THE EFFECT OF STRESS LEVEL ON SHORT CRACK GROWTH RATES IN NIMONIC AP-1

P. Woollin*, J.M. Kendall+, J.E. King* and J.F. Knott*

The effects of mean stress level on cracks initiated at aluminosilicate inclusions and in persistent slip-bands have been investigated in powder formed alloy AP-1. Whilst little effect was observed for either type of crack, those initiated at inclusions showed a tendency to grow more slowly at a higher mean stress level, which corresponded to shorter crack lengths. The cracks initiated at inclusions also grew more slowly than those initiated in persistent slip bands. These observations are consistent with the presence of a residual stress field associated with the inclusions due to the difference between the thermal expansion coefficient of the particle and that of the matrix.

INTRODUCTION

The use of increasingly large alloying additions in order to develop the required strength levels in Nickel based superalloys has led to segregation within castings and poor forgeability. To overcome these problems, powder routes, involving hot isostatic pressing (hipping) and forging operations, have been developed for the production of turbine discs (1). The lifetime of a turbine disc is determined by its resistance to the growth of fatigue cracks from pre-existing defects. A large proportion of the fatigue life is then spent developing a crack to a stage, at which it is typically a few hundred microns in length, when it will grow in a "long crack" fashion. During powder production a number of refractory particles may become incorporated in the alloy powder. When carried through to the final product these may behave as crack initiation sites (2). During hipping the alloy powder and inclusions are subjected to high temperatures (≥1000°C) and strains develop in and around the particles due to differential thermal contraction. The alloy matrix has the higher thermal expansion coefficient and tends to contract onto the inclusions. At temperatures above 800°C the strains can be relaxed by diffusion and plastic flow. At lower temperatures this is not possible and residual stresses develop in the inclusions and surrounding matrix (3). The

---

* Dept. of Materials Science and Metallurgy, University of Cambridge, England
+ Dept. of Metallurgy and Materials Science, University of Nottingham, England

inclusion is held in a state of uniform hydrostatic compression which is balanced in the matrix by a compressive radial stress and a tensile hoop stress. Theoretical calculations predict that a hoop stress as high as 700 MPa may be developed at the particle/matrix interface but that at one particle radius from the interface this would decrease to around 200 MPa (4,5). If the residual stresses are of such magnitude then they will substantially modify the local stress field associated with the refractory inclusions acting purely as stress concentrators.

MATERIALS

Testing was carried out on hipped and forged Nimonic alloy AP-1 provided by Inco Alloy Products Ltd. The composition of the alloy was: Co, 16.8%; Cr, 14.8%; Mo,5.0%; Al, 4.0%; Ti, 3.5%; W, 0.5%; Zr, 0.04%; C, 0.02%; Ni, Balance (figures in weight %). Two batches of material were used: first, a production standard alloy manufactured from clean powder; second "doped" material which had been manufactured from alloy powder doped with 0.188 g of crushed aluminosilicate refractory per Kg of alloy powder. All of the aluminosilicate particles observed in the doped material were of less than 300 microns in any dimension. Both batches of material were solution treated below the gamma prime solvus (approximately 1140°C) to give a mixed grain size and subjected to a two stage aging process. A few coarse grains of 30-40 microns diameter were surrounded by finer recrystallised grains of 10-12 microns diameter. The 0.1% proof stress of the alloy was approximately 1020 MPa. A typical microstructure, showing the coarse undissolved gamma prime and the mixed or necklace grain structure is shown in Figure 1 whilst Figure 2 shows a refractory particle.

EXPERIMENTAL

Testing was carried out on smooth bar specimens of cross section 10mm x 10mm in four point bending under a sinusoidal loading waveform of frequency 40 Hz. and at a loading ratio, $R = 0.1$. Tests were performed in air at 20°C. Both batches of material were tested at maximum top surface stress levels of 895MPa. The doped material was also tested at a maximum stress of 450MPa. When calculating stress intensity values for the cracks it was assumed that they were of semi-circular profile (6). Breaking open the specimens revealed that the longest cracks were approximately semi-circular at the end of the test. Chemical analysis was carried out using a LINK systems EDAX.

