INFLUENCE OF SILICON ON THE CYCLIC CRACK GROWTH IN A CAST ALUMINIUM ALLOY

L. E. CULVER, J. C. RADON and J. C. BALTHAZAR*

A fractographic study conducted on the fatigue cracked surfaces of BS 1490 LM-30 (U.S. A390) aluminium-silicon alloy revealed that the silicon-rich constituent determines the mode of fracture. The fracture mechanism was basically of a mixed cleavage-tearing type which consisted of cleavage of the silicon particles ahead of the crack front followed by tearing of the remaining ligaments of the aluminium-rich phase left in the wake of the crack. The fractured surfaces resulted from fatigue tests run at frequencies of 50 and 0.25 Hz, and at temperatures of 20 and 250°C. The ASTM compact specimens used were loaded at stress ratios, R, of 0.08, 0.5 and -1.0.

INTRODUCTION

Fracture mechanics concepts have been increasingly applied to engineering design in recent years but insufficient attention has been given to cast aluminium alloys despite their substantial use in structural applications where their high strength-to-weight ratio is advantageous. Components made from these alloys are often subjected to cyclic loading, the safe limits having been previously established by the usual range of fatigue testing. However, the results of such testing give little information about the mode of failure of the components or the effect on the mode of failure of factors such as the microstructure of the alloy and casting defects. The specific fracture mechanism of one micro-constitutent may change as a consequence of the fracture mode which is occurring in another one. Fracture mechanisms may change from ductile to cleavage or from intergranular to transgranular as a result of environmental conditions, changes in the microstructure, heat treatments, inclusions, alloying elements and other factors. It is essential to understand the features which characterise the fracture mechanisms caused by specific loading conditions if industry is to pursue improvements of cast components and refinements in design criteria.

Striation formation is generally accepted as the predominant mechanism of fatigue crack propagation (1,2), the formation of these striations being related to the intensive plastic deformation taking place around the crack tip during each load cycle. However, recent evidence of much more complicated processes in fatigue cracking has been described by Gerberich and Moody (3). They identified at least ten microscopic fracture modes of fatigue crack propagation as alternatives to ductile striations. McEvilly (4) and Beevers (5) have also shown that fatigue cracking is strongly dependent on stress intensity, microstructure, environment and other factors. In Al-Mg-Si alloys,

* Department of Mechanical Engineering, Imperial College, London

Edwards and Martin (6) found that the presence of dispersoids had the effect of changing the fracture path from intergranular to transgranular, while Ruch and Gerold (7) reported that microstructural features like second phase particles, small dispersoids and strengthening precipitates could markedly influence crack initiation and crack propagation.

A number of studies has been made on the most commonly used cast aluminium-silicon alloys (6-9) which are generally hypoeutectic alloys, i.e. alloys containing less than 12.6% Si. Published information on hypereutectic Al-Si alloys is very limited. Nevertheless, it may be inferred from the available information that the cracking of silicon particles appears to have a fundamental role in both the mode and the path of fracture under monotonic loadings. The occurrence of casting defects may also influence the mode of fracture but their effects on crack initiation, fracture path or energy required to fracture are not clearly established.

The present work reports an investigation on a cast hypereutectic aluminium-silicon alloy which has recently been in use in the automotive industry. Compact specimens were tested under fluctuating tensile and reverse loads at temperatures of 20 and 250°C producing the samples on which the subsequent fractographic study was conducted.

EXPERIMENTAL PROCEDURE

The material which has been examined is the BS 1490 LM-30 aluminium-silicon alloy. This alloy has excellent casting properties due to its high silicon content and the wear resistance is comparable to that of cast iron. It was originally designed as a substitute for gray cast iron in automobile cylinders where its wear resistance would allow the cylinder liners to be dispensed with. It may also be used in other applications where good wear properties associated with lightness are required. Its chemical composition is given in Table 1 and its similarity to the American alloys 390 and A390, Jorstad (10) will be noticed.

TABLE 1 - Chemical Composition of LM-30 (% by weight)

El.	Si	Cu	Fe	Mg	Mn	Zn	Ti	Ni	Pb	Sn	Al
%	16-18	4-5	1.1	0.4-0.7	0.3	0.2	0.2	0.1	0.1	0.1	bal.

