CREEP-FATIGUE INTERACTION IN TITANIUM TI-1100 AT 593°C

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The mechanism responsible for the low frequency crack growth acceleration observed in the near-α, Ti-1100 alloy at 593°C has been investigated. A series of high temperature fatigue and creep crack growth experiments have been carried out on both unaged and aged (containing Ti_3Al and Silicides precipitates) specimens. Results of this work, supported by observations related to fracture mechanisms of this alloy corresponding to low and high frequency loadings, show that the time-dependent crack growth acceleration is primarily due to modifications of the crack tip localized deformation as it is influenced by fatigue-creep interaction effects associated with the low frequency type of loading.

INTRODUCTION

Ti-1100 is a new near alpha titanium alloy that was developed for its high creep resistance. Under low frequency loadings or hold time fatigue conditions, this alloy has shown an increase in the fatigue crack growth rate (Ghonem and Foerch (1)). This increase was found to be insensitive to time-dependent environment related effects, see Foerch et al (2). It was then assumed in the work of Foerch (3) that the dwell fatigue crack growth enhancement is due to creep or creep-fatigue interaction effects.

The objective of this study was to experimentally investigate the nature of the low frequency fatigue crack growth acceleration in Ti-1100 as well as the involving mechanisms at 593°C. The next section of the paper is a description of the materials and experimental techniques utilized in this study. This is

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followed by results obtained from a series of creep-fatigue crack growth experiments. A crack tip deformation mechanism is then proposed to account for the increase in the crack growth rate seen under creep-fatigue conditions employed in this study. This mechanism is verified by examining the crack growth response of a microstructurally modified Ti-1100 material. Finally, results of the study are discussed on the basis of fracture mechanisms expected in the test materials.

EXPERIMENTAL PROCEDURES

The nominal chemical composition of the Ti-1100 used in this study is, in wt%, Ti-6Al-2.8Sn-4Zr-0.4Mo-0.45Si with 0.07 O and 0.03 Fe maximum. A more detailed description of the material can be found in (1,3). Test specimens of a compact tension geometry were fatigue tested using a servohydraulic testing machine with a computer controlled waveform generation and data acquisition system. The crack length was monitored using a DC potential drop technique (3). The loading cycle used in the testing program was of a 0.05 Hz with hold periods of 0, 150 and 300 seconds imposed at the maximum load level; these cycles are described hereafter as 10-10, 10-150-10 and 10-300-10, respectively. The load line crack opening displacement, COD_{LL}, was measured using a clip gage with a high temperature scissors extender while the near field crack opening displacement, COD_{IDG}, was measured using a laser interferometric displacement gage (IDG) described in details in the work of Rosenberger (4). Experiments were conducted at 593°C using a resistance furnace with the temperature controlled to ± 1°C during the test. Furthermore, in order to promote a different mechanism of creep-fatigue interaction, several tests were preformed on material that has been aged for 500 hours at 593°C prior to testing. Aging promotes the precipitation of Ti_3Al and silicide particles which was found, as will be detailed in this study and also discussed in the work of Lütjering and Weissmann (5), to cause a change in the fatigue deformation behavior of the alloy under investigation.

