THE INFLUENCE OF MATERIALS PROCESSING ON THE HIGH TEMPERATURE LOW CYCLE FATIGUE PROPERTIES OF THE CAST ALLOY IN-738LC

M.A.Burke[1], C.G.Beck Jr.[1] and E.A.Crombie[2]

Westinghouse Electric Corporation

1. Research and Development Center
Pittsburgh Pa. U.S.A.

2. Combustion Turbine Systems Division
Concordville Pa. U.S.A.

Summary

The influences of various conventional and D.S. casting practices, as well as subsequent HIPping, on the LCF behaviour of IN-738LC have been assessed by strain controlled testing at 650C and 850C. The greatest resistance to strain cycling was exhibited by the D.S. alloys and is attributed to the beneficial influence of the low elastic modulus which is associated with the <001> directional structure. The conventionally cast materials displayed significant scatter in fatigue life, while the data for the HIPped material lay towards the upper limits of the scatter band for the conventionally cast materials. Fatigue crack initiation occurred preferentially at casting porosity in the unHIPped alloys at both temperatures, although at 650C crack initiation at MC carbide particles was a competitive mechanism. The HIP cycle not only reduced porosity but also provided increased strength at 850C, both aspects contributing to the improved fatigue life of the HIPped material at this temperature. The beneficial effects of D.S. processing derive primarily from the accomodation of strain elastically by virtue of the low <001> elastic modulus but the reduced porosity and the absence of grain boundaries perpendicular to the applied stress also contribute to the improvement of fatigue life.

Introduction

The use of cast nickel-base superalloys greatly facilitates the fabrication of power generation turbine blades. The generation of microstructure under different solidification conditions and the derivation of mechanical properties from the microstructure dictate that the local mechanical properties depend upon the prior local solidification conditions, including those inherent in the mold and casting geometry. The microstructural features that are controlled by solidification, and which are known in turn to control certain mechanical properties, are dendrite arm spacing, carbide particle size, porosity and preferred crystallographic orientation (1). Since the properties of cast alloys are frequently determined by testing small,especially cast bars, it is important to recognize that the properties of large castings may differ from those specified from small bars.

While many nickel base superalloy components are fabricated simply by precision casting and subsequent heat treatment, it has been recognized that solidification control and post solidification processing (HIPping) can significantly improve behaviour. Wasielewski and Lindblad (2) reported in a previous Superalloys conference that HIPping significantly improved the LCF properties of IN-738 in the 700C temperature regime, while a similar beneficial effect of HIPping upon the HCF behaviour has also been recently published (3). In both cases, the improvement in the properties was ascribed to the elimination of casting porosity. The practical application of HIPping, however, does require an upper limit to be imposed upon the casting pore size because the largest pores, which are also those that most affect fatigue behaviour, sinter most slowly (4). That meticulous control of solidification processing can improve the performance of cast components has been demonstrated by the success of directionally solidified (D.S.) and single crystal components for aircraft applications (5). D.S. processing improves longitudinal properties at high temperature by suppressing intergranular fracture,although the deformation behaviour is expected to be most strongly influenced by the preferred crystallographic orientation (6). Wright and Anderson (7) proposed that the orientation dependence of the fatigue life of D.S. Rene 120 was determined by the crystallographic dependence of the elastic modulus.

In order to assess the effects of processing on the properties of IN-738LC, castings were prepared in the forms of D.S. and conventionally cast(c.c.) bars and D.S. and conventionally cast experimental power generation turbine blades. One of the c.c. blade castings was subsequently HIPped using the procedure described in ref.8. LCF tests were performed at 650C and 850C on specimens machined from these castings.

