ENVIRONMENTALLY ASSISTED SMALL FATIGUE CRACK GROWTH IN 7017 ALUMINIUM ALLOY

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ABSTRACT

Small and long fatigue crack growth tests on 7017 aluminium alloy are reported for air and in 3.5% NaCl solution. The results show that growth rates are significantly increased by NaCl solution. A more brittle crack growth process is associated with this aggressive environment. However, the growth behaviour both in air and in NaCl solution is insensitive to ageing treatment. Physically small cracks ($a \leq 1mm$) are able to propagate significantly below the long fatigue threshold for either air or NaCl solution. The crack growth in both environments can be described by the Paris Law based on a ΔK criterion.

KEYWORDS

Small crack, corrosion fatigue, fatigue crack growth, 7017 aluminium alloy.

INTRODUCTION

The Al-Zn-Mg(-Cu) or 7000 series aluminium alloys are widely used where a high strength/weight ratio is required, e.g. aerospace and transport. The 7017 alloy is one member which provides a good combination of medium strength, low weight, weldability and corrosion resistance. It has been used for the plating of armour vehicles and in the construction of lightweight bridges. Previous work on this important engineering alloy under corrosion fatigue has been reported by Holroyd and Hardie (1983). They, however, concentrated on long crack growth. Information on small crack behaviour is limited and comparison between long and small cracks under fatigue and corrosion fatigue (CF) have not been reported.

The aim of the present research was to establish the growth characteristics of small cracks in 7000 series aluminium alloy under fatigue and CF conditions by exploring the effects of mean stress and variations in microstructure. The work on 7017 is in support of the development of a damage tolerance design approach for "subcritical crack growth" in this important engineering material.

EXPERIMENTAL PROCEDURES

The material was supplied in the form of 2x1x0.075(m) rolled plate in a solution treated condition. The composition is listed in Table 1. The microstructure was fully fibrous with the anisotropic nature evident from elongated grains along the rolling direction. The material was cut into blocks of 40x40x18(mm) and 75x18x18(mm) for the different specimen geometries prior to heat treatment.

Table 1. Material composition.

Element	Zn	Mg	Cu	Mn	Fe	Cr	Si	Ti	Ni	Al
Wt%	4.75	2.50	0.14	0.32	0.25	0.17	0.14	0.04	0.01	bal.

In order to eliminate the influence of natural ageing, a solution treatment was carried out by heating the blocks in an air circulation furnace at $460 \pm 2^{\circ}C$ for 2 hours followed by quenching into cold water. The solution treated pieces were then aged in air at $150^{\circ}C$ for 3, 8, and 24 hours to give: underaged, peakaged and overaged microstructures respectively.

Two types of specimens, the modified Rolls Royce Corner Crack (CC) and the Compact Tension (CT), were used for the programme. The geometries of these specimens are shown in Figs.1 and 2. The essential difference is that the CC cracks are smaller and have a 'part through' quarter circle geometry whereas the CT specimens contain large through section cracks. The specimens were machined from the heat treated blocks to remove the surface layer. The starter slit for CC and notches for CT specimens were machined parallel to the rolling plane. This allowed the cracks to propagate along the longitudinal direction, the most susceptible orientation in an aggressive environment for 7000 series aluminium alloys.

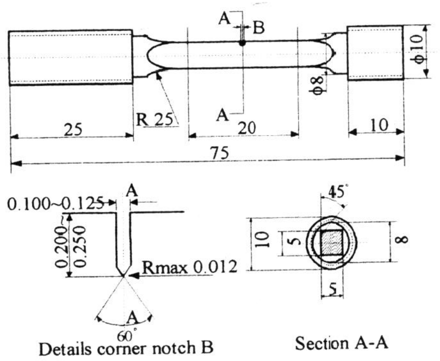

Fig.1. Drawing of the CC specimens.

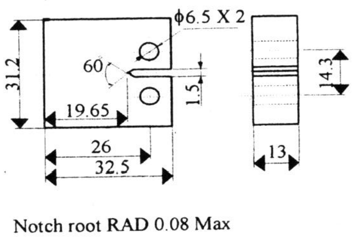

Fig.2. Drawing of the CT specimens.

