Dwell Fatigue Behavior of Two-Phase Ti-6Al-4V Alloy at Moderate Temperature

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Fatigue life of titanium and titanium alloys is frequently reduced if the load holds at the maximum stress are introduced, which is termed dwell debit. Many factors affect dwell fatigue of titanium alloys, like stress state, level and ratio, alloy chemistry, microstructure and microtexture, holding time and temperature. Dwell sensitivity of titanium alloys is usually attributed to the phenomenon of load shedding, which is a time-dependent redistribution of stress from so-called ‘weak’ to ‘strong’ grains possessing different crystallographic orientations. Stress concentration leads to crack initiation and formation of quasi-cleavage facets in ‘strong’ grains. Another factor contributing to dwell sensitivity of titanium alloys is time dependent strain accumulation, which is observed even at room temperature. In the paper, the effect of load holds and volume fraction of primary α phase on the fatigue behavior of Ti-6Al-4V alloy at the temperature of 150°C was investigated. Two variants of bimodal microstructure of the alloy were obtained by means of the heat treatment. Stress controlled fatigue tests with and without hold time at maximum load and constant load creep test were carried out.

INTRODUCTION

Near-α and two-phase α + β titanium alloys are routinely used in aerospace engineering for critical rotating elements of modern gas turbines because of their advantageous combination of strength and fatigue properties.1 However, under cyclic load conditions they show dwell sensitivity manifested in the reduction of fatigue life if load holds at high stresses are introduced.2-5

Elastic and plastic anisotropy of the hcp structure of α phase and different crystallographic orientation of neighboring α grains leads to load shedding, which involves a time-dependent redistribution of stress from ‘weak’ grains, in which dislocation slip in basal and prismatic systems dominates, to ‘strong’ grains, well-oriented for pyramidal slip. Interaction between ‘strong’ and ‘weak’ grains results in formation of quasi-cleavage facets on the basal planes of ‘strong’ grains and in turn in crack initiation.3,6

In CP titanium and titanium alloys large time-dependent strains develop even for stresses significantly lower than yield strength and at temperature < 0.25Tm, where Tm is the melting point of the alloy in Kelvin.7-9 Such behavior is often termed ‘cold creep.’ That phenomenon is disadvantageous under dwell fatigue conditions because it can contribute to increase in dwell sensitivity and lead to premature failure.10-12

It was reported that dwell sensitivity in titanium alloys usually disappears at temperatures > 200°C because the strength mismatch between differently oriented grains is reduced and ‘strong’ grains boundaries no longer cause dislocation pile-up.9,13 However, alloy chemistry and microstructure may have significant impact on the temperature range in which dwell sensitivity is observed.

Although two-phase titanium alloys are usually less sensitive to dwell fatigue than near-α alloys, some microstructural features can result in
moderate dwell debit.\textsuperscript{14–17} One of the microstructural factors having strong influence on the dwell sensitivity of two-phase titanium alloys is the volume fraction of primary $\alpha$ phase.\textsuperscript{18} Due to importance of dwell sensitivity of titanium alloys in the context of their strength properties, the following investigations were carried out. They were focused on the determination of the influence of the load hold time and volume fraction of primary $\alpha$ phase on dwell fatigue behavior of Ti-6Al-4V alloy with bi-modal microstructure at the temperature of 150°C.

**MATERIALS AND THE METHODS**

The material studied was two-phase titanium alloy Ti-6Al-4V having the following chemical composition (wt%): Al—6.30, V—4.23, Fe—0.079, C—0.011, O—0.154, N—0.013, H—0.0009, Ti—balance. This is martensitic $\alpha + \beta$ alloy, which is characterized by high strength at room and elevated temperatures and high fracture toughness.

The alloy was delivered in the form of rolled bars, 12 mm in diameter in mill-annealed condition. To determine the influence of the volume fraction of primary $\alpha$ phase on dwell fatigue behavior of the alloy, two schemes of heat treatment process were applied.

