

# THE CONCEPT OF THE FATIGUE POTENTIAL OF CAST ALLOYS

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## ABSTRACT

The cumulative failure distributions of fatigue life of cast alloys is shown to be characterised by a number of distinct platforms and steps. A technique to demonstrate the existence of an ultimate step has been discovered. This illustrates for the first time that some current cast aluminium alloys are, at best, only achieving 1% of their potential fatigue life. If the materials are subjected to poor casting techniques the fatigue life performance is expected to fall further. It is therefore concluded that by a combination of attention to metal quality and casting technique, the fatigue lives of most Al alloy castings can probably be improved by a factor of 100 to 10,000 times. It is speculated that the huge numbers of double oxide films which appear to be present in liquid aluminium and control the properties of castings will survive to be present in most wrought products and similarly control the failure behaviour throughout the greater part of aluminium metallurgy.

## INTRODUCTION

It is practically a truism to state that materials fail by the initiation cracks from their most serious defects. If these defects are eliminated by, for instance, careful and/or appropriate processing, then the performance of the material is likely to be improved. Strength and durability in service are then limited by the next most serious family of defects, and so on. When each family of defects has been removed, the material will fail from metallurgical features such as the large intermetallic compounds or other brittle phases which might be present. Finally, if these in turn are removed, we have the ultimate resistance to failure as dictated by the metallurgy of the alloy, for instance the ability of the alloy to resist slip, and once slip has been initiated, the slip distance as dictated by the grain size as enshrined in the Hall-Petch relation.

Thus in the hierarchy of features which initiate cracks which lead to failure, much exemplary scientific work has been carried out on the metallurgical aspects of failure initiation.

However, to understand the causes of failure in most manufactured components, particularly cast products, it is essential to identify those features which might have precedence in the hierarchy. In general, such features have been overlooked because of the statistical problem of finding large features in small specimens generally used in metallurgical studies. It is clearly easy to concentrate on those features such as grain size and subgrain structure which are the outstanding features of microscopical examination. However, on this scale, features measured in areas of square millimetres or square centimetres are often invisible to the naked eye because they are so thin, but also remain undetected by microscopical examination because they are statistically rare in such small samples. Regrettably, they are not rare on the scale of the whole cast component, and are likely therefore to be the very features which cause failure in service.

For this reason, work has been carried out at Birmingham /1/ to determine the nature of these elusive major faults at the top of the hierarchy of potential initiators.

Fatigue studies on a commonly-used aluminium cast alloy, Al-7Si-0.4Mg, were carried out. The alloy was prepared from good quality starting materials, carefully melted and cast, and subsequently subjected to a fatigue test in a pull-pull mode. This mode of stress was chosen so that the fracture surface would not be damaged during the fatigue process when the crack was opening and closing (because in this case the crack would always remain open) and especially during the final tensile failure. Thus the fracture surface could be examined in great detail, allowing the fatigue striations to be followed to their source with the use of the scanning electron microscope (SEM). In this way the crack initiator could be identified with certainty.

### **EXPERIMENTAL TECHNIQUE AND FINDINGS**

The work is described in detail elsewhere /1/. However, in brief, A356 alloy was cast into a urethane-bonded 60 AFS grade silica sand mold as shown in Figure 1. Each mold gives 10 test specimens. Five molds were cast without a filter in the base, and five were cast with a filter in the filter print.

The filling of the molds was studied by real-time x-ray radiography. To the surprise of all, the filling appeared at first sight to be little altered, being apparently good in both cases. However, a further frame-by-frame examination of the video recording revealed a few frames in the case of the castings made without the filter in which the metal suffered some surface turbulence in the empty filter print (Figure 2). Figure 3 illustrates the mechanism of the generation of both cracks and pores by surface entrainment.

