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EFFECTS OF STRENGTH AND GRAIN SIZE ON NEAR-THRESHOLD FATIGUE CRACK GROWTH IN ULTRA-HIGH STRENGTH STEEL

Robert O. Ritchie*

INTRODUCTION

Although much research in recent years has focused on fatigue crack propagation in metals, the slow growth of fatigue cracks at rates less than 10^{-5}-10^{-6} mm/cycle has received comparatively little attention. This is unfortunate since the majority of the lifetime of a cracked component would be spent in this region. Furthermore, a knowledge of a threshold stress below which fatigue cracks cannot propagate would be essential in the design of components which are subject to extreme high frequency, low amplitude loading for lifetimes of 10^{10}-10^{12} cycles [1].

The fatigue crack growth rate, da/dN, is generally related to the alternating stress intensity ($\Delta K = K_{max} - K_{min}$), developed at the crack tip, through a relationship [2] of the form

$$da/dN = C\ \Delta K^m , \qquad (1)$$

where C and m are constants. Although this expression adequately described behaviour for the mid-range of growth rates ($\sim 10^{-5}$-10^{-3} mm/cycle), it often underestimates the propagation rate at higher growth rates, where K_{max} approaches K_{Ic}, the fracture toughness, and is conservative at low growth rates ($< 10^{-5}$ mm/cycle) where ΔK approaches a threshold stress intensity, ΔK_o, below which crack growth cannot be detected.

For the mid-range of growth rates fatigue crack propagation has been shown to be largely insensitive to such variables as mean stress, microstructure and test-piece thickness [3 - 5], and this is consistant with a ductile striation mechanism [6] of growth. At higher stress intensities the predominance of additional superimposed fracture mechanisms on striation growth ('static modes') results in an increasing sensitivity to mean stress, microstructure and thickness [3 - 5].

At very low stress intensities, which approach ΔK_o, fatigue crack propagation again becomes dependent on mean stress and microstructure, although the precise fracture mechanisms are not well understood. It is clear, however, that the influence of environment is of paramount importance here. Previous studies in steels, aluminum and titanium alloys [eg. 7 - 10] have shown that ΔK_o is decreased and near-threshold crack propagation rates are increased at high stress ratios ($R = K_{min}/K_{max}$). This R dependence, however, can be severely reduced for tests *in vacuo* [7]. Mechanisms of growth in this region have been observed to be microstructually sensitive involving the occurrence of environmentally induced fracture modes, such as intergranular fracture in steels. However, exactly which microstructures are beneficial in improving resistance to near-threshold fatigue failure is

* Department of Mechanical Engineering, Massachusetts Institute of Technology, Cambridge, Massachusetts 02139, U. S. A.

still open to question. Limited data in the literature indicate that material strength and coarseness of microstructure (i.e., grain size) may be important variables. A reduction in threshold ΔK_o has been seen with increasing yield strength in steels [9], while lower near-threshold crack growth rates have been measured with increasing grain size in titanium alloys [8]. More recently, it has been suggested [11] that the threshold is related to grain size through an empirical Hall-Petch type relationship (i.e., $\Delta K_o = A + Bd^{-1/2}$, where A and B are constants and d the grain diameter), and that the yield strength dependence on ΔK_o simply arises from decreasing yield with increasing grain size (i.e., from the Hall-Petch relationship).

To examine this effect and to extend the data to much higher strength materials (where yield stress can be independent of grain size), fatigue tests were performed on ultra-high strength steels to assess (i) the influence of strength at constant grain size, and (ii) the influence of grain size at constant strength, on threshold stress intensity and near-threshold crack propagation rates. The results constitute part of a larger programme of research to characterize the microstructural influences on fatigue crack propagation at growth rates less than 10^{-5} mm/cycle in an attempt to provide a basis for the design of alloys resistant to low growth rate fatigue failure.