Experimental

The standard,two step heat treatment recommended for IN-738LC was applied to the cast and cast-plus-HIPped structures. Smooth sided fatigue specimens, with a cylindrical gauge length 0.25 inch in diameter and 0.625 inch in length, were machined from each batch of material. In the case of the D.S. alloys, the long axes of the specimens were always parallel to the solidification direction. LCF tests were conducted in air under the control of a gauge length extensometer at 650C and 850C using a 0.1 Hz. triangular waveform. Continuous load-time and inelastic strain-time chart recordings, as well as load-time hysteresis loops were available from each test. For each test, the load amplitude at half-life was taken to be the cyclic stress response to that imposed strain amplitude and the complete separation of the specimen was taken to be the fatigue life. Scanning electron microscopy was employed to examine the fracture surfaces after testing.

Curve 746199-B

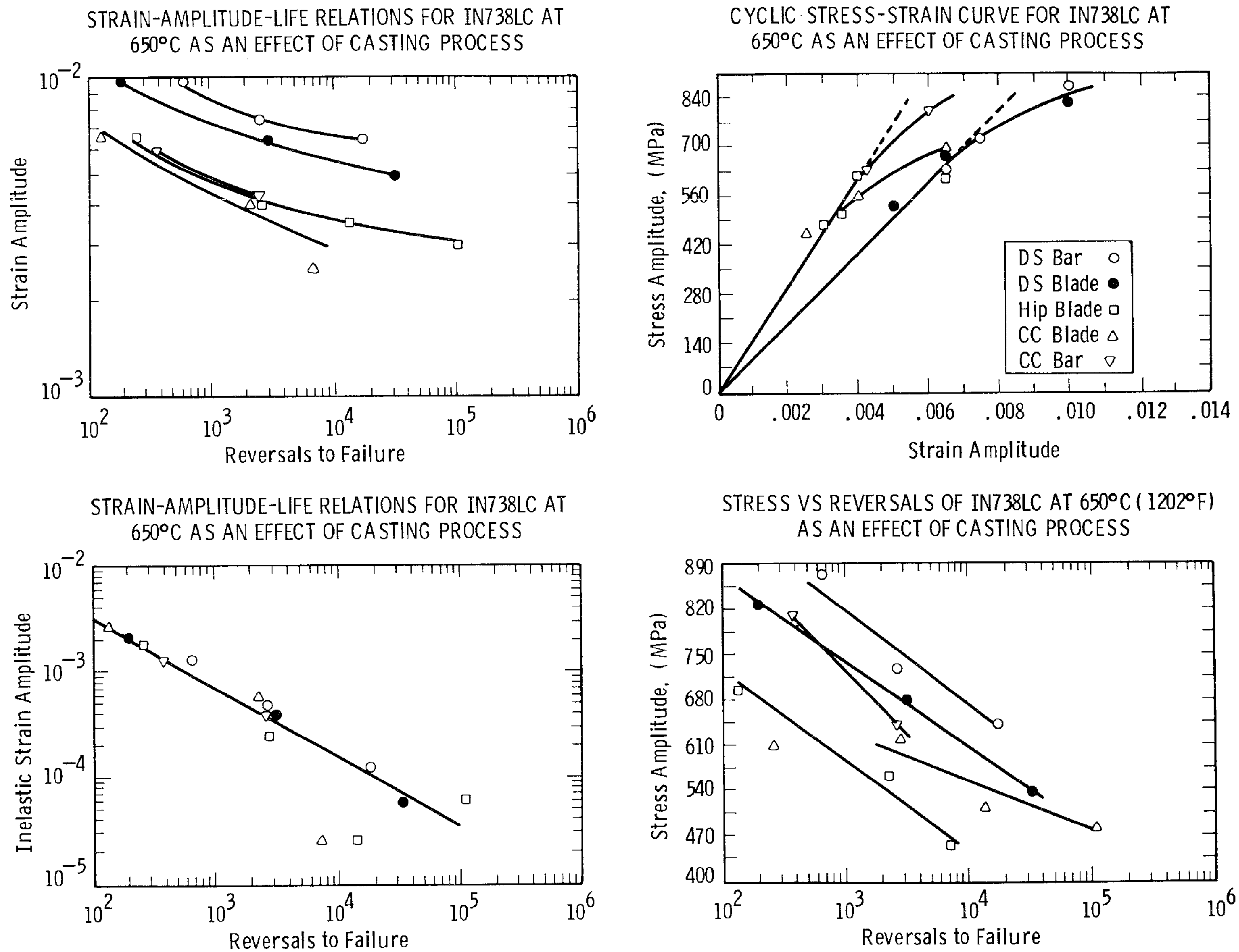

Figure 1 LCF Data for IN-738LC Castings at 650C