RESULTS

Crack growth rate versus the applied stress intensity range, ΔK is shown in Fig. 1 for various cyclic loading frequencies. Two distinct regions of crack growth behavior are observed. Cyclic frequencies between 10 Hz and 0.05 Hz result in a lower growth rate as compared to test frequencies equal or lower than 0.01 Hz. It could be noted that the 0.005 Hz and 10-300-10 cycle produce a crack growth rate approximately 4 times higher than that of the high frequency fatigue cycling. Similar results have been found for specimens tested in high vacuum environment (2). This latter result indicates that the crack growth acceleration is primarily due to creep induced effects rather than being a result of oxidation enhancement of the crack growth behavior. In addition, it was observed that in Ti-1100 the fracture mode changes from transgranular at high frequencies to an intergranular mode involving prior beta grain boundary cracking at lower frequencies. Furthermore, pure creep crack growth experiments conducted by Rosenberger (6) on as received specimens of Ti-1100 showed that only negligible crack growth was detected under sustained load conditions, see also the work of Ashbaugh et al (7). On the basis of these results, the increase in creep-fatigue crack growth rate observed in Fig. 1 is viewed to be a result of a process involving a synergistic interaction of both creep and fatigue components of the loading cycle. This interaction should result in a deformation mechanism that promotes the occurrence of intergranular fracture without the aid of environmental embrittlement or extensive grain boundary sliding as both phenomena play little or no role in controlling the fracture processes of Ti-1100 at elevated temperature. Recognizing that in near-α Ti alloys creep deformation is generally governed by dislocation creep (8), a deformation mechanism is proposed to account for the role of creep in the increase of the time-dependent fatigue crack growth of Ti-1100. This mechanism is based on the assumption that under high frequency loading, the alpha colony boundaries could act as effective barrier to active slip planes thus resulting in an increase in the local slip density due to the rapid hardening of these planes (4). Fracture in this case occurs along crystallographic planes within the crack tip grains thus resulting in a quasi-cleavage fracture mode as observed in fatigue crack growth experiments in the frequency range from 10 Hz to 0.05 Hz (1). With a decrease in the loading cycle frequency or with the application of a hold time period at the maximum load of the fatigue cycle, the movement of dislocations across the alpha colony boundaries is made easier through time-dependent climb and/or cross slip aided by the relatively close crystallographic alignment of alpha colonies within a single prior beta grain (3). In this case, the slip planes could impinge upon prior beta grain boundaries and result in dislocation pileups which would give rise to an intergranular fracture mode. Partial support for the proposed deformation mechanism is gained from the fact that slip steps of 15-20 μm have been observed on grain boundary facets obtained from tests involving 10-300-10 cyclic load conditions as shown in Fig. 2. Validity of this proposed deformation mechanism could be tested by examining the effects of modifying the degree of deformation homogeneity in the local crack tip region during low frequency cracking. One of the direct approaches to achieve such modification is through the thermal aging of test specimens prior to their fatigue loading. This is based on results of Madsen (8) which showed that aging Ti-1100 at 593°C produces two types of precipitate, Ti_3Al, small coherent particles and $(TiZr)_6Si_3$ (silicide), a relatively larger, ≈ 150 nm in length, incoherent precipitate. The later is generally formed along alpha colony and prior beta grain boundaries.

It has been found that these two types of precipitates do not modify either the macroscopic high frequency fatigue crack growth (8) or the creep properties of the alloy (6). This consistency in material properties between aged and unaged Ti-1100 could be utilized to identify the influence of the precipitates alone on the crack tip deformation during low frequency loadings. Here, it is assumed that during the loading portion of the fatigue cycle, the silicide precipitate, in particular, could act as sites for dislocation pileups within the prior beta grains. An expected consequence of this is the increase of the slip homogeneity within the crack tip grains leading to a decrease in the stress localization at their prior beta grain boundaries. Therefore, under loading conditions producing intergranular fracture in unaged material, the presence of silicides in the aged material should modify the crack tip deformation behavior and result in a quasi-cleavage fracture mode.

The first step in this investigation is to examine the continuum evolution of the time-dependent deformation as well as the crack tip driving force in both aged and unaged materials during low frequency loadings. If these two parameters are found to be similar in both types of materials, variation in their crack growth rates, if detected, should then be interpreted in terms of differences in their localized deformation processes. Measurements of the near field crack opening displacement on aged and unaged materials during the hold time of a loading cycle were thus, carried out, using an IDG system, in order to identify, as mentioned above, the effect of aging on the crack tip continuum deformation response. A typical COD profile obtained during the hold time, δ_{HT} for the 10-300-10 cycle in both the aged and unaged materials is shown in Fig. 3. It can be observed that the COD magnitude and its rate during the hold time are identical for both materials. Hence, it is acceptable to consider that the creep zone size developed in the crack tip regions in both materials at a particular crack length are similar. It should also be noted in Fig. 3, that for both materials the overall time dependent opening of the crack tip is small - accounting for less than 1.5 percent of the total COD measured during the cycle.

