreement with the observations of constant strain rate behaviour at 940° C (Fig. 6). The variation of stress and strain rate sensitivity ($m$ value) with strain rate is also shown in Fig. 6. It can be seen that slightly lower $m$ values are exhibited by Ti-6/2/4/2 and the maximum $m$ value for the alloy coincides with the lower strain rate end of the broad peak of maximum values of $m$ shown by Ti-6/4. Both alloys exhibit $m$ values greater than 0.7 close to the optimum strain rate.

The results of elongation to failure tests at 940° C carried out at various constant strain rates are shown in Fig. 7. At all strain rates in the range $2.33 \times 10^{-4}\,\text{sec}^{-1}$ to $2.33 \times 10^{-3}\,\text{sec}^{-1}$, Ti-6/2/4/2 exhibits higher elongations to failure. However, the elongations are not as high as those for Ti-6/4 at 880° C, the temperature at which this alloy contains 42% $\beta$ phase. An additional observation is that the optimum strain rate

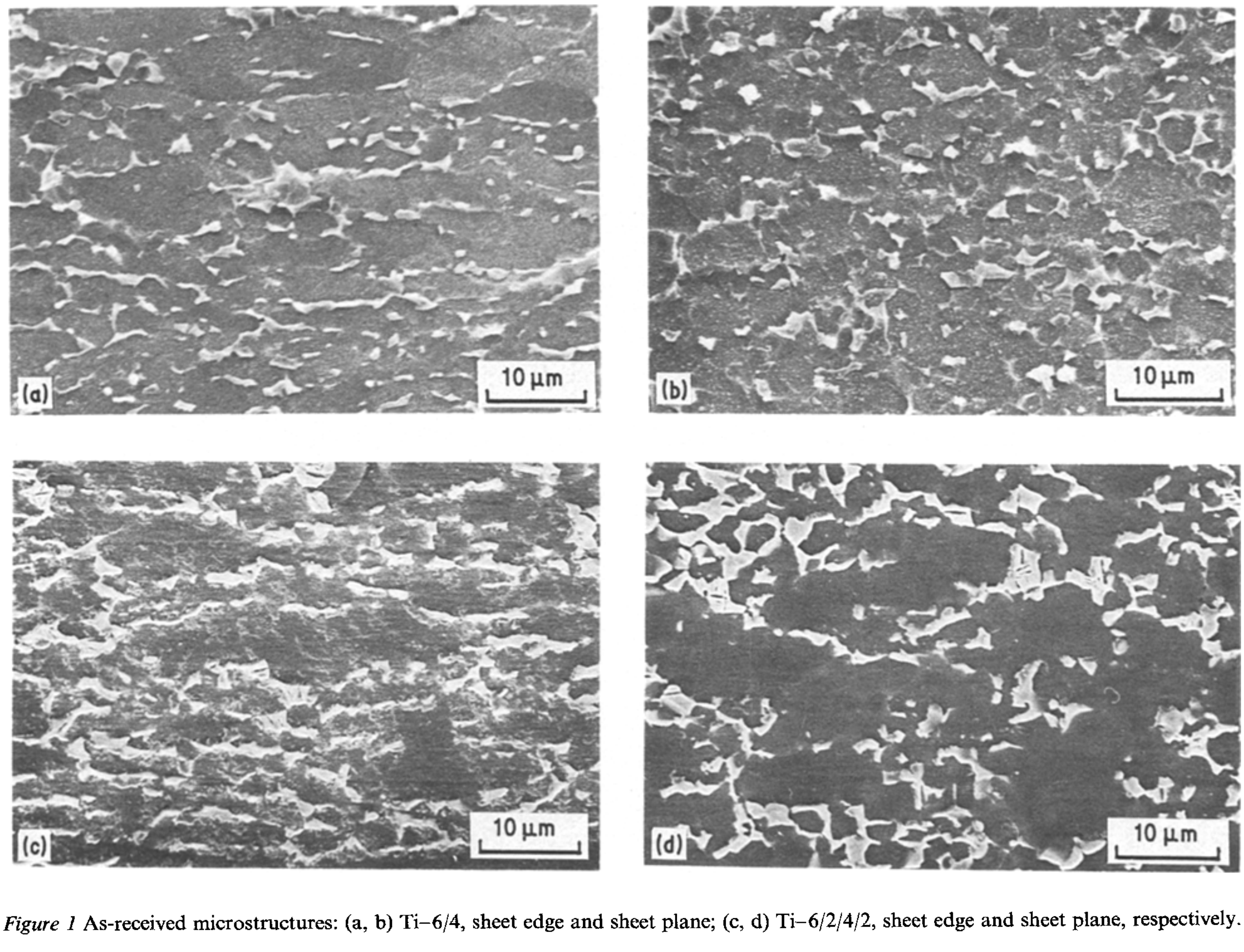

*Figure 1* As-received microstructures: (a, b) Ti-6/4, sheet edge and sheet plane; (c, d) Ti-6/2/4/2, sheet edge and sheet plane, respectively.