EXPERIMENTAL

The material studied was a vacuum-arc remelted, silicon-modified 4340 (300-M) low alloy steel, tested in the quenched and tempered condition. Fatigue tests were performed on 12 mm thick C.T.S. specimens, cycled at 50 Hz (sine wave) in air at controlled temperature (23° C) and relative humidity (45%). Growth rates were monitored using the electrical potential technique [12]. The threshold ΔK_o was calculated in terms of the stress intensity at which no growth occurred within 10^7 cycles. Since the crack monitoring technique is accurate to 0.1 mm, this corresponds to a maximum growth rate of 10^{-8} mm/cycle. Thresholds were approached using a successive load reduction followed by crack growth procedure, in increments of 1-1.5 mm of growth, to avoid residual stress effects. For comparison, tests were also conducted at intermediate and high growth rates to yield propagation rate data between 10^{-8}-10^{-2} mm/cycle.

RESULTS

To assess the influence of strength at constant grain size, specimens were austenitized at 870° C and oil quenched, giving a prior austenite grain size of 20 µm. Fatigue crack propagation rates were measured for 100° C, 300° C, 470° C and 650° C tempered conditions. The mechanical properties are listed in Table 1, and the fatigue results at R = 0.05 are shown in Figure 1. From these data it is clear that the influence of tempering treatment is maximized at high and low stress intensities. For the mid-range of growth rates, where equation (1) holds, there is little variation in propagation rate with tempering temperature. In this range, the mechanism of failure was observed to be solely ductile striation growth, which is consistant with the lack of microstructure sensitivity on the growth rate behaviour [4]. At higher stress intensities, faster propagation rates were measured for the lower tempering temperatures, consistant with the lower K_{Ic} of these structures. Fractography confirmed the presence of static modes (quasi-cleavage and microvoid coalescence) accompanying striation growth in the lower toughness structures.

The largest effect of microstructure was seen at low growth rates, where, for example, at $\Delta K = 9$ MPa$\sqrt{m}$, there was a factor of 100 times difference in growth rate between 100 and 650° C tempered conditions. As the tempering temperature was raised, the threshold ΔK_o increased from 3.0 to 8.5 MPa$\sqrt{m}$, concurrent with a two-fold reduction in tensile strength. Figure 2 shows this variation of ΔK_o, at R = 0.05 and 0.7, with tensile strength, the inverse dependence of ΔK_o on strength being far less marked at the higher R ratio. Examination of the literature confirms this trend of increasing threshold and decreasing near-threshold growth rates with decreasing strength in steels (Figure 3). The effect is not seen at higher growth rates. Little variation in fracture mechanism was observed at low growth rates. All structures failed by a flat, fine-scale ductile transgranular mode containing segments of intergranular fracture. Intergranular facets were not seen in the 650° C tempered structure.

To assess the influence of grain size, specimens were austenitized at 870° C and at 1200° C, oil quenched and tempered at 300° C. The higher austenitizing temperature results in a larger prior austenite grain size (160 µm), yet the yield and ultimate strengths remain largely unaffected (i.e., Hall-Petch relationship does not apply). The mechanical properties are listed in Table 1. The fatigue results at R = 0.05 (Figure 4) indicate that, at low stress intensities, near-threshold crack propagation rates were lower in the larger grained structure although the value of ΔK_o was unchanged. This difference in crack growth rate was not observed at higher stress intensities. Fracture surfaces were much rougher in the coarse grained structure, showing a facetted, ductile, transgranular mode with segments of intergranular fracture.

DISCUSSION

It is clear that, for fatigue crack propagation in air at low stress intensities in high strength steels, decreasing the strength of the steel leads to a marked increase in threshold ΔK_o and a reduction in near-threshold crack propagation rates. Coarsening grain size also leads to lower near-threshold crack growth rates although the threshold itself remains unchanged. Thus, the proposed interrelation [11] between ΔK_o, strength and grain size based on a Hall-Petch relationship does not apply in these ultra-high strength steels.