Typical mechanical properties of LM-30 in the "as cast" condition at 20 and 250°C, Table 2, show that hardness, tensile and yield properties are comparable to other commonly used cast aluminium alloys. The ductility is very low but, this is not considered a disadvantage as most of its potential applications are as a replacement for cast irons.

TABLE 2 - LM-30 Typical Mechanical Properties

Temp °C	Casting Method	UTS Mpa	YS Mpa	Elong. 2"G.L.	Hard.	Fat. Strength* Mpa
20	Die	239	231	<1	120 VHN	100
250	Die	156	155	<1		50

* - $5x10^8$ cycles

Details of the testing machines, instrumentation, specimens and test procedures have been described elsewhere, Culver et al (11). The fracture surfaces were examined using standard techniques of electron microscope fractography in a JEOL scanning electron microscope, in which an accelerating potential of 25 kV was applied to study randomly selected areas.

MORPHOLOGY OF THE ALLOY

The microstructure of LM-30, Figure 1, basically contains approximately 5% of primary silicon embedded in a ternary eutectic matrix which comprises 87.4% aluminium, a silicon-rich solid solution and thin platelets of the phase AlCuFeSi, which is a solid solution of the inter-metallic compounds FeSiAl5 and CuFeAl7, Mondolfo (12). The primary silicon crystals begin to precipitate from the liquid solution at about 650°C (liquidus temperature) and without refinement would grow comparatively large and irregular in shape causing a decrease in the tensile properties of the alloy. The refinement is usually achieved through the addition of phosphorus to the melt which combines with aluminium to form insoluble aluminium phosphide (AlP) which, due to its similar crystal habit and lattice constants, acts as suitable nuclei for primary silicon crystal formation (10).

The general macroscopic appearance of the fracture surfaces inspected was brittle. The preferential path of fracture was through the silicon particles either in the eutectic matrix or the primary precipitate, Figure 2. This trend was followed for fatigue crack growth at 20 or 250°C and under either monotonic or cyclic loading.

ROLE OF THE ALUMINIUM-RICH CONSTITUENT

The aluminium-rich constituent generally failed by ductile fracture, Figure 3, after extensive plastic deformation in the characteristic mode of a tearing mechanism. The occurrence of tearing is accompanied by the formation of sharp tear ridges which produces a bright contrast in the SEM image. The extensive plastic deformation of this constituent prior to fracture will absorb energy and tends to reduce the rate of fatigue crack propagation. This fracture mechanism is a consequence of the brittle behaviour of the silicon particles, which crack by cleavage ahead of the crack front, leaving small areas unbroken which will then fracture by tearing.

Although tearing had been the main fracture mechanism observed in the aluminium-rich constituent, patches of microvoid coalescence were observed at 250°C in the Paris regime of fatigue crack propagation, Figure 4, where cleaved facets are separated by arrays of dimples. This mechanism was also observed, but to a much lesser extent, in the threshold regime. Striations occurred mainly at 250°C for a stress ratio R=-1.0, Figure 5, but were irregular in their form. For the same conditions, reducing the frequency from 50 Hz to 0.25 Hz caused finer striations, though they were still irregular in shape and distribution, Figure 6.

ROLE OF THE SILICON-RICH CONSTITUENT

The silicon-rich constituent, whether the primary silicon or in the eutectic phase, fractured by cleavage on crystallographic planes giving rise to the smooth-faceted regions on the fracture surface, Figure 7. This appears to be the determinant fracture mechanism in this alloy and is supported by the observation that the preferred path of crack propagation was through the silicon particles. The cleavage facets change their orientation from one grain to another leading to crack branching along different planes and this is the cause of the very irregular appearance of the fracture surface.

Poniewierski and Linkowski (13) observed, in a hypoeutectic aluminium-silicon alloy, that the silicon particles, due to their brittle characteristic, fractured before the aluminium constituent and established the pattern for the whole fracture process.