The second method employed here to examine the continuum deformation at the crack tip of the aged and unaged materials is based on an evaluation of the cyclic J integral, ΔJ, as measured at different crack positions. This will show any change in the applied crack driving force and the relative path independence in the J evaluation. For a compact tension fatigue specimen, ΔJ can be estimated, following the work of Dowling and Begley (9), as:

$$\Delta J = \frac{2A}{Bb} \tag{1}$$

where A is the area under the load versus load line displacement curve, B is the specimen thickness, and b is the uncracked ligament, w-a. Creep deformation on loading or during the hold period will tend to increase the COD and hence the A term in equation (1). A similar expression can be considered for the near field displacements, COD_{IDG}, which tend to be more sensitive to the creep deformation of the crack tip material:

$$\Delta \bar{J} = \frac{2\bar{A}}{B\,b} \tag{2}$$

where $\Delta\bar{J}$ is related to ΔJ except that $\bar{A}$ is the area under the load versus crack opening displacement curve measured at a location other than the load line. This is similar to (1) above except that the COD measurement is referenced to a different location and does not accurately calculate the work input as does ΔJ. $\Delta\bar{J}$ can be converted to ΔJ by applying a geometrical correction to the displacement measurements. Following this approach, $\Delta\bar{J}$ calculated on the basis of COD_{IDG} values measured at location x_{IDG}/w along the crack length, can be compared with load line ΔJ via the expression (4,10):

$$\Delta J = \frac{2\bar{A}}{B\,b}\left(\frac{x_o/w}{x_o/w - x_{IDG}/w}\right) \tag{3}$$

Here x_o/w is the rotation center of a virtual crack. Using (1) and (3) above, the values of ΔJ were compared for the three cyclic conditions, 10-10 and 10-300-10 for the aged and unaged material. Fig. 4 shows results for $\Delta J_{COD\,LL}$ being plotted as a smooth regression through the data points. It is observed that the results are identical for the 10-300-10 unaged and aged condition as well as the 10-10 aged condition. The ΔJ_{IDG} results as corrected using equation (3) follow the line for $\Delta J_{COD\,LL}$ though they exhibit more scatter than the load line measurements which is expected considering the higher sensitivity of the measurement. These results show that the crack driving force in pure fatigue and the creep-fatigue cycles is the same for both the aged and unaged materials.

Turning attention to the crack growth rate versus ΔK for several test conditions, results are shown in Fig. 5 for 0.05 Hz (aged and unaged), 10-300-10 (aged and unaged), 10-150-10 (aged) and 100-100 (unaged). It is apparent that the aged material has produced a crack growth rate during the 10-300-10 cycle that is no different than that of the pure fatigue cycle, (10-10) and much lower than that of the unaged material during the same low frequency cycling (10-300-10). That is, aging has suppressed the creep-fatigue interaction. Furthermore, a scanning electron microscope examination of the fracture surfaces indicates that the fracture mode in the aged material remains transgranular even under the 10-300-10 hold time cyclic condition. The unaged material indicated some intergranular features at $\Delta K = 23.6$ MPa$\sqrt{m}$, and completely intergranular fracture at values of ΔK greater than 26 MPa$\sqrt{m}$ for the 10-300-10 cycling. Fracture in the aged material under the same 10-300-10 conditions showed less than 20% intergranular fracture features at $\Delta K = 28$ MPa$\sqrt{m}$ and under 10-150-10 conditions showed that intergranular fracture was delayed to $\Delta K = 30$ MPa$\sqrt{m}$ which also corresponds to the start of the stage III crack growth region.

DISCUSSION

The main outcome of this study is that the effect of creep-fatigue interaction during the low frequency fatigue crack growth in Ti-1100 is due to a time dependent modification of the localized crack tip deformation behavior. This conclusion has been reached by comparing the fracture processes in unaged and aged Ti-1100 subjected to identical hold time fatigue cycles. While both materials have identical macroscopic creep and fatigue behavior, the occurrence of intergranular fracture in the unaged material suggests that creep deformation, during the low frequency loading, influences the slip homogeneity and stresses at prior beta grain boundaries. This influence will be discussed here by emphasizing the basic difference in the fracture process occurring in aged and unaged Ti-1100.

In unaged Ti-1100, the high frequency fatigue process is controlled by the quasi-cleavage or striation fracture mechanisms which generally result in the lowest growth rate as indicated by the high frequency growth regime shown in Fig. 1. At lower loading frequencies, the crack growth rate tends to accelerate due to a switch of the fracture mode into an intergranular process. This process could be aided by the creep effects accompanying low frequency loadings in the following manner. In unaged Ti-1100, which is free of precipitates, the only barriers to slip are the α colony and prior β grain boundaries. The long cycle time of a 10-300-10 cycle, for example, ensures that dislocations will pass through the α colony boundary by means of climb and/or cross-slip, and move through the following colonies until encountering the relatively strong prior β grain boundary. In this manner, time-dependent creep effects ensures that dislocations pile up at prior β grain boundaries.

