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A simple model to account for the role of microtexture on fatigue and dwell fatigue lifetimes of titanium alloys

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A deterministic model to account for the effect of initial crack size, microtextured region size and aspect ratio on the mean cyclic and dwell fatigue lifetimes of titanium alloys is described. The model also quantifies variability in lifetime from sample to sample due to differences in the position of subsurface crack initiation. The results indicate that dwell fatigue life and cyclic fatigue life depend most on microtexture region size and initial crack size, respectively.
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Near-alpha titanium alloys are susceptible to early fatigue failure when a hold at peak stress is imposed between the unloading and reloading cycles. Prior research has shown that the primary source of this debit is the presence and attributes—including the size, shape, and intensity—of the microtextured regions (MTRs) [1,2]. In this context, MTRs are clusters of alpha phase with similar *c*-axis orientation. Recently, it was suggested that microtexture also contributes to variability in the low-cycle fatigue (LCF) regime [3]. Nevertheless, there is still no model which accounts for the effect of microtexture on either cyclic or dwell fatigue properties.

In this work, modifications were made to standard linear elastic fracture mechanics crack growth calculations in order to introduce microstructural sensitivity into both fatigue and dwell fatigue lifetime prediction. The model makes the following assumptions based on experimental observations:

(1) Dwell fatigue crack initiation occurs predominantly subsurface and the sites are characterized by large regions of faceted fracture oriented nominally perpendicular to the stress axis with dimensions commensurate with the size and shape of the MTRs in the material [2]. The dimensions of the individual facets correspond well within the dimensions of the primary alpha grains/colonies in the microstructure.

(2) As a result of subsurface initiation, the cracks initially grow in a high-vacuum environment. However, in contrast to typical vacuum fatigue crack growth behavior [4], where the growth rate per cycle is slower at an equivalent stress intensity range, ΔK ($\Delta K = K_{max} - K_{min}$; $K = F\sigma(\pi a)^{1/2}$, where F is a shape factor and a is the crack length), titanium dwell fatigue cracks have been measured to grow up to two orders of magnitude faster [5]. This acceleration, due in part to the absence of high-angle boundaries within the MTRs, depends on the specific misorientations within the MTRs and the acceleration is different for cyclic and dwell fatigue [6,7]. Under dwell loading, the subsurface, small crack growth rate is so fast that it can consume the entire MTR before the crack propagates into the remaining material outside the initiating MTR [2]. During cyclic fatigue, small cracks growing by the faceted mode [6,7] are only moderately accelerated within the MTRs, resulting in a potential for deviation from conventional thumbnail-shaped crack fronts [3]. In contrast to dwell loading, a small crack can convert to long-crack-like behavior before propagating out of the MTR in which it

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initiated. This occurs at a critical stress intensity range $\Delta K = \Delta K_{th,facet}$ where the cyclic plastic zone size exceeds a few average grain diameters and there is significant plastic flow at the crack tip during cycling. The exact value of $\Delta K_{th,facet}$ depends on many specifics including the microstructure, the applied stress, the load ratio and the texture. Until a physical model which takes these factors into account is developed, an approximate value obtained from the literature is $15\ \mathrm{MPa\ m^{1/2}}$ [8,9]. This value may also be obtained from a near-threshold long crack growth test utilizing the proper load ratio followed by careful fractography. A recent study of the role of microtexture on small crack growth rates [10] provides additional credence to the choice of $\Delta K_{th,facet}$. The measurements showed that small cracks in microtexture-free material converged with the long crack growth rates rather quickly at ΔK of $\sim$7–8 $\mathrm{MPa\ m^{1/2}}$. In heavily microtextured material, on the other hand, the small crack growth rates did not approach long crack growth rates until the crack length was larger than 1 mm at $\Delta K \approx 15\ \mathrm{MPa\ m^{1/2}}$ though the growth rates still remained marginally accelerated.

(3) Once outside of the MTR in which it initiated, the crack acts as a long crack and no longer propagates by the faceted mechanism such that the dwell at peak stress has very little influence on the growth rate [9,11].

(4) The shape of MTRs, and thus the shape of the subsurface crack, can be approximated by ellipses for the purposes of calculating ΔK [10].

