A study of high temperature deformation in the titanium – aluminum binary system was undertaken to support the development of titanium alloys for use over the temperature range 900F to 1200F. Specimens containing 4, 5, 6, and 8 weight percent aluminum were creep tested to compare single phase material with an alloy containing a second phase, Ti₃Al. The starting material is shown to have a strong texture which is related to a reduction in the difference in critical resolved shear stress between the basal and prismatic planes. Creep testing revealed that the activation energy for steady state creep is a function of stress and aluminum content over the temperature range of 800F to 1400F but was inconclusive in identifying the principal mode of deformation.
Introduction

An interest exists in developing titanium alloys for certain gas turbine engine applications where the normal use temperature would lie between 900°F and 1200°F. Although the intended environment may be regarded as hostile for many reasons, this report will concentrate on only one, the temperature, for a temperature in the neighborhood of half the melting point (calculated to be ~1100°F (1)) is a major concern for a material in a stressed state. The study reported here was undertaken to obtain a better understanding of the high temperature deformation behavior of promising alloys.

The titanium-aluminum binary system was chosen because aluminum is a major constituent in nearly all commercially significant titanium alloys, and because the system has been studied extensively with regard to low temperature deformation behavior. In 1957 Sparks et al (2) showed that small concentrations of aluminum changed the texture of cold-rolled sheet iodide titanium sufficiently to indicate a change in the principal slip system. Subsequently Blackburn and Williams (3) found by transmission electron microscopy that in low temperature deformation, aluminum additions were related to reduced twinning, increased planarity of slip, and decreased cross slip. Although this amounted to improved strength, the addition of aluminum has a practical limit in that the appearance of a second phase, Ti₃Al, is coincident with sharply decreased low temperature ductility (4,5,6). Aluminum concentrations over eight weight percent are not now considered for commercial purposes.

The effect of temperature has not been treated as thoroughly for titanium-aluminum alloys. Richardson and Grant (7) examined the stress-rupture behavior of two titanium-aluminum alloys over the temperature range 1000°F to 2000°F, but did not analyze their results with respect to deformation modes. Their study concerned itself only with alloys containing 3.32 and 5.79 weight percentages of aluminum and the two phase α-Ti/Ti₃Al region was never encountered. They did, however, point out the reason for favoring the hexagonal structure at high temperatures: a sudden decrease in one hour rupture strength is coincident with the appearance of the b.c.c. structure. This study of the titanium-aluminum binary system was undertaken to determine the results of deformation at high temperatures in the alpha phase field and to compare the creep resistance of single phase alpha titanium alloys with that of an alloy containing a second phase, Ti₃Al.

Experimental Procedure

All titanium-aluminum alloys studied were prepared in a similar fashion. One pound ingots of each alloy were cast in an argon-filled non-consumable arc melting chamber, remelting each ingot at
least six times. A combination of forging and rolling with intermediate reheating at the temperatures listed in Table I reduced the cross sectional area by approximately ninety percent and produced 3/4 inch square bars having a structure described in a general sense as equiaxed alpha. The bars were heat treated in argon at 1112°F (600°C) for periods varying from 168 hours to 285 hours. This is sufficient to develop the Ti₃Al particles in the eight percent aluminum alloy (3). Core samples were analyzed for interstitial content and all alloys were found to have an oxygen content of 780-1280 ppm, a hydrogen content of 42-66 ppm, and a carbon content of 250-350 ppm. No portion of any range was favored by any single alloy and the range of values is essentially that of the starting ingots.

Table I

Binary Alloy Reference Data

<table>
<thead>
<tr>
<th>Nominal Composition w/o Al</th>
<th>Chemical Analysis w/o Al</th>
<th>Working Temperature °F</th>
<th>Mean Linear Intercept (µ) Longitudinal section</th>
<th>Transverse section</th>
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<td>41</td>
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</tr>
<tr>
<td>5</td>
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<td>1700</td>
<td>55</td>
<td>56</td>
</tr>
<tr>
<td>6</td>
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<td>41</td>
</tr>
<tr>
<td>8</td>
<td>7.94</td>
<td>1790</td>
<td>38*</td>
<td>38*</td>
</tr>
</tbody>
</table>

* broad bands not considered

Creep specimens were cut from the core of each bar, eliminating the possibility of contamination from surface oxides. Gage section dimensions, in inches, were .200 diameter x .750 long. Specimens were tested for transient and minimum creep rates at pressures <2 x 10⁻⁵ torr with the temperature variation along the gage length held within ± 1°F. Loads were applied through a lever arm system; total elongation was measured with a linear variable differential transformer, the movable core mounted on the lever arm. Load changes were calculated from the continually reduced cross section area of the gage section in order that multiple tests might be run on the same specimen at comparable stresses. Corrections were included for thermal expansion of the linkage.