The fatigue tests in laboratory air at a humidity of 45~55% and in 3.5% NaCl solution were carried out on an Amsler Vibrophore machine. Specimens for corrosion fatigue were immersed in 3.5%(wt) sodium chloride solution with pH of 5.5~5.9 in a perpex corrosion cell. Tests were carried out under free corrosion conditions at room temperature. Only a small area near the starter slit for CC or near the notch area for CT was exposed to the corrodent. The region away

from the crack was protected by silicon rubber and Lacomit lacquer. The corrodent was made using Analar grade chemicals and distilled water, and added to the cell immediately prior to loading the specimen. The solution, during the test, was open to the atmosphere, i.e. naturally aerated. A range of heat treated and pre-cracked CC and CT specimens were tested under constant load amplitude at a stress ratio R=0.1 and a frequency of 40Hz.

Crack advance was monitored by an ASI 4088 Alternating Current Potential Difference (ACPD) Monitoring System operating at a frequency of 80KHz. The PD readings were converted into crack lengths through a calibration curve (Lu *et al*, 1995). All the specimens were insulated from the machine during the tests by specially designed grips.

EXPERIMENTAL RESULTS

The crack propagation behaviour for the various heat treated CC and CT specimens in air and NaCl solution is summarised in Figs.3 to 6. As a comparison, long fatigue crack growth data published by Holroyd and Hardie (1983) for 7017-T651 in dry air with a frequency of 4Hz and in sea water with frequencies of 20Hz and 70Hz are also plotted.

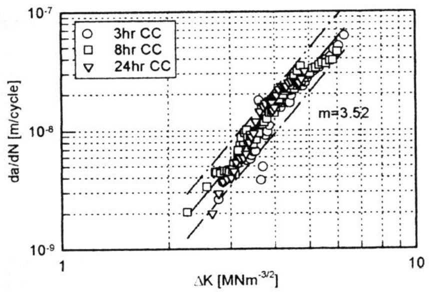

Fig.3. Crack growth in air for CC specimens.

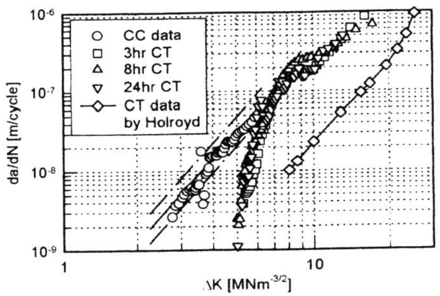

Fig.4. Comparison of crack growth in air for CC and CT specimens.

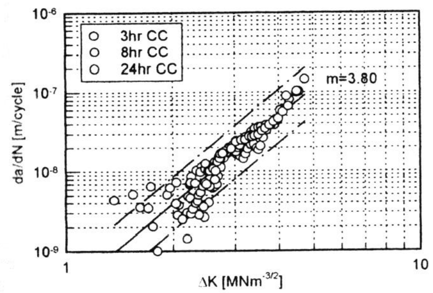

Fig.5. Crack growth in salt for CC specimens.

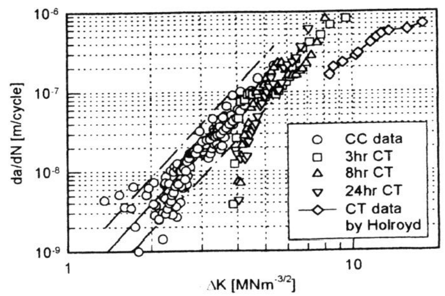

Fig.6. Comparison of crack growth in salt for CC and CT specimens.

It is apparent that on a log-log basis linear relationships can describe the crack growth behaviour in terms of ΔK for CC specimens. The best fit regression line and 95% tolerance limit at a 95% confidence level are plotted for the CC data. At higher ΔK, the CC data merge with the CT rates once the cracks in the former specimens exceed small crack status. Small cracks can propagate well below the fatigue thresholds calculated from long crack data. The ΔK value for CC at a crack growth rate of 10^{-9} m/cycle is approximately 40% of the long crack value obtained from CT tests at the same growth rate both in air and in NaCl solution (Figs.4 and 6). In addition, it is evident that heat treatment has little influence on the fatigue crack growth behaviour. A Paris relationship can be used to characterise fatigue crack growth behaviour for CC specimens:

$$\frac{da}{dN} = C\Delta K^m$$

with a the crack length, N the number of cycles, C and m material constants.

Comparing the above results with the published long crack data of Holroyd and Hardie (1983), it is obvious that there are significant differences. The present work gives growth rates that are about an order of magnitude higher at the same ΔK.

On comparing the CC and CT data in air and in NaCl solution, a strong environmental enhancement of crack growth rate is apparent. This is represented by a shift in the curves to lower ΔK at a given rate of growth. Alternatively, for the same ΔK, crack growth rates are increased by almost an order of magnitude in the aggressive environment, Figs.7 and 8. Compared with the published data by Holroyd and Hardie (1983) for long crack data, however, the differences between crack growth in air and in NaCl solution for CT specimens in the present work are not as significant (Fig.8).

