The Effect of Different Dispersoids on the Ductile Fracture Toughness of an Al–Zn–Mg Alloy

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ABSTRACT

A comparison has been made of the effect of incoherent (Mn-bearing) and semi-coherent (Zr-bearing) dispersoid phases on crack tip deformation processes during fracture toughness testing of a weldable (copper-free) Al-Zn-Mg alloy in the peak-aged condition. A third, dispersoid-free, alloy has also been studied, and all the alloys were processed to possess comparable grain sizes and yield stresses.

Fracture toughnesses were obtained from COD measurements, where it was found that the Mn-bearing alloy had significantly lower toughness than the other two. In all the alloys the crack extended intergranularly during the formation of the stretch zone, and the crack tip sharpness was thus independent of the degree of slip homogenization within the grains (in contrast to the behaviour of Al-Mg-Si alloys (Blind and Martin, 1983)). Crack tip plastic zone sizes have been measured as a function of applied stress intensity, and when allowances are made for crack orientation, the data for all the alloys correlate well with theoretical prediction.

It is concluded that crack propagation is associated with microvoid coalescence associated with incoherent grain boundary phases. In the Mn-bearing alloys the incoherent dispersoids are more closely spaced than the incoherent eta-phase, and this accounts for the reduced fracture toughness in that alloy. The semi-coherent Zr-bearing dispersoids have no deleterious effect.

INTRODUCTION

Commercial heat-treatable aluminium alloys contain dispersoids, which are fine particles of intermetallic phases arising from the addition of a transition element such as Cr, Mn, Zr or V. Their prime function is to control grain growth during heat-treatment. Cr- and Mn-bearing dispersoids have also been shown to promote homogenization of slip (Lutjering et al., 1973; Dowling and Martin, 1976) which, together with the refined grain size, increases the tensile ductility and also the fracture toughness (Prince and Martin, 1979; Blind and Martin, 1983). The ductile fracture toughness of Al-Mg-Si alloys aged to peak hardness is found to increase with increasing volume fraction of the incoherent Mn-bearing dispersoid phase. This increase is attributable to the homogenization of slip distribution by the dispersoid phase leading to blunting of the crack tip.

Little systematic work has been carried out in this area on the medium-strength Al-Zn-Mg alloys, and the object of the present work has been to compare the effect upon crack propagation of a semi-coherent (Zr-bearing) and a non-coherent (Mn-bearing) dispersoid phase in this alloy system. Only the latter dispersoid would be expected to give rise to slip homogenization; a ternary (dispersoid-free) alloy was studied for comparative purposes.

EXPERIMENTAL

The alloys were prepared from high purity (low Fe) material in order to minimize the occurrence of coarse residual particles, whose effect might be to swamp that of the dispersoids themselves. The composition of the alloys is given in Table 1.

TABLE 1: Composition of the alloys.

ALLOY	Wt % : Cu	Fe	Mg	Mn	Si	Ti	Zn	Zr
CTE	0.01	0.004	1.71	0.001	0.004	0.001	5.43	-
CMH	0.001	0.007	1.66	0.42	0.08	0.001	5.47	-
CZH	0.001	0.01	1.64	0.004	0.01	0.001	5.58	0.15

All the alloys were subjected to thermo-mechanical treatments such that the grain structures were equiaxed and recrystallized. They were given a common solution-treatment at 475°C, followed by water-quenching and 24 hr ageing at room temperature. Alloys CTE and CMH were then artificially aged at 130°C for 60 Hr, and alloy CZH was aged at 120°C for 210 Hr, which corresponded to peak hardness. The yield stresses obtained from the average of several tensile tests, and the grain sizes of the alloys, together with the dispersoid phase distribution parameters are given in Table 2.

Table 2: Grain sizes, yield stresses and dispersoid characteristics

Alloy	Grain Size (μm)	Yield Stress (MPa)	Mean Dispersoid Size (μm)	Dispersoid Vol. Fraction
CTE	105	389	-	-
CMH	101	388	0.05	8×10^{-3}
CZH	80	394	0.04	1.2×10^{-3}

Each alloy had similar dispersions of the ageing precipitate within the grains, similar precipitate-free zones of width 0.5 μm at the grain boundaries, and similar dispersions of the incoherent eta-phase in the grain boundaries.

