Creep failure mechanisms in a Ti-6Al-4V alloy

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ANNOTATION
The creep behaviour of a 17-mm thick plate in various heat treating conditions has been analysed. No difference in fracture topography was found between smooth and notched specimens in the same heat treating condition. Maximum creep lives were achieved in BA samples. Fractographic examination of the fracture surfaces revealed a clearly intergranular topography that has been attributed to grain boundary sliding along the former beta grains. The failure operating mechanism for the alpha-beta annealed was diffusional creep by the nucleation and coalescence of the creep cavities generated mainly at the alpha-beta interfaces. Material in the MA condition exhibits the lowest times to failure. The presence of a reduced number of voids, just sited in the neighbourhood of the fracture surface and the shape of these voids, suggested that they were not really creep cavities but voids generated by the plastic deformation. Consequently, failure is attributed to dislocation climb activated by the high testing temperature.

Key words: alpha-beta alloys, creep, failure mechanism, microstructure

1. INTRODUCTION

All materials creep under load at all temperatures, but a very marked difference in behaviour can be found between lead used as a roofing material that creeps under its own weight at room temperature and nickel base alloys that can operate successfully at temperatures higher than 1000°C. These differences are revealed when comparisons are made in terms of the three important parameters that describe the creep process; namely, stress, temperature and time [1].

The response of a material to applied stress depends on these variables but also on its microstructure. Although the generalities of the creep behaviour are well known, that is, an increase of the temperature or the stress lead to a greater creep rate and shorter time to failure but the complete quantitative description of engineering materials is often much more complex. Elastic deformation takes place at the instant of stress application and disappears in the instant of applied stress removal. Thus, the elastic strain is time independent and reversible its magnitude is a single function of the stress. On the other hand, plastic deformation is irreversible and thus leads to a permanent change in the form of the body. Generally, it consists of time independent and time dependent components. The time dependent component is designated as creep and the graphical representation of the time dependence of the strain as the creep curve [2].

At low temperatures the material work hardens and the dislocation density increases obstructing further dislocation motion, the creep rate, initially large, decreases in time and finally falls to a negligible value, practically zero. At higher temperatures (above 30% of the absolute melting temperature for pure metals and 40% of their melting points for alloys) a decrease in strain rate is observed reaching a minimum value but, as higher temperatures promote recovery, this is sufficiently important to compensate the effects of deformation strength. Even at higher temperatures (above 90% of the melting point) the state variables instead of tending to constant values may oscillate about more or less steady values because of dynamic recrystallisation. Then it is possible to define a quasi-steady state and, once more, stress, temperature and strain rate are approximately related.

Crystalline solids deform plastically by a number of alternative, often competing, mechanisms. Plastic flow is a kinetic process. Although it is often convenient to think of a polycrystalline solid as having a well defined yield strength, below which it does not flow and above which flow is rapid this is true only at absolute zero. Generally,
the strength of the solid is determined by the kinetics of the process occurring on the atomic scale: the glide motion of dislocation lines, their coupled glide and climb, the diffusive flow of individual atoms, the relative displacement of grains by grain boundary sliding (involving diffusion and defect motion in the boundaries), mechanical twinning (by the motion of twinning dislocations) and so forth [3].

Deformation maps are diagrams in the stress versus temperature plane in which the areas are indicated where a particular creep mechanism predominates [4]. The idea is that different mechanisms operate independently and the fastest one determines the deformation behaviour and occupies the respective regime in the stress-temperature plane. The boundaries between the different regimes are usually calculated, rather than measured, by comparing the strain rate equations for the different mechanisms [5].

These deformation mechanisms maps can be used to identify the mechanisms by which a component or structure deforms in service, thereby identifying the constitutive law or combination of laws that should be used in design. Moreover, they can give guidance in alloy design and selection. It must be kept in mind that strengthening methods are selective; alloying, for instance, may suppress power-law creep but leave diffusional flow unchanged; increasing the grain size does the opposite. In a previous work it was demonstrated that an increase in the alpha grain size led to an increase in the fracture toughness of 20 mm diameter alpha-beta annealed bars of a Ti-6Al-4V alloy [6]. A similar improvement in toughness was observed in a 17-mm thick plate of the same alloy in the same heat treating condition [7]. The possibility that this improvement in toughness would be accompanied by a better creep performance is interesting but this will be only true when failure is produced by diffusional flow. Consequently, a confirmation of the operation of this creep mechanism will constitute a significant advance in order to achieve better in-service performance.