The period of turbulence lasted approximately 0.02 second, compared to the total fill time of approximately 2 seconds. Thus, for convenience, this was termed a 1% turbulent fill (compared for instance with other designs of filling system where the extent of turbulence is 10 % or, very commonly, even 100 %). This method of quantifying surface turbulence is admittedly poor because the intensity of

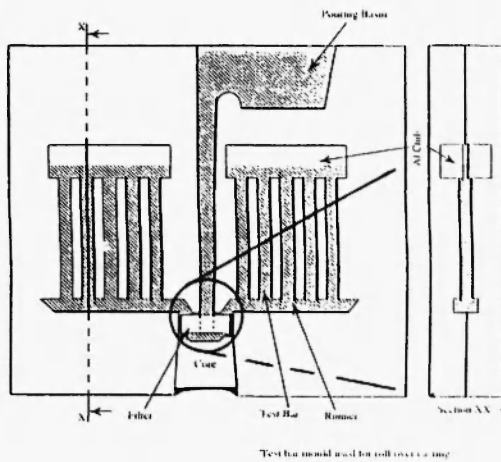


Fig. 1: The test mold as used in reference 1.

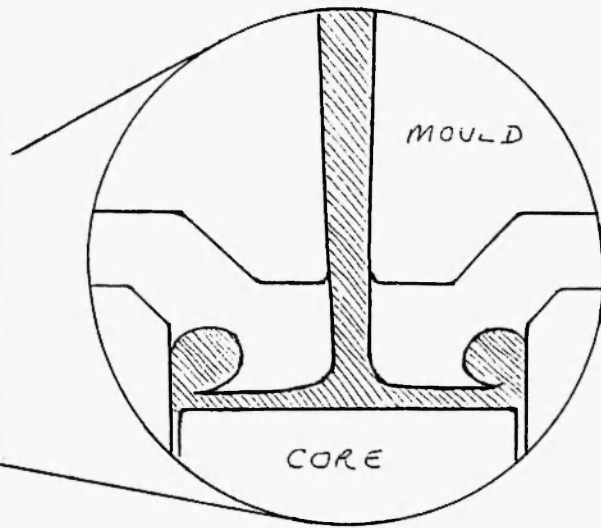


Fig. 2: A line drawing from a frame of the x-ray video of filling showing the brief surface turbulence.

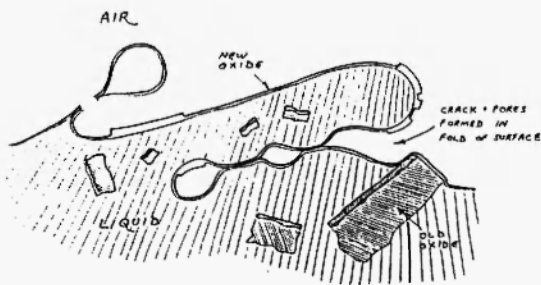


Fig. 3: Surface entrainment leading to the formation of cracks and pores.

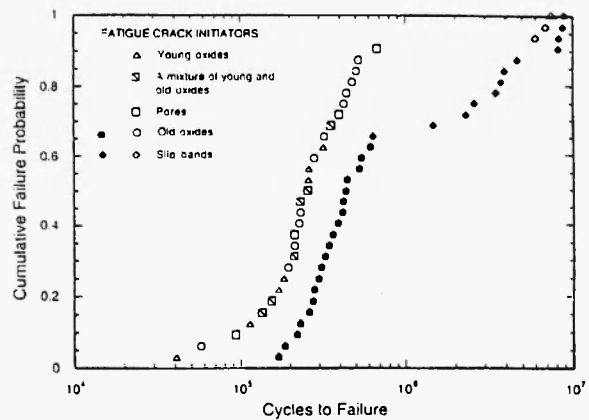


Fig. 4: Fatigue lives for (i) filtered and (ii) unfiltered and turbulently-cast metal.

turbulence is not included. However, it is proposed as an easy, if not especially accurate, measure of surface turbulence during pouring in the absence of any better measure at this time.

All 50 specimens were tested in fatigue in pull-pull mode (maximum stress = 150 Mpa and  $R = +0.1$ ) for both the filtered and unfiltered castings. The distribution of results is given in Figure 4.