(5) Finally, the model aims to capture mean-field behavior, i.e. that due to order-of-magnitude differences in lifetime due to large variations in microtexture, and thus does not capture the influence of all relevant microstructural parameters at this time. A more complete, probabilistic model which incorporates microstructure effects (grain size, compatibility between hard/soft grains, etc.) is the subject of future work.

Small crack [12,13] and closure-corrected long crack [14] growth rates for a similar alloy, Ti–8Al, were obtained from the literature. Growth rates for subsurface dwell fatigue cracks propagating by the faceted mode were not available. While progress has been made with microfocus X-rays on small fatigue samples [5] it remains difficult to locate and measure the growth rate of these subsurface cracks and this remains a fruitful area for future research. It was shown in prior work [5,7] that these cracks may propagate up to two orders of magnitude faster than long cracks at equivalent ΔK. Thus, for the present analysis, the dwell crack growth data were obtained by scaling the cyclic small crack growth data about an order of magnitude and decreasing slightly the ΔK dependence to be more consistent with experimental observation [5].

The model domain (Figure 1a) consists of an infinitesimal slice through the gauge section of an axially loaded fatigue specimen that is subjected to either cyclic or dwell fatigue loading. An elliptical domain is placed in the model and an elliptical crack with major axes a_i and c_i is assumed to be present at the center of this domain (this choice makes the solution more conservative) on the first cycle. The position of the initiation site and the encompassing ellipse affect the calculated lifetime. The initiation location can be selected by Monte Carlo methods to assess the expected variability among a limited population of samples, or systematically to aid fractographic inspections in assessing variability in fatigue life from sample to sample. When the crack is within the bounds of the ellipse (i.e. within the microtextured region in which it initiated) it is assumed to propagate at rates consistent with faceted growth while it propagates at long crack growth rates outside of this region (Figure 1b). For the case of cyclic fatigue, small crack growth behavior also ceases when $\Delta K > \Delta K_{th,facet}$.

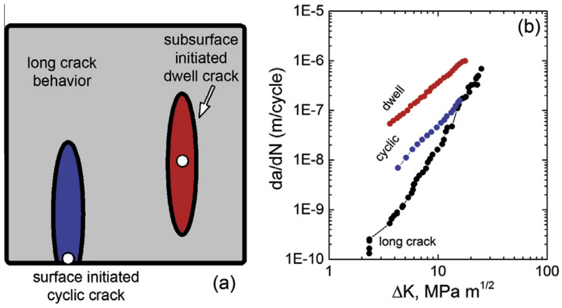

Figure 1. (a) Schematic illustration of model domain: the white circle indicates the initial crack, while the ellipses represent the bounds of the microtextured region in which the crack initiated. (b) The crack growth rates associated with each region (surface-connected cyclic small crack, subsurface-initiated dwell crack and long crack).

The initial crack is grown to failure, which occurs at a final crack length a_f by integrating:

$$N_f = \int_{a_i}^{a_f} \frac{1}{\frac{da}{dN}}\, da,$$

where the crack growth rate da/dN is either provided as a lookup table or as $da/dN = C\Delta K^m$, where C and m are constants that depend on material microstructure and texture, and a_f is defined as the crack length where $K_{max} \geqslant K_{IC}$ for the material of interest ($35\ \mathrm{MPa\ m^{1/2}}$ for the present investigation). In this framework, ΔK is calculated for each of the four crack tips and the corresponding growth rate for each is interpolated from the empirical crack growth curves or calculated from Paris's law. Using empirical data permits the near-threshold behavior to be captured more accurately and also introduces natural variability in small crack growth rates. Moreover, since the input data can be collected at any temperature and load ratio, the model is generally applicable, though it remains to be demonstrated whether or not crack growth remains significantly accelerated at elevated temperature. The four crack tips of the ellipse may be in different stages of growth, specifically inside or outside of the initiating MTR, and thus may follow different growth rate curves during crack extension. The K solution depends on whether or not the crack is surface, subsurface or corner connected. The implemented K solutions are those due to Newman and Raju (for surface, subsurface and corner cracks on square