Deformation mechanisms maps of a large number of metals, alloys and ceramics were presented by Frost and Ashby [3]. Among these materials commercial purity titanium is included. However, no information about other titanium alloys is given. More recently the fracture mechanism maps of commercial purity titanium and several alloys, namely an alpha alloy (Ti-5Al-2.5Sn), a near alpha alloy (Ti-6Al-5Zr-0.5Mo-0.25Si), an alpha-beta alloy (Ti-Al-4V), a beta alloy (Ti-15Mo) and two titanium aluminides (TiAl and Ti3Al) were constructed and compared with the fracture observations reported in the literature for each material [8]. However, in their study on Ti-6Al-4V two important aspects are missing. Firstly, all the data were obtained on specimens tested inside the temperature range where alpha and beta phases coexist and no modification in the boundaries between the different mechanisms was considered by the exclusive presence of beta at the highest temperatures. However, this aspect possesses only an academic interest as the maximum service temperature of this alloy is clearly inside the alpha-beta field. Much more practical application has the possibility to improve the creep performance of the alloy by means of a modification of its microstructure introduced by changes in the thermomechanical process. Consequently, there is a marked interest to study the changes in the failure mechanisms due to changes in the microstructure of the material, institutions not considered in the paper.

Recently, French researchers have published some papers on the influence of the stress triaxiality on the damage mechanisms of various titanium alloys [9-12]. In the first one the micromechanisms of tensile and creep behaviour of an alpha-beta Ti-6-2-4-6 alloy at 500° C were studied. Three stress domains are defined linked to particular viscoplastic mechanisms. The creep curve saturated and no fracture was observed below a threshold stress. In the range from this threshold stress and the yield stress the three usual creep stages were observed until failure. In this domain dislocation emission at subboundaries occurred in a few alpha grains and then steady state creep was observed. Above the yield stress the extent of stage II became negligible and dislocation emission occurred in many alpha grains, subboundaries did not remain stable and stage III occurred [9]. Three different types of fracture were observed in the second paper depending on strain and stress triaxiality. The first assisted by triaxiality leaded to fracture by void growth. The second appeared for lower triaxiality and under certain strain. In this case the quantity of voids is the cause of failure. Finally, when the stress triaxiality was too low no voids were created during loading but plastic strain was localised in shearing bands and the alpha grains rotated so as to adapt more plastic strain. The materials then developed plastic strain instability at a macroscopic scale [10].

The influence of the stress triaxiality on the damage mechanisms in an equiaxed alpha-beta Ti-6Al-4V alloy was the subject of their third paper. Void nucleation in the alpha phase was found to occur for a critical value of macroscopic plastic strain, whereas void nucleation at the alpha-beta interface depended on stress triaxiality. In a middle range of triaxiality and plastic strain voids nucleate in alpha because of the sufficient plastic strain and also at the alpha-beta interface because of the sufficient triaxiality but under high triaxiality voids nucleate preferably at these alpha-beta interfaces and grow perpendicular to the stress axis by a cleavage mechanism [11]. Finally, in the fourth one the fracture surfaces of specimens with different notched radii were studied.
Differences in the fracture topography of smooth and notched specimens were observed. Three fracture zones were identified in the notched specimens corresponding to a transgranular ductile fracture, a transgranular mixed zone and a shearing zone. Smooth specimens, however, only exhibited the facets corresponding to the second of these zones [12]. Once again a lack of information about the influence of heat treatment on failure mechanisms is deplored.

The aim of this paper is to investigate the influence of heat treatment on the creep behaviour of both smooth and notched specimens and the possible changes in fracture mechanisms of a Ti-6Al-4V alloy.

2. EXPERIMENTAL PROCEDURE

The material chosen for the present study was a 17 mm thick plate of a Ti-6Al-4V alloy whose chemical composition corresponds to 6.51% Al, 4.08% V, 0.16% Fe, 0.01% C, 0.19% O, 0.005% N, 0.0016% H, balance titanium. In the as-received condition this plate was in the mill annealed (MA) condition, that is a short maintenance at 720° C followed by air cooling as the final step of its thermomechanical process.

