THE INFLUENCE OF DISPERSOIDS ON FATIGUE CRACK PROPAGATION IN Al-Mg-Si ALLOYS

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## ABSTRACT

The fatigue behaviour of three Al-Mg-Si alloys in the peak aged conditon has been studied by means of both S - N curves, and fatigue crack propagation tests. Each of the alloys was free from coarse iron-bearing intermetallic phases, and each had similar values of yield stress and grain size, and showed a similar cyclic stress-strain response. One alloy was essentially of pure ternary composition, the others contained volume fractions of 0.22 and 0.61 respectively of 0.1μm diameter particles of intermetallic dispersoids, due to the addition of manganese. In contrast to previous work in this area, it is found that the dispersoids give rise to a decrease in crack growth velocity over a wide range of value of applied $\Delta K$, and at low $\Delta K$ values the dispersoids also have the effect of changing the fracture path from intergranular to transgranular. The results are interpreted in terms of the dispersoids causing a reduction in the slip band spacing by dislocation interaction with the particles, thus homogenising the slip distribution. This lowers the strain concentration at the head of the slip bands within the crack tip plastic zone.

## KEYWORDS

Fatigue crack propagation; aluminium alloys.

## INTRODUCTION

Commercial heat-treatable aluminium alloys contain, in addition to the coherent or semi-coherent precipitating phase which is the source of strengthening, two types of incoherent particles, namely the coarse (> 1μm) particles linked to the iron impurity content, and fine (∿ 0.1μm) intermetallic particles arising from the addition of the transition elements Cr, Mn, Zr or V. These latter particles are termed 'dispersoids', and have been shown to promote homogenization of slip (Dowling and Martin, 1976), and thereby under conditions of fatigue stressing to inhibit the nucleation of cracks (Lütjering, Döker and Muntz, 1973).

It has previously been shown, however, that the effect of the Cr-bearing dispersoids normally present in commercial 7075 (Al-Zn-Mg) alloys is to accelerate the rate of propagation of fatigue cracks (Albrecht and co-workers, 1976): it was observed that the particles fracture ahead of the crack tip and thus accelerate crack propagation. The present work is concerned with a series of Al-Mg-Si alloys:

one alloy was essentially of pure ternary composition, the remainder contained increasing volume fractions of particles of the $\alpha$-$Al_{12}Mn_3Si$ dispersoid phase. In previous work concerned with the influence of Mn-bearing dispersoids upon the fracture behaviour of these H30-type alloys, Dowling and Martin (1977) observed that particle fracture of the dispersoid phase seldom occurred, even in regions of high local matrix strain. This may be due to the roughly equiaxed shape of the phase, in contrast to the more plate-like morphology of the Cr-bearing phase present in alloys of the 7075-type. The object of the present work has therefore been to examine the effect upon fatigue crack propagation in an Al-Mg-Si alloy (aged to peak hardness) of the presence of such equiaxed dispersoids, in order to establish whether their beneficial effect as slip homogenizers outweighs their possible deleterious effect as crack accelerators.

## EXPERIMENTAL MATERIALS AND METHODS

The compositions of the three alloys are given in Table 1.

TABLE 1 Composition and Structure of the Alloys

| Alloy | wt%Mg | wt%Si | wt%Fe | wt%Mn | Grain size (μm) | Vol. fraction dispersoid(%) |
|---|---|---|---|---|---|---|
| MTS | 0.63 | 1.07 | 0.001 | 0.001 | 125 | - |
| ML | 0.59 | 0.99 | 0.001 | 0.21 | 100 | 0.22 |
| MH | 0.61 | 1.03 | 0.001 | 0.60 | 80 | 0.61 |

The alloys were DC cast, homogenized, and then thermo-mechanically processed to produce comparable grain-sizes in each alloy, as indicated in Table 1. 6.25mm thickness CT test-pieces were machines in the LT orientation to ASTM specification E647, and 5mm thickness plane-bending fatigue test specimens were also machined. For cyclic stress-strain response tests, cylindrical specimens of 5mm diameter and 13.5mm gauge length were prepared.

All specimens were solution-treated at 560°C, and aged to peak hardness at 185°C, and fig.1 illustrates the type of microstructure obtained. All three alloys had similar dispersions of the age-hardening precipitate, and contained similar precipitate-free zones adjacent to the grain boundaries; only ML and MH contained the $\alpha$-$Al_{12}Mn_3Si$ dispersoid - mainly in the form of the 0.1μm particles seen in fig.1, but also as residual particles ( about 1μm in diameter ), there being more in MH than in ML.