Annealing at the temperature of 850°C and 910°C, which is in $\alpha + \beta \rightarrow \beta$ phase transformation range, followed by air cooling was applied to obtain bi-modal microstructure with varying volume fractions of primary $\alpha$ phase. These two variants of the alloy are denoted in the paper with the following abbreviations: MAT850 and MAT910, respectively. Microstructure of the alloy was examined using Nikon Epiphot 3 light microscope equipped with a DS-1 camera. In both cases it consisted of equiaxed grains of primary $\alpha$ phase separated by transformed $\beta$ phase (Fig. 1). Average grain diameter ($d_{\text{sp}}$) and volume fraction ($V_{\text{sp}}$) of primary $\alpha$ phase ($x_p$) were determined using Aphelion v.2.3. image analysis software (Table I). Measurements were made on five randomly selected microscope images for each material variant. Average grain size was determined using the grain intercept lengths method. In each case $> 500$ grain intercept lengths (chords) were measured. Volume fraction of primary $\alpha$ phase was determined by automatic image analysis.

Static tensile tests were carried out at the temperature of 150°C on Instron 5982 testing machine at the constant strain rate of 0.005 min$^{-1}$. The strain was measured during the test by extensometer with 1-μm resolution. The values of 0.2% yield stress are the average of three tensile experiments for each alloy variant. Constant load creep tests were carried out at the temperature of 150°C in air on Walter + Bai AG LFMZ-30kN electromechanical creep machine. Two specimens of each alloy variant were tested. Creep tests were discontinued after approximately 50 h to determine material behavior during primary creep stage.

Stress controlled fatigue tests were performed at the temperature of 150°C on Instron 8801 servohydraulic testing machine in tensile mode at the load ratio $R = 0.05$. Loading and unloading ramps were realized at the constant time of 2.5 s in all tests. The maximum stress was selected to be equal to 0.2% offset yield stress at the test temperature (Table I). Fatigue tests were carried out without holding at peak stress and with dwell times $t_H$ at the maximum tensile stress of 10 s, 20 s and 60 s.

The level of maximum stress applied in fatigue and dwell fatigue tests is justified because the effect of load holds on fatigue life of near-$\alpha$ and $\alpha + \beta$ titanium alloys occurs at relatively high stresses, usually $> 0.6$ UTS (ultimate tensile strength).\textsuperscript{2} Similar test conditions concerning maximum stress applied in experiments were adopted in several research programs to assess the effect of typical load spectrum of the elements of compressor section of an aero-engine leading to the dwell fatigue life debit. Chandravanshi et al. applied maximum stress corresponding to 0.95 of yield strength in load-controlled fatigue and dwell fatigue tests at the

![Fig. 1. Microstructure of two-phase Ti-6Al-4V alloy after annealing at: (a) 850°C, (b) 910°C.](image-url)
stress ratio of $R = 0.1$ at room temperature.\textsuperscript{19} Maximum stress level of 0.95 of the yield strength was adopted in a number of other works, among others by Neal,\textsuperscript{20} Pilchak et al.\textsuperscript{21} and Qui et al.\textsuperscript{22} Mori et al. investigated Ti-6Al-4V alloy in stress-controlled fatigue and dwell fatigue tests at the maximum stress ranging from 0.93 to 1.0 of the 0.2\% proof stress.\textsuperscript{23} Lefranc et al. carried out fatigue tests on smooth specimens in the range of 0.92–1.05 of yield strength with load hold periods at maximum stress of 80 s.\textsuperscript{10} The load hold periods of 120 s are used frequently in dwell fatigue testing. However, the effect of other hold times was also investigated for example by Bache et al. (load holds of 15 s, 120 s and 300 s),\textsuperscript{2} Everaerts et al. (load holds of 30 and 120 s)\textsuperscript{17} and Lefranc et al. (load holds of 80 s).\textsuperscript{10}

In all tests smooth cylindrical specimens were used, with a gauge diameter and gauge length of 6 and 24 mm, respectively. All fatigue specimens were low stress ground and polished longitudinally to a surface roughness of $R_a = 0.20 \ \mu m$. For each set of conditions two specimens were tested. Fracture surfaces of selected specimens were examined using Hitachi S3400 scanning electron microscope.

