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Fatigue performance of a cast aluminium alloy Al-7Si-Mg with surface defects

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Fatigue performance has been studied in a cast Al-7Si-Mg alloy with regard to various casting defects, particularly surface or subsurface defects. Fatigue tests were carried out under four-point bending to investigate the stress-life relationship and fatigue fracture characteristics in order to understand crack initiation and growth behaviour in conjunction with the examination of surface roughness and porosity. Surface hollows were found to control crack initiation of as-cast specimens. The fracture surfaces of polished specimens revealed that surface or subsurface shrinkage pores replaced the hollows to act as crack initiators when the rough surface was removed. The importance of oxide films in crack initiation was also demonstrated. The effects of all these casting defects on fatigue life are discussed.

1. Introduction

The need for lighter, more fuel-efficient automobiles has instigated research into the replacement of fatigue-critical steel components with aluminium equivalents. An Al-7Si-Mg (LM25) alloy is a good candidate material for application in cast components in the automotive industry because of its low density, good casting properties and treatability. One problem in aluminium casting is that various casting defects, such as surface hollows or pores, are likely to be present using commercial metallurgical techniques. A further type of defect, not always recognized, is entrained oxide film, which results from turbulent mould-filling [1]. These defects are invariably detrimental to the fatigue properties [2–11]. For example, an increase in the degree of porosity can reduce the fatigue strength by 17% [2]. The surface roughness can also significantly reduce the fatigue strength [6–11]. Furthermore, in the presence of casting defects, the traditional fatigue design approach, based on the use of a material's stress-life (S-N curve), becomes unsuitable because the traditional approach has the inbuilt assumption that the materials are defect free. As most fatigue failures nucleate at the surface of a material, casting defects at or near the surface become an extremely important factor in determining the fatigue strength of cast components. For cast materials with inherent defects, the introduction of a damage-tolerant design approach based on crack initiation and growth and the establishment of risk levels is more appropriate. The application of such a new approach requires detailed characterization of defects (nature, position, size and shape), and statistical modeling of the defect size distribution and crack growth behaviour with particular emphasis on cracks initiating from these defects.

As part of an ongoing research programme on the study of probabilistic approaches for fatigue design and optimization of cast aluminium structures, the present work focuses on the effects of as-cast surface (high level of roughness), pore defects and oxide films on the fatigue performance of a cast Al-7Si-Mg alloy, in particular on fatigue crack initiation and small crack growth.

2. Experimental procedures

The material under investigation is a standard Al-7Si-Mg aluminium alloy (LM25) of composition as given in Table I. During alloy preparation all melts were modified with sodium, grain refined with Tibor and rotary degassed for 5 minutes.

Fatigue specimens with dimensions $10 \times 10 \times 100$ mm were produced by sand casting and die casting. Specimens were sand-cast in batches of six, but die-cast individually. All cast specimens were heat treated in the TF conditions: solution treatment at 520 °C for 8 hours followed by a warm-water quench and age hardening at 170 °C for 8 hours. Some sand-cast specimens were mechanically polished on the top and both side surfaces to give a final finish of 1 μm.

Topographical examination of the various surface textures has been carried out using a 3-D Talysurf stylus instrument. The surface roughness, Ra, was measured in two directions on the surface to be subjected to the maximum stress during fatigue loading. The definition of the parameter Ra is illustrated in Fig. 1.

Porosity was measured on the polished sample surface (unetched). The samples were cut from three locations in a fatigue specimen: A, near the filler end; B, in

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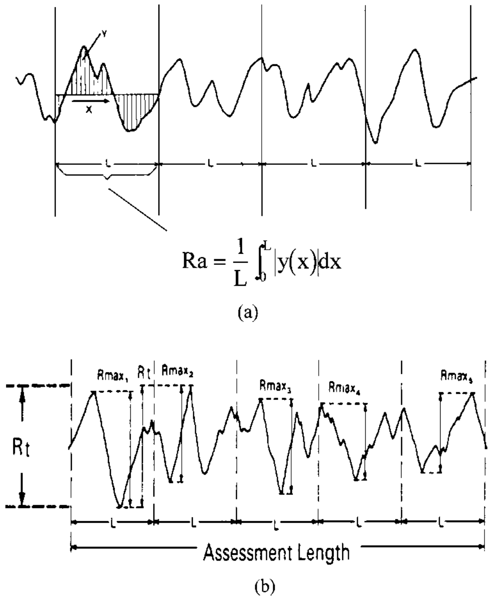


Figure 1 Definition of surface roughness parameters (a) Ra and (b) Rt.

the middle; and C, near the far end. The porosity was quantified using a computerized Quantimet 500+ image analyser in conjunction with an optical microscope. All images were measured at a magnification of $\times 50$ and a sufficient number of measurements were taken to cover the entire surface of the sample. Each image, however, was first adjusted at $\times 500$ before switching to $\times 50$ because precise focus is critical in the determination of pore shape and size.