It is not clear why threshold ΔK_o and near-threshold crack propagation rates show a large dependence on strength. It is felt that one or more of the following reasons may apply: (1) If the growth of a fatigue crack near ΔK_o relies on some critical fracture stress being exceeded near the crack tip, then, at some characteristic distance ahead of the crack tip, higher tensile stresses will be achieved in higher strength materials. The effect is enhanced in quenched and tempered steels by the greater work hardening capacity of structures tempered at lower temperatures. (2) In lower strength materials larger plastic zones are created ahead of the crack tip, and hence residual stresses, arising from a crack closure effect [13], may be more effective in retarding crack growth. However, electrical potential measurements were unable to detect any crack closure, except below K_{min}, supporting previous observations [14] of negligible closure for crack growth under plane strain conditions. (3) The effect is more likely to arise from the increased susceptibility of higher strength steels to environmental influences of hydrogen in moist air. Increased strength can result in a higher equilibrium solubility of hydrogen in the crack tip region (due to higher triaxiality), and a reduction of hydrogen necessary to cause cracking [15]. This is reflected in the reduction in K_{Iscc}, the threshold for hydrogen and stress corrosion cracking in steels, with increased strength. It is conceivable that the same effect influences near-threshold fatigue crack propagation rates in humid air.

The superior near-threshold fatigue crack propagation rates observed in coarse grained structures are, similarly, thought to be environmentally induced, since little grain size dependence is seen at higher growth rates. The environmental influence of hydrogen is greatest when the maximum plastic zone is of the order of the grain size [7], such that hydrogen atoms can be swept-in by dislocations into grain boundaries. In coarser grained structures the plastic zone remains small compared to the grain size until much higher stress intensities (in present study $\Delta K \sim 30$ and 90 MPa$\sqrt{m}$ in 870° C and 1200° C austenitized structures respectively when maximum plastic zone equals grain size), resulting in a reduced environmental influence. Experiments are now being performed in dry, inert argon atmospheres in an attempt to clarify the above effects.

The environmental influence of hydrogen from moist air can be enhanced, however, by varying the cooling rate after tempering. In 300-M steel, slow step-cooling (instead of oil quenching) after tempering at 650° C, to allow segregation of impurity elements to prior austenite grain boundaries, results in increased near-threshold fatigue crack propagation rates, and a reduction in the threshold ΔK_0 from 8.5 to 6.2 MPa$\sqrt{m}$ [16]. Here the grain boundaries are embrittled by impurity segregation, and are consequently more susceptible to intergranular hydrogen-assisted cracking during slow fatigue crack growth.

In summary, it is apparent that microstructure has a large influence on fatigue crack propagation behaviour at low stress intensities. Improved resistance to near-threshold fatigue crack growth in steels can be achieved through lower strength and coarser grained structures, although whether these effects are mechanically or environmentally induced remains an open question.

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Table 1 Mechanical Properties of 300-M Steel

Austenitizing Temp. (°C)	Tempering Temp. (°C)	Prior Austenite Grain Size (μm)	0.2% Proof Stress (MPa)	U.T.S. (MPa)	K_{Ic} (MPa$\sqrt{m}$)	Threshold ΔK_o (MPa$\sqrt{m}$) R = 0.05	Threshold ΔK_o (MPa$\sqrt{m}$) R = 0.70
870	100	20	1497	2338	35.5	2.98	2.28
870	300	20	1737	2006	65.1	3.08	2.30
870	470	20	1497	1683	68.9	5.10	2.46
870	650	20	1074	1186	185*	8.50	3.68
1200	300	160	1657	1986	80.3	3.00	2.39

*estimated from equivalent energy at maximum load.

