The Effect of Slip Distribution on the Growth of Small Fatigue Cracks in Al–Mg–Si Alloys

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ABSTRACT

The influence of slip distribution on small fatigue crack growth has been studied in a series of Al-Mg-Si alloys which contain differing volume fractions of dispersoids. Small cracks were found to propagate faster than long cracks when correlated with both ΔK and ΔJ. Although the slip homogenization caused by these dispersoids profoundly effect the behaviour of long cracks no effect on small fatigue crack growth could be detected. Thus the effect of slip distribution on fatigue life is shown to be essentially due to the initiation rather than the propagation of cracks.

KEYWORDS

Small fatigue crack; slip distribution; Al-Mg-Si alloy

INTRODUCTION

It is now well accepted that the homogenization of slip in heat-treatable aluminium alloys by the addition of dispersoid particles can result in substantial improvements in tensile ductility (Dunwoody et. al. 1973), fracture toughness (Blind and Martin 1983 a, b), and fatigue crack initiation resistance both under high cycle and low cycle conditions (Edwards and Martin 1981, 1982).

Homogenizing slip normally has a deleterious effect on long fatigue crack propagation. (Lindigkeit *et al*, 1981, Edwards 1983). This is because inhomogeneous slip promotes faceted crack surfaces which provide effective crack tip shielding through both crack deflection and roughness induced closure mechanisms. However, peak-aged Al-Mg-Si alloys exhibit intergranular embrittlement by microvoid coalescence along grain boundary precipitate free zones (Blind and Martin 1983, a,b). Thus, homogenizing the slip in Al-Mg-Si alloys decreases fatigue crack propagation rates and produces higher ΔK_{th}'s. This occurs despite the effective crack tip shielding created by the

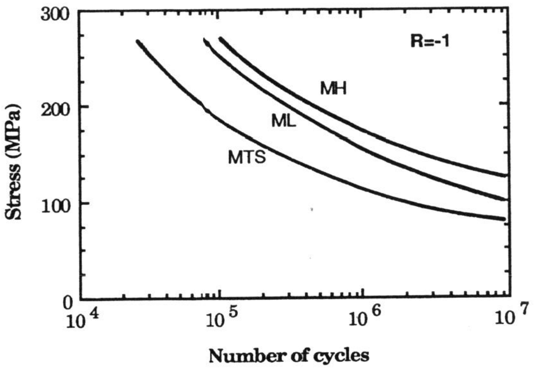


Fig. 1. Summarized S/N data from Edwards and Martin (1981).

rough fracture surfaces as the crack extension process is dominated by intergranular embrittlement which is K_{max} controlled (Edwards and Martin 1981, 1983). The present work is concerned with the effect of slip distribution on small fatigue crack growth in the same peak-aged Al-Mg-Si alloys. It forms part of a project developing small crack based fatigue life models for automobile components. Al-Mg-Si alloys are of particular interest to the automobile industry due to their high stress corrosion resistance.

Table 1. Composition of alloys under investigation

Alloy	Element, wt %				Dispersoid (vol. %)	Coarse particles (vol.%)
	Mg	Si	Fe	Mn		
MTS	0.63	1.07	<0.001	<0.001	..	<0.1
ML	0.59	0.99	<0.001	0.21	0.22	0.78
MH	0.61	1.03	<0.001	0.60	0.61	2.03

EXPERIMENTAL MATERIALS AND METHODS

The alloys were the same as used by Edwards and Martin, 1981, 1983 from which the long crack data presented in this paper were taken. Their composition and tensile properties are given in Table 1 and 2 respectively. All the alloys have similar dispersions of age hardening precipitate, and contain similar precipitate free zones (PFZ) adjacent to the grain boundaries; only ML and MH contain the α - $Al_{12}Mn_3Si$ phase in the form of 0.1µm dispersoids. All the alloys contained some coarse constituent particles between 1 and 10µm in diameter, but as may be seen from Table 1, their numbers increased markedly with dispersoid content.