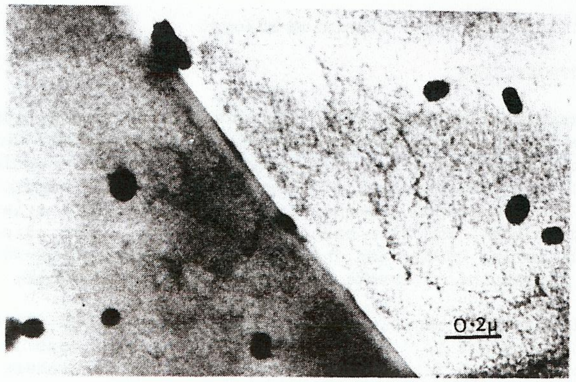


Fig. 1. Microstructure of alloy MH



S - N curves were determined using an Avery plane bending fatigue machine, and ds/dn vs ΔK curves were determined at a frequency of 40Hz and at an R of 0.1 in a servo-hydraulic machine at constant ΔK by a manual load-shedding technique. Crack detection was by an optical method, with a resolution of 0.1mm. The cyclic stress-strain response was investigated by using an incremental step test and also a multiple-step test (Landgraf and co-workers, 1969). In the former, the maximum applied strain was ±0.015, with about 30 strain cycles per block.

## RESULTS AND DISCUSSION

### (a) S - N Curves

The data are shown in fig.2, where it may be seen that the dispersoid-bearing material exhibits an increased life at all stresses. This is in agreement with the work of Lütjering and co-workers (1973) in 2024 alloys, and may likewise be accounted for by the homogenization of slip by the dispersoid particles. The increase in volume fraction of dispersoid in MH compared with ML is not associated with a large increase in fatigue life, although at longer fatigue lives the improvement is more marked.

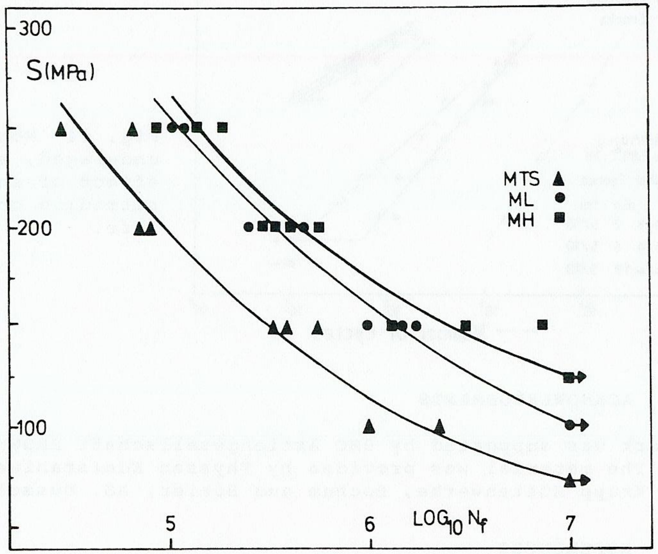


Fig. 2. S - N Curves

### (b) Cyclic Stress-Strain Response

The incremental and the multiple-step tests gave very similar results. In fig.3, therefore, only the multiple-step test data are presented. At all strain levels, each material cyclically stabilized in a similar number of cycles, and hence it may be seen from fig.3 that increasing the volume fraction of the dispersoid increases the "cyclic ductility" of the alloy.

This effect will arise due to the progressive homogenization of slip with increasing volume fraction of dispersoid (Dowling and Martin, 1976). In the

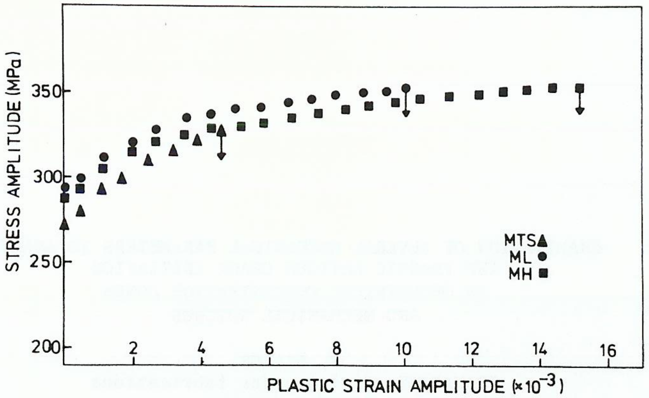


Fig. 3. Cyclic Stress-Strain Response

dispersoid-free material (MTS) the intense slip-bands present will be associated with high local strain concentration, leading to early failure. As the volume fraction of dispersoid increases, slip is progressively homogenized, so that a greater macroscopic strain is required to cause the critical local strain for fracture.

(c) da/dn vs ΔK Curves

The experimental results are summarized in fig.4, which are averaged data for several specimens of each material. Three modes of failure were identified: the two main modes were (a) striation growth (fig.5), which occurs in the reversed plastic zone ahead of the crack tip, and (b) intergranular separation (fig.6), which is a static mode of failure, and hence occurs in the monotonic plastic zone ahead of the crack tip. The mechanism of failure is by ductile rupture along the grain boundary precipitate-free zone, and so the intergranular surfaces may be seen to be covered in fine (∿ 0.1μm) dimples (fig.7). The third mode of failure was also a static mode process, involving ductile failure around coarse intermetallic particles (fig.8).