Results of the creep tests are shown in Figures 1, 2, and 3. Transient creep was found to be virtually non-existent below 1100°F and to diminish rapidly to a minimum creep rate at higher temperatures. The appearance of transient creep can be attributed to a recovery process since it can be found when the stress is lowered at constant temperature. A series of temperature changes at constant stress produced the strain rate variation with temperature shown in Figures
Fig. 1. Creep strain of titanium - 4 wt pct aluminum at 1300 F with applied stresses of 2100 psi and 11,500 psi.

Fig. 2. Minimum creep rate vs. $10^3/T$ (K$^{-1}$) for titanium, -4, 5, and 6 wt pct aluminum. Stresses determined from constant load testing.
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Fig. 3. Minimum creep rate vs $10^3/T \ (°K^{-1})$ for titanium-8 wt pct aluminum acted upon by an applied stress of 61,500 psi. The dashed line represents the $\alpha$-Ti/$\alpha+$ Ti$_3$Al phase boundary according to Blackburn (8).
2 and 3. In Figure 3 the dashed line represents the \(\alpha\)-Ti/Ti\(_2\)Al phase boundary as given by Blackburn (8). A least squares fit to the data points for the entire temperature range covered is shown to indicate the difficulty in differentiating between the single phase and two phase materials. If the data on either side of the boundary are fit as separate groups, the slope would be slightly less for the two phase region to the right of the dashed line.

Following creep testing, specimens representative of each alloy were sectioned for optical metallography. The mean linear intercepts (grain size) reported in Table I were calculated from separate measurements on longitudinal and transverse sections of the grip ends. Similar values are found for comparable sections within the gage length. One alloy, that containing eight percent aluminum, developed a unique microstructure in that large bands were found nearly free of grain boundaries. These bands appeared throughout the specimen and were obviously created in forging and rolling. Their greatest length was in the direction of the tensile axis and their breadth, often irregular, was considerable larger than the average size of the surrounding grains. One of these bands was found on the surface of the gage section of a specimen and was photographed in a scanning electron microscope following creep at 1100°F (Figure 4). Its breadth is approximate by 200 microns (compared to an average grain size of 38 microns). The specimen had a strong texture with [0001] normal to the tensile axis and [10\(\bar{1}\)0] parallel to the tensile axis. Since the bands did not account for the greater area fraction of the specimen and since the remaining grains were quite regular in size and shape, the mean linear intercept reported for the eight percent alloy in Table I did not take the bands into account.

**Discussion**

Alpha-titanium alloys readily develop a texture during deformation processing, and that texture can have a profound influence on subsequent mechanical property determination. Titanium rod, deformed at 1112°F (600°C) develops a texture similar to that identified in this study for titanium-aluminum at higher temperatures; [0001] normal to the working direction and [10\(\bar{1}\)0] parallel to the working direction (9). Cold working titanium-aluminum alloys containing more than 3.8% aluminum also produces this texture (2). Cold working iodide titanium sheet, however, moves the basal pole 27° from the sheet normal without changing the alignment of [10\(\bar{1}\)0] with the rolling direction.

Cold worked zirconium has a texture analogous of that of iodide titanium, and Picklesimer attributed the displacement of the basal pole from the sheet normal to slip in the <1123> direction (9). He also noted that when the basal slip was the primary mode of defor-
Fig. 4. Scanning electron micrograph of the gage section surface of a titanium - 8 wt pct aluminum specimen following creep at 1100F. See text for a discussion of grain size.
mation, a texture would develop in which the basal poles would be clustered about the sheet normal and spread toward the rolling direction. This type of texture is found in cold-worked magnesium (11) where basal slip is expected to be the primary mode of deformation due to the extremely low value of critical resolved shear stress on the basal plane (12). A similar argument for titanium would favor prismatic slip, but increasing the processing temperature or increasing the aluminum content tends to eliminate that favoritism. Both systems could by operating, however, and still not account for the observed texture since little tendency is noted for spreading basal poles toward the rolling direction. Unless all of the dislocations leaving a single grain in the polycrystalline specimen are absorbed at the grain boundary, thus rotating the grain, it is not possible to generate a texture with basal and prismatic slip alone while maintaining any semblance of an equiaxed grain size. If all dislocations were absorbed at the grain boundaries, it would not be possible to develop the banded structure in Figure 4. Therefore some other deformation mode must be operating which can transfer strain to neighboring grains.

