FATIGUE CRACK INITIATION AND PROPAGATION IN Ti-6Al AND TI-6Al-4V

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ABSTRACT

Studies of sustained-load cracking in titanium alloys show that highly accelerated crack growth occurs at sub-ambient temperatures in alloys containing 100-300 ppm hydrogen. Such accelerated crack growth is dependent on alloy composition and microstructure. This investigation addresses the combined effect of hold time (dwell at maximum load) and temperature on the fatigue crack initiation and propagation behavior of Ti-6Al and Ti-6Al-4V containing interstitial hydrogen of 100 to 300 ppm by weight.

Increases in crack growth rate up to a factor of twenty were observed for hold-time testing, the magnitude of the increase being dependent on microstructure, hydrogen content and test temperature. Such increases were often associated with a fracture mode transition, including: (1) striations to a cleavage-like fracture of primary alpha particles in the equiaxed alpha microstructures, and (2) transition from a tortuous to a more planar fracture path in the beta quenched microstructure.

The effect of hydrogen content and hold-time on fatigue crack initiation was studied in recrystallization annealed Ti-6Al-4V. The cyclic response of this equiaxed microstructure of low initial dislocation content was found to be unaffected by hydrogen contents of 100 to 300 ppm at low stresses amplitudes. This result suggests that over the range of variables studied, propagation is affected significantly by hold time though initiation is not.

KEYWORDS

Fatigue; fatigue crack propagation; titanium alloys; environmental effects, hold-time effects

INTRODUCTION

Titanium alloys are used in structural applications because of their good fatigue resistance combined with their excellent corrosion resistance and high fracture toughness. It has been shown that the comparatively good fatigue resistance of titanium alloys can, under certain circumstances, be modified considerably. For example, it is now the consensus that microstructure can have considerable influence on fatigue resistance of titanium alloys; in particular, beta processed or beta annealed microstructures exhibit a lower fatigue crack growth rate than comparable recrystallization annealed or solution-treated and aged microstructures (Chesnutt, Thompson and Williams, 1980; Paton and co-workers, 1975; Yoder, Cooley and Crooker, 1979).

The effects of a hold time, however, are generally detrimental in that fatigue life is decreased (Harrod and Manjoine, 1976; Eylon and Hall, 1977) or fatigue crack growth rate is increased (Eylon, Rosenblum and Fujishiro, 1980). In most of the systems investigated to date, the increase in crack growth rate or decrease in fatigue life is associated with an environmental interaction (where the environment degrades material properties) or, alternatively, is ascribed to a superposition of a creep phenomenon in addition to the recognized fatigue phenomenon normally addressed in cyclic loading. A further complication of the problem is that significant decreases in fatigue life have been observed in cases where neither environmental effects nor creep effects of any significance would be expected to be present (Eylon and Hall, 1977).

The dwell debit -- as it has been called -- has in some cases been ascribed to the presence of specific alloying additions, such as hydrogen, or specific alloy microstructures or defects in these microstructures. In the case of defects in the microstructures, some evidence exists for linking the dwell debit to internal initiation of fatigue cracks, as opposed to the more common surface initiation (Eylon and Hall, 1977).

In this investigation, an attempt was made to link the interacting effects of microstructure and alloy composition by studying a simple, single-phase Ti-6Al alloy and comparing its behavior to the more complex two-phase Ti-6Al-4V alloy in two different microstructural conditions. These two alloys have been investigated under various loading conditions and with initial interstitial hydrogen concentrations from near zero to 300 ppm by weight.

EXPERIMENTAL PROCEDURES

The compositions of the alloys studied are given in Table 1.

TABLE 1 Chemical Composition

	Material Composition, wt %						
	Al	V	Fe	C	N	O	H
Ti-6Al	6.2	-	0.01	0.050	0.007	0.172	0.005
Ti-6Al-4V	6.2	4.1	0.22	0.010	0.012	0.122	0.003

The Ti-6Al was produced as 14 mm thick plate and had a weak (2 times random) basal transverse texture. The Ti-6Al-4V was produced as 50 mm thick uniformly and weakly textured pancake forgings. Compact tension specimens 12.7 mm thick were machined in both the TL and LT orientations for the Ti-6Al and the CR orientation for Ti-6Al-4V. Ti-6Al was tested in a recrystallized condition with a 100 μm grain size. The Ti-6Al-4V was tested in a recrystallization annealed (RA) and a beta-quenched (BQ) condition, the RA condition having a 10-20 μm equiaxed α grain size. Details of the processing of the Ti-6Al-4V are given elsewhere (Chesnutt, Thompson and Williams, 1980). The microstructures tested are shown in

