Superplastic Behavior of Ti-6Al-4V and β-Enhanced Iron-Modified Ti-6Al-4V

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Abstract

The superplastic behavior of Ti-6Al-4V and iron-modified Ti-6Al-4V has been characterized in terms of: 1. a weighted average diffusivity for the β phase which emphasizes the faster diffusing species and takes into account alloy element partitioning, 2. a \( d^2 \) dependence on the α grain size and 3. a dependence on the volume fraction of the α phase which can be expressed as \( (1-f_\alpha)^q \). This behavior is in general agreement with Chen’s model except that \( q \) is approximately a factor of two low. This is probably due to grain boundary sliding which is in addition to the dislocation creep of the softer phase which Chen’s model assumes. Also, deformation which is dominated by either the α phase (\( q<0 \)) or the β phase (\( q>0 \)) has been observed.
Introduction

The industrial and commercial importance of Ti-6Al-4V is well known. It is well established in aerospace, marine, chemical and other industries and accounts for approximately half of the titanium produced. There is also an ever increasing interest in the superplastic properties of titanium alloys due to the savings in weight and cost and the potential for improved designs. In this respect Ti-6Al-4V is in effect a standard to which other superplastic titanium alloys are compared and efforts to exceed this standard include developing completely new alloys, finding new ways to process the existing alloy, or to make minor changes in the composition of Ti-6Al-4V. Regarding altering the composition of Ti-6Al-4V, two strategies have been tried: adding a fine dispersion of inert oxides which would stabilize grain growth and modifying specifically the properties of the β phase to lower forming temperature and increase the forming rate. The latter involves additions of 1-3 percent Ni, Co, Fe and/or other high-diffusivity β-stabilizers.

Aside from the potential for improved superplastic properties, these β-modified alloys offer a unique opportunity to more fully understand the superplastic behavior of the base Ti-6Al-4V alloy, for although Ti-6Al-4V is a commercial success, its relatively complicated and variable (at superplastic temperatures) microstructure have made it difficult to determine the fundamental rate controlling mechanism(s) responsible for its renowned superplastic properties. Other than the tendency for concurrent grain growth, the basic problem seems to be in the fact that it is a two phase alloy in which the properties of each phase are supposedly quite different and where the phase proportion varies significantly, particularly in the superplastic temperature regime. A series of β-modified Ti-6Al-4V alloys can provide different phase proportions at any given temperature. Also, since the β-modifying elements, which are eutectoid formers, generally have very low solubilities in the β phase the composition (and properties) of the α phase will be relatively fixed in all cases. As a result it should be easier to characterize the base alloy and a series of β-modified alloys than the base alloy alone. This is because it allows one to differentiate between the temperature dependence of phase proportion and composition and the temperature dependence of the kinetics of the deformation mechanism.

In this paper we describe the characterization of the superplasticity of Ti-6Al-4V and iron-modified Ti-6Al-4V in terms of the Mukherjee, Bird, Dorn [1] equation:

\[ \frac{\dot{\epsilon}kT}{D\sigma} = A\left(\frac{\sigma}{G}\right)^n\left(\frac{b}{d}\right)^p \]  

\[ n \] the stress exponent, \( p \) is the grain size sensitivity, \( d \) is the grain size, \( b \) is the magnitude of the Burger's vector, \( G \) is the shear modulus and \( D \) is the diffusivity. The dependence of the stress exponent \( n \) on the volume fraction of the β phase \( f_\beta \) is investigated and an estimate of \( D \) is made using published data on the tracer diffusivities, equilibrium phase diagrams and on our experimental measurements of \( f_\beta \). The remaining dependence of the deformation behavior on temperature is then analyzed in terms of the mechanics of deformation of two-phase materials. The grain size dependence is not investigated.
Procedure

Material

Fine grained 1.6 mm sheet Ti-6Al-4V was supplied by Northrop Aerospace (Los Angeles, CA). The microstructure of the as received material was elongated but after annealing at 925°C for 2 hours, the microstructure consisted of approximately 50% β phase and equiaxed α grains. Five Ti-6Al-4V alloys containing controlled amounts of iron were supplied by Timet (Henderson, Nevada). The concentration of iron ranged from 0.115 to 3.04 weight percent. The microstructure of the five alloys consisted of different amounts of the β phase but the α grain size ranged only from 3 to 6 µm. The dependence of $f_\beta$ on temperature and concentration of iron was determined and have been reported elsewhere [2].

Equipment

All testing was conducted using an automated MTS system which had been outfitted with a radiant heating furnace capable of rapid temperature changes and a retort which allowed testing to be done in an atmosphere of purified argon. The testing software and servo-hydraulic system was capable of providing up to 20 true strain rates in a single test with total strains as high as 2.0. Specimens used for testing had dimensions of 25.4 mm in length and 6.3 mm wide and 1.6 mm thick in the gage section.

