Long crack growth mechanisms in Ti-6Al-4V alloy

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A transition mechanism of long crack growth from structure sensitive (Regime I) to insensitive region (Regime II) was investigated at (i) room temperature under high humidity conditions, and (ii) elevated temperatures for Ti-6Al-4V alloy and other materials. Constant amplitude tests were conducted using several stress ratios \((R = 0.05, 0.1, 0.4\) and \(0.7)\) for the high humidity tests and 0.1 for room and elevated temperature tests. Conventionally forged Ti-6Al-4V alloy disks, processed to the solution treated and over-aged condition, were studied for both programs. An increase in stress ratio and temperature lowered the transitional stress intensity factor range where Regime I transitioned to Regime II. Higher stress ratios \((R = 0.7)\) accelerated the fatigue crack growth rates many times the crack growth rates obtained at lower stress ratios \((R = 0.05)\). However, higher temperatures \((345°C)\) influenced the crack growth rates only marginally. The mechanisms controlling elevated temperature fatigue crack growth in Ti-6Al-4V were by the secondary crack formation, striations and some cavity features on the fracture surface. The damage was localized on the \(\alpha\) platelets, where a type dislocations found with the Burgers vectors \<(a/2,12,0)\> \(g = 1010\) near \((10 I 0)\).

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The growth of long fatigue cracks in Ti-6Al-4V is a function of testing parameters and material parameters such as microstructure. The microstructure of Ti-6Al-4V, which is a \(\alpha-\beta\) alloy, is achieved by heat treatment summarized elsewhere\(^1\). A number of studies are found in the literature\(^2\)-\(^{20}\) describing the fatigue crack growth behaviour of Ti-6Al-4V alloy and establishing microstructure-property correlations. With increased attention given to high cycle fatigue\(^{12,17-20}\) being one of the potential failure modes in military aircraft engine disks and blades made of titanium alloys, room and elevated temperature fatigue crack propagation (FCP) behaviour is currently being re-evaluated. The evaluation of interaction effects of mixed fatigue (interactions between low and high cycle fatigue), foreign object damage and fretting aspects are discussed elsewhere\(^{17-20}\). This paper provides additional fatigue crack growth rate data at elevated temperatures for Ti-6Al-4V. Since a major percent of component life is spent in nucleation and growth of a crack to a detectable size, crack growth studies are conducted both in Regime I (low ranges of \(da/dN\) and \(\DeltaK)\) and Regime II or Paris region. Thus, the study of transition of Regime I to II and modeling this behaviour will be useful in structural integrity assessment of aircraft structural components.

Crack growth behaviour of Ti-6Al-4V shows a bilinear behaviour in that two slopes represent Regimes I and II respectively. Figure I schematically shows the stress intensity factor ranges as a function of the crack growth rates for the two Regimes. The mechanisms of FCP in Ti-6Al-4V are structure sensitive at the low \(\DeltaK)\) values and depend upon the \(\alpha-\beta\) packet size\(^6\)-\(^11\). The reversed plastic zone at the tip of a fatigue crack has been related with \(\alpha-\beta\) packet size\(^8\)-\(^11\), in that the reversed plastic zone size smaller than the \(\alpha-\beta\) packet size causes Regime I growth and vice-versa for Regime II. The transition

\[\frac{da}{dN}\]

Regime II

\[\Delta K_r\]

Stress intensity factor range

Fig. 1—Schematic representation of Regime I and Regime II crack growth behaviour in Ti-6Al-4V alloy
between structure sensitive Regime I behaviour to structure insensitive Regime II behaviour, occurs at intermediate range of stress intensity factor range. In the literature, Mode I ΔK was reported to be from 10 to 30 MPa\·m where transition occurs. This paper examines the parameters that influence the transition characteristics of long crack growth for a number of structural materials used in aircraft structure and develops an empirical model to predict the transition point.

Fatigue crack growth data from two test programs are presented revisiting the mechanisms of FCP under different stress ratios (R) at high humidity and elevated temperature environments.