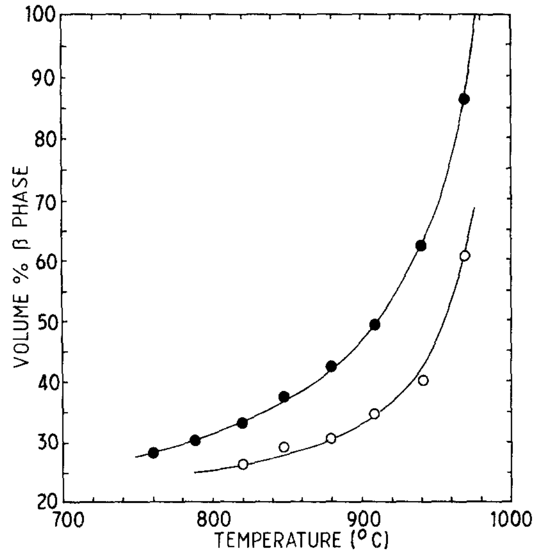

*Figure 2* Proportion of $\beta$-phase against temperature for (●) Ti–6/4, (○) Ti–6/2/4/2.

for maximum elongation for Ti–6/4 at 880° C is lower than the corresponding strain rate for Ti–6/2/4/2 at 940° C (the optimum deformation conditions for each alloy).

## 4. Discussion

The sheet materials chosen for the present work were both 0.8 mm thick. Hence, the superplastic elongations exhibited by each alloy could be compared directly in the absence of a specimen geometry variable, which has a significant influence on the elongation to failure in superplastic materials [3, 11]. Both alloys possessed similar as-received microstructures (Fig. 1). Therefore, any differences in superplastic behaviour must be attributable to differences in microstructure at elevated temperatures and to differences in alloy composition.

Metallographic studies of specimens of Ti–6/4 and Ti–6/2/4/2 which had been quenched from the $(\alpha + \beta)$ phase field revealed that the latter alloy possessed a considerably lower $\beta$ phase proportion than Ti–6/4 at any of the temperatures studied in the present work (Fig. 2). However, at high temperatures (> 940° C) it can be seen that the $\beta$ phase volume fraction is increasing rapidly with temperature for the Ti–6/2/4/2 alloy. One possible reason for this is that although Ti–6/2/4/2 has a lower concentration of $\beta$-stabilizing additions, molybdenum is a stronger $\beta$ stabilizer than vanadium at high temperatures [12].

TABLE II Mean linear intercept values for as-received alloys*

| Alloy | $L$ (μm) | $T$ (μm) | $ST$ (μm) |
|---|---|---|---|
| Ti–6/4 | 4.8 | 4.7 | 4.4 |
| Ti–6/2/4/2 | 4.7 | 4.5 | 4.2 |

*$L$ = longitudinal, $T$ = transverse, $ST$ = short transverse.

Previous studies of superplasticity have shown that the optimum deformation temperature is that at which approximately equal phase proportions exist [13]. Consequently, it would be expected that the optimum superplastic deformation temperature for Ti–6/2/4/2 would be significantly higher than for Ti–6/4. Elongation to failure studies for each alloy at a strain rate of $2.33 \times 10^{-4}\,\text{sec}^{-1}$ supported this view but also showed in each case that the optimum temperature coincided with approximately 40% $\beta$ phase (~880° C for Ti–6/4 and ~940° C for Ti–6/2/4/2), rather than equi-volume proportions (Figs 3 and 7).

At low deformation temperatures, the elongations to failure increased with increasing temperature due to the increase in diffusivity and proportion of $\beta$ phase. These two factors gave rise to a rapid increase in elongation to failure (Figs 3a and b). However, as the temperature was increased above the optimum, the further increases in diffusivity and $\beta$ phase proportion are likely to have resulted in excessive grain growth causing a reduction in strain rate sensitivity and, hence, in elongation to failure.