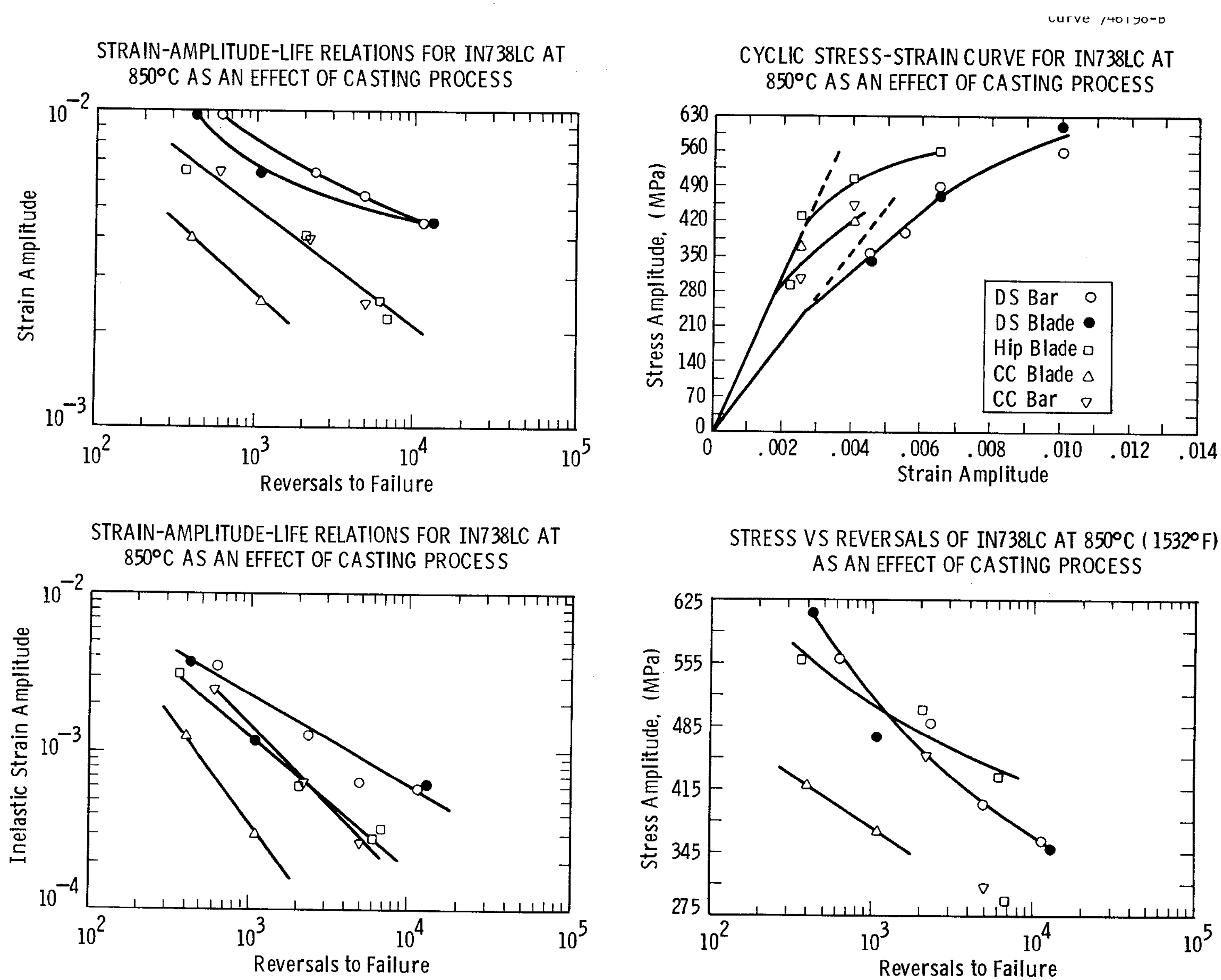

Figure 2 LCF Data for IN-738LC Castings at 850C

Results

The total strain-life, inelastic strain-life, cyclic stress-strain and stress-life data for the different materials at 650C and 850C are shown in figures 1 and 2 respectively. The curves are approximate fits to the data and are intended only to draw attention to the groups into which the datasets fall, e.g., at 650C, it appears that the inelastic strain-life data for all the tests fall upon a single line, whereas,at 850C, the D.S. materials show the highest resistance to inelastic strain cycling. The cyclic stress-strain curves at the two temperatures form two distinct sets which are characterized by the elastic moduli (which were available from the hysteresis loops). The mean values of the elastic moduli for the four sets are presented in Table 1 and are illustrated by the dashed and solid straight lines in figures 1 and 2.

Table 1 Elastic Moduli of IN-738LC Castings at 850C and 650C

Specimen Group	850C	650C
Directionally Solidified	89 GPa	103 GPa
Conventionally Cast	159 GPa	166 GPa

A further anomaly in the stress-strain data pertains to the HIPped material which displays very low cyclic work hardening at 650C but exhibits the highest cyclic strength at 850C. The solid lines in the relevant portions of figures 1 and 2 are intended to draw attention to this.

SEM fractography revealed that fatigue crack initiation and crack propagation were affected by both testing temperature and specimen microstructure. Furthermore, the fracture surfaces obtained at 850C were covered by thicker oxide films than those obtained at 650C and they appeared somewhat flatter and more featureless. At both temperatures fatigue crack initiation was associated with specimen porosity (figure 3). Even the directionally solidified alloys displayed fatigue crack initiation at casting porosity located close to the free surface of the fatigue specimens (figure 4). Crack initiation at deeper internal porosity was much rarer, although such a crack was observed to have grown to a very large size in one specimen taken from the c.c simulated blade casting (figure 5). In this case, while the