Small fatigue crack growth was measured at R=0.1 on electropolished 6.5mm thick four-point bend specimens using a two-stage replication technique. Stress intensities were calculated from the Newman and Raju (1981) solution for semi-elliptical surface cracks. Following the work of Starkey and Shelton (1981) and Dowling(1977), ΔJ was calculated from the following expression:

$$\Delta J = (3.2\ \Delta W_e + 6.5\ \Delta W_p).\ a \qquad (1)$$

where ΔW_e and ΔW_p are the elastic and plastic strain energies respectively. They are calculated thus :

$$\Delta W_e = \frac{(\Delta\sigma)^2}{2E} \qquad \Delta W_p = \frac{\Delta\sigma\Delta\varepsilon_p}{1 + n'} \qquad (2)$$

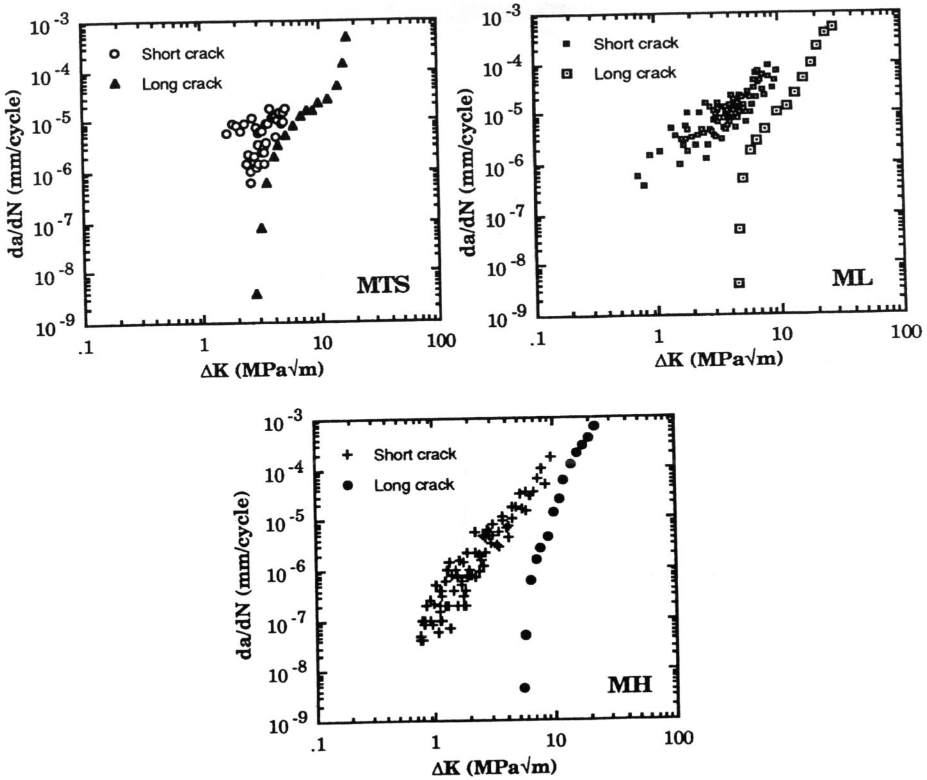


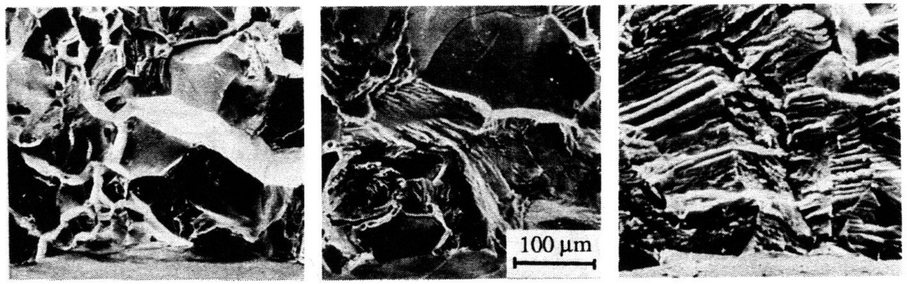
Fig. 2. Comparison of small and long fatigue crack growth for each alloy.

Table 2. Tensile properties and grain sizes of alloys under investigation

Alloy	Yield stress, (MN m^{-2})	Tensile strength, (MN m^{-2})	Fracture strain	Strain hardening coefficient	Grain size, (µm)
MTS	301	323	0.076	0.050	125
ML	318	350	0.244	0.051	100
MH	323	380	0.255	0.062	80

The plastic strain range was estimated using the cyclic stress-strain data given in Edwards and Martin (1982). Long crack ΔJ values were calculated from the formula:

$$\Delta J = \frac{\Delta K^2}{E(1-\nu^2)} \qquad (3)$$



(a) MTS (b) ML (c) MH

Fig. 3. Typical small crack fracture surfaces for each alloy (1<ΔK<3).

RESULTS AND DISCUSSION

S-N curves for the alloys are presented in Fig. 1. It can be seen that fatigue life improves as the dispersoid content increases and slip is progressively homogenized. Conventionally, this effect has been explained in terms of fatigue crack initiation as it is assumed that the larger slip offsets created by inhomogeneous slip provide easy crack initiation sites. However, if fatigue life is essentially controlled by the propagation of small cracks then there must be an effect of slip distribution on small crack fatigue growth.

