EXPERIMENTAL STUDY OF DAMAGE KINETICS IN $\alpha/\beta$ TITANIUM ALLOYS

A.L. HELBERT, X. FEUGAS, M. CLAVEL
Université de Technologie de Compiègne, Division Mécanique
Laboratoire LG2mS, URA CNRS 1505
BP 649 60206 Compiègne Cedex, France

ABSTRACT

Nucleation and growth kinetics in $\alpha/\beta$ titanium alloys have been experimentally and numerically investigated. Efforts were focused on the triaxiality effect. Experiments were performed on different types of specimens in order to use the local approach to fracture. It has been evidenced that internal stresses increase both nucleation and growth kinetics. Void growth has been divided into two stages. The first occurs in the $\alpha$-grains and the second through the surrounded $\beta$-matrix. In the second stage, growth rate strongly depends on the $\lambda/\phi$ ratio (where $\lambda$ is the center-to-center particle spacing and $\phi$ the particle diameter).

KEYWORDS

Void nucleation and growth kinetics, $\alpha/\beta$ titanium alloys, triaxiality, internal stresses, local approach to fracture, Gurson-Tvergaard’s model.

INTRODUCTION

Void nucleation and growth kinetics have been largely studied for alloys containing hard inclusions (Argon and Im, 1975; Le Roy et al., 1981; Beremin, 1981; Brownrigg et al., 1983; Marini et al., 1985; Becker et al., 1987; Gilormini et al., 1988,...). These authors have revealed the influence of mechanical parameters such as the hardening rate, $\sigma/d\sigma/d\varepsilon_p$ or the triaxiality, $\chi=\sigma/\sigma_0$, on the damage development, but also the role of the morphological parameters like the initial volume fraction of porosity or the $\lambda/\phi$ ratio (where $\lambda$ is the center-to-center particle spacing and $\phi$ the particle diameter). Only few studies have focused on damage kinetics for materials containing soft inclusions such as $\alpha$-particles in $\alpha/\beta$ titanium alloys. A few results (Margolin et al., 1980) report the effects of morphological parameters on the damage kinetics under low triaxiality, however the effect of triaxiality remains poorly studied. The purpose of this work is to focus on void nucleation and growth kinetics in $\alpha/\beta$ titanium alloys and to specify the influence of mechanical and microstructural parameters on these kinetics under a wide range of triaxialities.

EXPERIMENTAL PROCEDURE

Materials:

Three $\alpha/\beta$ titanium alloys were studied: the 6246 (773 K) alloy and the TA6V (300 K) alloy which presents two different microstructures (TA6Vg and TA6Vsa) resulting from various heat treatments. Chemical compositions and heat treatments for each alloy are given elsewhere (Helbert et al., 1996a, b). In particular, all the materials studied have the same content of aluminium (5% wt). The microstructure of these alloys is composed of a primary $\alpha$-phase (h.c.p. soft phase) surrounded by a $\beta$-phase (b.c.c. hard phase). The $\beta$-phase can be transformed (in 6246, TA6Vg) and is then composed of a mixture of residual $\beta$-phase and secondary $\alpha$-phase. The main mechanical and microstructural parameters are reported in table 1. The h.c.p texture is
Table 1: Mechanical and microstructural parameters of the three α/β titanium alloys. The hardening law has been chosen: σ = σ₀ + K(ε̇)² where σ₀ is the yield stress. λ is the mean center-to-center α-particle spacing.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>E (GPa)</th>
<th>G (GPa)</th>
<th>σ₀ (MPa)</th>
<th>K (MPa)</th>
<th>n</th>
<th>εₘ (µm)</th>
<th>λ</th>
<th>λ₀</th>
</tr>
</thead>
<tbody>
<tr>
<td>TA6Vα</td>
<td>110.0</td>
<td>90.00</td>
<td>1120.2</td>
<td>0.134</td>
<td>-0.23</td>
<td>0.3</td>
<td>479.5</td>
<td>50</td>
</tr>
<tr>
<td>TA6Vγ</td>
<td>130.2</td>
<td>912.2</td>
<td>1172.6</td>
<td>0.135</td>
<td>-0.23</td>
<td>0.3</td>
<td>455.4</td>
<td>47</td>
</tr>
<tr>
<td>6246</td>
<td>109.6</td>
<td>740.8</td>
<td>1004.0</td>
<td>0.10</td>
<td>-0.38</td>
<td>0.75</td>
<td>7646</td>
<td>33</td>
</tr>
</tbody>
</table>