Recent work [14–17] has shown that it is not only the volume fraction of phases present that is important for high superplasticity but also the composition of the titanium $\beta$ phase. It has been known for some time that many elements are capable of stabilizing the $\beta$ phase in titanium alloys [18], but not all elements improve the superplastic response. Paton and Hall [17] examined the effect of additions of the $\beta$ stabilizers iron and molybdenum to Ti–6Al alloy. The addition of 2 wt % iron, an element which has a diffusivity 32 times greater than that of vanadium in $\beta$-titanium at 871° C, reduced the flow stress and increased the strain rate sensitivity. However, molybdenum, which diffuses five times slower than vanadium, increased the flow stress and decreased the strain rate sensitivity compared to the base composition Ti–6/4. The authors did not report any $\beta$ phase proportion

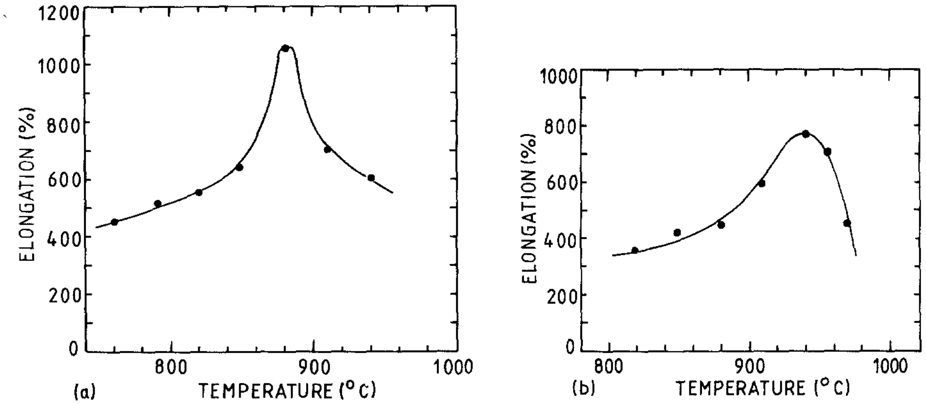

*Figure 3* The variation of elongation to failure at a constant strain rate of $2.33 \times 10^{-4}\,\text{sec}^{-1}$ against temperature for (a) Ti–6/4 and (b) Ti–6/2/4/2.

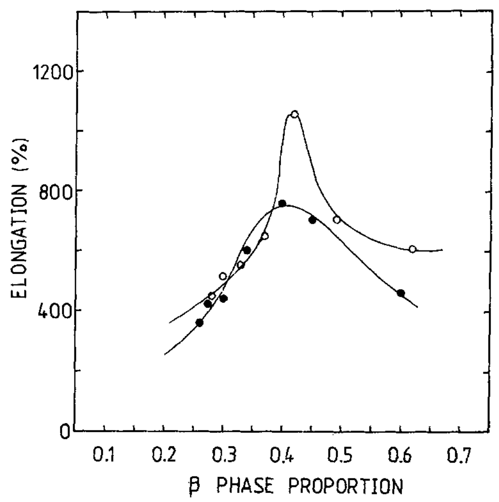

*Figure 4* Elongation to failure at a constant strain rate of 2.33 × $10^{-4}$ $\text{sec}^{-1}$ against $\beta$ phase proportion for (○) Ti–6/4, (●) Ti–6/2/4/2.

values for the alloys and assumed that the superplasticity differences could be attributed to diffusivity changes.

The elongation results obtained in the present work are in general agreement with the effects reported by Paton and Hall [17] and are consistent with the view that the presence of molybdenum, rather than a faster-diffusing element such as vanadium or iron, reduces the superplasticity of titanium alloys. A comparison of the flow stresses during both constant strain rate and strain rate change testing reveals that the flow stress for Ti–6/2/4/2 is higher than for Ti–6/4 for each test temperature and strain rate studied in the present work (Figs 5 and 6). The reasons for this observation are the lower proportions of highly deformable $\beta$ phase in Ti–6/2/4/2 and the presence of molybdenum. The influence of $\beta$ phase proportion on the superplasticity of Ti–6/4 has been discussed elsewhere [19]. The most important observation is that the flow stress decreases rapidly with increasing $\beta$ phase proportion.