All fatigue tests were conducted under four-point bending at a stress ratio of 0.1 ($R = \sigma_{min}/\sigma_{max} = 0.1$) using an Amsler Vibrophore resonant testing machine. The advantage of four-point loading is that it provides a constant maximum bending moment between the shorter loading arms (20 mm in the present case) so that a $10 \times 20 = 200\ mm^2$ area on the top surface is subjected to a constant maximum tensile stress. A surface replication technique was used for monitoring initiation and propagation of fatigue cracks on the top surface.

3. Results

3.1. Surface roughness and porosity

Fig. 2 shows the surface roughness measured on as-cast and polished surfaces produced by sand casting and die casting in terms of the central-line-average (Ra). This Ra is a widely used parameter for assessment of surface roughness [12], but the maximum peak-to-valley height (Rt in Fig. 1) may be a more suitable measure for fatigue analysis [10] so that the corresponding Rt values are illustrated as well. It can be seen from Fig. 2 that the average value of Rt was over 80 μm for the sand-cast surface, compared with order 20 μm for the die-cast surface. The polished surface showed extremely low roughness compared with the as-cast surfaces.

The porosity for sand-cast and die-cast samples are shown in Fig. 3. Metallographic examination of sand-cast and die-cast samples revealed that most cast cavities were shrinkage pores resulting from insufficient metal flow into the space between connected dendrites

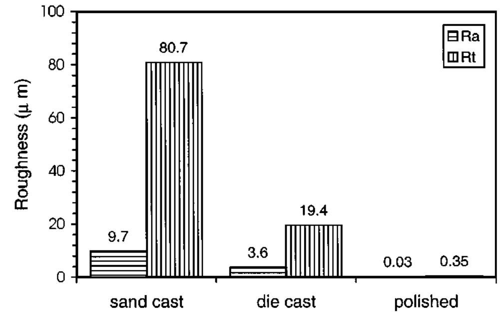


Figure 2 Measured surface roughness of various samples.

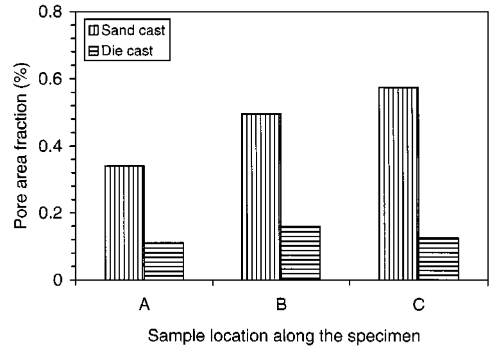
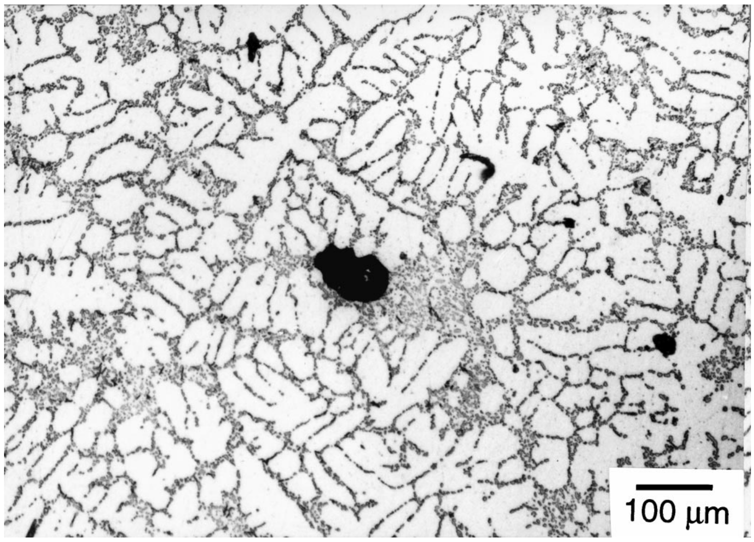


Figure 3 Measured porosity of various samples.