similar for all the alloys studied and was previously described (Helbert et al., 1996a, b). An important number of heterogeneity levels are generally observed in α/β titanium alloys (prior-β grains, two phases (α, and β), transformed β structure, α₅ (TlA), precipitation in the α-phase, ... (Margolin et al., 1980; Helbert et al., 1996c; Feaugas et al., 1995). Plastic strain incompatibilities which result from these microstructural heterogeneities, induce important internal stresses. This explains the high value of the back stress (X) generally observed in α/β titanium alloys (Feaugas et al., 1995). This particular component of the macroscopic stress is obtained using the Cottrell’s method (1953). This method has been previously applied for the three materials studied here (Helbert et al., 1996c) and the results are given in table 2.

Table 2: Saturated values of the macroscopic stress σ, the effective stress σₑᵣ and the back stress X.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>σₑᵣ (MPa)</th>
<th>G (GPa)</th>
<th>σ₀ (MPa)</th>
<th>X (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>TA6Vα</td>
<td>450.0</td>
<td>674.0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>TA6Vγ</td>
<td>467.0</td>
<td>643.0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>6246</td>
<td>391.2</td>
<td>612.8</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Experiments and Calculations:

The local approach of fracture has been used for the void nucleation study of the three α/β titanium alloys. This method was largely described on steels and aluminum (Beremin, 1981; Hancock and Brown, 1983; Walsh et al., 1989) and allowed to examine a wide range of triaxialities with the use of axisymmetric notched specimens labeled AE2, AE4 and AE10. As far as the damage is studied under high and low triaxiality, smooth specimens (TL) were also used.

The specimens were tested to fracture or interrupted before fracture. Besides, a finite element calculation was performed for each specimen design to provide the mechanical parameters distribution in the bulk of the specimen during loading. The mesh is reorganized so as to take into account displacement. For each material, an elasto-plastic law is identified, from the framework of the classical elastoplastic theory (Chaboche, 1989). It has been experimentally demonstrated that it was not necessary to take into account damage in the behavior law since, just before fracture, the hydrostatic part of strain remains lower than 0.0025 (Helbert, 1996c). The validity of such calculations was checked comparing the experimental and calculated loading curves (Fig. 1) of notched specimens. The identification of the plastic law as described above allows the determination, during necking, of accurate calculated loading curves for smooth specimens (TL) (Fig. 2). Figure 2 shows the good agreement between calculated and experimental displacements during loading up to fracture. This macroscopic validity can be completed by a microscopic analysis of the local plastic strain of the α-particles measured from their shape (Helbert et al., 1996a; Bourgeois et al., 1996). If this local plastic strain is close to the calculated one, calculations are satisfactory.

Metallography and damage observations:

To understand the damage development, midsections of the specimens were polished and etched to be examined in the SEM. The midsection is divided into surface elements of 0.025 mm², in which the number of voids, Nv, as well as their length, L, are measured. Each of these elements is characterized by mechanical parameters provided by calculations (Helbert et al., 1996a).