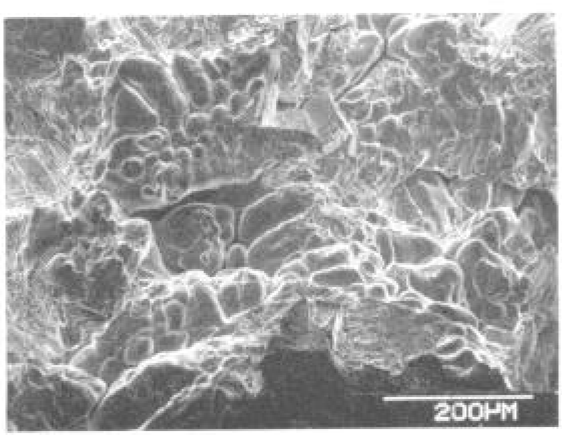

FIGURE 3 Fatigue crack initiation in D.S. IN-738LC tested at 850C

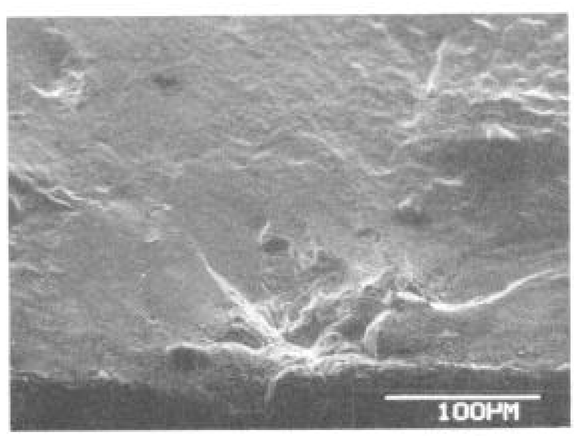

FIGURE 4 Fatigue crack initiation in D.S. IN-738LC tested at 850C

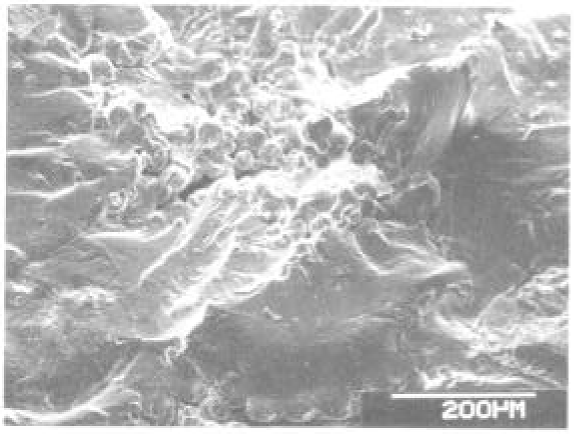

FIGURE 5 Fatigue crack initiation at internal porosity in c.c IN-738LC tested at 650C

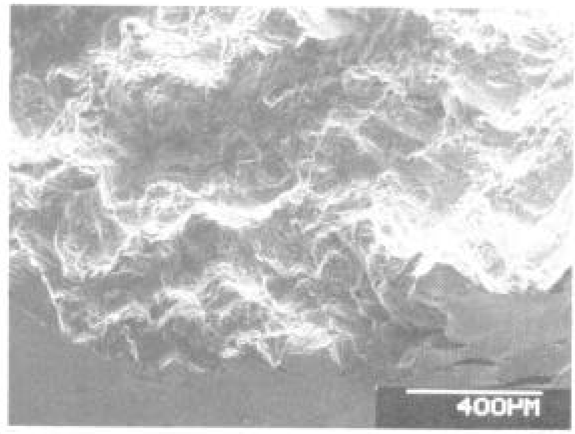

FIGURE 6 Fatigue crack initiation in HIPped IN-738LC tested at 650C

overall porosity level was still relatively low, the clustering of the pores has produced a significant internal defect from which the crack has grown. In the absence of porosity in the HIPped material, the fatigue crack initiation sites were relatively featureless but some specimens fractured at 650C showed evidence of crack initiation at MC carbides (figure 6). Longitudinal sectioning of these specimens showed that secondary cracking was found in all specimens and that, in those of low porosity, cracks could initiate at either MC carbide particles or casting porosity.