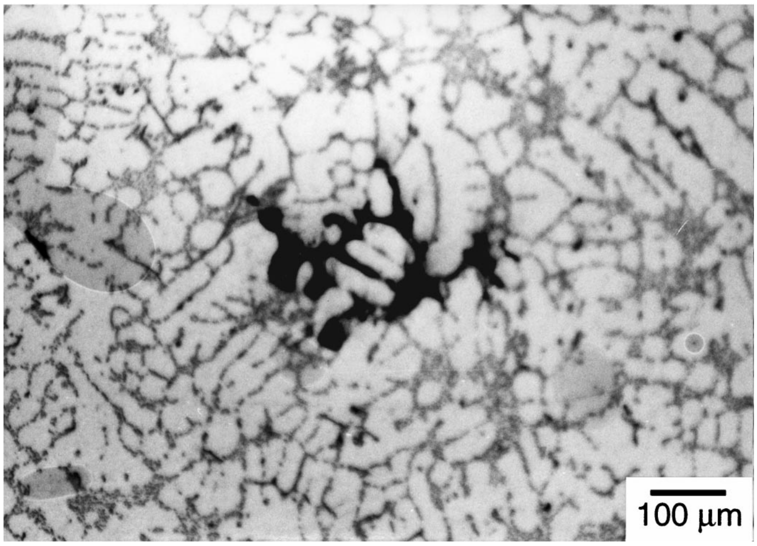
during the solidification process. Only a few defects were gas pores, whether the samples were produced by sand-casting or die-casting. The shrinkage pores had very irregular three-dimension shapes and varied sizes, whereas the gas pores were usually roughly spherical (circular in cross-section). Typical examples of a shrinkage pore and a gas pore are shown in Fig. 4. Quantitative analysis of the porosity indicated that the porosity at location C (far end) was higher than that at locations B (middle) and A (filler end) for sand-cast samples. Location A possessed the lowest porosity. The difference of porosity at various locations results from mould design: for the sand-cast specimens, location A had the best feeding of metal and location C had the worst metal feeding during the solidification process. The die-cast samples generally exhibited much lower porosity than sand-cast samples due to the faster solidification rate. A slight difference of porosity at various locations was also observed, but the highest porosity was at location B in die-cast samples. Again, this is attributed to difficulties in metal flow during the solidification process at different locations. Statistical analysis of pore size distribution for the sand-cast samples indicated that more than 90% of the pores were smaller than 50 μm and less than 1% of them were larger than 300 μm, as shown in Fig. 5. Die-cast samples exhibited similar pore size distribution with an even lower percentage of large pores (Fig. 5).

3.2. Fatigue tests

Fig. 6 shows S-N curves for various specimens where the number of cycles to failure (N_f) is plotted against stress range ($\Delta\sigma = \sigma_{max} - \sigma_{min}$). Although scatter in fatigue life is observed, there is no significant difference



(a)



(b)

Figure 4 Optical micrographs of (a) a gas pore and (b) a shrinkage pore.

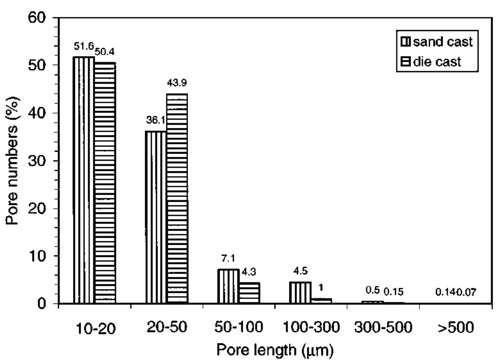


Figure 5 Pore size distribution of sand- and die-cast samples.