The dominant change in Ti-1100 due to aging is the formation of Ti_3Al and silicide precipitates as mentioned previously. These are responsible for the change in intrinsic fracture behavior under low frequency loading conditions. The influence of Ti_3Al precipitation on the mechanical behavior of binary Ti-Al alloys and ternary Ti-Al-Si alloys has been studied by a number of investigators primarily at the room temperature. Here the homogeneity of slip was found to be dependent on the precipitate/dislocation interaction (8,11). In Ti-1100, aging promotes the formation and coarsening of Ti_3Al and silicide precipitates (8). In the aged material, no intense shear bands were found on the crack profile as was found in the unaged material. A measurement of the secondary cracks found that the spacing was approximately twice as close in 10-300-10 test of the unaged material as compared to the aged material (4). This lower density of secondary cracks in the aged material qualitatively indicates that the slip character is more homogeneous. It is assumed, then, that precipitate coarsening tends to promote more homogeneous slip (direct proof of this assumption using observations involving transmission electron microscopy have not been attempted here). Precipitation will have little effect on the high frequency cycle fatigue behavior of the material as the fracture mode will remain transgranular with a torturous path. Under hold time conditions, however, the homogeneous slip will reduce the intensity of the dislocation pileups on the prior β grain boundaries thereby lessening the driving force for intergranular fracture. It is expected, then, that the FCGR under 10-300-10 conditions would be similar to the 10-10 condition as the fracture mechanism is identical and the bulk creep influence is minimal.

CONCLUSION

A series of experiments has been conducted to examine the mechanism responsible for the increase in the low frequency crack growth rate seen in Ti-1100 tested at 593°C. This increase is governed by a process involving a synergistic interaction of both creep and fatigue components of the loading cycle. This interaction results in a deformation mechanism that promotes the occurrence of intergranular fracture without the aid of environmental embrittlement or extensive grain boundary sliding as both phenomena play little or no role in controlling the fracture process of Ti-1100 at elevated temperatures.

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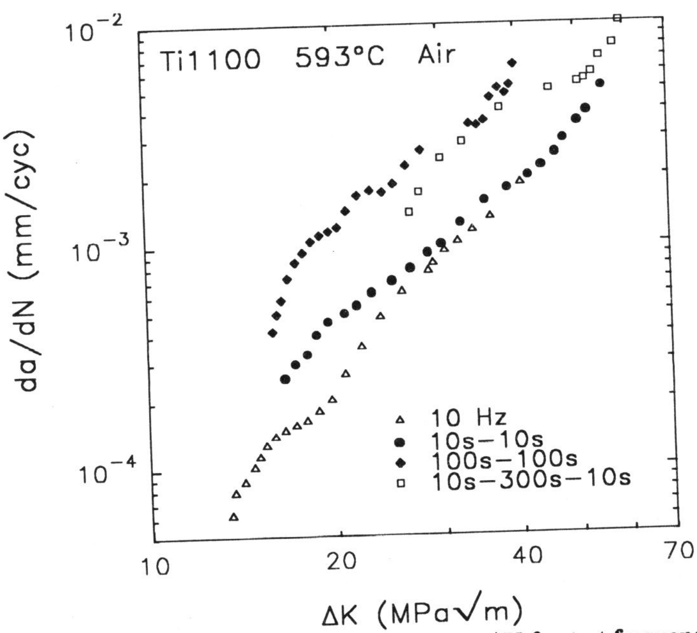

Fig. 1: **Fatigue crack growth rate versus ΔK for test frequencies ranging from 10 to 0.003 Hz**

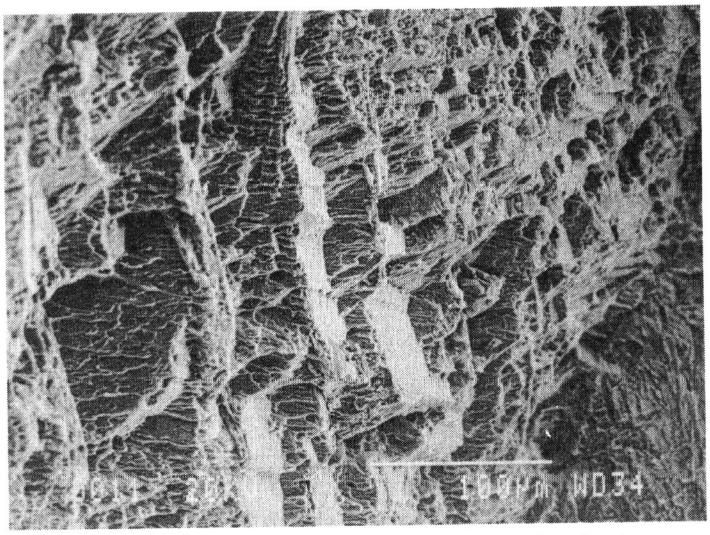

Fig. 2: Scanning electron micrograph of intergranular fracture surface of unaged material with 10-300-10 cycling showing steps due to heterogeneous slip.