Stepped-Strain Rate Tests

A series of stepped-strain rate tests was conducted to determine the value of the stress exponent for each alloy. Each test consisted of an initial 20% true strain at a strain rate of $2 \times 10^{-4}$ s$^{-1}$ after which the strain rates were changed incrementally, at intervals of 0.05 true strain. After reaching the maximum strain rate of $1 \times 10^{-3}$ s$^{-1}$ the strain rate was decremented until the original strain rate was obtained. The total strain on any specimen was only 0.54. This range of strain rates was chosen because it was expected, based on our previous results for Ti-6Al-4V, that region II behavior would be observed for all alloys. Three temperatures were investigated: 775°C, 825°C and 875°C. It was in this temperature range that, for all alloys, $f_\beta$ was relatively insensitive to temperature but was different for each alloy.

Stepped-Temperature Tests

A series of stepped-temperature tests was conducted at a strain rate of $2 \times 10^{-4}$ s$^{-1}$ and temperatures between 695 and 879°C. Each test consisted of an initial 20% true strain at 810°C followed by an abrupt increase in temperature of 10-15°C every 0.03 strain. After reaching a temperature of 879°C (or slightly less, depending on the alloy) the temperature was decremented by 10-15°C until the temperature reached 695°C or, judging from the stress-strain curve, the specimen appeared to have started to neck.

Results and Discussion

Strain Rate Dependence

For all six alloys the value of $n$ was consistently close to 2. Note that while the volume fraction of phases is very different for each alloy the value of $n$ was essentially constant. This showed that, in the present temperature range, $n$ is not effected by $f_\beta$. 1423
Temperature Dependence

Figure 1 shows an Arrhenius-type plot of the results of the stepped-temperature tests. The six lines are curved and approximately parallel, shifting to higher temperatures as the concentration of iron decreases. These results, however, require two corrections (or compensations) before one can begin to estimate the activation energy. The first is the effect of the small differences in the initial grain size and the expected variation in grain size with temperature. This was done by assuming $p = 2$ and that small variations of $d$ with temperature are such that the number of grains is constant but their size varies to produce the required value in $f_B$:

$$d = d_0 \sqrt[3]{\frac{f}{f_0}}$$  \hspace{1cm} (2)

where $f_0$ and $d_0$ are measured, reference values of the volume fraction and grain size of the same phase. In this the α phase is considered primarily because it is practically impossible to resolve the β/β grain boundaries. (Note that this adds an additional $f(T)^{-p/3}$ term to equation 1.)

The second correction concerns the expected variation of $D$ as a result of alloy composition and alloy element partitioning. This was done by estimating the composition of the β phase at different temperatures and replacing $D_0$ in the diffusivity term $D$ in equation 1 with the equivalent term from the average diffusivity for the β phase:

$$D^{\alpha\text{V}}_\beta = \sum_j x_j D_\alpha \exp\left(\frac{Q_j}{kT}\right)$$  \hspace{1cm} (3)

![Figure 1. Results of the stepped-temperature tests plotted in an Arrhenius fashion.](image)
where \( x_i \) is the concentration of the \( i \)th of \( j \) elements. (See reference 2 for details.) As a result of these two corrections (figure 2) the lines for all alloys at temperatures below 780°C have condensed to one line whose slope indicates that the activation energy is 210 kJ/mole, a value typical for the superplasticity of titanium alloys [3]. At higher temperatures the line for the base alloy does not deviate but those for the other alloys branch upwards with a slope suggesting the activation energy is 485 kJ/mole.

While the correlation among the different alloys is excellent, both of these activation energies are too high to associate with a conventional diffusion controlled mechanism. Note, however, that the lower value happens to be 1.6 times the activation energy expected from \( D^A \) and that this is equal to the sum of the activation energies for lattice and grain boundary diffusion, \( \phi^L \) and \( \phi^G \), respectively. Therefore we split this into two parts, the activation energy for lattice diffusion stays with the diffusivity term in equation 1, which is in agreement with Suery and Baudelet's model [4], and the remaining \( \phi^G \) represents the temperature dependence of the \( A \)-parameter. At this point this seems to be the only way to deal with such high activation energies. What we are basically proposing is that there appear to be two mechanisms operating and that they interact in such a way that the observed deformation behavior is the product of these two processes (the activation energies are additive) which is different from the series/parallel approach to accounting for the simultaneous contributions of multiple deformation processes. Following the conventional way on thinking, the most likely explanation is that these processes involve the deformation mechanism itself plus strain enhanced microstructural evolution.