We will consider the curves of fig.4 in three regimes of growth rate.

Low growth rates. Here it may be seen that the ΔK threshold increases with volume fraction of dispersoid in the alloys. The failure mode in this regime in predominantly transgranular, although the ternary alloy (MTS) also shows some intergranular failure. At these low stress intensities, the crack growth rates may be considered to be controlled by the cyclic stress-strain response, and so the increasing cyclic ductility shown in fig.3 with increasing dispersoid content is reflected in the increasing threshold values indicated in fig.4.

Intermediate growth rates. In this regime, the tendency for static mode failure ahead of the crack tip increases in all three alloys. In the dispersoid-free material (MTS) the degree of intergranular failure increases markedly with increasing stress intensity range. In the dispersoid-containing alloys this tendency is suppressed to a large degree, the fracture being predominantly transgranular. Thus the low volume fraction dispersoid-containing alloy (ML), which has a low number of coarse residuals, exhibits a lower growth rate at all ΔK values. In MH, however, a large increase in growth rate is seen when ΔK exceeds 10 MPa m$^{\frac{1}{2}}$. This is associated with ductile failure, arising from the coarse



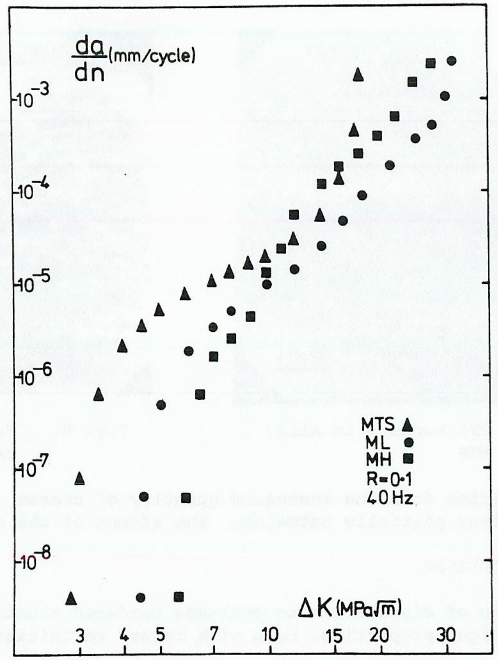


Fig. 4. da/dn vs ΔK Curves

residual particles, and it reflects the higher volume fraction of these particles in this alloy.

High growth rates. In this regime, all the alloys fail predominantly by static processes. Alloy MTS is wholly intergranular in fracture, giving a final low-toughness fracture behaviour. ML shows the lowest growth rate, which can be accounted for by the effect of the dispersoids on crack-tip processes, as has been described by Prince and Martin (1979). In MH, although the volume fraction of dispersoids is higher, there is no corresponding decrease in growth rate. This is


Fig. 5. Striation growth in Alloy MH

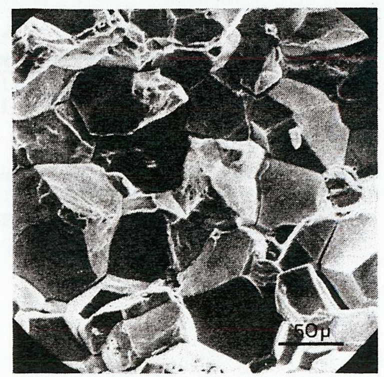


Fig. 6. Intergranular separation in Alloy MTS

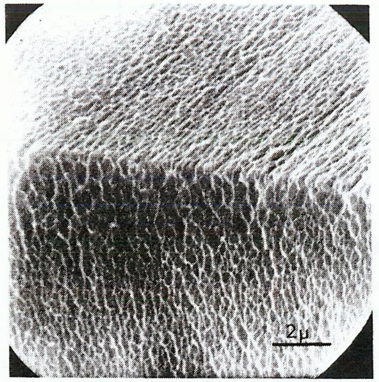


Fig. 7. PFZ rupture in Alloy MTS

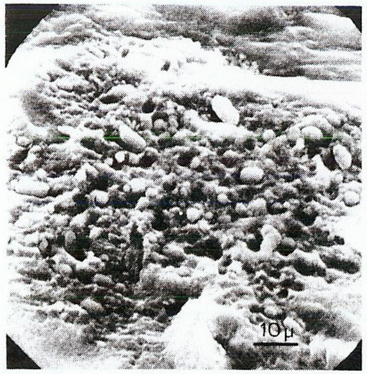


Fig. 8. Failure around coarse residuals in Alloy MH

considered to arise from the increased quantity of coarse (∿ 1μm) particles whose deleterious effect partially outweighs the effect of the dispersoids.

## CONCLUSIONS

1. The addition of dispersoids to peak-age hardened aluminium alloys can result in improved fatigue properties, both with regard to initiation and propagation of cracks.

2. The deleterious effect of coarse residual particles upon crack propagation can, if present in sufficient quantity, outweigh the beneficial effect of the dispersoids.

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