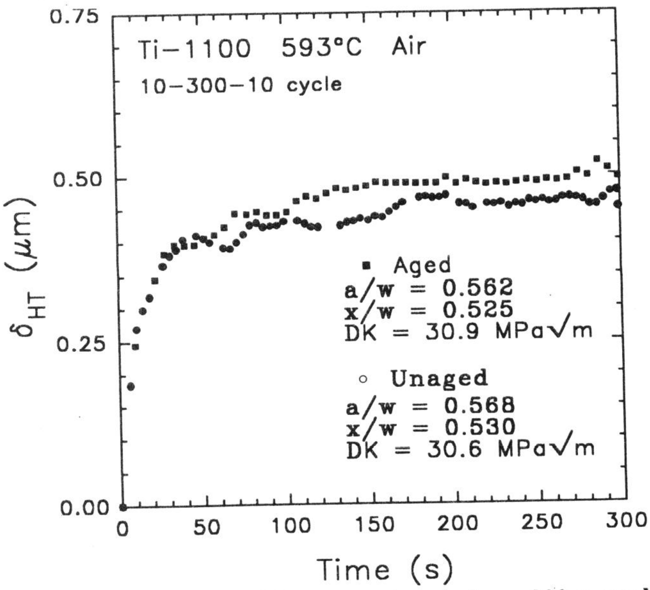

Fig. 3: Evolution of COD versus time during a 300 second hold time at maximum load.

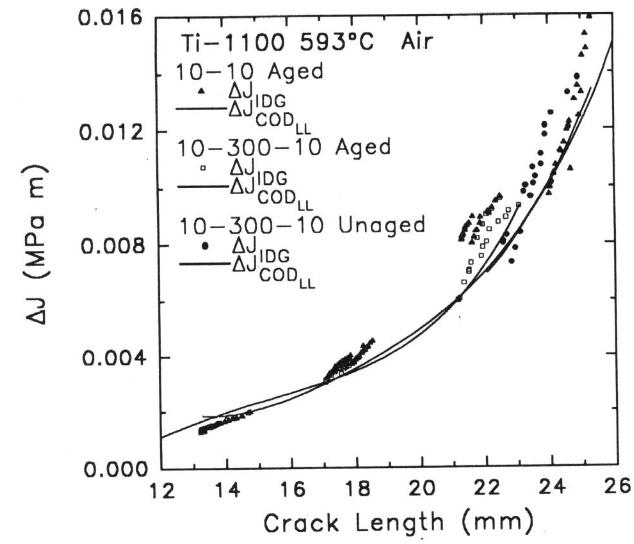

Fig. 4: **ΔJ-integral measured by the load line COD and near-tip COD versus crack length.**

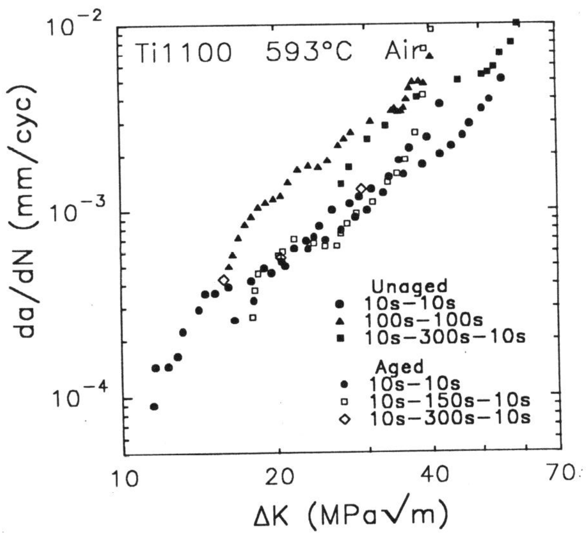

Fig. 5: **Fatigue crack growth rate versus ΔK for aged and unaged Ti-1100**