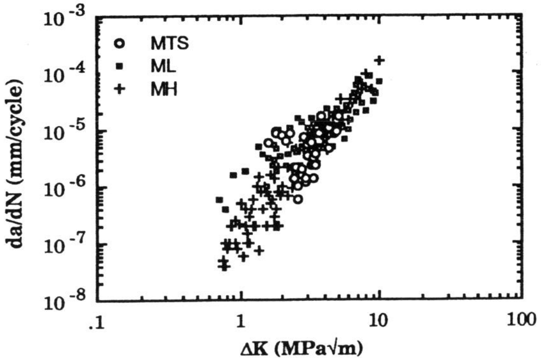


Fig. 4. Comparison of small fatigue crack growth of all alloys.

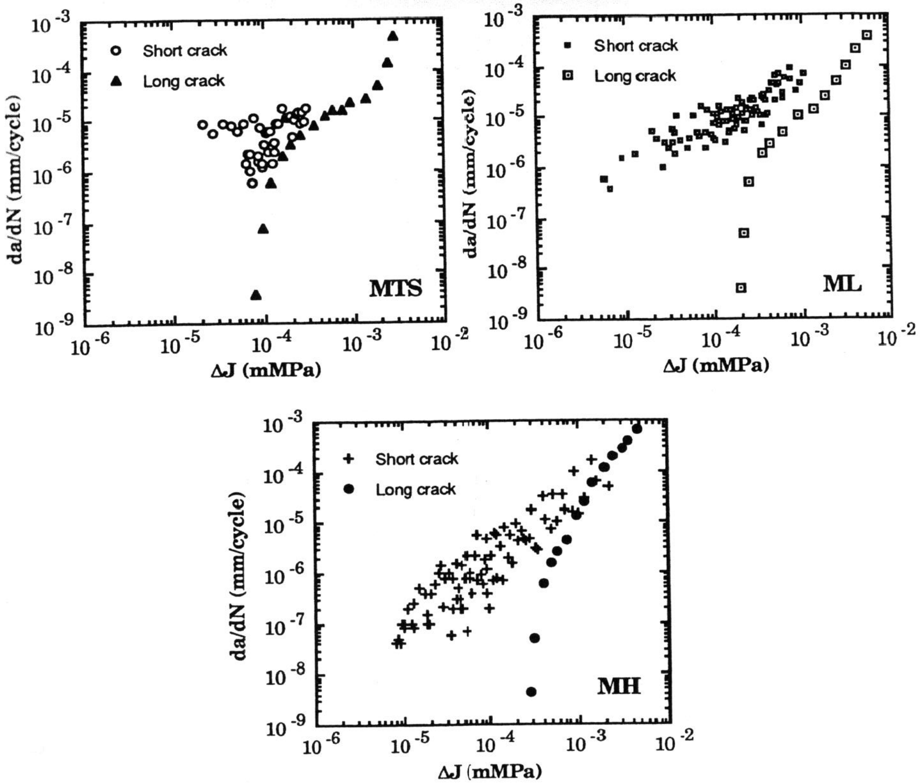


Fig. 5. Comparison of small and long fatigue crack growth using ΔJ.

Small crack growth experiments were initially conducted at stress levels that produced lives of approximately 10^5. Figure 2 compares the growth of long and small fatigue cracks for each alloy when correlated with the LEFM parameter ΔK. In each case, it can be seen that the small cracks grow significantly faster than long cracks. There is increased scatter in the results as you go from MH > ML > MTS. This is due to crack growth becoming more intergranular. The growth of intergranular cracks was much more irregular than that of transgranular cracks. In particular, intergranular cracks slowed when grain boundary triple points were encountered. Transgranular crack growth occurred along slip bands so that all the alloys exhibited faceted fracture surfaces. Striation growth was only seen at higher growth rates (> 10^{-4} mm/cycle) when the cracks were typically ≈ 1 mm in length and >80% of the fatigue life had occurred. The crack growth micromechanisms were similar to those observed in long cracks. As may be seen from Fig. 3, at low ΔK's the degree of intergranular failure decreased as the dispersoid content and hence homogeneity of slip increased. As with long cracks the degree if intergranular failure also increased with ΔK in all three alloys.