RESULTS AND DISCUSSION:

Void nucleation:

Previous works on the α/β titanium alloys studied here have evidenced the nature of voids created in such materials, their physical origins and the associated nucleation criterion (Margolin et al., 1980; Helbert et al., 1996a, b, c). Voids form at the α/β interfaces or in the α-grains. The former voids are observed for the three materials whereas voids in α-grains are only present in the TA6V alloy. Observations and calculations allowed to conclude that both plastic strain and hydrostatic stresses are necessary to create cavities (Fig. 3). A void nucleation criterion at the α/β interface, based on these two parameters, was identified for each material (Table 3 and eq.1).

\[ \varepsilon_{\text{lim}} = -A \ln(BX - C) \]  

The plastic strain needed for nucleation, \( \varepsilon_{\text{lim}} \), is a function of triaxiality for α/β voids and \( \varepsilon_{\text{lim}} \) is constant and equals 0.05 for α-grain voids. In fact, it has been shown that the plastic strain needed for nucleation depends on the back stress X and its rate DX/DE₁ (the kinematic hardening rate) under a given hydrostatic stress (Helbert et al., 1996c). The purpose of the present work is not to detail the nucleation criterion part since it is done elsewhere (Helbert et al., 1996c), but more to focus on void nucleation and growth kinetics.

Table 3: α/β voids nucleation criteria for the three alloys studied.

<table>
<thead>
<tr>
<th>Alloys</th>
<th>A</th>
<th>B</th>
<th>C</th>
</tr>
</thead>
<tbody>
<tr>
<td>TA6Vα</td>
<td>0.128</td>
<td>1</td>
<td>0.5</td>
</tr>
<tr>
<td>TA6Vγ-α/β</td>
<td>0.133</td>
<td>0.66</td>
<td>0.26</td>
</tr>
<tr>
<td>6246</td>
<td>0.143</td>
<td>0.45</td>
<td>0.31</td>
</tr>
</tbody>
</table>
Void nucleation kinetics:

Using all specimens strained, it is possible to plot the number of voids per unit area, \( N_v \), versus plastic strain for different triaxialities. This plot reveals an exponential increase (Fig. 4) whatever the alloy. Both non-linear plots of \( N_v \) vs \( \varepsilon_p \) and dependence of this plot on triaxiality have already been observed on iron-based materials (Brownrigg et al., 1983) and aluminium alloys (Walsh et al., 1989). The exponential increase has also been reported on spheroidized steels (Le Roy et al., 1983) and Ti-5-5 alloys (Greenfield et al., 1972). As previously shown (Chu et al., 1980), the rate of voids nucleated per unit area \( dN_v/d\varepsilon_p \) can be expressed as:

\[
dN_v/N_v = D d\varepsilon_p
\]

where \( D \) is a function of triaxiality. In \( \alpha/\beta \) titanium alloys, \( D = E \exp (\chi) \) seems to be well adapted to describe the experimental results (Fig. 4). So as to account for the change in \( N_v \) with \( \varepsilon_p \) (Fig. 4), the following relationship is obtained:

\[
N_v = N_v^{\infty} \exp [D(\varepsilon_p - \varepsilon_p^{\infty})]
\]

Table 4: Values of \( E \) and \( F \) parameters used for the nucleation kinetics of three different alloys.

<table>
<thead>
<tr>
<th></th>
<th>E</th>
<th>F</th>
</tr>
</thead>
<tbody>
<tr>
<td>TA6Vα</td>
<td>1.11</td>
<td>4.21</td>
</tr>
<tr>
<td>TA6Vγ-α</td>
<td>4.12</td>
<td>1.96</td>
</tr>
<tr>
<td>TA6Vγ-β</td>
<td>1.21</td>
<td>2.39</td>
</tr>
<tr>
<td>6246</td>
<td>1.28</td>
<td>2.26</td>
</tr>
</tbody>
</table>

Experimental Study of Damage Kinetics in \( \alpha/\beta \) Titanium Alloys

\( N_v = N_v^\infty + N_v^\alpha \), where \( N_v^\infty \) and \( N_v^\alpha \) are both described by equation (3). An example of the good agreement between equation 3 and experimental results is shown in Fig. 4 for the TA6Vα alloy. The comparison of the rate of nucleated voids, \( D \), versus triaxiality for the different alloys is shown in Fig. 5. Whatever the stress triaxiality, \( D \) increases with the back stress \( \chi \) (see table 2). The plastic strain incompatibilities associated to \( \alpha/\beta \) interfaces assist the nucleation kinetics.