DISCUSSION

Detailed information on the fatigue fracture characteristics of cast aluminium alloys is very limited. Some information about fracture under monotonic loading for hypoeutectic alloys is available (6,8,9,13,14) and the determinant role of the silicon particles in establishing the pattern of the mode of fracture is clear. The present investigation confirms that this role is very likely to have the same importance in hypereutectic alloys under cyclic loading.

The intermetallic compound AlCuFeSi which fails by cleavage, and which in hypoeutectic alloys has the effect of accelerating the crack growth, Ogilvy and Robinson (14), probably has its effects relatively diminished in hypereutectic alloys due to the massive presence of silicon through which the crack propagates preferentially. In LM-30 no major effect on this compound was observed.

Casting defects such as shrinkage porosity were observed in the crack path, Figure 8. Beyond reducing the effective load bearing area and affecting fatigue crack growth rates, Culver et al (15), they did not show any significant effect on the fatigue cracking mechanisms of LM-30. They could, perhaps, have a more active role in crack initiation but that was not the object of this study.

The extensive plastic deformation followed by ductile fracture of the aluminium-rich constituent probably has the effect of moderating the crack growth rate which is stimulated by the cleavage of the silicon-rich solid solution.

CONCLUSION

The mechanisms of fatigue crack propagation in the LM-30 aluminium-silicon alloy have been investigated and it has been shown that the silicon precipitate determines the mode of fracture of such an alloy. The general fracture mechanism did not change essentially with temperature between 20 and 250°C and consisted basically of a mixed cleavage-tearing mechanism with cleavage occurring in the silicon-rich constituent and ductile fracture by tearing in the aluminium-rich constituent.

REFERENCES

1. McIntyre,D., J.Eng.Mat.Tech. - Trans. ASME, 97, (1975) 194
2. Engel,L. and Klingele,H., "An Atlas of Metal Damage", (Wolfe Pub., London, 1981)
3. Gerberich,W.W. and Moody,N.R., in "Fatigue Mechanisms",(ASTM-STP 675, 292, 1979)
4. McEvily,A.J., Metal Science 11, (1977) 274
5. Beevers,C.J., Metal Sciences 11, (1977) 362
6. Edwards,L. and Martin,J.W., "Proc. 5th Int. Conf. on Fracture", (Cannes, 1981, 323)
7. Ruch,W. and Gerold,V., "Proc. 4thE.C.F. Conf.", (Leoben-Austria, 1982, 383)
8. Stofanak,R.J., Hertzberg,R.W., Leupp,J. and Jaccard,R., Engng. Fract. Mechanics, 17, (1983) 541
9. Saunders,D.S., "Proc. of Conf. of the Australian Fracture Group", Melbourne, 1974, 24)
10. Jorstad,J.L., Trans. Met. Soc. AIME, 242 (1968) 1217
11. Culver,L.E., Balthazar,J.C. and Radon,J.C., "Proc. 4th E.C.F. Conf.", (Leoben-Austria, 583, 1982)
12. Mondolfo,L.C., "Metallography of Aluminium Alloys", (Wiley, New York, 1943)
13. Poniewierski,Z. and Linkowski,J., Archiwum Hutnictwa, 25, (1980) 465
14. Ogilvy,I.M. and Robinson,P.M., "Proc. 8th SDCE Int. Die Casting Cong" (Detroit, 1, 1975)
15. Culver,L.E., Balthazar,J.C. and Radon,J.C., Trans. CSME, in print, (1983)

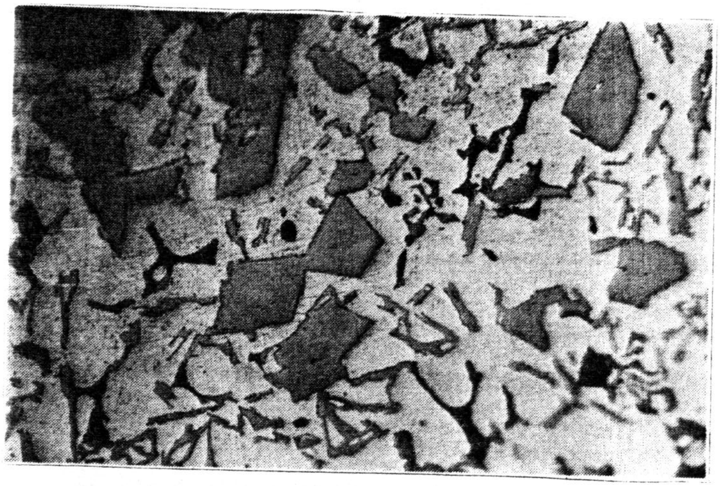

Figure 1 - Microstructure of LM-30,
HF etching, Mag. 200 x.