**RESULTS AND DISCUSSION**

Results of the fatigue tests show that introduction of hold periods at the maximum stress leads to decrease in the number of cycles to failure (Fig. 2). However, fatigue life was reduced only by the factor around 2 for the longest load hold (60 s). The trend can be observed that longer hold time causes increase in dwell debit (Fig. 2b). The most pronounced change in this effect was observed for hold time between 10 s and 20 s. Further increase of dwell time caused only moderate decrease of the average number of cycles to failure. Similar trend was observed also in the case of other alloys, e.g., IMI834.\textsuperscript{2} Some deviations from this trend may be due to natural scatter in the fatigue data. The alloy with lower volume fraction of primary $\alpha$ phase shows slightly higher fatigue life, in both the cyclic and dwell load conditions. These differences are bigger in the case of longer hold times (20 s and 60 s) (Fig. 2b). It indicates that the effect of volume fraction of primary $\alpha$ phase on the fatigue properties of the alloy is more significant in the case of longer dwell.

The second effect following the introduction of load holds at the maximum stress was considerable increase in plastic strain accumulated during the test (Fig. 3). It indicates significant influence of time-dependent process of cold creep on deformation behavior of the investigated alloy.

The maximum strain obtained in the case of both investigated variants of Ti-6Al-4V alloy, MAT850 and MAT910 was similar for all hold times applied. The maximum strain was slightly higher for MAT850. The abrupt change of total strain value in the final stage of some tests was linked to the effect of crack opening on the extensometer reading. That part of the fatigue data was not taken into account in the analysis.

During some dwell fatigue tests, after reaching a certain value of strain, decrease in total strain with growing number of load cycles was recorded. This behavior was observed during investigation of the first set of specimens when a standard extensometer, attached to the specimen at one side, was used for strain measurement. The decrease in measured strain values was attributed to the effect of the crack initiated at a random position relative to the extensometer. During testing of the second set of specimens, averaging extensometer was used instead, which effectively reduced the influence of the crack location on strain readings. Therefore, the previously observed decrease in strain was no longer detected (dashed lines in Figs. 3, 4, 5).

To capture the differences in kinetics of plastic strain accumulation, fatigue data were plotted also against test time (Fig. 4). It is evident that the hold time has a decisive influence on the total time to failure. For tests with hold time equal to 10 and 20 s, the total time to failure is almost the same in both cases. However, due to large proportion of loading/unloading time in total test time in these cases such comparison is not fully reliable. That was the reason for calculation of the 'effective' creep time for the dwell fatigue tests as the total time of load holds at the maximum stress level (Fig. 5). It was found that during the tests with shorter hold time the observed strain was increasing faster in terms of the effective creep time than in the case of longer dwell.

### Table I. Ti-6Al-4V alloy variants used in investigations (SD denotes standard deviation)

| Material variant | Annealing temperature, °C | $d_{pp}, \mu m$ Average | $d_{pp}, \mu m$ SD | Volume fraction of $\alpha_p$ phase $V_{\alpha_p}, \%$ Average | $V_{\alpha_p}, \%$ SD | 0.2\% offset YS (150°C), MPa Average | 0.2\% offset YS (150°C), MPa SD |
|------------------|--------------------------|-------------------------|-------------------|-----------------------------|-------------------|--------------------------|--------------------------|
| MAT850           | 850                      | 8.8                     | 1.2               | 67                          | 2.3               | 739                      | 6.2                      |
| MAT910           | 910                      | 8.2                     | 1.6               | 48                          | 1.9               | 752                      | 6.6                      |
Fig. 2. Fatigue life (a) and average dwell debit in terms of load hold time (b) for the two investigated variants of Ti-6Al-4V alloy at 150°C ($N_F$ and $N_D$ denote the number of cycles to failure in fatigue and dwell fatigue tests, respectively).

Fig. 3. The maximum strain as a function of the number of cycles: (a) MAT850, (b) MAT910.

Fig. 4. The maximum strain as a function of the test time: (a) MAT850, (b) MAT910.
hold time. That fact may be attributed to the larger number of load cycles, as immediately after reaching the maximum stress level the observed strain rate was the highest and then decreased during load hold.

Comparison of strain changes during dwell fatigue and creep tests shows that although the strains attained after 50 h in creep test and dwell fatigue test with 60 s hold time were similar, the course of deformation process was different (Fig. 6). Initially, strain increased more rapidly in crept specimens. It can be explained by the fact that during the dwell fatigue test the strain attained at the beginning of hold period was always smaller than the strain at the end of preceding load hold. However, after a certain number of load cycles the rate of creep strain decreased noticeably and the creep strain was surpassed by the strain developed in dwell fatigue test. The character of the curves obtained in dwell fatigue tests suggests that:

- Cyclic unloading/loading sequence accelerates strain accumulation and
- The time required to stabilize strain level decreases as the hold time is reduced.