**Void growth kinetics:**

Void growth at the earlier stage of plastic strain produces spherical voids independent of the nucleation location (\( \alpha \) or \( \beta \) interface). As strain increases, the longest voids are most often elliptic in shape. Assuming that voids width, \( \varepsilon \), stays constant (2 \( \mu \)m) during void growth, this latter will be characterized in the following by the change in the void length, \( L \). From scanning observations, it appears that void growth begins in the \( \alpha \)-phase and then carries on in the \( \beta \)-phase with difficulties (Fig. 6 a, b). This second stage of growth can be assisted by the fracture of the \( \alpha \)-particles in the vicinity of the principal void. It is well established that void growth rate increases when the yield stress and the hardening rate decrease (Gilorimini et al., 1988). This explains why the void growth firstly occurs in the \( \alpha \)-phase where the yield stress and the hardening rate are lower than in the \( \beta \)-phase. An experimental analysis of the specimens leads to the plot of the longest void length, \( L \), versus triaxiality for given plastic strains or to the plot of \( L \) versus plastic strain for fixed triaxialities. These two curves are determined for each alloy and show exponential increases (Fig. 7). Such a result has been previously observed in aluminium alloys (Walsh et al., 1989) and in titanium alloys (Nurendrath and Margolin, 1988).

![Fig. 6: Micrographs of void growth in the \( \alpha \)-grains a). Void growth seems to be slowed down by the \( \beta \)-phase b).](image)

![Fig. 7: Increase of the longest void length \( L \) with plastic strain a), with triaxiality b).](image)

Numerous studies have provided analytical laws to describe a single-void growth. Various void geometries (cylindrical, spherical) and matrix behaviors (rigid-plastic, linearly or power-law viscoelastic) have been used (Gilorimini et al., 1988). These approaches agree with the following...
classical growth law:

$$\frac{dR}{dt} = f(\gamma) \text{d}_{\text{eq}}$$

(4)

where $R$ is the mean void radius. In the present work, $L$ will be used instead of $R$ since it characterizes void growth. The void volume equivalence between an ellipsoidal and a spherical void allows the use of an isotropic Gurson model. Among these analyses, the Gurson-Tvergaard model, where the second order terms are neglected, seems to be more appropriate to describe void growth in the $\alpha/\beta$ titanium alloys since it takes into account a randomly distributed volume fraction $f$ of voids. Gurson’s yield function is written as follows:

$$\phi = \Sigma_{\text{eq}}^2 - \sigma_0^2 [1 - 2 q_1 f \cosh(\frac{3}{2} q_3 \chi)]$$

(5)

where $q_1$ and $q_3$ were introduced by Tvergaard (Tvergaard, 1981) to bring predictions of the model into closer agreement with full numerical analyses of a periodic array of cylindrical voids. Assuming macroscopic normality, the hydrostatic strain rate, $\text{d}e_{\text{eq}}$, is expressed as:

$$\text{d}e_{\text{eq}} = \frac{3}{2} f q_1 \sinh(\frac{3}{2} q_3 \chi) \text{d}_{\text{eq}}$$

(6)

with $q_1 = q_3 q_2$. Besides, taking into account the plastic incompressibility of the matrix and the mass balance, the void growth law can be written as:

$$\frac{dL}{dt} = (1-f) \frac{dL}{d\text{eq}} = (1-f) \frac{3}{2} f q_1 \sinh(\frac{3}{2} q_3 \chi) \text{d}_{\text{eq}}$$

(7)