**Experimental Procedure**

**Room temperature tests**

Forged bars were used in this program from which the specimens were machined in the R-L orientation. The bar was processed to solution treated and aged condition (STOA). The processing parameters for STOA were kept same for both programs (Table 1). Tests were conducted to study more fully the transition mechanisms from Regime I to Regime II at different stress ratios. Since the environment influences the FCP behaviour and a wide range of transitional stress intensity (10-30 MPa\·m) reported in the literature, tests were conducted below Mode I stress intensity factor range (ΔK) of 20 MPa\·m to document the onset of the transition. Both pre-cracking and FCP tests were conducted in high humidity (over 85% relative humidity) using a frequency of 10 Hz. All the samples were prepared per ASTM E-647-95 by inducing an EDM notch at the center of the specimen. Crack growth measurements were made using resistance gauges mounted on each face of specimens. A total of twelve specimens were tested, 5 each with R = 0.05 and 0.7, and 2 with R = 0.4.

**Elevated temperature tests**

Forged disks were supplied by AlliedSignal in the solution treated and over-aged (STOA) condition. Details of STOA parameters are outlined in Table 1. Ti-6Al-4V is typically solution treated between 955 and 970°C and water quenched. When the material was subsequently held at or near the annealing temperature, which was above normal aging temperature (700°C), an over-aged condition resulted. The applied heat treatment resulted in the development of duplex microstructure of primary α, and platelets of α in β (Fig. 2), containing acicular alpha with an aspect ratio of approximately 10:1. However, a very different microstructure was observed when a specimen was made from a different area (flange area) (Fig 3). This microstructure (Fig. 3) contains slightly distorted, coarse, plate like alpha grains with randomly equiaxed alpha grains (light) in beta phase (seen dark), with an aspect ratio of 10:1. Specimens were machined such a way that the loading was in the radial direction, whereas the crack growth occurred in longitudinal direction (R-L), representative of actual stress distribution in an engine disk from the blade.
attachment root or from bolt-holes. The dimensions of the compact tension (CT) specimens used for the tests were in conformance with ASTM E647.88. A direct current potential difference (DCPD) system was used to record the crack growth of CT specimens and procedure reported elsewhere.13,14

A total of sixteen CT specimens were tested in this program, eight from each disk and each test was duplicated. Specimens numbering 1 and 2 represent Disk I and 3-4, Disk II, respectively. It is likely that variations in the data may arise from locations from where specimens were machined, product form (disk and bar), forging ratio (different % volume reduction ratio) used for bar and disks, and methods used to monitor crack growth. Since the plane strain fatigue crack growth rate data were generated for both high humidity testing and elevated temperature testing, the data were compared to derive trends in behaviour and relating the transitional stress intensity factor range for each condition. Monotonic properties are presented in Table 2 for different temperatures. The material became slightly ductile (measured by % elongation) followed by lower yield strength, and modulus with temperature increase.

Additionally, in a separate test effort, cyclic strain accumulation tests were conducted under total strain control and was recorded for several hourglass specimens in Fig. 4 showing behaviour under repeated cyclic plastic deformation for room temperature, 175, 230, and 290°C. At the same cycles to failure (400 cycles) the strain accumulated at room temperature was 0.004, whereas at 290°C, 0.0055 (Fig. 4), which is over 35% higher, meaning higher strain accumulation and reduced cyclic life. As a result, the low cycle fatigue behaviour of Ti-6Al-4V at high temperature is inferior from its high cycle fatigue resistance. Similar trends are shown for other combination of test parameters.

For the fatigue crack growth tests, the tests were continued with the crack length to specimen width ratio, (a/W) of 0.24 to 0.7 within the range specified in ASTM E-647-88. Fatigue crack growth rates were calculated from incremental measurements of crack growth.

The incremental crack growth rates were correlated with ΔK values derived from the mean of the crack growth interval, (a_i+1 + a_i)/2.