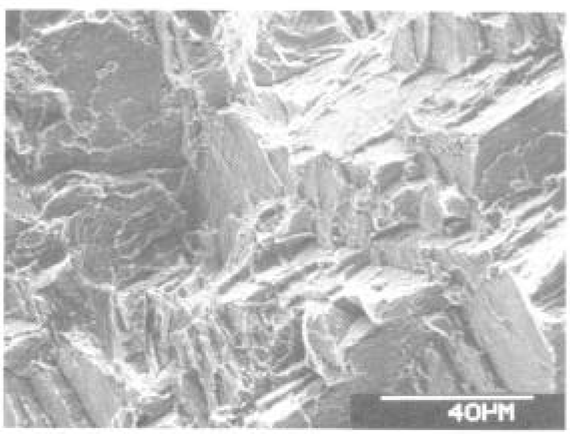

FIGURE 7 Faceted fatigue fracture of HIPped IN-738LC tested at 650C

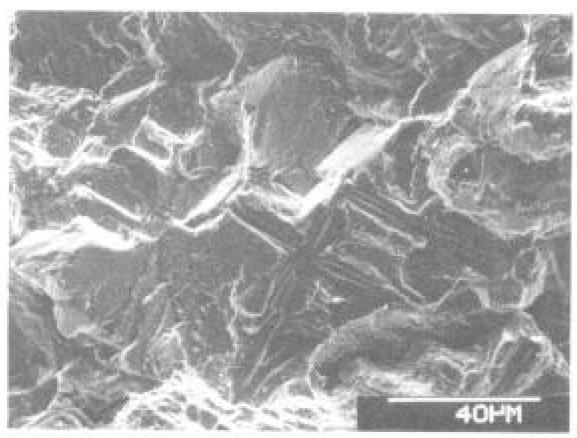

FIGURE 8 Fatigue fracture facet in D.S. IN-738LC tested at 650C

Fatigue crack propagation at 650C produced very different fracture surface appearances to those obtained at 850C. At 650C, the fracture surfaces of all the specimens displayed prominent angular facets, especially those of the HIPped material (figure 7). In the coarser D.S. structure, formed in the simulated blade casting, the dendrite arms were clearly visible in an orthogonal pattern across some facets, indicating a (001) fracture facet morphology (figure 8). At 850C, the fracture surfaces were more heavily oxidized than at 650C. The surfaces were relatively flat and featureless. The HIPped material showed mixed fracture (figure 9). In the cast alloys, however, the crack path was predominantly intergranular or, in the case of the D.S. alloys, interdendritic (figure 10). Longitudinal sectioning of these specimens revealed profuse crack branching along interdendritic paths and profuse oxidation at the crack tips.

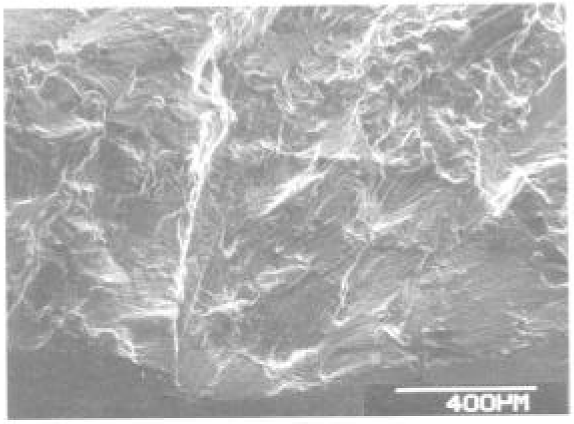

FIGURE 9 Fatigue fracture surface of HIPped IN-738LC tested at 850C

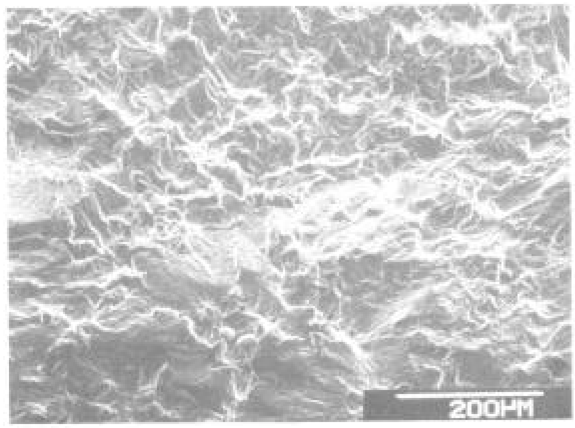

FIGURE 10 Fatigue fracture surface of D.S. IN-738LC tested at 850C

Discussion

The ranking of materials based on the results of mechanical testing is always affected by the method of testing. In the present case, it is obvious that the total strain-life representation tends to magnify the superiority of the D.S. castings, over the other castings, when compared to the other representations of figures 1 and 2. The differences in materials behaviour can be rationalized on the basis that fatigue damage accumulation is related to the inelastic strain amplitude (9), which is in turn related to the imposed total strain or stress amplitude through the cyclic stress-strain curve (10). In nickel base alloys the <001> directions are elastically soft, and the longitudinal elastic modulus of a D.S. <001> structure is expected to be approximately 62% of that of a true polycrystalline sample of the same alloy (11). This prediction agrees very well with the data of Table 1. During cycling under total strain control, therefore, the D.S. structures were able to accommodate elastically a greater fraction of the total strain than the conventionally cast structures and, thereby, provide extended fatigue life as a result of the correspondingly reduced inelastic strain amplitude. This mechanism is different to that proposed by Wright and Anderson (7), who claimed that the improved fatigue life resulted from a reduced maximum tensile stress as a result of the lower elastic modulus. In this investigation, in fact, the D.S. alloys exhibit superior fatigue behaviour even when the materials are compared on the basis of stress-life data.