Fig. 1. Test specimens were charged in hydrogen in a Sieverts apparatus at 900°C for Ti-6Al and 700°C for Ti-6Al-4V. Fatigue crack propagation tests were conducted in an electro-hydraulic closed loop test machine at a frequency of 10 and 20 Hz. A block of cycles with a 5 min hold at K_{max}, using the same loading and unloading rate as for the "no-hold" cycles, was imposed periodically. Room temperature testing was performed in laboratory air; sub-ambient testing was done in a refrigerated box with temperature controlled to ±5°C. Crack length measurements were optically made by using a cathetometer. Fatigue crack initiation was studied in smooth-bar, axial fatigue specimens with a 5.1 mm gage diameter tested at 30 and 120 Hz in laboratory air.

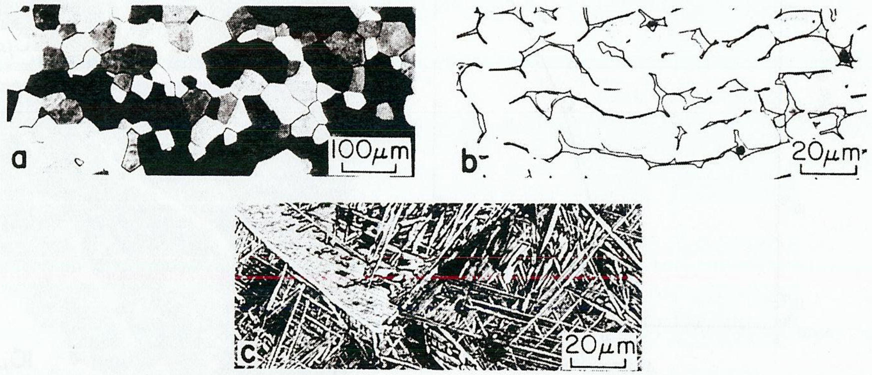

Fig. 1. Light micrograph of conditions studied, (a) Ti-6Al, (b) RA, Ti-6Al-4V, (c) BQ, Ti-6Al-4V.

RESULTS

The fatigue crack propagation results for Ti-6Al in the TL orientation are shown in Fig. 2. Significant increases in crack growth rate (da/dN) were observed at both -40° and -70°C, with the largest effect occurring during high K_{max} loading. Note that the specimens used were the increasing ΔK, compact tension type; for long crack lengths, ΔK and K_{max} may be increasing quite rapidly, which would result in non-conservative average ΔK and K_{max} values. At intermediate and high, no-hold growth rates, for which striation formation is the major mode of crack propagation, the increase in crack growth rate with the hold at K_{max} was accompanied by a distinct fracture mode transition from striations to transgranular cleavage, as shown in Fig. 3. No change in crack growth rate with hold was observed for specimens tested in the TL orientation, a result consistent with sustained-load cracking SLC results (Pardee and Paton, 1980) for this material; average crack growth rates (da/dt) calculated for each hold-time increment were also equivalent to the SLC data. Similar behavior was observed for RA, Ti-6Al-4V containing approximately 100 ppm hydrogen as shown in Fig. 4. A hold at K_{max} produced large increases in crack growth rate at -70°C, in contrast to an absence of a hold-time effect at 20°C. Increases in crack growth were less than a factor of two in both Ti-6Al and RA Ti-6Al-4V for hold time tests at 20°C. A similar fracture mode transition to that in Ti-6Al was observed in the RA Ti-6Al-4V, which consists of a large volume fraction (approximately 90%) of equiaxed, recrystallized primary α. The results for RA Ti-6Al-4V and Ti-6Al at approximately 100 ppm hydrogen are shown in Fig. 5; hold time effects appear larger in Ti-6Al-4V. When the hydrogen content is increased to 300 ppm, increases in crack growth rate ranging from a factor of three to thirty occur at both 20°C and -70°C.

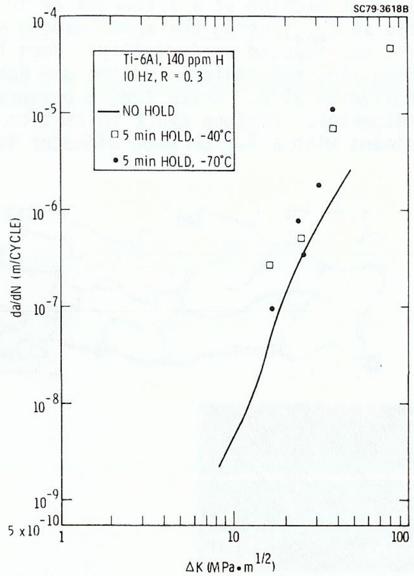

Fig. 2. FCP results for Ti-6Al containing 140 ppm hydrogen tested at -40 and -70°C.