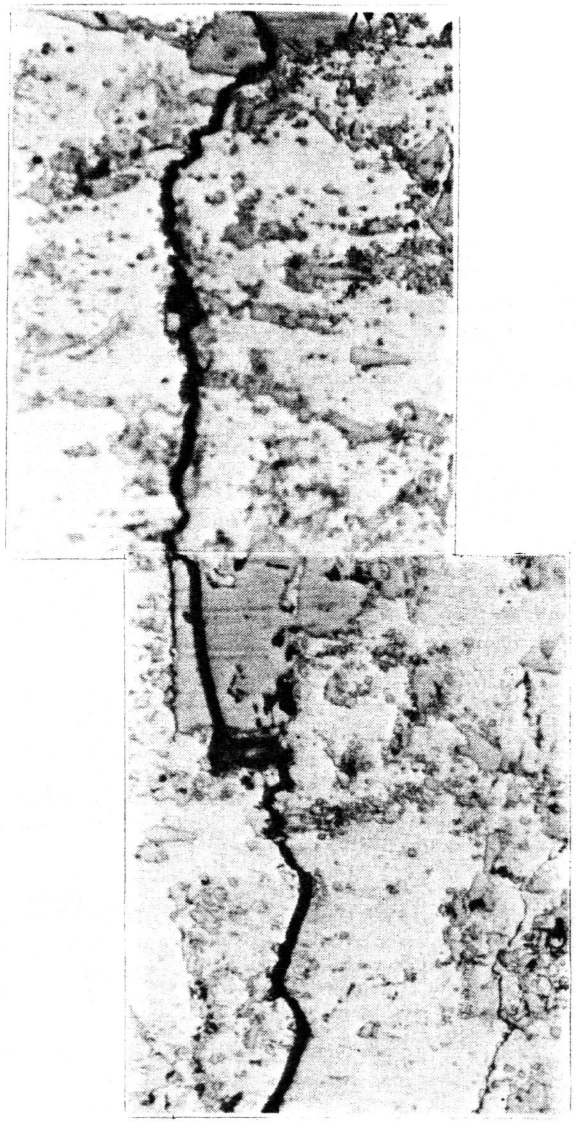

Figure 2 - Optical Micrograph Illustrating a Typical Example of Fatigue Crack Path on LM-30

Figure 3 - Fracture by cleavage in platelets of the intermetallic compound AlCuFeSi.
(1500x)

Figure 4 - Ductile fracture by microvoid coalescence of the
aluminium-rich constituent. (1000x)

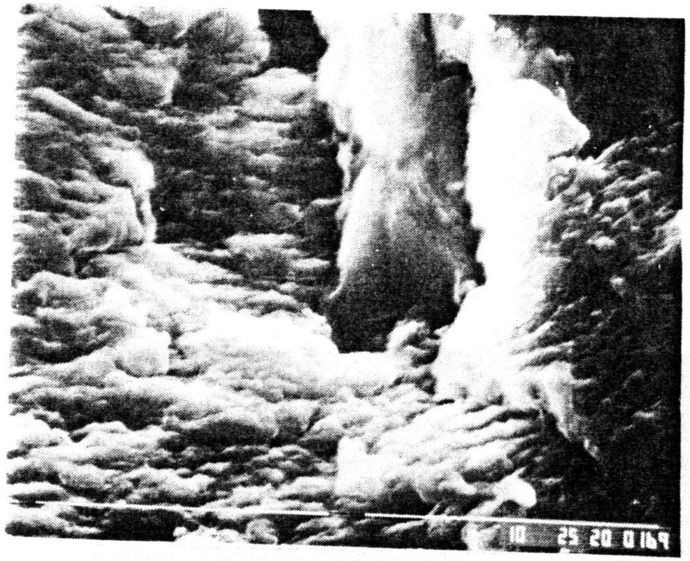

Figure 5 - Striation formation in the aluminium-rich constituent at high frequency. (5000x)