Every fracture surface was subsequently studied in the SEM, following backwards the fatigue striations (Figure 5) to identify the nature of the initiation of the fracture. Mostly this was a defect such as an oxide film as in Figure 6, but occasionally was a slip plane as in Figure 7. These SEM images were quite distinctive.

The failures in the case of the 1 % turbulent-filling were found to be a mixture of pores, old oxides and new oxides. The failures in the case of the filtered casting was exclusively from old oxides (apart from a few failures which were found to be the result of metallurgical mechanisms such as the operation of slip bands as will be discussed below). Clearly, in spite of the use of a filter, inclusions remained the dominant cause of failure.

In the previous publication of this work /1/ the emphasis was on the significance of the separation of the two experimental curves. Thus on average there was about a factor of 3 benefit in fatigue life, although this improved to a factor of 5 at lower lives, for castings which had eliminated turbulence during filling, and which therefore had eliminated the occurrence of new oxide films and pores. It was further confirmed in this study that new films generated during the pour were finely folded when viewed under the SEM (Figure 6). In addition they were associated with pores, and were formed from substantially pure aluminium oxide. In contrast, old films carried over from the melt were simply thick chunks of refractory oxide, high in magnesium because of the development of the spinel structure with time, and were not associated with pores because they had consumed their entrained layer of air (Figure 3). Thus old and new oxides were quite distinct.

In this paper, the discussion is developed to show how the results of this important study actually indicate much more, and how much further we appear to have to go in achieving the real potential of cast materials.

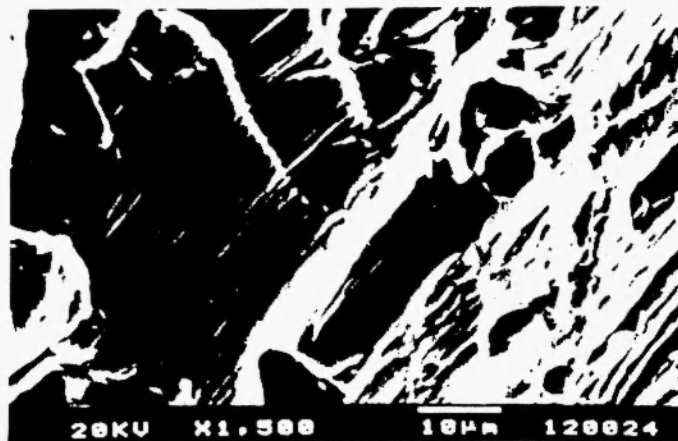


Fig. 5: SEM observation of fatigue striations on the fracture surface.

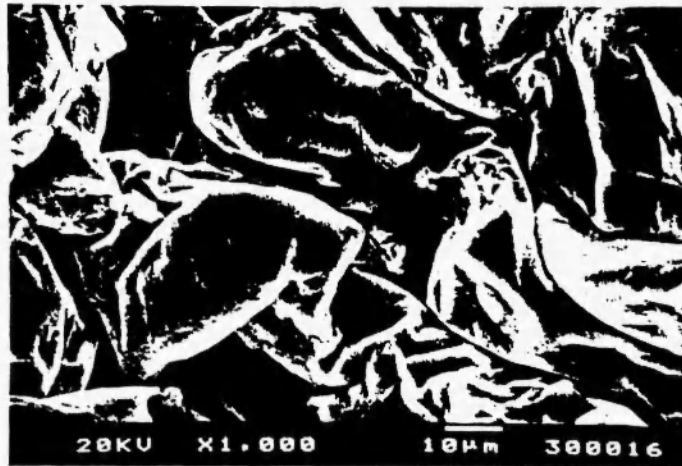


Fig. 6: SEM observation of a folded oxide film observed on a fracture surface.