RESULTS

(a) Stress Level Effects In the Undoped Material

Figure 3 shows the results of testing the undoped material at a top surface stress of 895 MPa. These short cracks were initiated from persistent slip-bands at the surface of the bars and compare well with earlier results reported for similar cracks but at a higher maximum stress of 980 MPa, which is close to the yield stress of the alloy. In common with short cracks in many materials the growth rates are higher than for corresponding long cracks at a load ratio of 0.1 (7).

(b) Stress Level Effects in the Doped Material

In Figure 4 the crack growth data for cracks initiated at the refractory particles at two different stress levels, are shown and compared with the data for slip-band initiated cracks. Testing at two different stress levels provided an opportunity to observe any mean stress level effect over a much greater range of stress than would be possible without the presence of such crack initiators, as cracks would not initiate from slip-bands at stresses as low as 450 MPa. The two sets of data from the doped material appear to fall within much the same scatter band. However, where the two data sets overlap in applied ΔK the lower crack growth rates were recorded at the higher applied stress level. These correspond to shorter crack lengths than at the lower stress level.

(c) Comparison of Particle and Slip-Band-Initiated Cracks

There is considerable overlap between the growth rates of cracks initiated at inclusions and those initiated in persistent slip bands. However, at the lower applied ΔK levels i.e. at the shorter crack lengths, the inclusion initiated crack data fall increasingly below the slip band initiated crack data. At the higher ΔK levels the growth rates for particle-initiated cracks lie above those for slip-band-initiated cracks but both sets tend to merge with the long crack data above a ΔK of 15 $MPam^{\frac{1}{2}}$.

DISCUSSION

(a) Mean Stress Effect on Slip Band Initiated Cracks

The crack growth rate data from slip band initiated cracks at the two different stress levels fall into a single scatter band. Therefore, at these relatively high stress levels there appears to be no significant effect of mean stress. In both cases the maximum applied stresses (895 MPa and 980 MPa) are below the 0.1 % proof stress of the alloy (approximately = 1020 MPa) and so in each case general yield of the top layer of the bar is avoided. Hence, only localised yielding in favourably oriented grains or at stress concentrators is possible. Furthermore, in this material, no bursts of static failure would be expected at the stress levels in question. It is therefore not surprising that there is no apparent mean stress effect: this supports the suggestion that the distribution of locally yielded regions is similar at the two stress levels.

(b) Mean Stress Effect on Particle-Initiated Cracks

The presence of relatively large refractory particles in the doped material has enabled cracks to initiate at much lower stress levels than would otherwise have been possible. This has allowed the comparison of crack growth rates over a wider range of mean stress level than is possible for slip-band-initiated cracks. Comparison of the data shows that overlap between the two sets of data is good and that they appear to be part of the same scatter band. This indicates that there is no significant difference between crack growth at these two different stress levels, for cracks initiated at inclusions. Also fractographic examination of specimens tested at the two stress levels revealed a very similar mode of crack growth in each case. The early growth of the crack away from the particle is in a faceted mode and there is a transition to a stage II type of crack growth at higher applied ΔK levels.