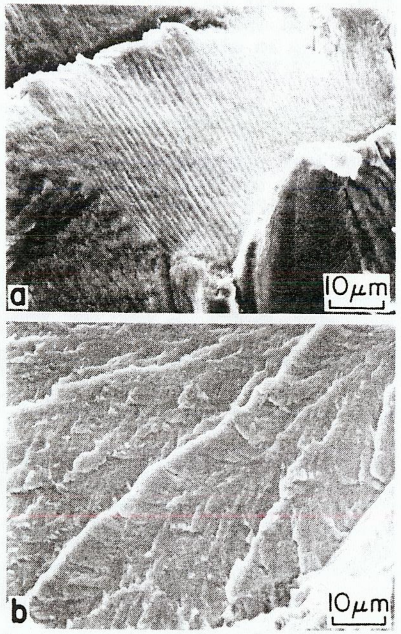

Fig. 3. Scanning electron micrograph of Ti-6Al (140 ppm hydrogen) tested at -70°C, (a) no hold, (b) 5 min hold at K_{max}.

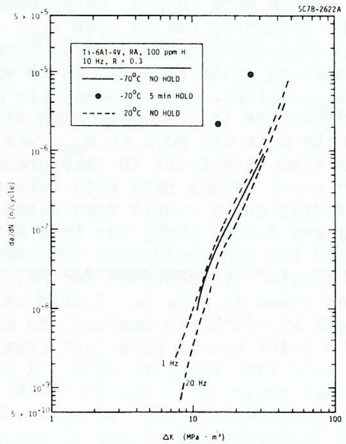

Fig. 4. FCP results for RA Ti-6Al-4V Ti-6Al (140 ppm hydrogen) tested at -70°C, (a) no hold, (b) 5 min.

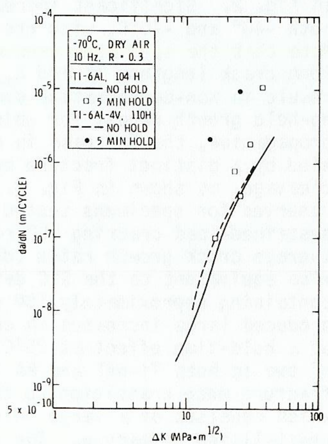

Fig. 5. Comparison of FCP results for Ti-6Al and RA Ti-6Al-4V containing 110-140 ppm hydrogen tested at -70°C.

Fatigue crack propagation data for BQ Ti-6Al-4V is shown in Fig. 6. Modest increases in crack growth rate occurred at both 20 and -70°C; these were accompanied at both temperatures by a fracture mode transition, as shown in Fig. 7. The crack path in the hold-time case became more transgranular than in the no-hold case.

Fatigue crack initiation studies on RA Ti-6Al have shown that this microstructure exhibits significant tolerance to increased hydrogen content and hold-time in the range of hydrogen content and hold-times studied in this program. Baseline fatigue life (S/N) data are shown in Fig. 8, with data for two additional hydrogen levels, 100 and 300 ppm. No effect of frequency in the range 30 to 120 Hz was observed. For longer life, lower stress amplitude tests, increased hydrogen content had no effect on fatigue life or equivalently on crack initiation. Initiation sites were at the surface of the gage section for all three hydrogen

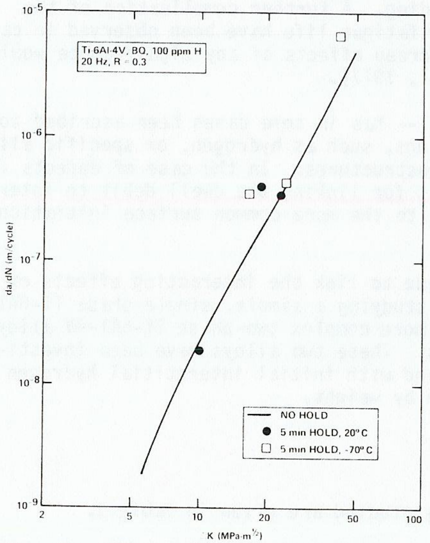

Fig. 6. FCP results for BQ, Ti-6Al-4V containing 100 ppm hydrogen tested at 20 and -70°C.

