# Effects of Temperature and Environment on the Long Crack Growth Behaviour of Inconel 718

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## ABSTRACT

The long crack fatigue behaviour of Inconel 718 has been investigated in vacuum over the temperature range 20°C to 650° C and in both air and vacuum environments at 20°C. The behaviour of conventionally treated material with a 30-40 micron grain size and that of direct aged material with a fine 3-5 micron grain size have been compared.

The coarse grained material showed substantial increases in crack growth rates and reduced threshold values as the temperature was increased. This is associated with a decrease in the roughness of the crack surface as the temperature increases. The fine grained material, which exhibits much smoother crack surfaces, showed little change in crack growth rates and threshold value as the temperature increased.

In air at 20°C both structures exhibited very similar crack growth data. However the fine grained material showed little change in crack growth mode from that observed in vacuum whereas the coarse grained material showed a much reduced facet size when tested in air.

## KEYWORDS

Inconel 718; Fatigue Crack Propagation; Temperature; Environment; Faceted Crack Growth.

## INTRODUCTION

Inconel 718 is a precipitation hardening nickel-iron based superalloy which has been used extensively in a number of gas turbine applications, in particular as a high pressure turbine disc material. The lifetime of a turbine disc is determined by its resistance to the development and subsequent growth of fatigue cracks either from pre-existing flaws or from microstructural features of the alloy. Once a crack has developed to a size of several grain diameters it tends to behave in a "long crack" fashion.

In service a turbine disc is subjected to a wide range of temperatures. In steady flight the temperature at the bore of the disc may be as low as 200°C whilst at the rim, near to the hot gas stream, it may reach or exceed 600°C. Cracks growing at the surface of a disc will be exposed to the air whilst subsurface defects will grow in a reduced pressure environment.

In order to understand the growth of fatigue cracks from their initiation to the critical size, both the early stage of "short crack" growth and the later "long crack" growth need to be considered.

The long crack growth behaviour of two batches of Inconel 718 with different grain sizes have been investigated over a range of temperatures in vacuum and in both air and vacuum at 20°C.

## MATERIALS

Material was taken from sections of two Inconel 718 turbine disc forgings made from double vacuum melted billets. The composition of the alloy is shown below in Table 1.

Table 1. Composition of Inconel 718 (Weight %)

| Ni | Cr | Fe | Ti | Al | Mo | Nb | C |
|---|---|---|---|---|---|---|---|
| 51.5 | 19.0 | 20.0 | 0.9 | 0.6 | 3.0 | 5.0 | 0.04 |

The two disc forgings used were:

(a) A conventional forging which was subsequently solution treated and then aged. This was fully recrystallised and had a grain size of 30-40 microns. This conventional microstructure is shown in Fig. 1.

(b) A disc forging which had been subjected to a "direct ageing" process which involved no further solution treatment between the forging and ageing procedures. This material was not fully recrystallised during the forging process and had a grain size of 2-5 microns. Its microstructure is shown in Fig. 2. Both batches of material were given the same double ageing treatment:

720°C (8 hours) → 620°C (For total ageing time of 18 hours)

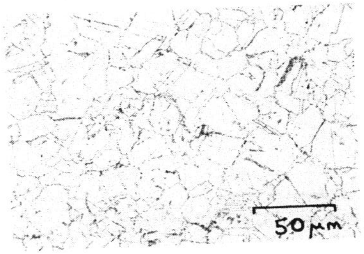


Fig. 1.

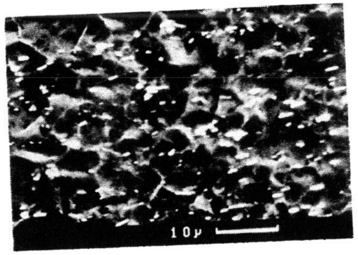


Fig. 2.

Fig. 1. Conventional microstructure (coarse grain size)

Fig. 2. Direct aged microstructure (fine grain size)

## EXPERIMENTAL

Testing was carried out on through cracked SEN bend specimens in four point bending. These were cut from the discs such that crack propagation occurred in the radial direction on a radial plane. The specimen dimensions were: depth W=15 mm, thickness B=12.5 mm and loading span S=15 mm. Testing took place inside an environmental chamber mounted on a servohydraulic testing machine of 30kN capacity) so that it could be conducted either in air or at a reduced pressure of < $1 \times 10^{-6}$ mbars (later referred to as vacuum). Tests were conducted at temperatures of 20°C, 550°C and 650°C under vacuum and at 20°C in air.