Various coupons were obtained from this plate and heat-treated in a small laboratory furnace under argon protection. BA samples were annealed in the beta field at 1030° C for ½ hour, air cooled and aged 2 hours at 730° C. In a previous study maximum fracture toughness was observed in specimens that were given this treatment but with the drawback of very poor ductility [7]. Those samples marked 940/4 were treated in the alpha-beta field at 940° C for 4 hours, furnace cooled to 700° C and air-cooled to obtain a good combination of ductility and toughness.

Smooth round 6.35-mm diameter creep specimens were machined at half thickness of these different heat-treated coupons in both the longitudinal or transverse directions. Moreover, one Bridgman semicircular-notched specimen was obtained from each coupon in the longitudinal direction of the plate. All creep tests were performed according to ASTM E 139 [13] in air at 455° C. Two stress levels were considered for the longitudinal smooth specimens and only one for the transverse or notched ones. Strain was continuously measured during the tests by means of an extensometer attached to the specimens and the graph of strain rate versus testing time recorded. Following failure, a fractographic analysis was carried out to determine the operating mechanism. This analysis consisted in the examination of one of the fracture surfaces of each broken specimen by means of scanning electron microscopy and the metallographic examination of longitudinal sections obtained from the other half of the failed specimens including the fracture surface.

3. RESULTS AND DISCUSSION

Table I exhibits the creep lives (\( t_i \)) for various smooth, longitudinal (L) and transverse (T) or notched (N) specimens together with the steady state creep strain rate, \( \varepsilon' \):

<table>
<thead>
<tr>
<th>Ref.</th>
<th>Orient.</th>
<th>( \sigma ) (MPa)</th>
<th>( \varepsilon' ) (h(^{-1}))</th>
<th>( t_i ) (h)</th>
<th>Ref.</th>
<th>Orient.</th>
<th>( \sigma ) (MPa)</th>
<th>( \varepsilon' ) (h(^{-1}))</th>
<th>( t_i ) (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>MA</td>
<td>L</td>
<td>489</td>
<td>0.349</td>
<td>36</td>
<td>BA</td>
<td>T</td>
<td>489</td>
<td>0.044</td>
<td>178</td>
</tr>
<tr>
<td>MA</td>
<td>L</td>
<td>345</td>
<td>0.017</td>
<td>714</td>
<td>BA</td>
<td>N</td>
<td>489</td>
<td>0.0007</td>
<td>2377</td>
</tr>
<tr>
<td>MA</td>
<td>T</td>
<td>489</td>
<td>0.163</td>
<td>61</td>
<td>940/4</td>
<td>L</td>
<td>489</td>
<td>0.057</td>
<td>107</td>
</tr>
<tr>
<td>MA</td>
<td>N</td>
<td>489</td>
<td>0.0012</td>
<td>828</td>
<td>940/4</td>
<td>L</td>
<td>379</td>
<td>0.0033</td>
<td>1619</td>
</tr>
<tr>
<td>BA</td>
<td>L</td>
<td>489</td>
<td>0.0037</td>
<td>209</td>
<td>940/4</td>
<td>T</td>
<td>489</td>
<td>0.101</td>
<td>58</td>
</tr>
<tr>
<td>BA</td>
<td>L</td>
<td>379</td>
<td>0.0046</td>
<td>1744</td>
<td>940/4</td>
<td>N</td>
<td>489</td>
<td>0.0018</td>
<td>828</td>
</tr>
</tbody>
</table>

Table I. Results obtained in creep tests

Minimum creep lives are recorded in tests carried out on specimens in the as-received (MA) condition. On the opposite side BA samples exhibited the maximum creep lives in both longitudinal and transverse directions and in smooth or notched specimens. A very great isotropy of properties was found and the life recorded in the specimen machined in the longitudinal direction was only a 15% longer than that in the transverse one. Moreover, the alloy in this heat treating condition possesses the highest fracture toughness but the main drawback is the poor ductility of the material [7]. Intermediate results obtained in alpha-beta field annealed samples (940/4) although a very low live, even lower than those in the mill annealed (MA) condition, was recorded in the transverse specimen.
Scanning electron microscope examination of the fracture surfaces of the broken specimens revealed the absolute similitude between the fractographic facets detected in both smooth and notched specimens corresponding to the same heat treating condition. Even if a thorough examination of the fracture surfaces was performed it was not possible to distinguish the various zones reported in reference [12]. This result seems to contradict the conclusions reached in that paper, where marked differences between the fracture topography of smooth and notched specimens were claimed. This it is even more surprising considering that the microstructure of those samples studied was very similar to that obtained after 4 hours annealing at 940° C and furnace cooled, condition analysed in the present work.

