EFFECTS OF MICROSTRUCTURE ON CREEP BEHAVIOR OF Ti-6211

R.T. Chen, W.H. Miller, Jr., and E.A. Starke, Jr., Department of Materials Science, University of Virginia, Charlottesville, Virginia 22901, U.S.A.

Introduction
Many titanium alloys have been shown to exhibit appreciable amounts of creep at ambient temperatures (T < 0.2 Tm) (1). This may occur at stress levels far below the tensile yield strength. Room temperature creep for Ti-5Al-2.5Sn and Ti-6Al-4V has been extensively studied by several authors (2,3,4). Creep properties and stress relaxation at room temperature for Ti-6Al-2Nb-1Ta-0.8Mo (Ti-6211) have been measured by Chu (5). Mechanistic studies related to microstructure and deformation behavior, however, have not been made. Since Ti-6211 is a weldable alpha-beta titanium alloy and its microstructure strongly depends on the processing history and thermal treatments, various microstructures can be developed in the heat-affected zone of the weld (6). As pointed out by Imam and Gilmore (3), microstructure is an important variable in influencing creep strain in Ti alloys. Therefore, it is of practical importance to study the effects of weld zone microstructures on the creep behavior of the Ti-6211 alloy.

Low temperature creep of Ti alloys is of the transient type; i.e., the creep curve tends to level off and reach creep saturation (5,7). At elevated temperatures (T > 0.4 Tm), primary creep is generally followed by steady state creep with a minimum creep rate (8). The contribution of diffusion to creep deformation and the amount of sliding at alpha/beta interfaces, colony boundaries and prior beta grain boundaries is more significant at high temperatures. A study of the creep behavior of Ti-6211 at high temperature will serve as a comparison with that at ambient temperatures and help in understanding creep deformation mechanisms. In this investigation, the effects of different microstructures, creep stress levels, and creep temperature on creep deformation were examined. Based on the analysis of creep curves as well as the examination of surface deformation and deformation substructures, the deformation behavior involved in creep in Ti-6211 will be discussed.

Experimental
Two heats of the Ti-6211 alloy were received in the form of one-inch thick plates. The chemical compositions of both heats are listed in Table 1.

The alloy in the as-received condition had a Widmanstätten alpha + beta structure which is typical of beta-processed alpha + beta titanium alloys. Different heat treatments were selected to simulate the microstructures in the weld zone. All of the heat treatments included heating above

<table>
<thead>
<tr>
<th>Heat</th>
<th>Al</th>
<th>Nb</th>
<th>Ta</th>
<th>Mo</th>
<th>O</th>
<th>N</th>
<th>H</th>
<th>C</th>
<th>Y</th>
<th>Mn</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>I</td>
<td>5.6</td>
<td>2.16</td>
<td>0.95</td>
<td>0.79</td>
<td>0.066</td>
<td>0.007</td>
<td>0.0046</td>
<td>0.02</td>
<td>&lt;0.002</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>-</td>
<td>-</td>
<td>Bal</td>
</tr>
<tr>
<td>II</td>
<td>5.9</td>
<td>2.0</td>
<td>0.93</td>
<td>0.60</td>
<td>0.056</td>
<td>0.011</td>
<td>&lt;0.01</td>
<td>0.02</td>
<td>-</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.06</td>
<td>&lt;0.01</td>
<td>Bal</td>
</tr>
</tbody>
</table>
the beta transus and cooling at different rates. The heat treatments and corresponding microstructures are listed in Table 2. The laboratory heat treatments were conducted in high purity flowing argon atmosphere. The basic microstructures of the different heat treatments were characterized by optical metallography with the modified Kroll's reagent as an etchant.

Cylindrical tensile and creep samples were machined such that the axis of loading corresponded to the rolling direction of the as-received plates. Tensile tests were performed with a servo-hydraulic MTS machine at a constant strain rate of $2 \times 10^{-3}$/s. Creep tests were performed on lever-arm-loading type creep machines. Creep strains at ambient temperatures (298K and 453K) were monitored with foil strain gages. High temperature creep was performed in a purified argon atmosphere and creep strains were measured using an LVDT in unison with a high temperature extensometer. Fracture surfaces of the crept specimens were examined using SEM. TEM foils prepared from crept specimens were examined with TEM operating at 100 kV.