The specimens were subjected to cyclic loading at a frequency of 40 Hz at the lower two temperatures and 5 Hz at 650°C. The reduced frequency was employed at 650°C because of the greater elastic deflections of the specimen required to apply the loads at that temperature. In all cases the loading ratio (R) was equal to 0.1. Crack lengths were measured by a D.C. potential drop technique using a current of approximately 30 Amps.

A load shedding (and subsequent increasing) sequence, involving 5% load reductions (or increments) after a crack extension of twelve times the previous maximum plastic zone size, was used. This was controlled by a computerised feedback system inorder to allow the use of a reproducible loading sequence.

## RESULTS

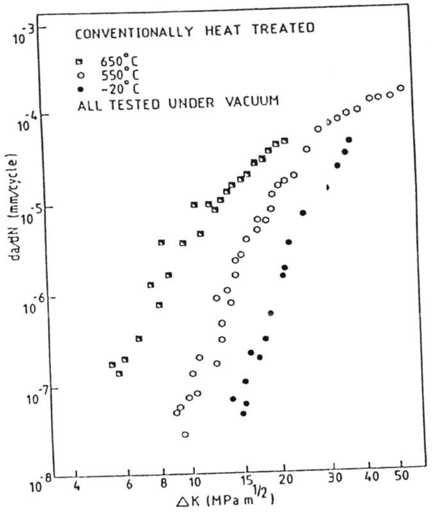
### (1) Conventional Material. Effect of Temperature

Figure 3 shows the crack propagation behaviour of the conventionally processed alloy at the three testing temperatures in vacuum. With increasing temperature the crack growth rates increased significantly and the corresponding long crack thresholds decreased from 13.7 MPam$^{1/2}$ at 20°C to 8.4 MPam$^{1/2}$ at 550°C and to 5.2 MPam$^{1/2}$ at 650°C.

The differences in crack propagation rate are greatest in the low $\Delta K$ regime. The data from all three temperatures tend to merge at the higher $\Delta K$ levels.

### (2) Direct Aged Material. Effect of Temperature

Figure 4 shows the crack growth rates from the fine grained, direct aged material. Here all three sets of data fall very close to each other and very similar threshold values were recorded at the three temperatures (7-8 MPam$^{1/2}$). Any effect of temperature on growth rates and threshold values is not sufficiently large to be observed given the amount of scatter associated with the data.



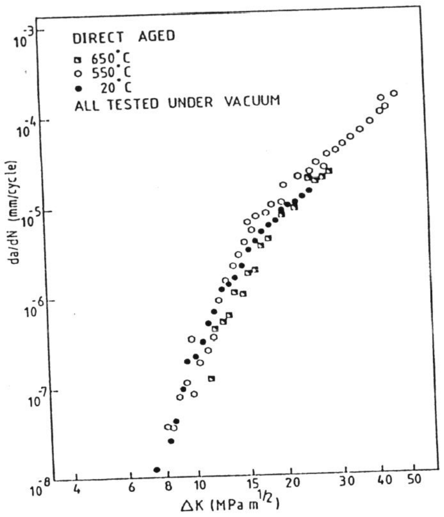


Fig. 3.

Fig. 4.

Fig. 3. Effect of Temperature.
Coarse grained material. In vacuum.

Fig. 4. Effect of Temperature.
Fine grained material. In vacuum.

### (3) Effect of Environment at 20°C

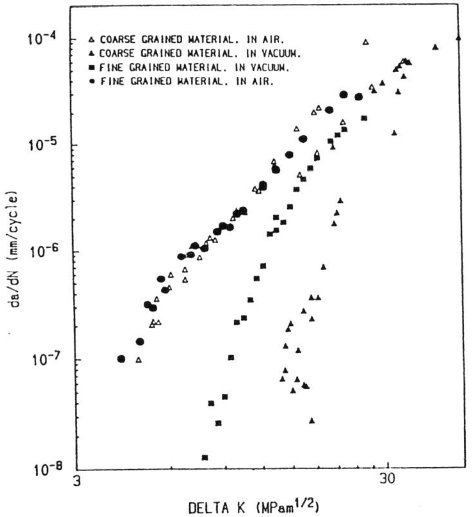
Figure 5 shows a comparison of crack propagation rates in the two types of material at 20°C under vacuum. The coarser grained, conventionally processed alloy shows considerably lower growth rates and a higher long crack threshold than the finer grained, direct aged material. However, when the tests were carried out in air the data from the two types of material showed very close agreement. In both cases the long crack threshold value fell to approximately 5.0 $\mathrm{MPam}^{1/2}$ which is considerably lower than the values measured in vacuum. Similarly, the crack propagation rates were increased relative to measurements made in vacuum, particularly at the lower applied ΔK levels.