Moreover, metallographic analysis of the longitudinal sections of the other half specimens confirmed this lack of significant differences. The only difference that must be pointed out is the distribution of the creep voids confined to the notched section in notched specimens versus the more homogeneous distribution of voids in the smooth ones, where some voids were observed relatively far-away from the fracture surface. A plausible explanation for this absence of differences could be based in the mild stress concentration introduced by the Bridgman notch, used in the present work, compared with the higher ones produced by the sharper notches used in reference [12]. This mild stress concentration factor of the notch is in very good agreement with the longer lives recorded in notch specimens. It has been claimed that the rupture strength ratio between notched and smooth specimens for a given life initially increases with the notch sharpness, but reaches a peak and then increasing sharpness decreases this ratio [14]. In-depth research, considering various notch geometries, is recommended before a conclusion about the influence of the notches on the fracture topography could be reached.

Besides this similitude of the fractographic facets in the smooth and notched specimens, longitudinal and transverse specimens also exhibited a near identical fracture topography. The only difference was found in the microstructural analysis of 940/4 samples, as it will be commented later. This is even true if marked differences in creep lives between longitudinal and transverse MA specimens tested under the same stress level were obtained. The opposite trend was found in the fracture toughness where slightly higher values in the L-T specimens were obtained [7]. Consequently, it seems reasonable to associate these differences with the basal type texture of the material in this heat treating condition.

Fractographic examination of the broken BA specimens revealed a clearly intergranular topography. This result indicates that the grain boundaries are the preferential points for crack nucleation. Metallographic analysis of longitudinal sections confirmed this hypothesis as the microcracks were formed at the former beta grain boundaries, that now delineate the colonies of alpha needles, and at the triple-point corners. Figure I allows checking the accuracy of these comments. According with these observations grain boundary sliding along these former beta grain boundaries seems to be the operating failure mechanism.

The action of this mechanism has not been indicated in any of the previously mentioned papers but it must be indicated that in no one of them had analysed the alloy in this condition. However, the presence of creep cracks at these locations was observed in two alloys with an acicular microstructure; a near alpha alloy Timetal 1100 [15] and a cast Ti-6Al-4V alloy [16].

According to this mechanism a decrease in the length of these former beta grain boundaries will improve the creep life of the alloy in this condition and must be considered as positive. Moreover, an intergranular fracture was also observed in the beta annealed fracture toughness specimens where fracture propagated along a tortuous path, surrounding the individual large needles or the whole colony when the needles are finer [7].

As it has been indicated a strong directionality was found in alpha beta annealed 940/4 creep lives, being in the longitudinal specimens near twice those recorded in the transverse direction. Ductile dimples cover fracture surfaces of both longitudinal and transverse specimens and no significant differences were found. However, the optical microscope examination of the metallographic samples helps to explain the reasons for these different lives. Firstly, as it has been previously indicated, voids were detected at relatively long distances from the fracture surfaces, supporting their identification as real creep cavities and towards diffusional creep as the failure operating mechanism. Secondly, it was observed that creep cavities were nucleated at the alpha-beta interfaces sited at some points of the alpha grain boundaries and grew perpendicular to the loading direction. Thirdly, the use of higher magnification allowed detecting the intergranular character of these cavities and of the fracture surfaces. Finally, some microstructural differences between longitudinal and transverse specimens that could justify the different creep behaviour were observed.
Against the action of diffusional creep it is the very high Norton's law exponent (11.25) much higher than the lineal relationship between strain rate and stress proposed in the Coble and Nabarro-Herring models [5]. Two explanations could be given for this discrepancy. The first one would attribute the high value of the exponent to a change in the failure mechanism from diffusional creep for the lowest stresses to one of dislocation climb controlled regime. This change in failure mechanism would mean that there is not a unique relationship between strain rate and stress but one different for each mechanism. Consequently, the Norton's law exponent determined from these data will not be valid. However, the absolute similitude of the facets observed in both specimens points towards the operation of the same failure mechanism. The other possibility attributes this high exponent to the very reduced number of data (only two) used for its calculation, decreasing its accuracy. Even if more testing is recommended before a definitive conclusion could be reached, this explanation seems more plausible.