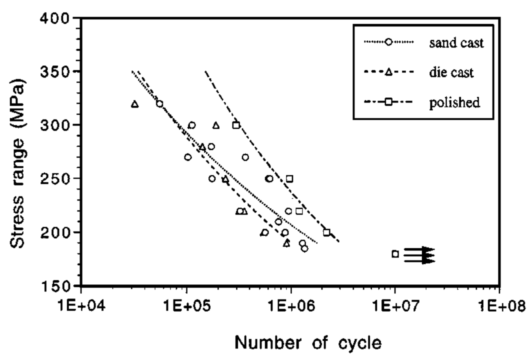


Figure 6 Fatigue life of various specimens.

in fatigue life between sand-cast and die-cast specimens. The die-cast specimens possess much lower surface roughness and porosity than the sand-cast specimens. Although these two factors would be considered to be important in controlling the fatigue life of a cast material, this is not demonstrated by the experimental results. The polished specimens, however, exhibited a significant improvement in fatigue life (Fig. 6). This result indicates that removal of the rough (as-cast) surface can increase resistance to either crack initiation or to crack growth or to both. Fig. 7 shows crack growth curves for an as-cast and a polished specimen produced by sand-casting. The crack length was measured from surface replicas. It is evident that the polished specimen needed a much longer time (cycles) than the as-cast specimen to develop a microcrack of 0.2 mm. The

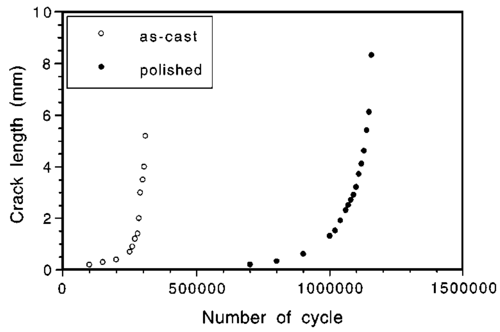


Figure 7 Fatigue crack propagation of an as-cast and a polished specimen.

cycles for reaching this crack length accounted for 60% of the total fatigue life for the polished specimen but only about 30% for the as-cast specimen. After reaching 0.2 mm the crack growth behaviour was similar in both as-cast and polished specimens, implying that the improvement in fatigue life of polished specimens could be mainly attributed to an increase in the (engineering) "initiation life" of the fatigue crack.

3.3. Fractography

Three different types of crack initiator were found for the various specimens. Examination of the fracture surface of sand-cast specimens revealed that in all cases the fracture initiated from the rough surface (hollows) as shown in Fig. 8. The widths and depths of the hollows which acted as initiators were in the range of 200–300 μm and 70–100 μm respectively. The stress concentration on the rough surface resulted in initiation of a fatigue crack which grew and finally caused fracture. In fact, cracking on the rough surface occurred not only from the sites of hollows but also from other surface defects such as micropores and inclusions. The microcracks initiating from these other defects, however, never grew to become a main crack which resulted in final fracture, implying that the stress concentration in the sites of hollows was much higher than in other areas so that the microcracks from the hollow grew quickly and became a main crack. A different type of fatigue crack-initiation origin was found in polished specimens where shrinkage pores were observed on fracture surface just below or at the specimen surface for all stress levels. The size of pore which was found to act as an initiation site was about 100–200 μm. A typical example of such a pore initiator is shown in Fig. 9. However, few such shrinkage pores were observed on the fatigue crack path in spite of the fact that large pores ($>300\,\mu$m) were often found in the final fracture area. This observation in polished specimens suggests that crack initiation is controlled by surface or subsurface pores, that crack growth is rather less affected by such defects, and final failure is dominated by rupture of the relatively weak regions where large shrinkage pores located.

The fracture surfaces of the die-cast specimens revealed a completely different type of crack initiation origin from the sand-cast specimens. In almost all cases, cracks initiated from oxide films or oxide inclusions located at or near the specimen surface. The oxide film was usually folded or tangled. A typical example of such a oxide film is shown in Fig. 10a. Microanalysis using EDX with ultra-thin window techniques indicated that the oxide was in the form of $MgO\cdot Al_2O_3$, known as spinel, because the Mg peak observed on an EDX profile was much higher than that of the matrix (Fig. 10b). This type of oxide film is thought to be transformed from MgO during solution heat treatment [1]. The MgO initially formed on the surface of the liquid alloy because Mg is surface active in liquid Al and concentrates at the surface of the liquid alloy [14].