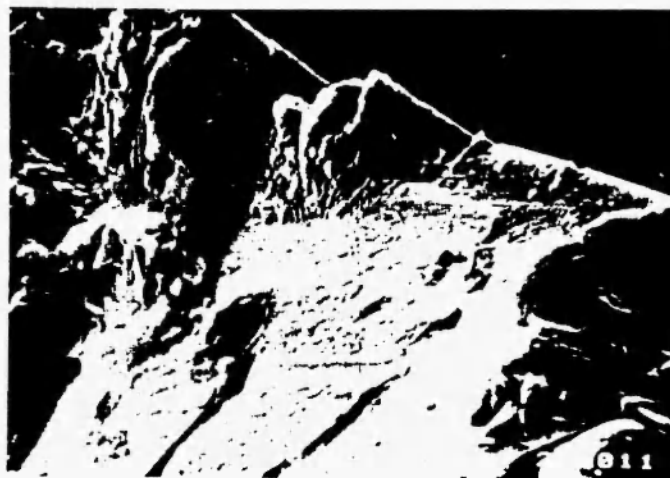


Fig. 7: SEM observation of the fracture surface showing a slip plane at the site of the initiation of a fatigue failure.

## DISCUSSION

To understand the problem of the statistics of defect populations, it is useful to consider the case of the plate casting in Figure 8 which contains a number of cracks. Clearly the plate will fail from one or more of these defects if subjected to a severe strain such as a 3-point bend test. However, if the plate is tested for the integrity of its material by cutting it into a number of tensile test bars, Figure 9, it is clear that in this instance only one of the ten bars happens to show any reduction in strength; the result is “90% good”.

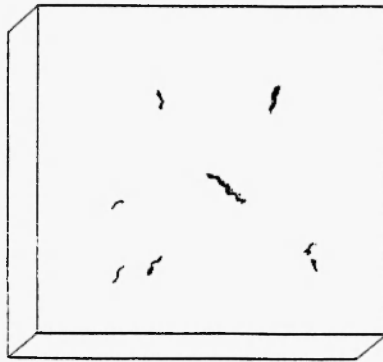


Fig. 8: Plate casting containing a number of cracks.

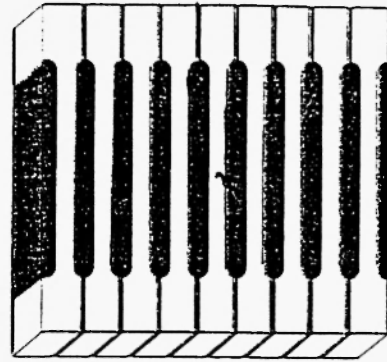


Fig. 9: The plate casting sectioned and machined to make tensile test pieces illustrating how only one test piece happens to contain a fault.

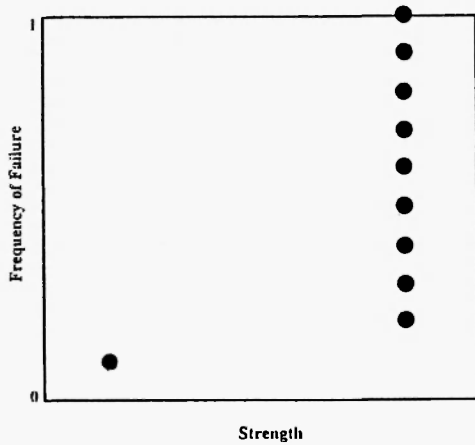


Fig. 10: A figurative plot of the cumulative probability of failure of the test results.

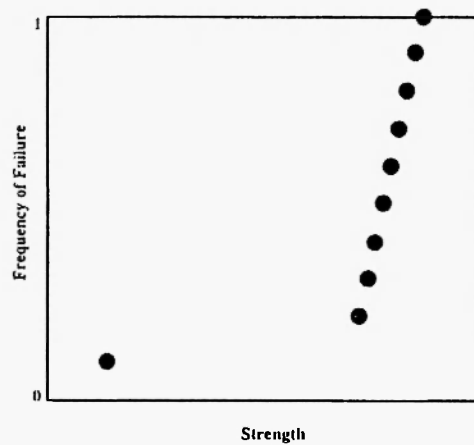


Fig. 11: A more realistic plot allowing for small errors and/or defects.