The superior S/N data of the D.S. material are consistent with the data of Tien and Gamble (12), who showed that progressive elimination of porosity and script-like carbides raised the fatigue limit of Mar M-002 at 650C. Similar behaviour has been presented here, where the deleterious effect of porosity upon structural integrity has been illustrated in figures 3,4 and 5. The premature initiation of fatigue cracks at stress raisers i.e. specimen porosity, will reduce LCF life for all castings. Schneider et al.(3) have ascribed \`size effects´ in the HCF data for different bar castings of IN-738LC to the variation of porosity size with casting cross-section. In the present investigation the properties measured on cast bars were superior to those of the simulated blade casting, presumably, due to the more rapid solidification of the bars, the porosity in the bars was more finely divided. The testing method does not evaluate absolute, or even average porosity levels, since the development of a crack from a flaw is sensitive to the size of the pore and its location: porosity which is located close to the free surface of the fatigue specimen is more deleterious than more pronounced internal porosity. HIPping, in eliminating casting porosity, caused fatigue crack initiation to occur at large carbide particles. Similar crack initiation at carbide particles was also observed in some c.c. bars, and ,in the absence of porosity, fatigue crack initiation occured at the sites of the next most favorable stress raisers, the MC particles.

At 650C, fatigue crack propagation occurred in the crystallographic manner for all the materials. This mechanism has been observed previously for IN-738LC at room temperature (13) but was considered to occur only at low rates of crack growth. The present investigation has shown profuse facetting in all the alloy specimens tested at 650C, although high levels of internal porosity caused the cracks to wander and to link up these regions of porosity. From the inclination of the different variants of the facets and the dendrite arm markings on the facets it appears that they are of the (001)-type, in contrast to the report of Scarlin who proposed that they were (111)-type facets. At 850C, facets were not observed and oxidation seemed to play a much more important role in the failure process. In the c.c., materials the fracture was intergranular, with some size effect being visible in that the coarser structure displayed larger regions of fracture surface porosity. In the D.S. alloys, intergranular fracture was not practicable but crack propagation occurred interdendritically. It is probable that the mechanical compatibility across the interdendritic region will provide a lower driving force for crack growth compared to that across grain boundaries.

In the HIPped material, fatigue crack propagation occurred in a transgranular manner, in contrast to that of the other materials. The crack path links up and follows the interfaces of the large MC carbides. The HIP processing appears to have caused some agglomeration of these carbides, which are formed on casting, but does not significantly affect their spatial distribution and they remain in the prior interdendritic regions. Therefore, while the HIP processsing removes the porosity due to casting, the carbide particles are left in place and may provide a path for crack propagation.

For the purposes of this investigation, the standard heat treatment was applied to all the castings and the HIPped material. It is recognized that the effect of this thermal cycle upon the microstructure is dependent on the materials previous thermal history i.e., cooling after solidification and HIPping. In particular, the HIP cycle reduces the compositional gradients developed during casting and subsequent thermal treatment will induce considerably different gamma-prime distributions to those observed in the cast structures. Although such variations are usually regarded as having less effect on fatigue properties than on creep properties (3), the cyclic stress-strain data of figures 1 and 2 indicate that the fatigue properties are also subject to this influence. At 650C, the HIPped material data points lie below the cyclic stress-strain data for the other c.c. alloys,whereas at 850C the HIPped material shows the strongest response. Because the cyclic stress-strain response forms the link between the damage accummulation rule and the imposed strain state,the HIP cycle has a double influence upon the materials resistance to fatigue.