4. Discussion

For sand-cast specimens, failure initiated from the rough surface (hollows) in all cases. Cracking was found to occur at other surface defects such as micropores and inclusions, but microcracks initiated from these defects never grew to become a main crack which

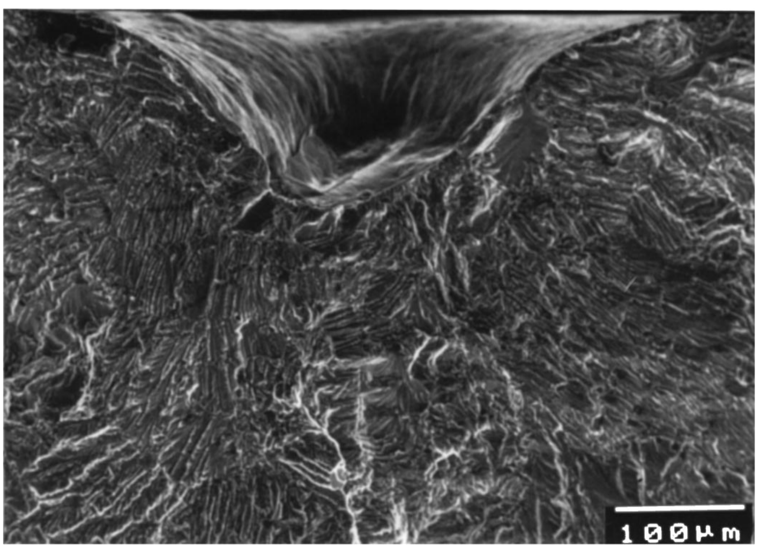


Figure 8 Fatigue fracture surface of a sand-cast specimen showing initiation from a surface hollow.

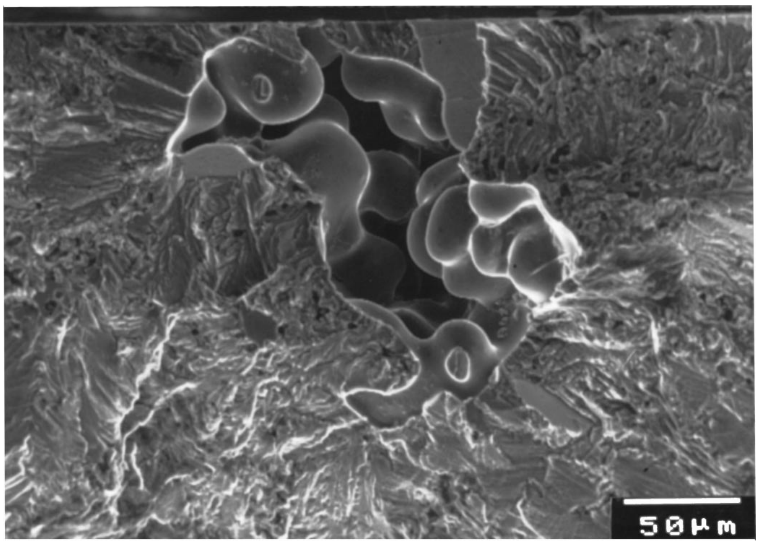
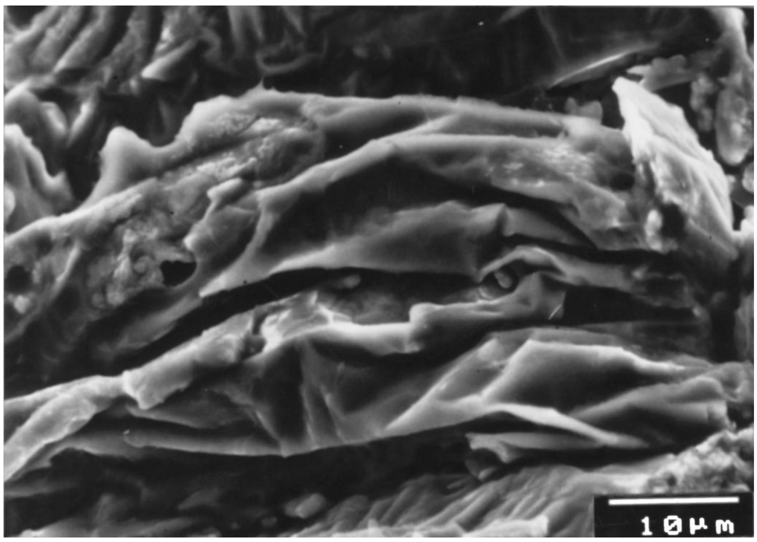
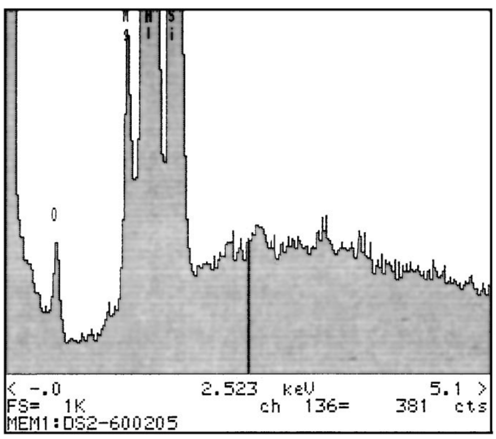