Compact Tension (CT) specimens were machined in the LT direction, so that the crack plane was perpendicular to the rolling direction. The specimen thickness was 6.35 mm, with a ratio of width (W) to breadth (B) of 4, so that plane strain conditions were satisfied. After heat-treatment the CT specimens were precracked according to the ASTM standards with a final ΔK of

11 MPa $\sqrt{m}$. A clip gauge was attached to the precracked specimen via knife-edges and connected to the strain gauge amplifier of the testing machine.

COD tests were performed according to BS 5762 (1979) at a constant crosshead speed of 0.5 mm/min. The critical displacement value corresponding to the maximum applied force was taken at pop-in, or at the discontinuity on the graph when no pop-in occurred. The elastic (δ_e) and the plastic (δ_p) parts of the total opening (δ_t) were calculated from the expressions:

$$\delta_e = K^2(1 - \nu^2)/\ 2\,\sigma_y\, E$$

$$\delta_p = 0.4\ (W - a)\ V_p/\ (0.4W + 0.6a + Z)$$

where ν is Poisson's Ratio, σ_y is the yield stress, E Young's modulus, W the specimen width, a the initial crack length, Z the distance of the clip-gauge location from the specimen surface and V_p clip-gauge displacement

Plastic Zone Size Measurements.

With each alloy a series of four interrupted tests were conducted by unloading the specimen after the application of different, known values of load. The unloaded specimens were sectioned on their mid-plane by spark-machining, ground and electropolished. The crack tip plastic zones were mapped by the use of electron channelling patterns (ECPs) in the SEM (Tekin and Martin, 1988; Davidson, 1984).

RESULTS AND DISCUSSION

Fracture Toughness Parameters.

The average values of the total COD are given in Table 3. The COD values have been converted to J_{Ic} using the relation:

$$J_{Ic} = M\,\delta_t\,\sigma_y$$

where M, the constraint factor is a material-dependent parameter which has been taken as 1.3 (Shih, 1981). This gives a calculated values of K_Q close to those derived from the COD graphs using the 5% secant line (see Table 3).

Table 3: The fracture toughness of the alloys.

Alloy	COD μm	J_{Ic} N/mm	K_Q (Calc) MPa$\sqrt{m}$	r_p crit. (calc.) μm	K_Q (Measured) MPa$\sqrt{m}$
CTE	50.9	25.7	45.0	435	44.4
CMH	37.2	18.8	38.4	319	40.3
CZH	52.4	26.8	45.9	441	44.2

It is apparent that the ternary alloy CTE and alloy CZH, which contains the semi-coherent zirconium dispersoid phase, have closely similar toughnesses, whereas the effect of the non-coherent manganese dispersoid is to bring about a reduction in toughness.

This behaviour may be contrasted with that observed by Blind and Martin (1983) in a series of Al-Mg-Si alloys containing increasing volume fractions of Mn-bearing dispersoids. Here it was found that the toughness increased with increasing volume fraction of dispersoid, and this effect could be accounted for by the progressive homogenization of slip by the increasing volume fraction of dispersoid phase. This effect was apparent from measurements of the crack tip plastic zone size as a function of applied stress intensity, when it was found that the rate of increase in plastic zone size with applied stress intensity decreased as the volume fraction of dispersoid increased. The effect of increasing slip homogenization was to increase the crack tip radius, and hence to decrease the plastic zone size for a given COD. It was therefore decided to examine the crack tip plastic zone as a function of applied stress intensity in the present series of alloys.

Plastic Zone Size Measurements.

Fig.1 illustrates a set of crack tip plastic zone maps obtained from ECP examination of the three alloys. It was apparent that the initial crack (typically 15 mm in length) had extended intergranularly by a few tens of micrometres during stretch-zone formation, so that in general the crack tip profile was deflected from the Mode I orientation. This effect is illustrated for the three alloys in fig.2, and this contrasts with the Al-Mg-Si alloys of Blind and Martin (1983) in which the crack tip blunted without extending intergranularly.