Most national specifications would allow the low result to be designated as a “rogue”, and thus would allow it to be discounted. The fact that the casting would almost certainly have failed from this source makes this procedure for ignoring results curiously inappropriate, if not ominously dangerous.

If we do not discount the low strength result, but include it in a plot of the cumulative probability of failure, we obtain Figure 10 assuming the material has perfect strength, except for the one specimen which contains the defect. If we make the results rather more realistic, allowing for the fact that the

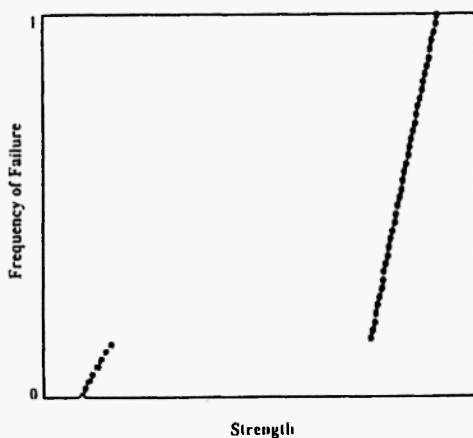
results are not likely to be perfectly identical, and even if they were, could not be measured to be so by a testing procedure which inevitably contains errors (even if very small), then we obtain Figure 11. Furthermore, if instead of only 10 results we had taken 50 specimens cut from similar plates, we might have finished with a result like that in Figure 12 where the different parts of the curve have now developed distinctive slopes which are characteristic of the fracture mechanisms at work.

If we now further progress our argument to consider an initial cast plate which now contains a high volume density of defects, it is clear that we could generate a cumulative probability of failure relation as illustrated in Figure 13. Clearly, to generate this "90% bad" result we would require a very high density of defects.

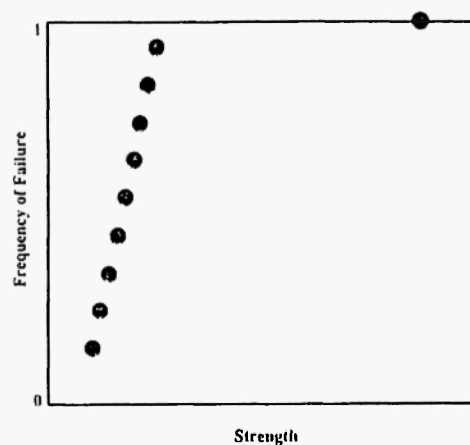
If now we return to the findings of our earlier study, we see that Figure 13 bears a remarkable similarity to the results from the non-turbulent filtered casting shown in Figure 4. (In Figure 6 it was fortunate that sufficient test samples had been made to ensure that there was a statistical probability that one or two samples would lie on the "good material" curve. Any fewer test samples might have never revealed the existence of a portion of the curve up at the top right hand corner.)

We can deduce immediately therefore that there is an ultimate curve somewhere in the vicinity of the broken line shown in Figure 14. It follows that the majority of the observed fatigue failures are approximately a factor of 100 lower than could be obtained in a defect-free material. Thus in general we can conclude that most of the test specimens of the observed material achieved only 1% of the potential fatigue performance of the material, if no defects had been present.

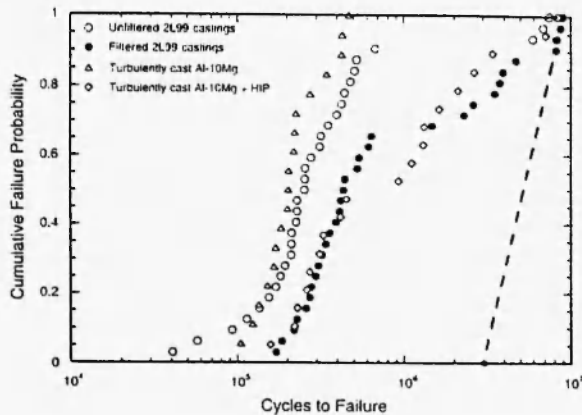
It is further significant that in Figure 14 the position of the projected defect-free curve is also confirmed by the results on a quite different aluminium alloy, Al-10Mg, by independent workers in a laboratory on the other side of the world /2/. (They used fatigue conditions of maximum stress = 106 Mpa and R = -1)