Various methods were used to carry out surface deformation studies. The specimens were electropolished and lightly etched prior to testing. Fine gold grid patterns were plated onto the surfaces of specimens to be crept at room temperature using a photolithographic technique. Fiducial scratches were inscribed on high temperature creep specimen surfaces with diamond-tip pencils. The distortion and offsets of fiducial grids or lines, which indicated deformation and sliding, were examined using SEM.

<table>
<thead>
<tr>
<th>Designation</th>
<th>Heat Treatment</th>
<th>Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>W1, W11*</td>
<td>As-received</td>
<td>Widmanstätten alpha + beta</td>
</tr>
<tr>
<td></td>
<td></td>
<td>(colony type)</td>
</tr>
<tr>
<td>W8, W12*</td>
<td>$1050^\circ C/40 \text{ min}/\text{AC}$ + $950^\circ C/4 \text{ h}/\text{AC}$</td>
<td>Widmanstätten alpha + beta</td>
</tr>
<tr>
<td></td>
<td></td>
<td>(basket-weave type) + thick grain boundary alpha</td>
</tr>
<tr>
<td>W9</td>
<td>$1050^\circ C/40 \text{ min}/\text{WQ}$ + $950^\circ C/4 \text{ h}/\text{AC}$</td>
<td>Tempered martensite</td>
</tr>
</tbody>
</table>

*samples prepared from Heat II

Results and Discussion

A. Microstructure and Tensile Properties

As summarized in Table 2, two microstructures from Heat I and one from Heat II were produced by thermal treatments to simulate microstructures in the weld zone. Representative features of these microstructures are shown in Fig. 1. The microstructure of the as-received condition, W1, W11, has a Widmanstätten alpha + beta structure exhibiting large colony size and a pancake grain shape due to hot working in the beta field. The W8 and W12 structures consist of a basket-weave type Widmanstätten alpha in a beta matrix with thick alpha platelets and grain boundary alpha (GBα). TEM micrographs revealed that small colonies are present in the basket-weave structures (10). The W9 structure consists of primary...
alpha in a matrix of tempered martensite structures. Microstructures resulting from excursions above the beta transus such as W8, W9 and W12 exhibited equiaxed grains due to recrystallization occurring in the beta field.

In the colony type Widmanstätten alpha + beta structure, there exists a Burgers orientation relationship between alpha and beta plates located in the same colony. The (0001)[1120] slip system in the alpha phase is coplanar with the (110)[111] slip in the beta phase. This enables the planar slip to pass through beta platelets and traverse the entire colony. In the as-received condition, large colony sizes are the result of slower cooling rates from the beta field, which is the case for air cooling of an one-inch thick plate. In the basket-weave type Widmanstätten structure, small colonies are scattered among interspersed long alpha plates. Grain boundaries in the equiaxed grain structure are straighter than those in the as-received structure. The alpha phase in the W9 structure results from the tempering of martensite in the alpha + beta field; no Burgers relationship exists between the alpha and beta phases. The thickness of grain boundary alpha is non-uniform in W9.

Tensile properties of the selected microstructures are listed in Table 3. It is noted that the variation of yield strength for microstructures produced by thermal modification is small; the difference between the maximum and the minimum is less than 10%. The tensile strength of W11, however, is substantially lower than that of W1. This difference is primarily due to differing oxygen contents in the two heats, as shown in Table 1. An increase in the content of the interstitial atoms such as oxygen will increase the flow stress in titanium alloys (11).

**TABLE 3. Tensile Properties of Selected Microstructures**

<table>
<thead>
<tr>
<th>Specimen Designation</th>
<th>Yield Stress (MPa)</th>
<th>UTS (MPa)</th>
<th>Proportional Limit (MPa)</th>
<th>Ductility (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>W1</td>
<td>760</td>
<td>864</td>
<td>592</td>
<td>13.4</td>
</tr>
<tr>
<td>W8</td>
<td>751</td>
<td>885</td>
<td>629</td>
<td>12.2</td>
</tr>
<tr>
<td>W9</td>
<td>784</td>
<td>884</td>
<td>619</td>
<td>7.5</td>
</tr>
<tr>
<td>*W11</td>
<td>702</td>
<td>805</td>
<td>516</td>
<td>13.1</td>
</tr>
<tr>
<td>*W12</td>
<td>742</td>
<td>855</td>
<td>549</td>
<td>7.9</td>
</tr>
</tbody>
</table>

*From Heat II.