Figure 2. Results of stepped-temperature tests corrected for minor variations in grain size, differences in alloy composition and alloy element partitioning.
Volume Fraction Dependence

As stated in the previous section we find that the A-parameter has its own temperature dependence given by:

\[ A = A_0 \exp\left(-\frac{Q_{AV}}{kT}\right) \]  \hspace{1cm} (4)

Figure 3 shows that the pre-exponential term in the A-parameter, \( A_0 \), follows the \((1-f_\beta)^q \) relation suggested by Chen [5] except in this case the values of \( q \) are low by a factor of nearly 2. Chen's model says that when \( n=2 \), the values of \( q \) should be -3.1 or 3 depending on which phase is considered the matrix. A value close to 4 (3.785) was found to fit Suery and Baudelet's data on alpha brass [4] and this tolerable discrepancy was attributed to the nonspherical shape of the grains and the resulting suppression of rheological flow around sharp corners, etc. By the same token lower values should be expected for equiaxed grains especially if there is a significant contribution from grain and interphase boundary sliding. Chen's model considers flow in the matrix phase and not in the grain boundaries.

Figure 3 also shows the transition from a behavior typical of soft particle is a hard matrix \((q<0)\) to hard particles in a soft matrix \((q>0)\) which Chen's model also describes. (For Ti-6Al-4V-xFe alpha is the hard phase and beta is the soft phase.) This transition, however, is observed only for the alloys where \( f_\beta \) can fall below approximately 0.25, roughly the void space in a
close-packed arrangement of spheres. When this is the case, lowering the concentration of iron shifts this transition to lower $f_B$. This transition can be seen as the intersection of two lines, one where $q < 0$ ($B$ in a matrix of $\alpha$) and the other where $q > 0$ ($\alpha$ in a matrix of $B$). The first line probably represents deformation through the $\alpha$ grains as well as $\beta$-assisted sliding between these grains, consistent with the TEM observations of Gurewitz [6] for Ti-6Al-4V. The second line probably represents deformation which takes place entirely in the $\beta$ phase (especially regions adjacent to the $\alpha$ grains) but modified by the presence of the harder $\alpha$ phase.

The pre-exponential term in the $A$-parameter can now be written as:

\[ A_0 = A_f (1-f_\alpha)^q \]

(5)

For the case where $q < 0$ $A_f$ has a unique value which one can think of as a "coefficient of friction" between sliding $\alpha$ grains, a structural constant for the $\alpha$ phase, or both. For the other case where $q > 0$ $A_f$ has the same meaning but in this case $A_f$ decreases with increasing concentration of iron until 0.5% iron and then is constant. This suggests a saturation-type behavior which is in addition to the rheological effects described above. This might be the behavior described by Chen [5] where the non-steady state nature of diffusion across sliding interphase boundaries leads to the saturation of these boundaries with the fast diffusing species. If this is the case then iron has the effect of lowering the "coefficient of friction" or at least increasing its importance in $A_f$ when $q > 0$. It also suggests that this is the strain-dependent process which results in high value of the activation energy.

Conclusions

1. The superplastic behavior of Ti-6Al-4V and five iron-modified Ti-6Al-4V alloys can be described using a single function based on the Mukherjee, Bird, Dorn equation:

\[ \frac{\varepsilon k T}{DGB} = A \left( \frac{\alpha}{G} \right)^n \left( \frac{B}{d} \right)^p \]

(1)

where $n = 2$, $p = 2$, $d$ refers to the $\alpha$ grain size, $D$ is replaced by the weighted average diffusivity for the $\beta$ phase which includes the effect of alloy element partitioning:

\[ D_\beta^{AV} = \sum_i x_i D_{\alpha_i} \exp \left( \frac{-Q_{\beta_i}}{kT} \right) \]

(3)

where $x_i$ is the concentration of the $i^{th}$ of $j$ elements and the $A$-parameter which is a function of temperature and phase proportion:

\[ A = A_f (1-f_\alpha)^q \exp \left( \frac{-Q_{AV}}{kT} \right) \]

(6)

2. Excellent agreement between experimental and computed stresses has been obtained, figure 4.

3. The results also suggest that the high values of the activation energy (often reported for the superplasticity of titanium alloys) may be due to an interphase boundary sliding related saturation of the interphase boundaries with the faster diffusing elements, particularly iron.
Acknowledgements

This work was supported by grant number 91-NC-251 from AFOSR (Alan Rosenstein) and Lawrence Livermore National Laboratory (Don Lesuer and Ed Dalder).

References


Figure 4. Comparison of stresses measured from the stepped-temperature tests and stresses computed using equations 1, 3 and 6.