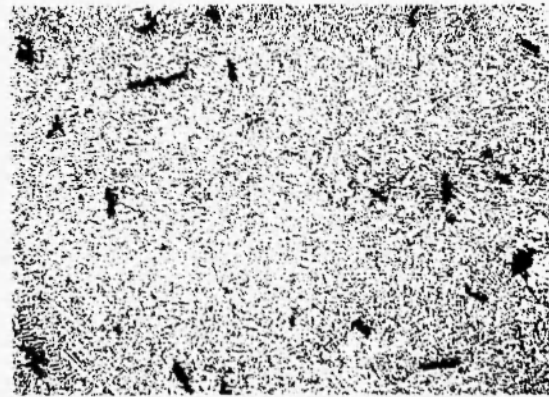
**Fig. 12:** A cumulative probability of failure relation for a reasonably defective material, with 50 test results.



**Fig. 13:** The cumulative failure probability for a plate with a very large number of defects, cut into 10 test bars.



**Fig. 14:** The results shown in Figure 4 illustrating the projected results for defect-free material.



**Fig. 15:** An optical micrograph of a polished section showing typical microporosity  
 ───────── 1 mm

### FURTHER PROJECTIONS

To take these thoughts further, it follows that this carefully melted and cast laboratory material contained a very high density of defects to have exhibited a cumulative failure curve of the type observed. A typical optical microstructure is shown in Figure 15, illustrating a defect density of 10,000 pores per ml. Thus it is to be expected that most industrially available materials will have at least as many defects. If this is true, it appears that most engineering aluminium alloys which are available to us at this date are probably highly defective because of the presence of a high density of folded oxide film defects.

Grain boundaries are one of the expected locations for double films because the growing dendrites will not be able to cross a double oxide crack because of the entrained layer of air. Most of the crack-like defects in Figure 12 are confirmed by anodic etching to be sited on grain boundaries.

A second likely location will be on or in many intermetallic compounds. This is because those that precipitate as a result of peritectic reactions at temperatures above the general freezing point of the alloy have been clearly demonstrated to deposit on oxide films floating in the alloy /3/.

If oxide defects permeate all of aluminium metallurgy, and if such defects are likely to survive significant amounts of processing, such as rolling, and possibly even extrusion, it is likely that they will continue to control the failure behaviour of most wrought products also. Thus cracks which appear to initiate from boundaries of grains might easily have arisen from thin films located on the boundaries. Superplastic flow leading to cavitation at grain boundaries and grain corners, creep failures from grain boundaries and many other instances of failure might originate from this cause.

Similarly, the cracking of intermetallics is conventionally attributed to their brittleness, or even their weakness. However, most such compounds, even if lacking ductility, are strong. Thus the cracks which are often observed travelling through such particles, or separating the particle/matrix interface may actually be travelling along a non-bonded oxide interlayer. Such non-bonded double layers are to be expected to be present and must be expected therefore to influence failure.

It appears therefore that much of our existing thinking in aluminium metallurgy may require to be re-thought.

#### REFERENCES

- 1 C Nyahumwa, N R Green and J Campbell; "Effect of mold -filling turbulence on fatigue properties of cast aluminium alloys" Proc AFS Congress 1998 paper 98-058.
2. G R Wakefield and R M Sharp; "Hot isostatically pressing of Al-10%Mg alloy castings" Science and Technology 1992 vol 8 1125-1129
3. N D G Mountford and R Calvert; J Institute of Metals 1959-1960 vol 88 pp 121-127