B. Creep at Ambient Temperatures

Fig. 2 shows creep curves for W1, W8 and W9 microstructures tested at the same stress level, 684 MPa, which is equal to 90% of [ys] of W1. These curves are typical of those previously observed for low temperature creep; i.e., creep was initially rapid and then decreased quickly at lower stresses, or gradually at higher stresses. Eventually, the creep curves appear to level off. This type of creep saturation (or creep exhaustion) is commonly observed for titanium alloys at low temperatures (2,3,4,5).

Generally, creep at low temperature is of the transient type, and the creep strain/time relationship is of the form: $\varepsilon_c = A t^n$, where $\varepsilon_c$ is the creep strain (total strain minus elastic strain), $A$ is a constant
depending on microstructure and creep stress and \( n \) is the time exponent. A least square fit method was used to determine the time exponent \( n \); the results for Wll are shown in Fig. 3. The \( n \) values of the Ti-6211 alloy vary with microstructure and stress, and generally fall below 0.3.

The creep rate and creep strain varied for specimens with different heat treatments as can be seen in Fig. 2. Since no strong texture was observed for the microstructures of interest in this study (10), the difference in creep response can most likely be attributed to the differences in phase morphology or slip length which is determined by the spacing of potential slip barriers. TEM examination of creep deformed specimens indicated that planar array and straight dislocations were the typical features in the alpha phase, and no cross slip was observed. One example of these features for the Wl sample is shown in Fig. 4. Due to the Burgers orientation relationship existing between alpha and beta plates within the same Widmanstätten colony, planar slip was able to pass through the beta platelets and the slip length was equivalent to the colony size.

When comparing the room temperature creep response of different microstructures (refer to Fig. 2), it is evident that the Wl structure shows the highest creep rate and creep strain to saturation, while the W9 structure shows the lowest. In the Wl structure, the large colony size results in long slip lengths and greater deformation. Moreover, sliding at alpha/beta interfaces and straight colony boundaries can contribute to the creep strain. Only small colonies are interspersed among the large alpha platelets in the W6 basket-weave Widmanstätten microstructure. Small colony size results in short slip length. This restricts the possible sliding along colony boundaries. As a result, W6 exhibited low creep strains relative to Wl. W9 consists primarily of alpha platelets in a beta matrix. As mentioned earlier, no Burgers relationship exists between alpha and beta phases so the slip length is short and, consequently, this microstructure exhibits a low creep strain. The stress during creep testing was usually below the yield strength and is low compared to that of a tensile test. The chance for a slip band to break the barrier is very small and, consequently, the slip length limits the extent of creep deformation.

Creep tests for the Wl microstructure were also conducted at 453 K to compare temperature effects. The creep strain at 453 K is much greater than that at room temperature although creep saturation still occurred at this temperature (10). A study of deformation substructures revealed that thermally-activated cross slip occurred at 453 K, as shown in Fig. 5. Distortion of grid patterns on room temperature creep specimens can also illustrate deformation modes. The as-received structure exhibits greater amounts of surface deformation at low stress levels than the modified structures. Fig. 6 is an example showing sliding along colony boundaries and slip traversing the whole colony. Sliding at alpha/beta interfaces and at colony boundaries have also been previously reported in Ti alloys undergoing deformation at ambient temperatures (4,12,13). At stress levels below 80% of\( \sigma_y \), only very minor grid distortion along colony boundaries is visible in the Wll structure. No grain boundary sliding was observed for the three types of microstructures, as illustrated...
in Fig. 7 for the Wl2 structure. These observations coupled with the large grain size in Ti-6211 suggest that the contribution of grain boundary sliding to room temperature creep is minimal.

Fig. 1. Optical micrographs showing typical features of (a) W11; (b) W12; and (c) W9.

Fig. 2. Room temperature creep curves for various microstructures at a stress level of 684 MPa.

Fig. 3. Log creep strain vs. log time plots for W1 structure crept at room temperature.
Fig. 4. TEM micrographs showing dislocation structure for W1 crept at 298K/684 MPa.