It indicates that both fatigue and creep processes influence to varying extents the whole process of material deformation and damage.

Significantly longer effective creep time in the case of dwell test with 60 s hold time did not result in equally distinct reduction of the number of cycles to failure. It may be attributed to the fact that after approximately 50 h of the creep test the strain rate decreased almost to zero.
Fatigue striations were observed on the fracture surfaces of specimens tested in both fatigue and dwell fatigue conditions (Fig. 7). The spacing between striations formed in dwell fatigue tests was generally larger than in the case of fatigue test. The biggest striations spacing was determined for the load hold time of 60 s. That observation correlates with the largest plastic deformation occurring during single load cycle.

The quasi-cleavage facets were not observed in the area of crack initiation, which supports the conclusion that for the examined alloy microstructure and in the test conditions applied cold creep deformation could be the primary cause of dwell fatigue life reduction. Another reason for that fact can be related to the macro- and microtexture of the examined alloy. Pilchak et al. characterized macro- and microtexture regions (MTR's) in the rods of various diameters of the alloy Ti-6Al-2Sn-4Zr-2Mo-0.1Si. Although macrotexture was rather weak, microtexture regions were present in all analyzed rods. The MTRs were found to be high aspect ratio ellipsoids with their semi-major axis aligned with the rod’s longitudinal direction. They became narrower and more elongated with diameter of the rods decreasing down to 57 mm. Therefore, it could be expected that the MTR aspect ratio on the fracture surface of the bars tested in longitudinal direction would be very small especially for bars with even smaller diameter. The second consequence of such character of MTRs would be the presence of very few basal planes aligned favorably for formation of facets on the fracture surface.

It can be expected that MTR size and crystallographic texture in the Ti-6Al-4V bars used in reported investigation could be similar to that described by Pilchak et al.24 Such assumption can be justified also by results concerning the texture in Ti-6Al-4V bars presented by Mori et al.23 Also, Everaerts et al. 17 investigated the texture and dwell sensitivity of Ti-6Al-4V alloy in the form of the wire, 1 mm in diameter, and reported the virtual lack of grains oriented with their c-axis parallel to the axis of the wire.

Therefore, the lack of quasi-cleavage facets on the fracture surface of the specimens tested could be justified also in the context of the results described in above-mentioned references. To clarify this issue, EBSD analysis should be performed. However, the authors have had no access to such equipment during realization of the reported research and plan to include this method in the future investigations.

**SUMMARY**

Two-phase titanium alloy Ti-6Al-4V with bimodal microstructure and various volume fractions of primary α phase was examined in constant load creep, low-cycle fatigue and dwell fatigue tests at 150°C. Based on the analysis of the test results, the following findings were formulated:

- Introduction of load holds at the maximum stress resulted in moderate reduction of the fatigue life of the alloy by a factor of around 2 in the worst case. The most pronounced change in the dwell debit was observed for hold time between 10 s and 20 s. Further increase in dwell time led to only moderate reduction of the fatigue life.

- The alloy with lower volume fraction of primary α phase exhibited longer life in both fatigue and dwell fatigue conditions.

- The plastic strain level attained was significantly higher in dwell fatigue tests compared with pure fatigue. Its value was similar for all hold times applied. It was also comparable with the strain developed in the creep test at the constant stress equal to the maximum stress in the fatigue tests. This may indicate the dominant influence of cold creep on dwell fatigue life of the examined alloy.

- The difference in dwell fatigue sensitivity of the two variants of the investigated Ti-6Al-4V alloy...
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with primary α phase volume fraction of 49 and 67% was not very significant but consistent for all load hold times applied and more pronounced in the case of longer dwell times (20 s and 60 s).

- The lack of quasi-cleavage facets on the fracture surfaces of the specimens tested in dwell fatigue conditions and increase of fatigue striations spacing with growing load hold time indicate that in analyzed case the primary cause of dwell fatigue sensitivity of the investigated alloy was cold creep.
- Potential influence of the crystallographic texture in examined alloy on the formation of the quasi-cleavage facets on the fracture surfaces was not clarified, as it would require use of EBSD technique, which is planned to be done in future investigations.

CONFLICT OF INTEREST

The authors declare that they have no conflict of interest.

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