### Results

#### Room temperature tests

A number of stress ratios were studied ranging from 0.05 to 0.7. Pre-cracking was performed with sinusoidal waveform at 10 Hz in a high humidity environment. Each specimen was pre-cracked 6 mm on the middle tension (MT) specimens 101 mm wide and 6.25 mm thick on the average. The crack typically propagated to a length of 45-46 mm, where ΔK_i was 13-16 MPa√m, and cycles applied ranged from 2 × 10^5 to nearly 7 × 10^6. The number of cycles to pre-crack the specimen varied from a few hundred thousand cycles to several million cycles.

### Table 2—Mechanical properties of Ti-6Al-4V alloy disks

<table>
<thead>
<tr>
<th>Temperature °C</th>
<th>Yield strength MPa</th>
<th>Young modulus GPa</th>
<th>% elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td>Room</td>
<td>813</td>
<td>120</td>
<td>12.4</td>
</tr>
<tr>
<td>175</td>
<td>800</td>
<td>110</td>
<td>17.1</td>
</tr>
<tr>
<td>230</td>
<td>772</td>
<td>109</td>
<td>18.7</td>
</tr>
<tr>
<td>290</td>
<td>760</td>
<td>100</td>
<td>20</td>
</tr>
</tbody>
</table>

![Fig. 4—Strain accumulation in Ti-6Al-4V alloy at room temperature, 175, 230 and 290°C.](image)

![Fig. 5—Crack propagation behaviour of Ti-6Al-4V in high humidity environment.](image)
Environment for all the tests was high humidity for both pre-cracking and fatigue crack growth phase. Tests were performed under constant load.

Fatigue crack growth behaviour of Ti-6Al-4V alloy is presented in Fig. 5. The data presented in Fig. 5 shows the \( \frac{da}{dN} \) and \( \Delta K \) for various stress ratios. All the data were below \( \Delta K \) of 20 MPa/\( \sqrt{\text{m}} \) to show the two Regimes in crack growth and capture the onset of transition.

**Elevated temperature tests**

Pre-cracking was performed with sinusoidal waveform at 20 Hz using a stress ratio \((R)\) of 0.1. Upon the specified pre-crack length was achieved, the specimens were tested for fatigue crack growth phase using the same test parameters except the maximum load reduced. A \( \Delta K \) gradient of 4 MPa/\( \sqrt{\text{m}} \) was used for load increments as the crack progressed. The fatigue crack growth behaviour is presented in Fig. 6 for elevated temperature tests at a constant stress ratio of (0.1), however, temperature varied from 175 to 345°C.

**Discussion**

**Room temperature tests**

The data presented in Fig. 5 shows the structure sensitivity effect for the lower stress ratio tests conducted at \( R=0.05 \) and 0.4. For these tests, microstructure has been found to dictate the mechanics of crack growth at the lower ranges of \( \Delta K \). As a result, slope of the \( \frac{da}{dN} \) and \( \Delta K \) changes early into the cracking. This effect is also interpreted in the literature in terms of texture effect\(^9,10\). As the stress ratio increases to \( R=0.7 \), the crack growth rates become much larger (a 3 to 5 times) at comparable \( \Delta K \) found in this study. However, an order of magnitude variation at comparable \( \Delta K \) has also been reported in the literature\(^2\). Several factors contribute to this accelerated crack growth, viz., the packet size of \( \alpha-\beta \) structure, and reversed plastic zone effect\(^7\) which depends on packet size. For the higher stress ratio tests, as \( R \) increases, structure sensitivity (causing multiple slopes in \( \frac{da}{dN} \) versus \( \Delta K \) below 10 MPa/\( \sqrt{\text{m}} \)) tends to diminish as observed in Fig. 6. For all the tests conducted in high humidity, the transition of structure sensitive to structure insensitive mode (Regime I to Regime II) was found to be lower than the tests conducted in laboratory air environment\(^3\). The transitional \( \Delta K \) was found to be from 3 to 10 MPa/\( \sqrt{\text{m}} \), also found in some studies in literature and higher within 30 MPa/\( \sqrt{\text{m}} \) summarized elsewhere\(^8,9\). Elevated temperature tests show distinct regions where Regime I and II exist and transition effects. High humidity in pre-cracking phase has been found to initiate pits from where cracking was observed in high strength aluminum alloys and a typical crack growth behaviour is presented in Fig. 7 for 7075-T76511. A pronounced texture and/or microstructure effect is exhibited for this material at lower stress ratios. However, as the \( R \) became higher (0.7) the transitional \( \Delta K \) reduced considerably. Figure 7 also shows the pronounced crack acceleration at the
Regime I with low $\Delta K_i$. Corrosion, as found in aluminum alloys, may be one of the contributing factors for accelerated crack growth in Ti-6Al-4V that lowers the transitional $\Delta K_T$ for Regime I phase at higher $R$ ratios (e.g., $R = 0.7$).