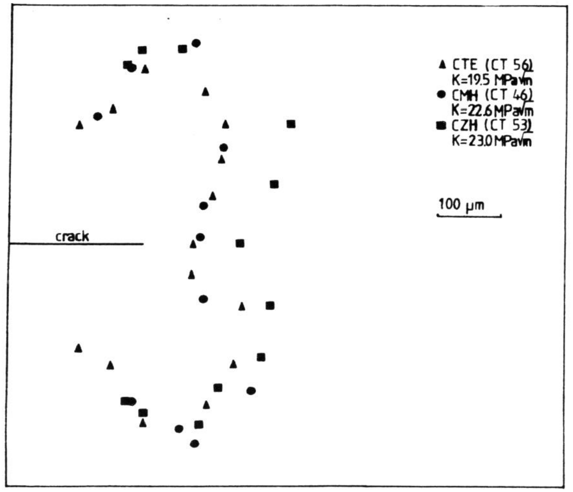
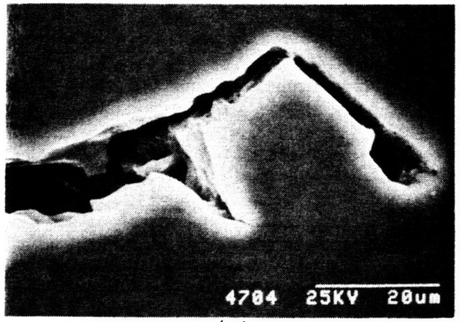
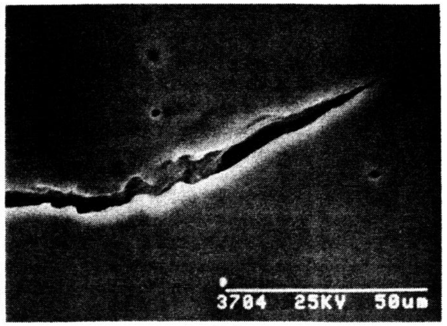


Fig.1. Crack tip plastic zone maps for the three alloys obtained from Electron Channelling Patterns.



(a)



(b)

(c)

Fig.2. SEM images of the crack tips seen in mid-thickness sections of (a) CTE, (b) CMH and (c) CZH. The specimens had been loaded to below the critical stress intensity, and intergranular crack extension with microvoid coalescence is evident.

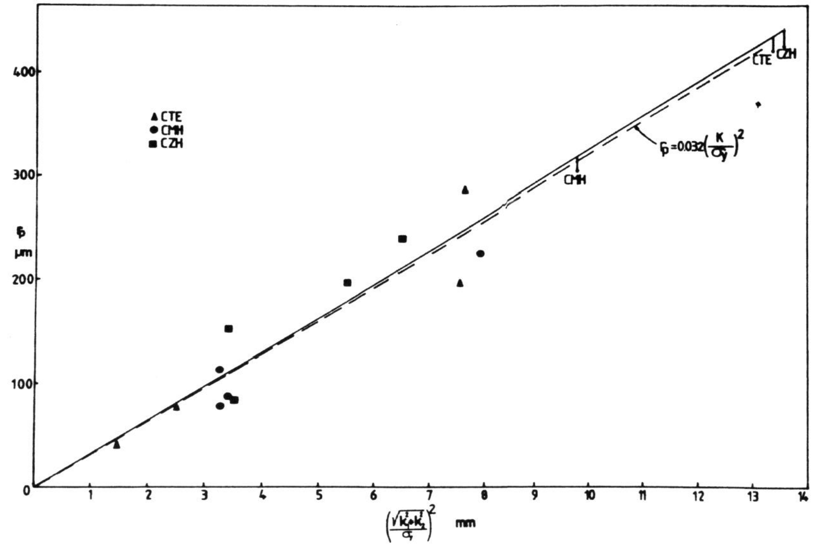


Fig. 3. Crack tip plastic zone sizes (r_p) as a function of $(K_{eff}/\sigma_y)^2$. The dotted line represents the theoretical slope.

When the crack tip is deflected through an angle θ , we can write:

$$k_1 = K_I \cos^2\theta \quad \text{and} \quad k_2 = K_I \sin\theta \cos\theta$$

where K_I is the applied stress intensity, and k_1 and k_2 are the tensile and shear stress intensity factors at the crack tip. The effective stress intensity factor may therefore be written (Suresh, 1983):

$$K_{eff} = (k_1^2 + k_2^2)^{1/2}$$

In fig.3 we have plotted the crack tip plastic zone sizes measured ahead of the crack tip (r_p) against the function $(K_{eff}/\sigma_y)^2$. It can be seen that all the data points for the three alloys fall in a narrow band. Also shown are the critical values of r_p for each alloy obtained by extrapolation of the experimental line from the calculated K_Q values shown in Table 3. The dotted line in Fig.3 represents :

$$r_p = 0.032\ (K/\sigma_y)^2$$

which is the Rice and Johnson (1970) solution for the plastic zone size ahead of the crack tip based on the von Mises yield criterion, and close agreement is apparent.

The Micromechanisms of Fracture.