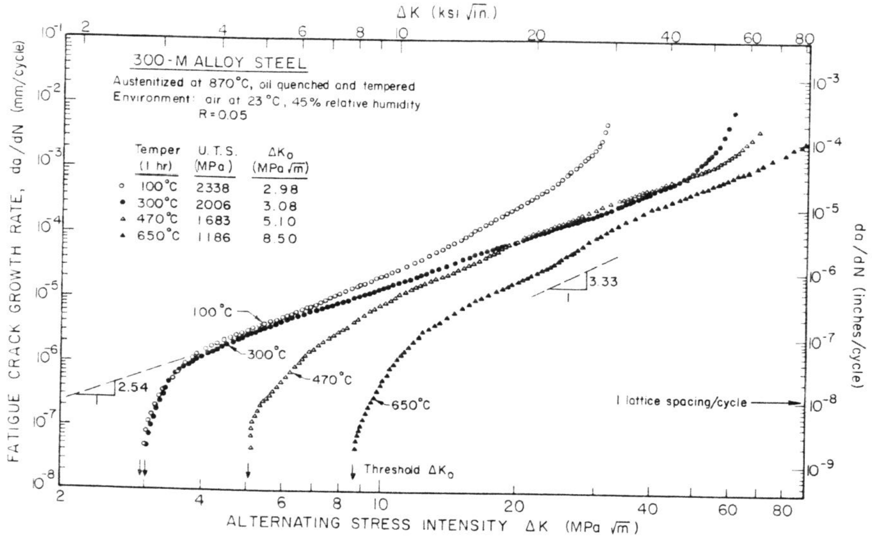

Figure 1 Variation of Fatigue Crack Growth Rate (da/dN) with Alternating Stress Intensity (ΔK) at R = 0.05, for 300-M Steel, Quenched and Tempered Between 100°C and 650°C, Showing Influence of Material Strength

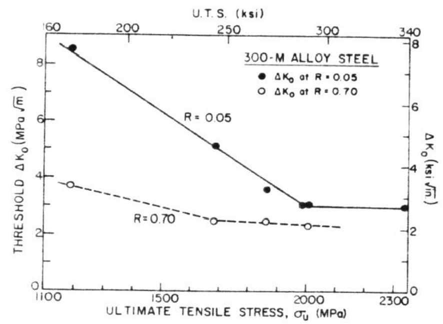

Figure 2 Influence of Ultimate Tensile Strength on Threshold for Fatigue Crack Growth (ΔK_o) in 300-M Steel, Tested in Air

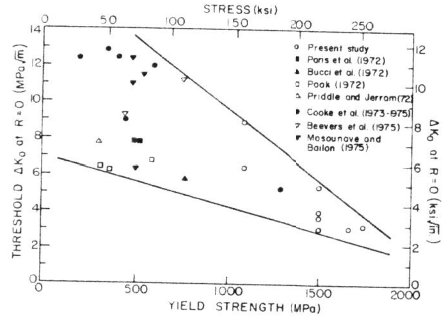

Figure 3 Summary of Results Showing Variation of Threshold (ΔK_o) at R = 0 with Yield Strength for Steels

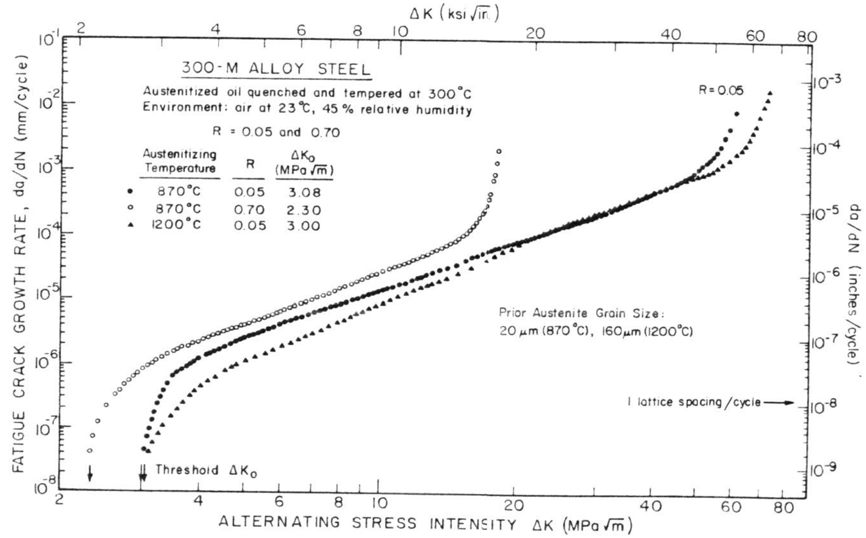

Figure 4 Variation of Fatigue Crack Growth Rate (da/dN) with Alternating Stress Intensity (ΔK) for 300-M Steel, Austenitized at 870°C and 1200°C, Showing Influence of Grain Size