Elevated temperature tests

Crack growth rates vary from $10^{-9}$ to $10^{-5}$ (m/cycle) (Fig. 6) within the range of Mode I stress intensity factor 10 to 53 MPa$\sqrt{m}$. Straight line fits for the data were determined within the range of crack growth rates from $10^{-8}$ to $10^{-6}$ m/cycle, representative of linear or power law regime. The parameters of the Paris equation, $d\alpha/dN = C(\Delta K)^m$ (slope of linear line, $m$, and material parameter, $C$), for each disk are presented elsewhere\(^b\), for all temperatures for both the disks (these parameters are material dependent). A two-parameter analysis of variance method was used to correlate the disk-to-disk variation in crack growth rates and appeared to be very small\(^b\). Also FCP rates obtained in this study for elevated temperature tests were compared with other studies\(^3,4,6,8\) in earlier work. A marginal variation was observed for different studies\(^5,6,8,15\). It is evident from Fig. 6 that crack growth rates within the mid-range ($>10^{-7}$ m/cycle), no substantial influence of temperature has been observed. Crack propagation rate at 345°C was slightly higher than at room temperature test. This may be a result of lower strength high ductility exhibited by Ti-6Al-4V at high temperatures. Similar results were reported for Ti-6Al-4V and other Ti-8Al-1Mo-1V at elevated temperatures\(^1,3,7,10\). The transition of Regime I cracking to II occurred for all the cases above 10 MPa$\sqrt{m}$ (refs 1-3, 7-10).

Each region, Regime I and II, has distinct features, shown schematically in Fig. 1. For the materials studied in this paper, the structural sensitivity in the crack growth rates occurs for stress intensity factor below 10 MPa$\sqrt{m}$ beyond this point the crack growth process is likely to be independent of microstructure. The slope of the crack growth rate curve in Regime I is higher than the Regime II, drawn schematically in Fig. 1. This behaviour has been interpreted in terms of cyclic plastic zone becoming equal to the $c\beta$ packet size\(^9,10\). This point also describes the boundaries within which structural sensitivity occurs. In the literature, several expressions have been proposed empirically\(^7,11\). Describing cyclic plastic zone in aluminum and titanium alloys.

Reversed plastic zone size $= 0.132 (\Delta K_i/F_y)^2$ \hspace{1cm}(1)

where, $\Delta K_i$ is the range of Mode I stress intensity factor in MPa$\sqrt{m}$, and $F_y$ the tensile yield strength in MPa of the material.

For titanium alloys transition point was determined to be

Transition point $(\Delta K_T) = 5.50 F_y (d)^{0.5}$ \hspace{1cm}(2)

where, $d$ is the grain diameter using the ASTM specifications.

A number of aluminum and titanium alloys were analyzed by the author, in a separate effort, that are used in aircraft structures (work performed at Cessna Aircraft Co., Wichita, KS)* and the correlations were made with the following equations.