Considering Figure 4, it is clear that the texture, once established, is "locked-in". Subsequent creep deformation resulted in a rumpled surface due to dislocation movement within the grains and to grain boundary movement but did not alter the texture. The banded area in Figure 4 appears to be slightly faceted: the facets separated by ribs running parallel to the tensile area. Since the surface flatness of the band is fairly well developed, all of the facets must be nearly parallel to the basal plane. In certain areas along the edge of the band faint slip traces can be seen inclined at 35°-40° to the tensile axis. Both \( \langle 11\overline{2}0 \rangle \) and \( \langle 11\overline{2}3 \rangle \) slip on \{10\overline{1}0\} planes could account for the observed traces. An examination of longitudinal and transverse sections of the gage region under polarized light failed to reveal an areal fraction of twins that could be considered significant in accounting for strain along the [0001] direction, thus favoring the existence of \( \langle 11\overline{2}3 \rangle \) slip. Texture development in titanium-aluminum alloys deformed in the alpha phase field can therefore be attributed in part to a reduction in the difference in critical resolved shear stress between the basal and prismatic planes. The operation of \( \langle 11\overline{2}3 \rangle \) slip to accommodate the other shears has little effect on the orientation of the basal pole relative to the working direction, and therefore the texture, once developed, is "locked-in".

From the previous discussion it is clear that sufficient deformation modes can be identified to account for homogeneous deformation. However, the accumulated results from creep testing titanium-aluminum alloys over the temperature range 800°F to 1400°F have not revealed any single rate-controlling mechanism. The activation energies determined from the variation of strain rate with temperature were found to be a function of aluminum content and the applied
stress as shown in Table II. No systematic variation in activation energy with temperature was found so it would appear that deformation is not diffusion controlled. The appearance of transient creep upon reloading a specimen at constant temperature but reduced stress suggests the occurrence of dislocation climb as part of a recovery process, but even this is not completely unambiguous since anelastic strain relaxation would produce similar results although probably of lesser magnitude. On the other hand dislocation glide cannot be shown to be exclusively controlling. Strain rates at fixed composition, stress, and temperature have been found to be reproducible within the limits of experimental accuracy without regard for prior strains.

Table II

<table>
<thead>
<tr>
<th>w/o Al</th>
<th>4</th>
<th>5</th>
<th>6</th>
<th>8</th>
</tr>
</thead>
<tbody>
<tr>
<td>Q (Kcal/mole)</td>
<td>46</td>
<td>51</td>
<td>49</td>
<td>59</td>
</tr>
<tr>
<td>(KSI)</td>
<td>9.4</td>
<td>19.8</td>
<td>9.2</td>
<td>9.3</td>
</tr>
</tbody>
</table>

The problem in trying to identify a rate-controlling mechanism seems to be related to an unfortunate choice of temperature and stress ranges. While the ranges were dictated by practical interest, all of the test points are clustered around a value of $\frac{\dot{\varepsilon}}{D} = 10^9 \text{cm}^{-2}$ where $D$ is the self diffusivity of titanium (13). The cluster is simply too small to be examined for a meaningful relationship between $\frac{\dot{\varepsilon}}{D}$ and $\sigma$, the applied stress.

For the same reason the apparent reversal in stress dependency of the activation energy for the eight percent aluminum alloy (see Table II) cannot be examined for a possible change in mechanism. However, it is possible to interpret figure 3 as matrix-controlled creep over the entire temperature range. If the second phase was beneficial to creep resistance, the plot would show a higher activation energy inside the two phase region (to the right of the dashed line in the figure) than outside. If the data points in figure 3 are fit to give a change in slope at the phase boundary, the higher activation energy would be in the single phase field. For the temperature range of interest, the amount of $\text{Ti}_3\text{Al}$ present is low. Increasing that amount will result in a loss in room temperature ductility, an impractical consideration without just cause. From the results of this experiment a further increase in the amount of $\text{Ti}_3\text{Al}$ is not warranted.

Conclusions

The following conclusions are drawn from this study:
1. Texture development in titanium-aluminum alloys deformed in the alpha phase field reflects a reduction in the difference in critical resolved shear stress between the basal and prismatic planes.

2. The texture, once developed, is not disturbed by further deformation.

3. Creep of a two phase $\alpha$-Ti/Ti$_3$Al alloy is controlled by a matrix mechanism, as yet not identified.

This report has examined only two aspects of the problem of high temperature deformation of titanium alloys. In so doing, the basis for future research has been established. Texture will be altered through the use of the allotropic transformation, and the temperature range for creep will be extended to isolate the controlling mechanism for deformation. It is quite apparent from the work completed that the principal deformation modes must be established to determine the relative effectiveness of measures employed to improve high temperature strength and creep resistance.

References