This point deserves further analysis as if the operation of diffusional creep is confirmed, according to the proposed models of Coble and Nabarro-Herring, a decrease of the secondary creep strain rate with the cube or the square of the grain size, respectively, can be expected. In addition to this increase in the creep lives an improvement in fracture toughness with no significant losses of the mechanical strength or the ductility will be achieved [7]. In a previous paper a direct relationship between the stress intensity factor and the beta phase mean free path (or the alpha phase grain size as the beta island are sited at the alpha grain boundaries) was obtained in specimens machined from 20 mm diameter bars of Ti-6Al-4V that were in a very similar heat treating condition [6].

About the intergranular character of these failures it must be indicated that in a previous paper where the fracture mechanism map of this same alloy in a very similar heat treating condition was constructed, only transgranular creep failure was reported. However, intergranular creep was found in the same work in an alpha Ti-5Al-2.5 Sn and a near alpha Ti-6Al-5Zr-0.5Mo-0.2Si alloys for lower stresses than those that produced a transgranular fracture and in the whole stress range of a beta Ti-15Mo alloy [8]. The possibility that intergranular failure could be also found in a Ti-6Al-4V alloy under a certain stress level seems reasonable and it is strongly supported by the metallographic analysis where cracking along the grain boundaries is evident (figure 2).

In a work where the damage micromechanisms during creep and tensile in an alloy Ti-6-2-4-6 at 500°C was studied it was indicated that the length of the alpha-beta interface did not affect the secondary creep strain rate [9]. However, it must be emphasised that the alloy microstructure in that study was different to the equiaxic alpha grains and beta phase sited in some of these grain boundaries observed in the present work. Besides, the nucleation of voids at the alpha phase, when a critical strain is reached, and/or at the alpha-beta interface, under high triaxiality, in a Ti-6Al-4V alloy has been reported. Three different types of fracture were found to occur depending on strain and triaxiality. Under very high triaxiality fracture is produced by the growth of the voids that were nucleated preferably at the alpha-beta interfaces. For lower triaxiality, voids were generated inside the alpha phase because of sufficient plastic strain and also at the alpha-beta interfaces because of sufficient triaxiality and the number of voids that were present in the material caused failure. Finally, when stress triaxiality is too low, no voids were nucleated during loading but plastic strain was localised in shearing bands and the alpha grains rotate so as to adapt more plastic strain. Materials then developed plastic strain instability at a macroscopic scale [10,11]. These observations seems to support the hypothesis given in favour of the nucleation of voids at the alpha-beta interfaces although it must be kept in mind that this type of nucleation was found under medium to high triaxiality and the specimens used in the present study are smooth or just have a mild stress concentration, being the stress triaxiality probably low. Once again testing of specimens with different notch geometries is needed to confirm the validity of the above-proposed hypothesis.

Microstructural differences between longitudinal and transverse specimens, that could justify the longer lives recorded in the former ones, where also observed in the metallographic analysis. As it is clearly seen by comparison between the micrographs of figures 3 and 4, longitudinal specimens exhibited longer mean free path between the points where cracks are nucleated (alpha-beta interfaces), rendering more difficult their growth and coalescence. Moreover, in this longitudinal sample a certain number of needles of alpha phase were found. The presence of these relatively large needles that must be surrounded by the cracks in their progression increases the time to failure.