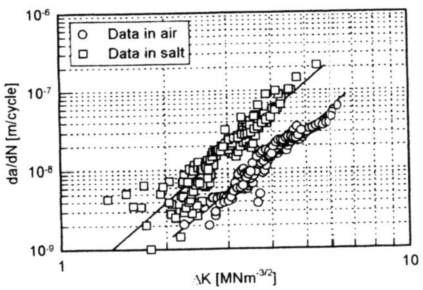

Fig.7. Influence of environment on crack growth for CC specimens.

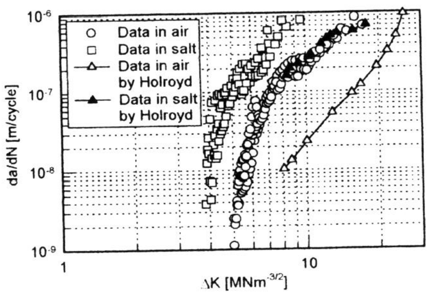

Fig.8. Influence of environment on crack growth for CT specimens.

For fatigue crack propagation of CC specimens in air, transgranular features are apparent (Fig.9). Fig.10 shows the fracture surfaces of CC specimens under corrosion fatigue. A mixed crack propagation mechanism appears to operate including transgranular pseudo-cleavage facets and intergranular cracking. The fracture surfaces of long cracks under fatigue and

corrosion fatigue are similar to those reported for the small cracks. This indicates that the same crack propagation mechanisms are operating.

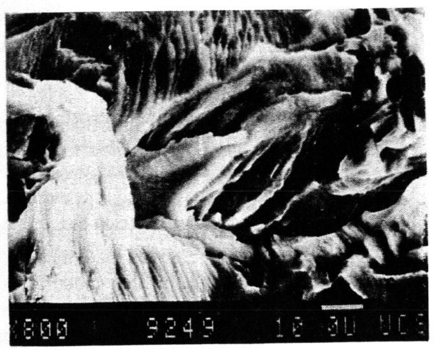

Fig.9. Fracture surface for CC in air.

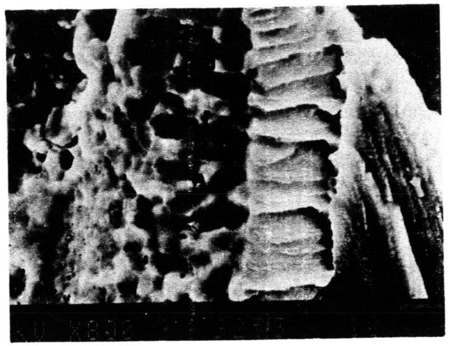

Fig.10. Fracture surface for CC in salt.

DISCUSSIONS

The crack growth rates for CC specimens either in air or in NaCl solution merge with the long crack growth data from CT specimens once the crack becomes long (a>1mm) as shown in Figs.3 and 6. This fact supports the use of CC specimens for evaluating the interaction between small and long crack growth behaviour. At low ΔK (below the long crack fatigue threshold), significant small crack growth occurs. This is crucial for damage tolerant design calculations because using this threshold as a "cut-off" criterion will inevitably result in a non-conservative design.

Some discrepancies are apparent between the present data and that of Holroyd and Hardie (1981). Crack growth is found to occur at a lower ΔK range with a reduced fatigue threshold. Furthermore, the difference between CT in air and in salt for the present results is not as large as Holroyd and Hardie reported. These differences can be ascribed to the following factors. Firstly the K calibration formulas are different, so that the calibration of Holroyd et al. under-estimates the crack growth rates at lower ΔK, but over-estimates at higher ΔK. Secondly the environmental conditions are different. The previous tests were carried out in dry air (complete immersion in anhydrous magnesium perchlorate) and natural seawater, whereas the present tests were in NaCl solution and laboratory air with some degree of humidity.

For fatigue tests in NaCl solution, crack propagation is accelerated by the aggressive environment (Figs. 7 and 8). Fractographic results reveal that the fracture mechanism is changed by the presence of the aggressive environment, which causes both pseudo-cleavage transgranular cracking and intergranular cracking (Fig.10).