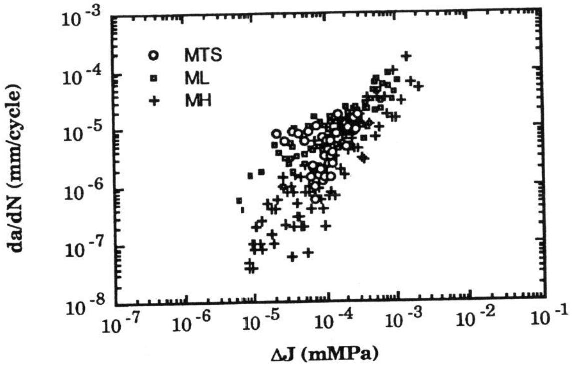


Fig. 6. Small crack growth rate for each alloy compared using ΔJ.

However, as may be seen from Fig. 4 which plots small crack growth from all three alloys, there is no significant difference in propagation rate with dispersoid content and hence slip distribution. This is in sharp contrast to the behaviour of long cracks and despite differences in the micromechanisms of crack growth between the alloys.

Crack growth rates were also correlated using the elastic-plastic parameter ΔJ. Small cracks still grow faster than long cracks as may be seen from Fig. 5 but there is an increase in the scatter of the results for all alloys. This increase in scatter can best be seen by comparing Figures 6 and 4.

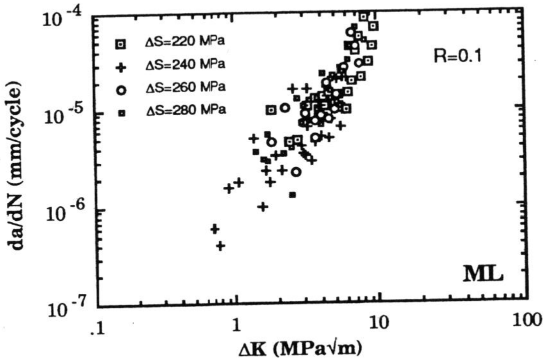


Fig. 7. Effect of mean stress on small fatigue crack growth in alloy ML.

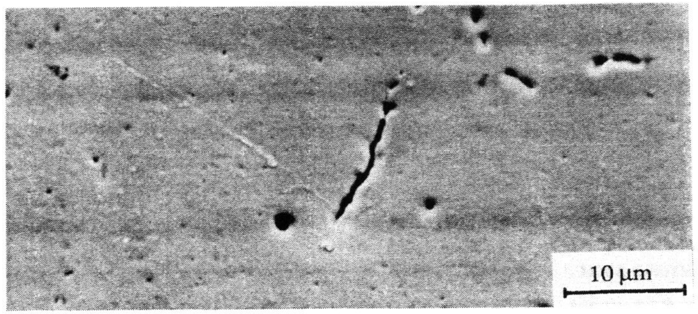


Fig. 8. Fatigue crack nucleated at a coarse particle.

As there is little correlation between long and small cracks it is difficult to tell whether ΔJ provides a more appropriate correlating parameter than ΔK for small fatigue cracks in peak-aged Al-Mg-Si alloys. Although LEFM is often held not to be suitable for use with small cracks, the correlation with ΔK is extremely good. This may be because, due to the low stress intensities associated with small cracks, the calculated plastic zone size was always less than 5% of the crack size. ΔK is also easier to calculate and is potentially of greater use in predicting the lives of engineering components.

The effect of mean stress on growth rate was investigated by further testing the ML alloy at four different stress levels. As may be seen from Fig. 7, mean stress does not appear to affect small fatigue crack growth rates. Thus the improvement in fatigue life with slip distribution in Al-Mg-Si alloys must be due to initiation rather than crack growth.

In all three alloys, cracks initiated from coarse particles (Fig. 8) and subsequent growth dominated the fatigue life of the specimen. The number of cycles required to initiate a crack varied between the alloys. Taking the production of a 50μm crack as 'initiation', this stage took typically 5% of the life in MTS, 25% of the life in ML and up to 80% of the life in MH. Since most fatigue specimens (and engineering components) are too small for long crack growth to significantly affect their fatigue lives, any effect of slip distribution must be due to initiation as no difference in propagation rate could be detected between cracks even smaller than 50μm.

CONCLUSIONS

(1) Small fatigue cracks grow faster than long fatigue cracks in peak-aged Al-Mg-Si alloys.

(2) Both the LEFM parameter ΔK and the elastic-plastic parameter ΔJ successfully correlate small fatigue crack growth in Al-Mg-Si alloys but neither can rationalize the differences in crack growth rates between small and long fatigue cracks.

(3) Slip distribution has no effect on small fatigue crack growth in Al-Mg-Si alloys.

(4) The increase in fatigue life seen with increasing homogenization of slip must therefore be due to the relative ease of fatigue crack initiation in alloys exhibiting inhomogeneous slip.

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