Figure 9 Fatigue fracture surface of a polished specimen showing initiation from a pore.



(a)



(b)

Figure 10 (a) A typical example of oxide film observed in the fatigue crack initiation area of a die-cast specimen; (b) EDX profile obtained from an area of oxide film shown in (a).

resulted in final failure. This result implies that the driving force for crack growth at the sites of hollows is much higher than that at the other sites. The size of the hollow which acted as a crack initiator is 200–300 μm in width and 70–100 μm in depth, which is nearly three times as large as other surface defects. The stress concentration at the hollows is therefore much higher than that at the other defects and hence it is easy for the microcracks at the hollows to initiate and grow. The depth of the hollows is about the value of Rt, indicating that only the largest hollows are able to act as crack initiators. A decrease in surface roughness of cast materials, and hence a decrease in initiator size, should increase the difficulty of crack initiation. However, if surface roughness is decreased, other types of defects play a increasingly important part in crack initiation, and eventually can become the critical crack initiators. This situation is demonstrated in the polished specimens. When the rough cast surface is removed, failure under all stress levels studied initiated from surface or subsurface pores.

The improvement of fatigue life in polished specimens was not attributed to the decrease in the size of defects which acted as crack initiators. The pores observed in the initiation sites have a size (100–200 μm) similar to that of the surface hollows. However, these pores were usually located partially or completely under the surface. In most cases, the part intersected by the surface is only 20–50 μm or even smaller. The growth of a surface crack from the intersected part to the overall pore size is not as easy as might be expected. It is evident from Fig. 7 that the surface microcrack spent about 60% of its total life in growing from 20 μm to the pore size (200 μm). On the other hand, it is relatively easy for the microcrack initiating from a hollow to reach 200 μm. It can be seen that the number of cycles required for developing a microcrack of 200 μm from a pore is nearly 7 times as great as that from a hollow. This suggests that in the case of microcracks initiating from a surface or subsurface pore, there is a degree of pessimism in predicting fatigue life from an identification of the pore size (as the crack initiation size) if "initiation life" is neglected. It should also be noted that the pores which acted as fracture initiators have a size of 100–200 μm while most of the pores (>99%) are less than 100 μm. Under the stress levels in the present study a critical crack size should be far less than 100 μm. A possible explanation for this observation could be that the probability of large pores (100–200 μm) being located at or close to the specimen surface is so high that the large pores are almost always available for crack initiation. The probabilistic analysis of cast defects and fatigue life will be discussed in another paper to be published.

Die-cast specimens did not show longer fatigue lives, though their surface roughness and porosity are much lower than those of sand-cast specimens. The reason is that oxide films replaced the rough surface and pores as fatigue crack origins (Fig. 10a) so that neither low roughness nor low porosity improve fatigue life, particularly crack initiation life. The oxide films observed on the fracture surfaces usually exhibited folded or tangled forms. Due to lack of wetting, they constitute cracks when they are entrained into castings so that they can significantly decrease the strength of the cast materials. This damage effect has been reported by Green. *et al.* [1] who demonstrated a significant decrease and scatter in ultimate tensile stress caused by oxide films in a Al-7Si-Mg alloy. The formation of oxide films results from aluminium alloy exposed to an atmosphere containing oxygen. The alloy reacts with the oxygen to form oxide film on the surface. When the meniscus is damaged due to surface turbulence, the oxide film can be entrained into the bulk of the alloy. The entrained film may be then folded and tangled by bulk turbulence. The oxide films can be divided into two types: "young" and "old" [1, 13]. The "young" oxide films form in a short oxidation time during mould filling due to in-mould filling surface turbulence. They appear as fine wrinkles which are controlled by the thickness of the films. The "old" oxide films with the appearance of coarse wrinkles are thought to be entrained with the metal coming from the melting crucible. Both types of oxide films were observed in the present tests, but most of the films observed on fracture surfaces are "young" oxide films, implying that they are not associated with the original liquid metal.