Fig. 5. TEM micrograph showing dislocation structure for W1 crept at 453K/608 MPa.

Fig. 6. SEM micrograph showing sliding at colony boundary for W11 crept at 298K/632 MPa.

Fig. 7. SEM micrograph showing no evidence of sliding along GBα/matrix interface for W12 crept at 298K/668 MPa.

C. Creep at High Temperatures

Creep tests of specimens with W11 and W12 structures were conducted at various stress levels in a temperature range around 0.4 Tm. For high temperature low strain rate creep tests, a steady state creep with constant strain rate is generally achieved (8). The steady state creep rate, $\dot{\varepsilon}_s$, is often considered to vary with stress $\sigma$ and temperature, $T$, according to: $\dot{\varepsilon}_s = k\sigma^m \exp\left(-\frac{Q}{RT}\right)$, where $k$ is a constant, $m$ is the stress exponent and $Q$ is the apparent activation energy for creep. The activation energy $Q$ can be estimated from the log $\dot{\varepsilon}_s[\sigma]$ vs. $1/T$ plot. The Q value was determined to be 143 KJ/mole for the W11 structure, as shown in Fig. 8. The estimated activation energy was very close to that of self-diffusion in Ti (150 KJ/mole) (14). This suggests that the rate-controlling process in high temperature creep is diffusion-controlled. Dislocation substructure of high temperature creep for the W11 structure is shown in Fig. 9. In contrast to the straight dislocations and planar slip developed during low temperature creep (Fig. 4), the wavy dislocations are tangled and more homogeneously distributed. Consequently, no distinct coarse slip bands were observed on the surface of the heavily crept samples. Furthermore, dislocation dipoles and loops were frequently
observed in the samples subjected to high temperature creep similar to that reported by Hall for Ti-6-4 (15). The dislocation configuration can result from the jog or climb of dislocations, which becomes easier at high temperature.

As far as creep rupture is concerned, the W12 structure exhibits a shorter creep life at high temperature than the W11 structure (10). This could be due to the difference in microstructure and the crack initiation mechanisms. W12 has equiaxed grain shape and thick GBα which allows more grain boundaries to be subjected to higher shear stress and consequently results in greater grain boundary sliding and strain localization in the grain boundary alpha phase. W11 has an elongated grain structure with the longer edge parallel to the loading axis and, consequently, a small percentage of the grain boundaries are subjected to high shear stress. The fracture mechanism involved in creep rupture at high temperature is similar to that of the hot tensile test (6). Diffusional void growth and sliding at various interfaces, however, will play a more important role in the creep rupture since they both are time dependent processes. Creep rupture in the W11 structure occurred by microvoid formation and coalescence at colony boundaries, similar to that of tensile fracture at room temperature (9). Sliding and cracking along alpha/beta
interfaces and colony boundaries in the W11 structure become more significant at high temperatures. Fig. 10 shows such an example. On the other hand, in the W12 structure, sliding and deformation along GBa/matrix interfaces are more severe and result in intergranular fracture. The distinct grain facets consisting of dimples were identified in W12 samples by SEM fractography (10). It is also noted that short secondary cracks which appeared on the well-polished specimen surfaces more frequently occurred along grain boundaries oriented 90° to the stress axis. Since the diffusional growth of cavities is driven by tensile stresses across the boundary, voids nucleated at the transverse boundaries tend to grow faster and coalesce into cracks.

Summary
1. The Ti-6211 alloy shows transient type low temperature creep in which creep saturation occurs. Work hardening and lack of recovery result in creep saturation. For room temperature creep, planar arrays of straight dislocations and a single operative slip system were observed. For creep tests run at 453 K, dynamic recovery and thermally-activated cross slip enhance the creep process.

2. Creep strain at ambient temperatures is strongly dependent upon creep stress and microstructure. Colony type Widmanstätten structures exhibit the highest creep rate. The long slip length associated with large colony size as well as the contribution of sliding at colony boundaries or alpha/beta interfaces account for enhanced creep rates.

3. For creep at high temperatures, the temperature dependence of steady state creep rate indicates that diffusion controlled dislocation movement is the rate-controlling process of creep.

References