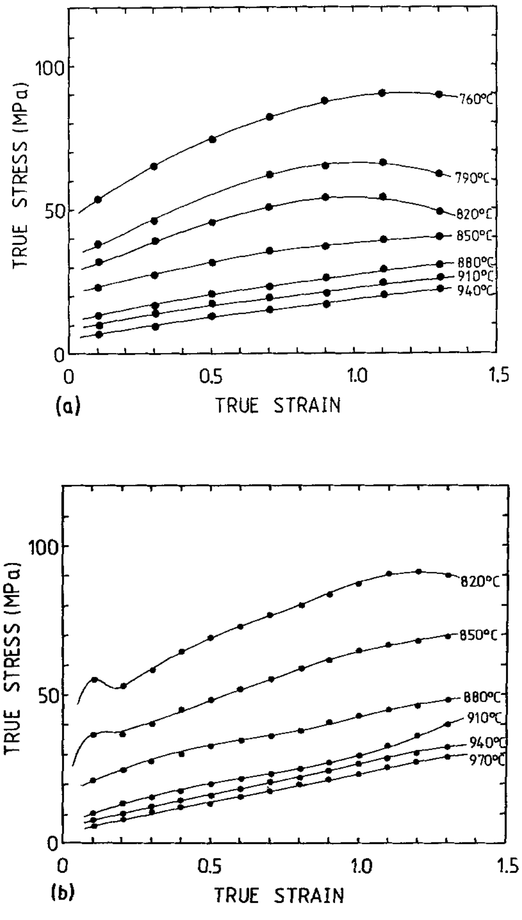

*Figure 5* Stress–strain behaviour at a constant strain rate of 2.33 × $01^{-4}$ $\text{sec}^{-1}$ for (a) Ti–6/4 and (b) Ti–6/2/4/2.

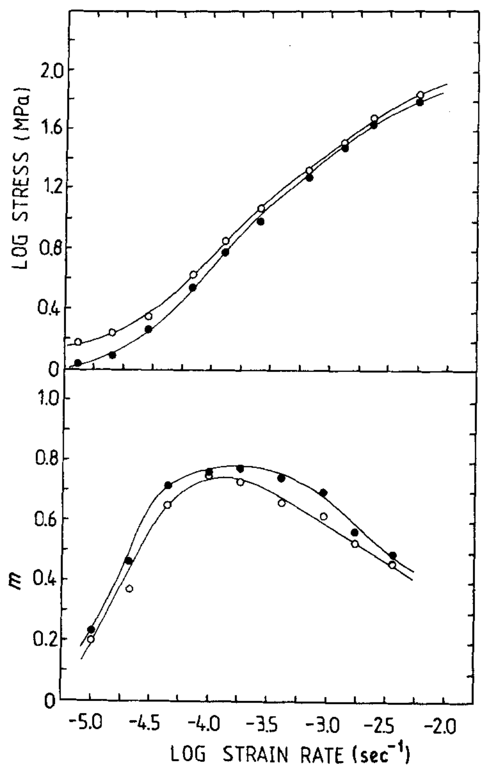

*Figure 6* Flow stress and strain rate sensitivity, *m*, against strain rate at 940° C for (●) Ti–6/4 and (○) Ti–6/2/4/2.

At 820 and 850° C, for a strain rate of 2.33 × $10^{-4}$ $\text{sec}^{-1}$, the Ti–6/2/4/2 alloy exhibited strain softening up to a true strain of 0.2 (Fig. 5b). It is thought that this observation could be due to the breakdown of clusters of $\alpha$ grains which existed in Ti–6/2/4/2, but which were removed during the preheat period in Ti–6/4 alloy. At higher temperatures, a strain softening effect is not observed, or is masked by the more rapid occurrence of grain growth which leads to strain hardening. Fig. 7 shows the variation of elongation with strain rate at the optimum superplastic temperature for each alloy. It may be seen that the highest elongations are exhibited at higher strain rates for Ti–6/2/4/2 than for Ti–6/4. This observation may be attributed to a higher effective diffusivity at 940° C for Ti–6/2/4/2 than at 880° C for Ti–6/4. When the elongations of the two alloys are compared at 940° C, it is observed that the optimum strain rate is approximately 7.5 × $10^{-4}$ $\text{sec}^{-1}$ in each case. As a consequence of the increased effective diffusivity at 940° C for Ti–6/4, the optimum strain rate is displaced to a higher value as reported for other alloy systems [20]. However, the situation will be complicated by grain growth, particularly for the Ti–6/4 alloy since 940° C is above its optimum deformation temperature.