We have shown that the intergranular extension of the crack during the stretch-zone formation leads to the crack tip sharpness being independent of the degree of slip homogenization, and thus to the presence and type of dispersoid phases. However, a significant decrease in toughness was observed in alloy CMH in comparison with alloys CTE and CZH.

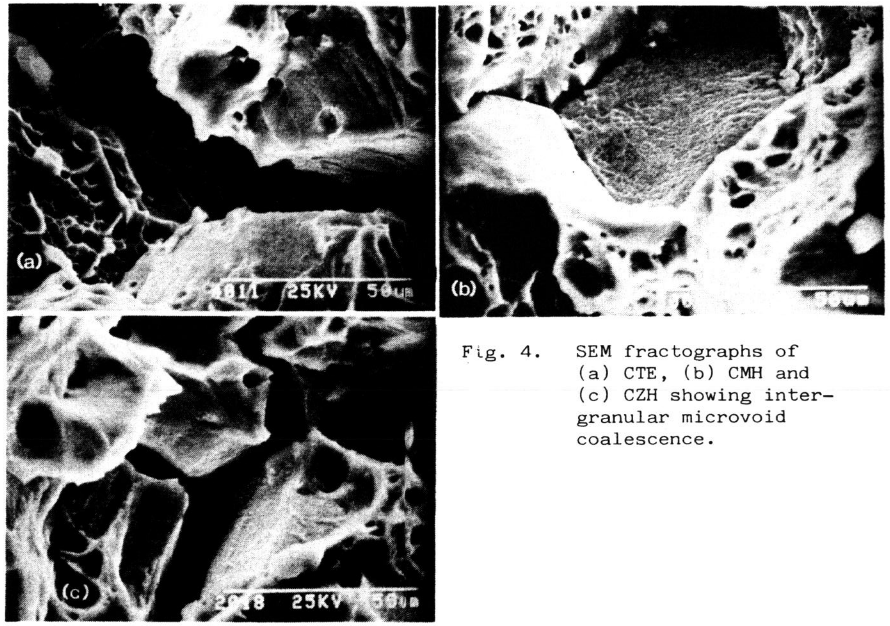


Fig. 4. SEM fractographs of (a) CTE, (b) CMH and (c) CZH showing intergranular microvoid coalescence.

Fig.4 shows the fracture surfaces obtained in the three alloys, where it is clear that intergranular failure predominates, with prolific subsidiary intergranular cracks. There is also evidence of considerable local microvoid coalescence associated with second-phase particles, and this is also apparent within the cracks of fig.2.

Hahn and Rosenfield (1975) suggest that in ductile rupture processes, the critical COD will be proportional to the mean spacing of the second phase particles responsible for the microvoid coalescence process. From the data of Table 2, the mean planar interparticle spacing of the dispersoid phases may be readily calculated, and this is found to be 0.5 μm in CMH and 0.9 μm in CZH. The COD values of the dispersoid-free alloy CTE and alloy CZH are almost identical, which suggests that the semi-coherent Zr-dispersoid particles do not readily nucleate microvoids, but that the fracture process is associated with cavitation at the incoherent eta phase particles present in the grain boundaries in all the alloys. The mean size of the grain boundary eta-phase particles is 0.2 μm, and estimating their volume fraction as 0.06, a mean planar spacing of 0.7 μm is obtained which is consistent with the metallographic evidence.

The ratio of the CODs of CMH to that of CTE or CZH is about 0.71. The ratio of the mean planar particle spacings of the dispersoids in CMH and the eta particles in CTE and CZH is about 0.73, which is consistent with the Hahn and Rosenfield model.

CONCLUSIONS

1. In COD testing, the starting crack extends intergranularly during the formation of the stretch zone. The crack sharpness is thus independent of the degree of slip homogenization within the grains.

2. The crack tip plastic zone sizes as a function of applied stress intensity are similar in all the alloys, and their magnitudes agree well with those theoretically predicted when allowances are made for the local orientation of the crack tip.

3. The fracture toughnesses of the ternary (CTE) and zirconium-bearing (CZH) alloys are closely similar, but that of the Mn-bearing alloy (CMH) is lower. This is attributed to the incoherent dispersoid phases promoting microvoid formation in CMH. The semi-coherent dispersoids in CZH do not appear to act in this way, and the toughness in that alloy and in the ternary alloy appears to be related to the volume fraction of the intergranular incoherent eta-phase.

4. In this system, therefore, it is suggested that grain size control is better achieved by the presence of semi-coherent dispersoid phases such as those formed from Zr-additions, rather than by the addition of incoherent dispersoids which may promote microvoid formation.

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