From $dL/d\text{eq} = f \text{d}V_{\text{eq}}/\text{V}_{\text{eq}}$ and equation 7, $dL/L$ can be easily deduced assuming that $dV_{\text{eq}}/\text{V}_{\text{eq}} = 3$ $dL/L$ since the void width, $\chi$, is a linear function of $L$. Then, it comes:

$$\frac{dL}{L} = \frac{3}{2} q_1 \sinh(\frac{3}{2} q_3 \chi) \text{d}_{\text{eq}}$$

(8)

From experimental results on void growth kinetics, the constants $q_1$ and $q_3$ have been determined for each material (see table 5). $q_2$ is different from an alloy to the other since, in this first approach, the hardening nature has not been clearly expressed in eq.5. More recently, Mear and Hutchinson (1985) proposed to introduce the effect of the isotropic and kinematic hardening of the matrix in the Gurson’s yield function. As it can be seen in table 2, the values of the internal stresses developed in the $\alpha/\beta$ titanium alloys are quite high. In order to give a better description of the void growth process, it seems then necessary to take into account these internal stresses and to define the local triaxiality rate, $\chi_{\text{int}}$, which is involved in void growth. In $\alpha$-grains, the local triaxiality is written as: $\chi_{\text{int}} = \sigma_1/\sigma_0$, where $\sigma_1$ is the hydrostatic pressure and $\sigma_0$ is the effective stress without damage, $\sigma_{\text{eff}} = \sigma - \chi(1 + \chi/\sigma_0)$. Then, $\chi_{\text{int}} = \chi(1 + \chi/\sigma_0)$.

<table>
<thead>
<tr>
<th>Table 5: Growth parameters</th>
<th>$q_1$</th>
<th>$q_2$</th>
<th>$q_3$</th>
<th>$(1+\chi/\sigma_0)$</th>
</tr>
</thead>
<tbody>
<tr>
<td>TAV6a</td>
<td>1.2</td>
<td>3.5</td>
<td>1.4</td>
<td>2.498</td>
</tr>
<tr>
<td>TAV6g</td>
<td>0.3</td>
<td>3.3</td>
<td>1.4</td>
<td>2.777</td>
</tr>
<tr>
<td>6246</td>
<td>0.999</td>
<td>3.6</td>
<td>1.4</td>
<td>2.567</td>
</tr>
</tbody>
</table>

The flow rule is then written as:

$$\phi = \frac{1}{2} \sigma(\Sigma - \sigma_{\text{eq}})^2 - \sigma_0^2 \chi_{\text{int}} (1-2 q_1 f \cosh(\frac{3}{2} q_3 \chi_{\text{int}}))$$

(9)

and the growth law form is:

$$\frac{dL}{L} = \frac{3}{2} q_1 \sinh(\frac{3}{2} q_3 \chi_{\text{int}})$$

(10)

where $I_1(\Sigma - \sigma_{\text{eq}}) = \frac{3}{2} (S - X) + (S - X)\chi^2$ ($S$ is the deviatoric part of the macroscopic stress $\Sigma$) and $X$ is deviatoric. For the alloys studied, the hardening is saturated when damage appears. So, $\chi_{\text{int}}$ can be expressed as: $\chi_{\text{int}} = \chi(1 + \chi/\sigma_0)$. Comparing eq. 8 and 10, it comes:

$$q_1 \chi_{\text{int}} = q_3 (1 + \chi/\sigma_0) = q_3$$

(11)

From the present calculations, it is found that $q_1 = 1.4$ and remains constant for the alloys studied. This value is in agreement with results on the Ti40 alloy where the internal stresses are low (Huez et al., 1995). $q_1$ is associated to void growth in the $\alpha$-phase. As shown in Fig. 8, the void growth in the $\alpha$-phase increases with the internal stresses. Previous studies report that $q$ increases with the initial volume fraction of porosity in the material (Gillorini et al., 1988). In particular, for alloys containing hard inclusions, $q$ has been directly linked to the amount of inclusions (Marini et al., 1985). This dependence on the volume fraction of porosity is often explained as the consequence of interactions between voids which increase the growth rate (Zhang and Niemi, 1995). In the present study, the effect of $\alpha$-phase percentage on $q$ has not been clearly evidenced, on the contrary, $q$ decreases with increasing $\lambda/\phi$ ratio (center-to-center $\alpha$-particle spacing over the $\alpha$-particle diameter) (Fig. 9).