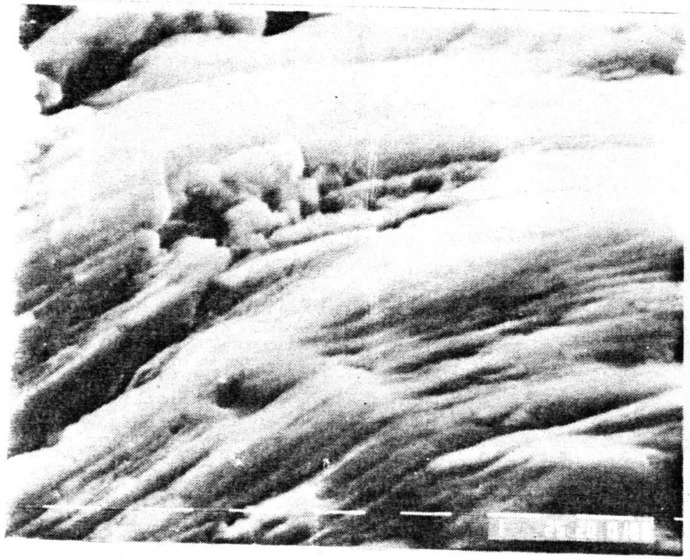

Figure 6 - Fine striation formation in the aluminium-rich constituent at low frequency (10000x)

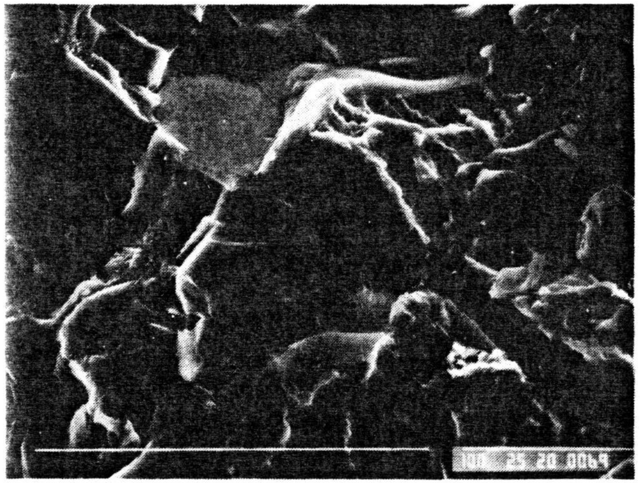

Figure 7 - Cleavage fracture of the silicon-rich constituent. (750x)

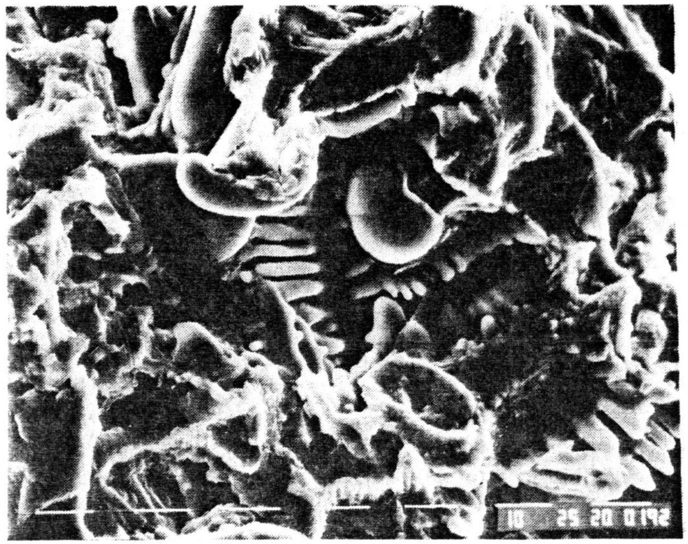

Figure 8 - Fracture across a region of porosity originated in the casting process. (1000x)