Material in the as-received, mill annealed (MA) condition exhibited the minimum creep lives, together with the lowest fracture toughness [7]. Scanning electron microscope observation of the fracture surfaces of these specimens revealed that they were completely covered with shallow ductile dimples, pointing towards a microvoid coalescence mechanism as the responsible of the failure. These ductile dimples were observed in smooth or notched longitudinal and in transverse specimens indicating that the operating mechanism was always the same.
Metallographic analysis of longitudinal sections of these specimens allows observing a reduced number of voids, just sited in the neighbourhood of the fracture surfaces. It was considered the possibility that if these voids were too small they could pass undetected in the optical microscope even if they were really present in the specimens. To avoid this possible mistake metallographic specimens were analysed in the scanning electron microscope. This examination confirmed the complete absence of voids far away from the fracture surface. All these observations and the fact that these voids were markedly elongated in the loading direction (figure 5) suggested that they are not really creep cavities but voids that were generated by the plastic deformation of the material and dislocation climb activated by the testing temperature was responsible of these failures. This hypothesis is supported by the strong elongation of the alpha grains in the loading direction, notably stronger than that of the as-received microstructure, by the presence of very small equiaxed grains along the elongated grain boundaries, easily observed in the micrograph of figure 6, which have been associated with the dynamic recrystallisation of the material, and by the large reduction in area recorded in these specimens (over 55% compared with 19 to 27% in the beta annealed BA ones and from 25 to 35 in the alpha-beta annealed 940/4 ones).

Dynamic recrystallisation was also found during tensile testing of this Ti-6Al-4V alloy, annealed in the alpha-beta field, at temperatures above 700° C and specimens necked down nearly to a point with values of reduction in area exceeding 95% [8]. Voids near the fracture were elongated in the loading direction. It is evident that neither the loading conditions for both tests nor the heat treatments given to the alloy in each study are the same, but this similarities constitute another support for the hypothesis in favour of plastic deformation of the material and the action of dislocation climb as the mechanism responsible of the failure.

Fractographic facets and microstructure of the longitudinal and transverse samples are near identical and no reason for the anisotropy in creep lives was found. As it has been previously pointed out the opposite trend was observed in fracture toughness but, once again, no difference in the fracture topography of both, longitudinal and transverse specimens, was detected. Consequently, the more logical explanation for this anisotropy will associate the differences with the basal type texture that possesses the material in this heat treating condition. In any case, the very reduced number of tests does not allow reaching a conclusion about this anisotropy of creep lives.

4. CONCLUSIONS

a. - The creep behaviour of a 17-mm thick plate in various heat treating conditions has been analysed. All the tests were performed out at 455° C on smooth and notched specimens, machined in the longitudinal direction of the plate and on smooth specimens machined in the transverse one.

b. - No difference in fracture topography was found between smooth and notched specimens in the same heat treating condition. These results indicated that the operating failure mechanism were identical for both smooth and notched specimens.

c. - Maximum creep lives were achieved in BA samples independently of their orientation or load level and in both smooth and notched specimens. Fractographic examination of the fracture surfaces revealed a clearly intergranular topography that has been attributed to grain boundary sliding along the former beta grains.

d. - The failure operating mechanism for the alpha-beta annealed was diffusional creep by the nucleation and coalescence of the creep cavities generated mainly at the alpha-beta interfaces. Directionality in the creep lives could be attributed to the longer mean free path and the presence of a certain number of alpha phase needles in the longitudinal samples that were not observed in the transverse ones.

e. - This point deserves further analysis as if the operation of diffusional creep is confirmed, according to the proposed models, a decrease of the secondary creep strain rate can be expected. Previous work has demonstrated that larger grain size leaded to an increase in the fracture toughness of the alloy in this heat treating condition. Heat treatments that induce larger grain size could then be recommended.

f. - Material in the MA condition exhibits the lowest times to failure. The presence of a reduced number of voids, just sited in the neighbourhood of the fracture surface and the shape of these voids, suggested that they were not really creep cavities but voids generated by the plastic deformation of the material and dislocation climb activated by the high testing temperature the failure operating mechanism.

g. - No clear explanation has been found for the longer lives recorded in the transverse specimen as no fractographic or microstructural differences between the longitudinal and the transverse specimens have been
observed. More plausible explanation will associate these differences with the basal type texture that possesses the material in this as-received condition.

5. REFERENCES


Figure 1. Grain boundary sliding in BA specimen

Figure 2. Failure surface in 940/4 specimen

Figure 3. 940/4 specimen in the longitudinal direction

Figure 4. 940/4 specimen in the transverse direction

Figure 5. Voids in the MA specimen

Figure 6. Small grains and elongated voids in the MA specimen.