As mentioned above, void growth first occurs in the $\alpha$-grains. Then, the propagation of damage from an $\alpha$-grain to the other is quite dependent on the microstructure, especially on the $\lambda/\phi$ ratio. Indeed, the interactions between voids are higher and the local necking is easier when voids are closer. If $L/\lambda$ is the ratio of the center-to-center $\alpha$-void spacing out of the $\alpha$-void length, it has been demonstrated that the constraint factor, $\alpha/\phi$ (where $\phi$ is the mean stress in the intervoid matrix), increases when $L/\lambda$ decreases (Thomason, 1985). Then, during loading, the fracture criterion of the matrix is reached first for small $L/\lambda$ ratios. In $\alpha/\beta$ titanium alloys, the $\lambda/\phi$ ratio corresponds to the lowest $L/\lambda$ ratio and it is at the origin of the first stage of $\alpha$-voids coalescence. Besides, according to the alloy, this stage of matrix fracture depends on the nature of the $\beta$-phase. On the fracture surface of TAV6g, a dimple ductile rupture process is observed between two $\alpha$-voids (Fig. 10a). The fracture is due to the presence of $\alpha$-inclusions in the $\beta$-matrix. On the contrary, for the TAV6a, when no inclusions are observed in the $\beta$-matrix, the $\alpha$-void coalescence results from a plastic failure of the $\beta$-phase (Fig. 10b).

Fracture process:

The mechanical conditions of fracture, in the space $(\chi$, $\epsilon_{\phi})$, obtained for different specimens (AE, TL) of each alloy are reported in Fig. 11. Using the nucleation and growth kinetics laws identified above, the volume fraction of voids associated to the nucleation of new voids, $f_0$, ($df_0/dt$ = d$t/\tau$) with $G_0 = 3.14 \times 10^4$ a geometrical parameter) and the one associated with the growth of actual voids, $f_\phi$, have been calculated at fracture for the three alloys. The $f_0/f_\phi$ ratio at
fracture is reported as a function of triaxiality for the three alloys (Fig. 12). For the TA6Vg and the 6246 alloys, three domains must be distinguished with the triaxiality:

- Under low stress triaxiality, \( \tau_0/\tau_n < 1 \). The growth process is negligible and the nucleation kinetics is the cause of fracture.
- For high triaxialities, \( \tau_0/\tau_n > 1 \). The nucleation process can be neglected in front of the growth process which involves fracture.
- In a middle range of triaxialities, the nucleation and growth processes occurs at the same time. Fracture results from a coupling of the two processes.

These domains depend on the material and act in more or less large range of triaxiality. The TA6Vg, for example, presents nucleation and growth coupling in a large range of triaxiality.

**CONCLUSION**

Void nucleation and growth kinetics have been studied for three \( \alpha/\beta \) titanium alloys. It has been evidenced that the internal stresses increase both nucleation and growth rates. The first growth stage occurs in \( \alpha \)-grains, and the second stage is mainly dependent on the \( \beta \)-matrix fracture process. The growth rate of this second stage increases when the microstructural \( \lambda_0/\lambda_0 \) ratio decreases. Three fracture modes have been identified depending on triaxiality and on the alloy. Under low triaxiality, fracture occurs by a nucleation process. Under high triaxiality, void growth and coalescence process lead to fracture. Finally, in the middle range of triaxiality, both nucleation and growth processes occur at the same time.

**REFERENCES**