Conclusions

The fatigue behaviour of IN-738LC is greatly affected by porosity and carbide distribution, both of which are dependent upon casting and subsequent processing conditions. The behaviour of the conventionally cast materials was a strong function of casting size, with cast-to-size bars giving the most favorable results. Post-cast HIPping of the experimental blade structure caused this difference to be eliminated, but the stress-strain response after HIPping was markedly different to that of the cast material. Directionally solidified material displayed superior resistance to strain cycling by virtue of the soft elastic response parallel to the <001> directions. S/N data also confirm that the D.S. material performs as well as, or better than cast and HIPped material.

Acknowledgements

The authors wish to thank J. Warnke of, and R Garlock formerly of TRW for D.S. casting of the alloys, K. Willett of Westinghouse Canada for provision of the HIP facilities and M. Huber and R Stentz of Martest for the high temperature fatigue testing. Interpretation of the results was greatly facilitated by the skilled assistance of J. Prohaska.

References

1. J.E. Northwood, "Improving Turbine Blade Performance By Solidification Control" Metallurgia 46 (1979) pp 438-442
2. G.E. Wasielewski and N.R. Lindblad "Elimination of Casting Defects Using HIP" ppD1-D24 in Superalloys-Processing, MCIR Colmbus Oh.1972
3. K. Schneider, G. Gnirss and G. McGlovin, "High Cycle Fatigue of Cast Nickel Base Superalloys IN-738LC and IN-939" pp319-344 in High Temperature Alloys for Gas Turbines 1982, R.Bruneland et al. eds.; Reidel Pub. Co. Dordrecht 1982
4. R.A. Stevens and P.J.E. Flewitt, "Removal of Casting Porosity in Superalloy IN-738 Using Hydrostatic Pressure Sintering" Metal Science 14 (1980) pp81-87
5. F.L. Versnyder and E.R. Thompson, "Superalloys for Gas Turbines- Another Twenty Years" pp6.1-6.18 in Alloys for the Eighties R.Q. Barr ed.; Climax Molyb. 1980
6. D.A. Woodford and J.J. Frawley, "The Effect of Grain Boundary Orientation on the Creep and Rupture of IN-738 and Nichrome" Metall. Trans. 5 (1974) pp2005-2013
7. P.K. Wright and A.F. Anderson "The Influence of Orientation on the Fatigue of Directionally Solidified Superalloys" pp689-698 in Superalloys 1980, J.K.Tien et al. eds.; ASM Metals Park OH 1980
8. G. VanDrunen, J. Liburdi, W. Wallace and T. Terada, "Hot Isostatic Pressing of IN-738 Turbine Blades" presented at AGARD Structures and Materials Panel Florence Italy September 1978
9. L.F. Coffin and J.F. Tavernelli "The Cyclic Straining and Fatigue of Metals" Trans AIME-TMS 215 (1959) pp794-806
10. M.R. Mitchell "Fundamentals of Modern Fatigue Analysis for Design" pp385-425 in Fatigue and Microstructure M.Meshii ed.; ASM Metals Park Oh. 19979
11. C.H. Wells "The Elastic Constants of a Directionally Solidified Nickel Base Superalloy" Trans ASM 60 (1967) pp270-271
12. J.K. Tien and R.P. Gamble "The Microstructure of a PLane Front Solidified Nickel-Base Superalloy" pp1037-1041 in Proc. 2nd. International Conference on the Strength of Metals and Alloys, ASM Metals Park OH. 1970
13. R.B. Scarlin "Some Effects of Microstructure and Environment on Fatigue Crack Propagation" pp396-414 in Fatigue Mechanisms J.T. Fong ed.; ASTM Phila. 1979
14. R.B. Scarlin "Fatigue Crack Propagation in a Directionally Solidified Nickel Base Alloy" Metall. Trans. A 7A (1976) pp1535-1541