Reversed plastic zone $= 0.5 (\Delta K_i/F_y)^2$ \hspace{1cm}(3)

Whereas,

Transition point was at $(\Delta K_T) = 2.8 F_y (d)^{0.5}$ \hspace{1cm}(4)

These expressions were verified with the materials used in this study (Figs 5-7) and other materials and results show very good agreement for both high-humidity and elevated temperature tests. All the data analyzed with Eq. (2) were tested in high humidity environment only. A two-parameter analysis of variance (ANOVA) was performed on the crack growth rate data and parameters of crack growth rate

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*Part of an internal study conducted in 1997-98 for Durability and Damage Tolerance Group.
equation \((C \text{ and } m)\) were analyzed\(^{15}\). The crack growth rate equation parameters (Paris equation) do not show disk to disk variation and both \(C\) and \(m\) values appeared to be very close.

Regime I fatigue is normally observed on high cycle-low stress fractures and is frequently absent in low cycle-high stress fatigue (Fig. 4) where fracture topography contained mainly striations (Fig. 9). The crack follows crystallographic planes but changes direction at discontinuities such as grain boundaries and other particles for Regime I. A typical Regime I fracture is shown in Fig. 8, showing the crack path through the crystallographic planes. The fracture topography contained such features as facets often resembling cleavage where striations do not occur. Figure 8 shows the Regime I behaviour observed at 345°C, implying that these features were common, more pronounced for lower temperature tests.

Figure 9 shows transgranular fracture, which is more influenced by the magnitude of the alternating stress than by mean stress or microstructure. Regime II fracture surface topography contained such features as striations in Fig. 9 ranging from 175 to 345°C. A variation in stress, temperature, microstructure, frequency and environment can change the orientation of the plane of fracture and alter the direction of striation alignment. These features are observed for all the tests as the alignment of striation planes were different for each case. These alignments (Fig. 9) are due to the alignment of \(\alpha-\beta\) planes in the microstructure and have not been investigated in this study.

Fatigue crack growth process is a complex interactive phenomenon among material, test,
Fig. 13—TEM micrographs of the elevated temperature fatigue crack growth mechanisms in Ti-6Al-4V alloy: (a) and (b) typical microstructure showing platelets of $\alpha$ in $\beta$; (c) Persistent slip band features at 290°C; (d) Persistent slip band features at 345°C; (e) a type dislocations with Burger's vector $(a/2)<11\overline{2}0>$ and $g = 10\overline{1}0$ near $(10\overline{1}0)$; (f) a type dislocations with Burger's vector $(a/2)<11\overline{2}0>$ and $g = 10\overline{1}0$ near $(10\overline{1}0)$; (g) dislocation network at 345°C; (h) dislocation network at 175°C
environment and other parameters. Therefore, accurate prediction of FCGR is a challenge. Higher temperature and aggressive environment may reduce greatly the crack growth resistance of a material. Cyclic cleavage, secondary cracks and voids formed as the temperature increased (Fig. 8). Secondary cracks often reduce the stress severity at the tip of the main crack leading to increased crack growth resistance as the test temperature increased. However, there are competitive mechanisms among secondary cracks, secondary cracks interacting with the primary crack, and other interactions are very difficult to interpret and quantify. Some voids were observed at 290 and 345°C as shown in the fractograph Fig. 10, which may provide crack linking path for the advancing crack.

Secondary cracks were observed in the Regime II of the fatigue crack growth region (see Fig. 11). In that area fine fracture features were seen compared to the early Regimes of crack growth process where fracture surface features were very rough showing cleavage, crack path changes and crack growth rate interacting with microstructure.

The room temperature crack growth process was dictated by the nature of the α-β grain structure, and distribution of alpha platelets in beta grain, in which cyclic cleavage was more deleterious and increased FCP rates. For higher temperature tests (at 345°C) some cavities formed indicative of creep effect. At some point towards the end of Regime II, in Regime III, dimples were observed and fracture was by intergranular mode as shown in Fig. 12. These features were similar to that reported earlier1-3,10.