(c) Comparison of Particle-Initiated and Slip-Band-Initiated Cracks

Both cracks initiated at particles and in slip bands grew at higher rates than equivalent long, through cracks and at ΔK levels below the long crack threshold (7). At applied ΔK levels of 10 $\mathrm{MPam}^{\frac{1}{2}}$ and above there was a tendency for the two sets of data from both particle-initiated and slip-band-initiated cracks to overlap. At high ΔK levels the data are beginning to merge with the long crack data (7). At the lower applied ΔK levels the cracks initiated at refractory particles show slower growth rates than those initiated in persistent slip-bands. For the cracks initiated at the refractory particles at two different stress levels it appears that where the data sets overlap, the lower growth rates are associated with the shortest cracks i.e. those whose tips are closest to the particle from which they were initiated (these were also those tested at the higher stress level). These two observations are consistent with the presence of residual stresses in and around the initiating particle. For a crack running through the particle and into the matrix the nett effect of these stresses will be compressive out to a distance of at least one particle radius away from the particle/matrix interface. This effect can be expressed as a compressive contribution to the effective stress intensity at the crack tip as shown in figure 5 (4,8). If this compressive stress intensity is superimposed upon the stress intensity due to the applied loading then it will reduce the effective loading ratio of the fatigue cycle. This reduction will be greatest when the crack tip is close to the particle and will decrease as the crack extends. Hence, for a crack initiated at an inclusion, whilst the crack tip is close to the particle it will be growing at a reduced load ratio. A reduction of the load ratio will tend to enhance any closure effects which may occur.

During the early stages, crack growth is via a faceted mode and contact of the crack faces at above minimum load is likely due to the roughness of the crack surfaces. This effect will be enhanced, leading to a reduction of the effective ΔK at the crack tip when it is close to the initiating particle. This would be expected to reduce crack growth rates, relative to cracks initiated in slip bands whilst the crack tip is close to the initiating particle, i.e. at lower ΔK levels. Similarly, this may explain the apparently lower growth rates observed for cracks grown at the higher stress level in the mid-ΔK region, as they are shorter than corresponding cracks at the lower stress level. At the higher ΔK levels the data from particle-initiated cracks lie above those for the slip-band-initiated cracks. This may be partly due to the fact that the particle-initiated cracks were able to grow to greater lengths than the slip-band cracks without interfering with each other, as they are less numerous than the slip band cracks. Hence, it is difficult to compare them in this region.

(d) "Halos" around refractory particles

Around some of the particles were regions or "halos" in which no coarse gamma prime was present and grain growth had apparently occurred to a greater extent than was generally possible (figure 2). These regions were of the order of 50-100 microns in size. Chemical analysis of these "halos" showed that typically they were rich in Aluminium relative to the matrix. This suggests that during high temperature processing the particles were able to react with the matrix. This may have lowered the gamma prime solvus temperature in the region of the "halo" thus allowing solution of the coarse gamma prime during solution treatment. Consequently, grain growth would be possible in the "halo". Purely in terms of the increased grain size in the "halo", higher short crack growth rates might be expected as has been previously reported for a coarse grained microstructure relative to a fine grained one (7). This is opposite to the effect observed so does not explain the results. However other effects of the "halo" cannot be ruled out as possible causes for the reduced growth rates observed for cracks initiated at the inclusions.

CONCLUSIONS

(1) For cracks initiated at persistent slip-bands in undoped alloy AP-1 there was no apparent effect of increasing the maximum stress level during the loading cycle from 895 to 980 MPa.

(2) For cracks initiated at refractory inclusions in the doped alloy little effect of maximum stress was observed between levels of 450 and 895 MPa. The data from the two stress levels essentially formed a single scatter band.

(3) Data from both slip-band and particle-initiated cracks were higher than those recorded for long through cracks. Both types of crack also grew at ΔK levels below the long crack threshold.

(4) The cracks initiated at refractory particles grew more slowly than equivalent slip band initiated cracks particularly at the shorter crack lengths (i.e. lower ΔK levels). This is consistent with the presence of a residual stress field associated with the particles.

(5) A variation of chemical composition and microstructure in the matrix around some of the particles was observed which indicates that they have reacted with the matrix. It is possible tht this also influences crack propagation rates in material adjacent to the particle.

ACKNOWLEDGEMENTS

Professor D. Hull F.Eng and Professor J.S.L. Leach are gratefully acknowledged for the provision of laboratory facilities at Cambridge and Nottingham respectively. Rolls-Royce plc and the Science and Engineering Research Council are thanked for financial support for two of the authors (PW and JMK) and British Gas and the Fellowship of Engineering for a third (JEK).