Oxide films have not been observed on the fracture surfaces of sand-cast specimens. There may be two possible reasons for this result. One is that the damaging effect of oxide films on fatigue crack initiation was overridden by other large casting defects such as rough surface (hollows) or pores. The other is that the number of oxide films in the sand-cast specimens is small. The first possibility can be ruled out by the fact that the sand-cast specimens do not exhibit shorter fatigue lives than the die-cast specimens. From the above discussion, the formation of oxide films is not associated with the casting method, i.e., sand-cast or die-cast. If a difference of oxide film level exists between the sand-cast and die-cast specimens, the difference is believed to be caused by the different designs of the mould filling system for these casting specimens. As the detrimental effect of oxide films on fatigue life is as important as that of rough surfaces or pores, it is important and necessary to decrease the level of oxide films by the design of the mould filling system while simultaneously making an effort to improve surface quality and porosity.

5. Conclusions

Fatigue tests and investigation of cast defects have been carried out. The various defects demonstrated significant effects on fatigue life, particularly the crack initiation behaviour. The following conclusions may be drawn from this study.

1. Surface hollows dominated fatigue crack initiation and became the sole failure origins in sand-cast specimens with high surface roughness. The depth of the hollows is identical to the maximum value of peak-to-valley height obtained in the measurement of surface roughness. Decrease in surface roughness would reduce the size of large hollows and increase resistance to (engineering) crack "initiation" and hence improve fatigue life.

2. Surface and subsurface shrinkage pores were found to be the crack initiators in the polished (sand-cast) specimens. Improvement in fatigue life was attributed to a longer crack "initiation" period. It is not as easy as expected for a microcrack initiating from a pore

to grow up to the pore size. It is suggested that a crack "initiation" life could be in the fatigue life prediction of pore defected materials.

3. In the die-cast material, oxide films were found to be significantly important in controlling fatigue life. The oxide films acting as crack origins are mainly "young" oxide in the present case, believed to form due to surface turbulence during mould filling. Improvement in the design of mould filling system for reducing surface turbulence would decrease the level of oxide films in the cast materials and hence improve fatigue strength.

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References

1. N. R. GREEN and J. CAMPBELL, *AFS Transactions* **114** (1994) 341.
2. C. M. SONSINO and J. ZIESE, *Int. J. Fatigue* **15** (1993) 75.
3. B. SKALLERUD, T. IVELAND and G. HARKEGARD, *Engineering* Fracture Mechanics **44** (1993) 857.
4. A. A. DABAYEH, R. X. XU, B. P. DU and T. H. TOPPER, *Int. J. Fatigue* **18** (1996) 95.
5. M. J. COUPER, A. E. NEESON and J. R. GRIFFITHS, *Fatigue Fract. Engng Mater. Struct.* **13** (1990) 213.
6. M. R. BAYOUMI and A. K. ABDELLALIF, *Engineering Fracture Mechanics* **51** (1995) 861.
7. Y. BERGENGREN, M. LASSON and A. MELANDER, *Fatigue Fract. Engng Mater. Struct.* **18** (1995) 1071.
8. D. TAYLOR and O.M. CLANCY, *ibid.* **14** (1991) 329.
9. L. WAGNER and G. LUTJERING, in "Fatigue 90" (Hauxii, Honolulu, 1990) p. 323.
10. P. GREENFIELD, D. H. ALLEN, P. BYRNE and D. TAYLOR, *ibid*, p. 391.
11. S. GUNGOR and L. EDWARDS, *Materials Science and Engineering* **A160** (1993) 17.
12. H. DAGNALL, in "Exploring Surface Texture" (Rank Taylor Hobson, Leicester, 1980).
13. J. CAMPBELL, in "Castings" (Butterworth-Heinemann, Oxford, 1991).
14. A. FURNESS and D. TALBOT, *Foundryman* August (1996) 226.

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