One possible consequence of the lower $\beta$ phase

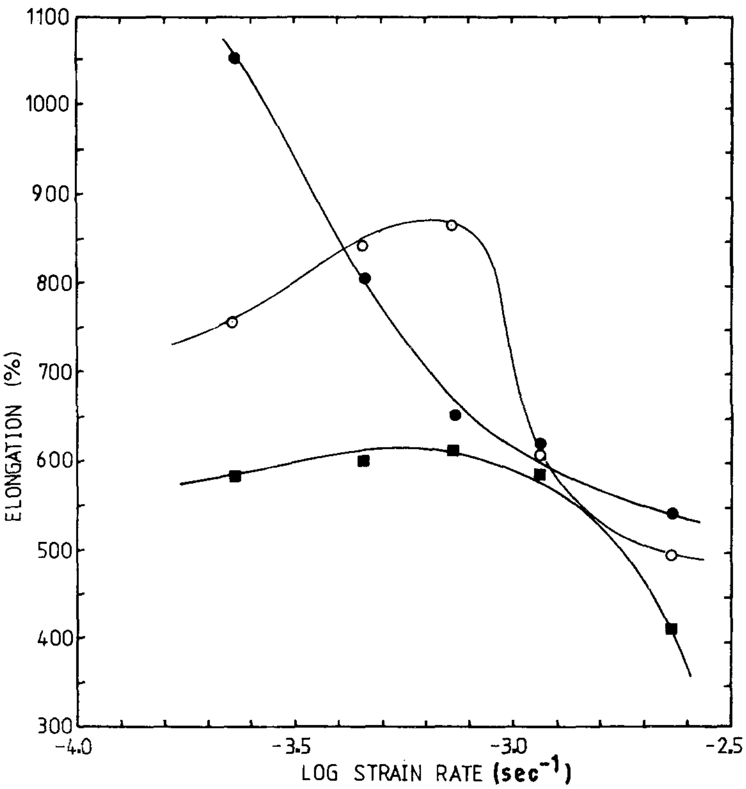

*Figure 7* Elongation to failure against strain rate at (●) 880 and (■) 940° C for Ti–6/4 and (○) at 940° C for Ti–6/2/4/2.

proportion in Ti–6/2/4/2 alloy for a given temperature is that cavitation may occur more readily during superplastic flow than for Ti–6/4. Preliminary studies have shown the presence of cavities situated at $\alpha/\alpha$ and $\alpha/\beta$ boundaries away from the fracture surface in Ti–6/2/4/2 deformed to failure at 880° C [21]. This compares with a maximum temperature of approximately 850° C below which cavities have been observed in Ti–6/4 alloy [11]. It must be emphasized that the cavitation levels observed in failed specimens are extremely low compared with those which have been reported for iron, aluminium and copper alloys [22], and would be even lower for the temperatures and strains generally used in commercial practice.

The results of the present work show that Ti–6/2/4/2 can undergo substantial superplastic deformation provided that the material is deformed under the optimum conditions of temperature and strain rate. For the materials used in this study, the optimum temperature of 940° C for superplastic deformation of Ti–6/2/4/2 was approximately 60° C higher than that for Ti–6/4. However, it is the $\beta$ phase proportion at the deformation temperature that appears to be the important parameter rather than the temperature itself. For example the $\beta$ phase proportion at 940° C has been observed to vary from 48 to 62% for various batches of Ti–6/4, and from 38 to 45% for Ti–6/2/4/2 alloys [11, 21].

## 5. Conclusions

1. Large superplastic strains were exhibited by microduplex Ti–6/4 and Ti–6/2/4/2 alloys when the materials were deformed under the optimum conditions of temperature and strain rate.

2. Although the optimum deformation temperature for the Ti–6/4 alloy was approximately 60° C lower than for Ti–6/2/4/2, the highest elongations are obtained at temperatures where each of the alloys contains approximately 40% by volume of $\beta$ phase.

3. For each temperature and strain rate the flow stress of Ti–6/2/4/2 was higher than that of Ti–6/4 studied. This was attributable to the lower proportions of $\beta$ phase in Ti–6/2/4/2 and to the slow diffusivity of molybdenum atoms in the alloy.

4. The high superplastic strains observed for the Ti–6/2/4/2 sheet showed that it was a suitable material for superplastic forming into components which required an improved creep resistance over that given by Ti–6/4.

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