Mechanisms of fatigue crack growth

The fatigue crack growth mechanism of Ti-6Al-4V has been evaluated with transmission electron microscope (TEM) of specimens tested at R = 0.1, 10 Hz frequency and tested at 175, 230, 290 and 345°C21. A parallel, thin layer of material was removed containing the fracture surface to prepare the samples for TEM studies. Disk samples were removed at various locations along the fracture path simulating slow, steady and rapid crack growth. Special care was taken to locate the TEM foil within the crack tip plastic zone to capture the mechanisms. Figure 13 (a-b) shows a typical alpha-beta microstructure under TEM. The features of persistent slip bands were visible in Fig. 13 (c-d) at two temperatures shown. The dislocations were mainly concentrated in α-grains with only a small fraction present in β-grains. Within the α-grains, <a> type dislocations formed with Burgers vectors (a/2) <1120> g = 1010 near (1010) (Fig. 13e). The same location was seen in Fig. 13f with a tilt g = 0002 near (1010). Some twinning together with other features appeared to be <c> type dislocations in planner arrays with g = 0002 near (1210) was found.

The observation of thin foils also revealed an increase in the number and density of both dislocations and twins as the cycles and crack length increased. However, the observed arrangement of dislocations was generally random with lines of dislocations emanating from the grain boundaries. The presence of deformation twins suggests that the intersecting twins are potential sites for micro-crack formation. As the temperature increased, 290-345°C, increased activity in the secondary crack formation perpendicular and/or at some angle to the load line was documented in the fracture surface (Fig 9). For a Ti-48Al-2Cr-2Nb intermetallic similar observations were made22 and found that presence and interaction of the deformation twins is strongly dependent on the mutually competitive influences of (i) orientation, (ii) grain size, and (iii) texture of the specimen. At the lower cyclic strain amplitude the resultant lower response stress, a greater number of dislocations was observed in regions of single phase, γ-grains and coarse twin-related γ-lamellae with no α2 present, with only a marginal increase in the γ lamellae.

The marginal effects of temperature on the fatigue crack growth rates may have been due to the following: (i) greater activity of dislocations Fig. 13 (g-h) and (ii) an increased activation of the dislocation sources.

A major amount of strain energy, from the crack growth test, may have been utilized in deformation process to initiate twins. However, dislocation mobility may have supplemented this process latter on the crack growth process showing the marginal differences in crack growth rates (Fig. 6).

Conclusions

The following conclusions may be drawn from this study: (i) The transition mechanisms in Ti-6Al-4V were a function of environment in which tests were performed. (ii) High humidity and elevated temperature environments reduced the stress intensity
range where Regime I transitioned to Regime II. This was also influenced by stress ratio. (iii) At higher stress ratio, $R = 0.7$, the transition phenomenon diminished in high humidity environment. Tests conducted at elevated temperatures showed bilinear slopes in the crack growth rate showing distinct Regime I and II, respectively. (iv) Cyclic plastic zone size calculated empirically was found to predict transitional stress intensity factor satisfactorily for a number of aluminum and titanium alloys. (v) Complex interactions of environment, microstructure, and test parameters were not as significant at lower temperatures and stress ratios. However, more secondary cracks and cavities formed at 20 Hz at or above 290°C and had a mixed effect on fatigue crack growth rates as secondary cracks may have reduced the stress intensity at the crack tip of a dominating crack and voids may have accelerated the growth of main dominating crack, needs to be investigated further. (vi) The dislocations were mainly concentrated in α-grains with only a small fraction present in β-grains. Within the α-grains, $<a>$ type dislocations formed with Burgers vectors $(aI2)_g = 1010$ near $(1010)$ (Fig. 13 (e-f)). Some twinning together with other features appeared to be $<a>$ type dislocations in planner arrays with $g = 0002$ near $(1210)$ was found.

References