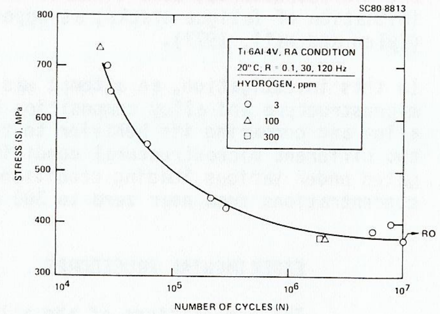

Fig. 8. S-N curve for smooth bar, Ti-6Al-4V in the recrystallization annealed (RA) condition.

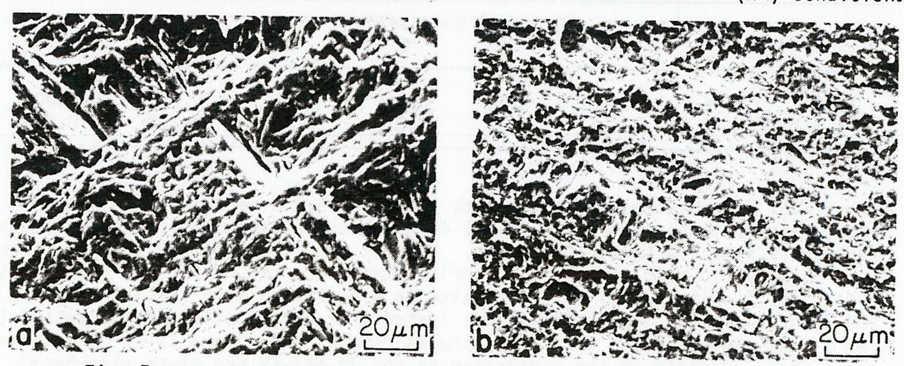

Fig. 7 Scanning electron micrograph of BQ, Ti-6Al-4V tested at -70°C, (a) no hold, (b) hold.

levels and the fracture surfaces were characterized by transgranular, cyclic cleavage of the alpha phase. The only notable difference in the three cases was an increased tendency to secondary cracking perpendicular to the fracture in the material containing 300 ppm hydrogen. At higher stress amplitudes, hydrogen contents of 100 ppm had no effect on fatigue life. Tests are currently in process to evaluate the possibility of combined hold-time, hydrogen effects. Two specimens, one containing 3 ppm hydrogen and one containing 300 ppm hydrogen, are being tested under dwell loading with a one minute hold at maximum load. The specimen containing 3 ppm hydrogen has run 10^4 cycles at 670 MPa, and the specimen containing 300 ppm hydrogen has run more than 8,000 cycles at the same stress level. The dwell debit in the former case would be less than a factor of 2.7 and in the latter case less than a factor of 3.3, in sharp contrast to the dwell debit found by Eylon and Hall (1977) for IMI 685.

DISCUSSION

These results show that for the Ti-6Al alloy investigated, increases in fatigue crack growth rate occurred with a 5-minute hold at either -40 or -70°C. The conditions required for an increased growth rate are: (1) more than the base level hydrogen content, i.e., 100-300 ppm hydrogen (2) a temperature below 20°C, and (3) a significant tensile hold time in the loading cycle. Without these three elements present at one time, increases in fatigue crack growth rate were not observed. For example, a sample containing 100 ppm hydrogen tested with a hold time at 20°C did not exhibit any acceleration in crack growth rate. An additional important observation was the fracture mode transition that occurred during hold-time to cycling, as shown in Fig. 3. The transition from ductile striation formation in standard fatigue cycling to a quasi-cleavage, brittle fracture with hold time appears to be consistent with a model suggesting hydride formation at the crack tip during the hold cycle. Calculations made recently by Pardee and Paton (1980) suggest that a hold time of the order of one minute or more is sufficient (because of the high triaxiality) to drive a significant quantity of hydrogen to the crack tip, increasing the local hydrogen concentration. The high tensile stress and local deformation occurring at the crack tip aid the nucleation of hydrides; thus, conditions required for hydride nucleation may exist at the crack tip even though the average hydrogen concentration is insufficient for hydride formation. This process is greatly favored at temperatures somewhat below room temperature in the calculations by Pardee and Paton (1980); temperatures on the order of -70°C were found to provide conditions for maximum crack growth rate in sustained load cracking experiments. Crack growth rates (da/dt) calculated from da/dN data are comparable to those calculated by Pardee and Paton (1980), suggesting that sustained load cracking during the tensile hold is the dominant mechanism.

It is suggested that the hold time on the tensile part of the fatigue cycle provides the right conditions to diffuse hydrogen to the crack tip, nucleate hydrides and cause quasi-cleavage type fracture as a result of the hydride formation. This process is aided by high hydrogen contents, long tensile hold times and temperatures below room temperature, as found experimentally in the present results.