Several mechanisms have been proposed to describe CF crack propagation in Al-Zn-Mg alloys. These include hydrogen embrittlement, film rupture and anodic dissolution (Gangloff, 1989). In the present work, the repair of the surface film is hindered due to the high frequency of the tests. The hydrogen embrittlement and anodic dissolution could operate simultaneously. The low Cu content for the present material will lead to a higher rate of dissolution or hydrogen absorption at the crack tip (Sarkar *et al*, 1981). However, most of the CF tests lasted only 2 to 3 hours. It is difficult for anodic dissolution to act as a controlling mechanism but it could be a source of hydrogen. During fatigue tests, dislocation movement would accelerate the diffusion of hydrogen thereby promoting embrittlement. Thus it is argued that hydrogen embrittlement is more likely to be the dominant mechanism.

The present results, which indicate physically small cracks can display significant propagation below the fatigue threshold obtained from long crack data, can be attributed to crack closure that decreases fatigue crack growth rates by reducing the effective stress intensity range (ΔK_{eff}). Due to the shorter plastic wake for cracks in CC tests compared with those in CT specimens, closure effects would be more apparent in the latter. However, even small cracks can close above the minimum load in the fatigue cycle. Evidence has been reported by James and Morris (1983) for the growth of small cracks in titanium alloys. They found that closure, especially that induced by roughness, decreased with decreasing crack length for cracks less than approximately 0.16mm long. The fact that the threshold for small cracks in the present work must be more than 60% lower than that found in CT specimens at R=0.1 is consistent with this argument. This is similar to the work of McCarver and Ritchie (1982) on a nickel alloy in which threshold values for physically small cracks (a=0.01~0.20mm) at low mean stresses (R=0.1) were found to be 60% smaller than for long cracks (a=25mm). At higher mean stress (R=0.7) where closure effects were minimal, they did not observe a similar difference. In summary, therefore, at equivalent nominal ΔK levels, physically small cracks may be expected to propagate faster and show lower thresholds than the corresponding long cracks simply because of a smaller effective stress intensity range in the case of the latter. On this basis, it is anticipated that the sub-threshold small crack growth behaviour will be brought into direct correspondence with the long crack data simply by using an appropriate ΔK_{eff} (Ritchie and Lankford, 1986).

In aggressive environments, corrosion debris can enhance crack closure. Furthermore, occluded environments within the confines of small cracks can differ from the bulk solution and from those associated with long cracks. Experimental results show that crack tip pH is more acidic than that of the bulk solution (Holroyd, 1989). This promotes CF crack propagation because the concentration of hydrogen adsorbed on the metal surface is proportional to the square root of the local hydrogen ion concentration (Gangloff, 1985). However, dissolved oxygen within an occluded crack leads to the production of hydroxyl ions and an increased pH (less acidic), thereby decreasing the concentration of adsorbed hydrogen and inhibition of the hydrogen embrittlement component (Gangloff, 1985). With small cracks, the oxygen is easier to move from the crack tip through convection and diffusion. Thus, it is reported that small crack environments appear to be more deleterious (Wei and Gangloff, 1989). With this respect, the characterisation of the local chemical conditions at the crack tip becomes the most important issue.

It is important to point out that the threshold of small crack growth only applies to the physically small regime, because the discrepancy is caused by crack closure. The present work appears to be consistent with a correlation through ΔK_{eff}. However, this correlation and the threshold do not apply to microstructurally small cracks. Some anomalous crack growth behaviour was observed during the course of the programme. Thus, in the small crack regime at lengths less than 0.5mm, initially high growth rates tended to decrease with increasing ΔK until a threshold condition was reached (Fig.6). This can also be explained by crack closure and the interaction of the small crack with microstructural features (Suresh and Ritchie, 1984). On one hand, closure increases with increasing crack length leading to a reduction of driving force (ΔK_{eff}), and hence crack growth rates. On the other hand, cracks which are microstructurally small can be impeded locally by grain boundaries (James and Morris, 1986; Tanaka, 1986). In the present tests, the longitudinal grain dimension is less than 0.5mm. Thus, when crack lengths are less than 0.5mm, anomalous crack growth behaviour is not unexpected.

CONCLUSIONS

1. Fatigue crack growth rates in under-, peak and over-aged 7017 aluminium alloy are significantly increased by the presence of a 3.5% NaCl solution. A more brittle crack growth mechanism is associated with this aggressive environment.

2. Fatigue crack growth behaviour both in air and in NaCl solution is insensitive to ageing treatment.

3. Physically small cracks ($a \le 1mm$) are able to propagate significantly below the long fatigue threshold in air and in 3.5% NaCl solution. Lower fatigue thresholds are associated with small crack growth in both environments. The crack growth in both environments can be described by a Paris Law in terms of ΔK, i.e.

$$\frac{da}{dN} = C\Delta K^{m}$$

4. Part through cracks in corner crack specimen provide a convenient means of investigating crack growth behaviour for both long and small cracks.

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