REFERENCES

(1) Wildgoose, P., Turner, N.G., Davies, H.F., Helliwell, B.J., Ubank, R. and Harrison, H., Powder Met. No. 2, 1981 pp 75-86.
(2) Jablonski, D.A., Mat. Sci. and Eng., 48, 1981 pp 189-198.
(3) Brooksbank, D. and Andrews, K.M., J. Iron and Steel Inst., vol 210, 1972 pp 246-255
(4) Tsubota, M., King, J.E. and Knott, J.F., 1st, Parsons Institute Turbine Conference, Trinity College, Dublin. Parsons Press, 1984, pp 189-196.
(5) James, M.N. and Knott, J.F., AGARD Conf. Proc. No. 393, 1985, pp 10-1 to 10-12
(6) Rooke, D.P. and Cartwright, D.J., "Compendium of Stress Intensity Factors":, HMSO, 1976
(7) Brown, C.W., King, J.E. and Hicks, M.A., Met. Sci., Vol. 18, 1984 pp 374-380
(8) Woollin, P. and Knott, J.F., :"Fatigue '87", Eds. R.O. Ritchie and E.A. Starke Jr., EMAS, 1987, V2, pp 1087-1097

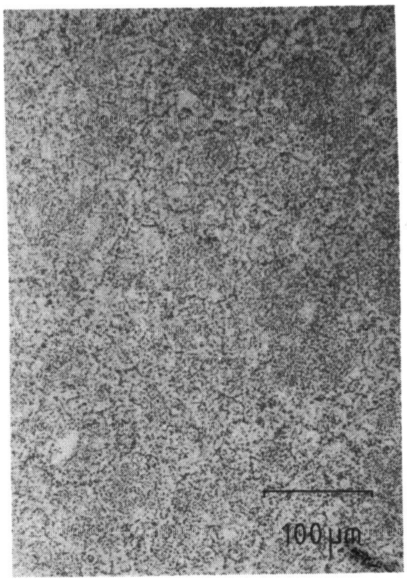


Fig. 1 Alloy AP-1
Necklace Microstructure

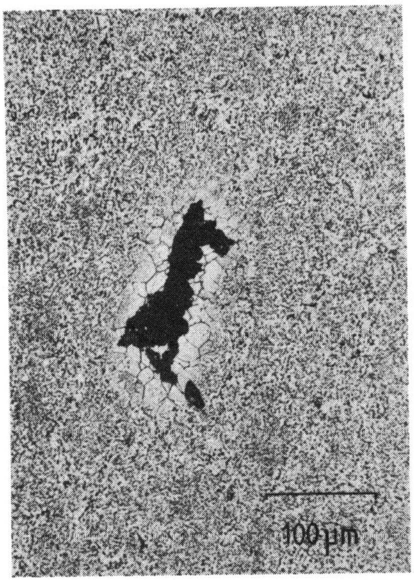


Fig. 2 Aluminosilicate Inclusion
with "halo"