In the recrystallization annealed Ti-6Al-4V, similar behavior to the Ti-6Al alloy was observed with 100 ppm hydrogen. However, the effect was larger than for the completely alpha-phase Ti-6Al alloy. Greater acceleration in an alloy such as Ti-6Al-4V containing some beta phase is assumed to be evidence for the contribution of beta phase to an increased effective diffusion rate of hydrogen, permitting hydrogen to reach the crack tip more rapidly than in an all alpha alloy. With 300 ppm hydrogen in the Ti-6Al-4V alloy in the RA condition, the effect was large even at 20°C, with no significant effect of sub-ambient temperature on further increases in crack growth rate. This is consistent with the notion that the beta phase accelerates hydrogen diffusion, and is also consistent with the data of Boyer and Spurr (1978) which showed that in da/dt tests below room temperature acceleration was at a maximum at about 0°C and then decreased below that temperature.

The beta-quenched Ti-6Al-4V alloy is different from the recrystallization annealed RA condition in that there is no continuous primary alpha path, as shown in Fig. 1. Hold time effects were equivalent at 20° and -70°C, and an acceleration was found in both cases where both hydrogen and hold time were present. For all cases in this present series of tests, the effect depended significantly on the local crystallographic orientation, and the texture had significant influence on whether or not a hold time effect was observed. In tests of a moderately textured plate of Ti-6Al, crack propagation was normal to the principal orientation of the basal planes; no hold-time effect was observed under any conditions in this case, whereas when the crack was propagating along the basal planes, a significant hold time effect was observed.

The hydrogen effect and hold-time behavior in smooth-bar fatigue specimens of recrystallization annealed (RA) Ti-6Al-4V was in sharp contrast to the crack propagation behavior of the precracked fatigue crack propagation specimens. It is well documented by Evans and Gostelow (1979), Eylon and Hall (1977), and Postans and Jeal (1977) that a hold-time at maximum load (dwell) results in an increase in fatigue crack propagation rate and a decrease in fatigue life in IMI 685 when compared to simple cyclic loading. Evans and Gostelow (1979) propose a model for "dwell sensitivity" and attendent faceted fracture which is dependent on a combined creep and hydrogen locking mechanism. Lankford, Davidson and Leverant (1980) have studied the low cycle fatigue (LCF) "dwell-debit" in IMI 685 at -70°C, 20°C and 200°C and have demonstrated that both creep deformation and the ability to precipitate hydrides are required for brittle fracture in the absence of pre-existing flaws. The microstructure in all the forging studies in IMI 685 consisted of coarse grained, Widmanstätten α with aligned α platelets providing large planar slip distances. The "dwell sensitivity" and "dwell-debit" models both invoke this type of microstructure.

Fatigue crack initiation results at 20°C for RA Ti-6Al-4V have shown that this microstructure is insensitive hydrogen contents up to 300 ppm. In addition, a one-minute hold did not produce any significant dwell-debit for a specimen containing 300 ppm hydrogen that was tested at a high stress amplitude (670 MPa). The specimen has run in excess of 8000 cycles (8400 minutes) and would be expected to have a significant creep component. Furthermore, even though the recrystallization annealed microstructures contain nearly continuous primary α (Fig. 1(b)), the fact that the material is weakly and uniformily textured and, therefore does not contain large unrestricted planar slip distances, results in reduced sensitivity to a dwell-debit in the absence of a pre-existing flaw or crack.

Thus, the recrystallization annealed microstructure of Ti-6Al-4V, that is specified for fracture critical applications in modern military aircraft, also exhibits considerable tolerance at 20°C to hydrogen content and dwell loading when fatigue crack initiation is a dominant mechanism.

CONCLUSIONS

1. Accelerations in fatigue crack growth rate with a 5-minute tensile hold were observed in Ti-6Al containing 100 to 300 ppm hydrogen. Growth rate was increased up to a factor of ten at -40° and -70°C, but was less than a factor of two at room temperature.

2. Ti-6Al-4V in the RA condition also exhibited an acceleration in fatigue crack growth rate with 100 to 300 ppm hydrogen; the effect was most pronounced at -70°C with accelerations of up to 30 times being observed.
3. Ti-6Al-4V in the beta-quenched condition, in which there is no continuous primary alpha path, showed only a minimal fatigue crack acceleration under conditions of high hydrogen, hold time, and sub-ambient temperature. Decreasing the temperature from 20° to -70°C had little or no effect on accelerating crack growth rate.
4. Fatigue crack initiation in smooth-bar specimens of Ti-6Al-4V in the RA condition was unaffected at 20°C by hydrogen contents of 100 to 300 ppm hydrogen and hold-times of one minute at 20°C.

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