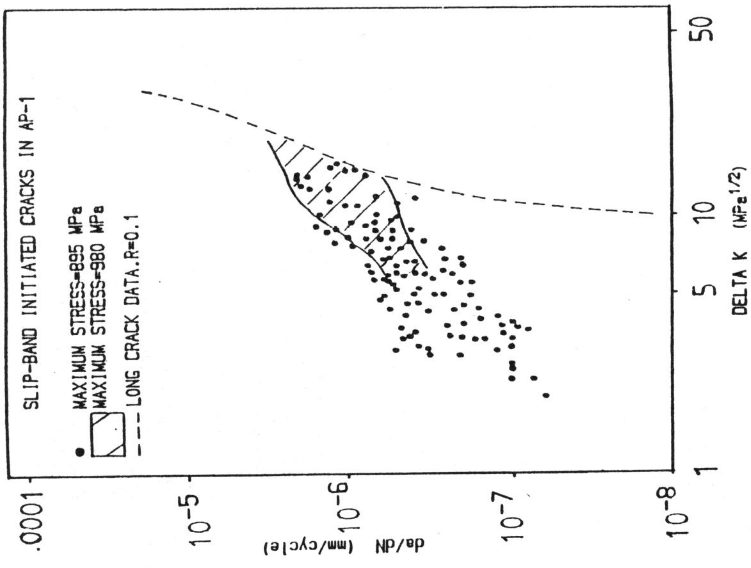


Fig. 3 Slip-band-Initiated Crack data
R = 0.1 In air, 20°C

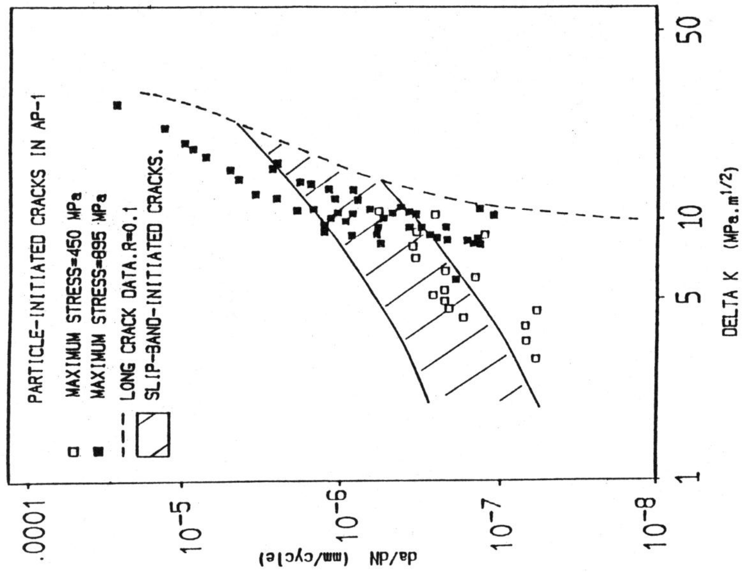
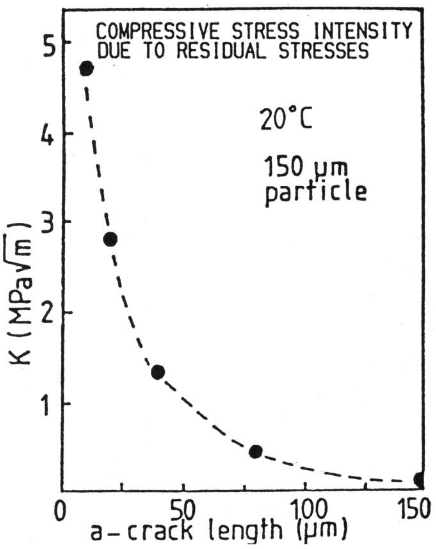


Fig. 4 Particle-Initiated Crack data
R = 0.1 In air, 20°C



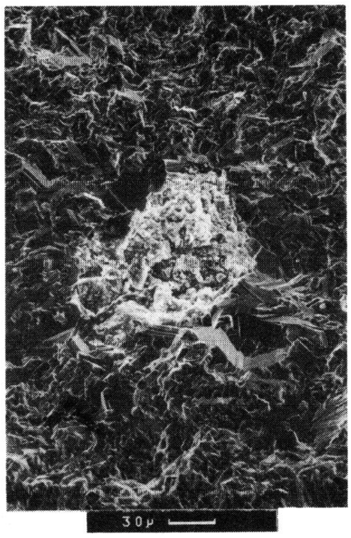


Fig. 5 Theoretical stress intensity due to residual stresses (8)

Fig. 6 Faceted crack growth away from initiating particle.