## DISCUSSION

### (1) Effect of Temperature on the Coarse Grained Material

Over the temperature range 20°C - 650°C, Inconel 718 shows only a small reduction in yield strength and elastic modulus (approximately 15 % reductions in each (King, 1986)). Therefore, the corresponding increase (approximately 40 %) in the crack-tip opening displacements (CTODs) with increasing temperature is not sufficient to explain the rise in crack growth rates observed, although it will contribute to some extent.

Fractographic examination of the fatigue crack surfaces of the conventionally treated material revealed that at near threshold ΔK levels a faceted crack growth mechanism was operative and at higher ΔK levels there was a transition to a typical stage II type of crack growth which showed regions of clear fatigue striations (Fig. 6). At 20°C and 550°C in particular, large facets of approximately the grain size were formed. At 650°C the facets were much less numerous and less angular in appearance (Fig. 7).



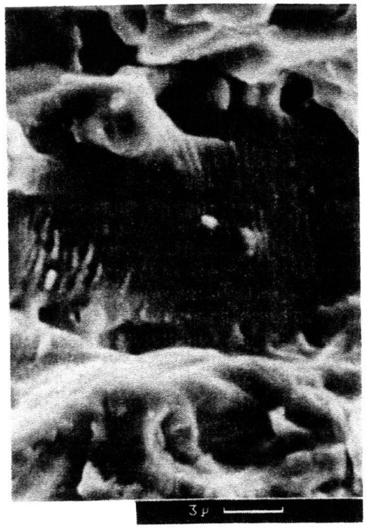


Fig. 5.

Fig. 6.

Fig. 5. Effect of environment at 20°C.
Both microstructures.

Fig. 6. Fatigue striations in coarse grained Inconel 718.
High ΔK.

The formation of planar facets during fatigue crack growth has been reported in a number of alloys and is common amongst precipitation strengthened nickel based superalloys (King, 1986, Hicks and King, 1983, Antolovich and Jayaraman, 1983) due to their planar slip characteristics. Their formation has been linked to the confinement of plastic deformation at the crack-tip to a single slip system within a single grain. It has been suggested that the transition between faceted and striated crack growth occurs when the reversed plastic zone size at the crack tip is equal to the grain size (King, 1986). This criterion would predict that the transition should occur at an applied $\Delta K$ of between 20 and 30 $MPam^{1/2}$ for the coarse grained, conventionally treated material. This corresponds well with the observed position of the transition at 20°C and 550°C although at 650°C the transition was not well defined. It is of interest to note that the transition appeared to occur at higher $\Delta K$ levels at the edges of the specimen, where the extent of crack-tip plasticity is greatest, contrary to the predictions of the above criterion.

At room temperature Inconel 718 shows planar (or heterogeneous) slip characteristics due to the presence of very fine, shearable gamma double prime precipitates. With increasing temperature cross-slip and climb processes can occur more easily. This leads to less planar slip and a generally more homogeneous slip distribution. This will have the effect of reducing the tendency for crack growth to occur along single planes and form well defined facets. Between 20°C and 550°C there was little change in the appearance of the crack surfaces generated. At 650°C the crack surfaces were generally much flatter than those generated at the lower temperature although the crack growth mechanism was still transgranular and lead to angular surface features.

At the lower temperatures roughness induced closure effects (Suresh and Ritchie, 1984), due to contact of the highly faceted crack faces at above the minimum load of the fatigue cycle, is likely to be important in contributing to the relatively high threshold values and low crack growth rates observed. As the crack surfaces become smoother these effects are likely to be decreased, particularly at 650°C. This is consistent with the increases in crack growth rates observed; firstly from 20°C to 550°C when there is a small reduction in the angularity of the crack surfaces generated and secondly the larger increase in growth rates between 550°C and 650°C when the crack surfaces become much flatter.

At high $\Delta K$ levels, when crack growth is via a ductile striation formation mechanism at all three temperatures, the data tend to merge. Ductile striation formation leads to a much flatter fatigue crack surface than the facet formation mechanism operating at near threshold $\Delta K$ levels. Roughness induced closure would be expected to be much less important in this regime due to the flatter crack surfaces and increased CTODs. An increase in temperature would only be expected to influence crack growth in this regime via the associated decrease of yield strength and modulus. As these properties decrease by approximately 15 % each over this temperature range they would be expected to lead to a relatively small increase in crack growth rates. This is consistent with the merging of the data at high $\Delta K$ levels.

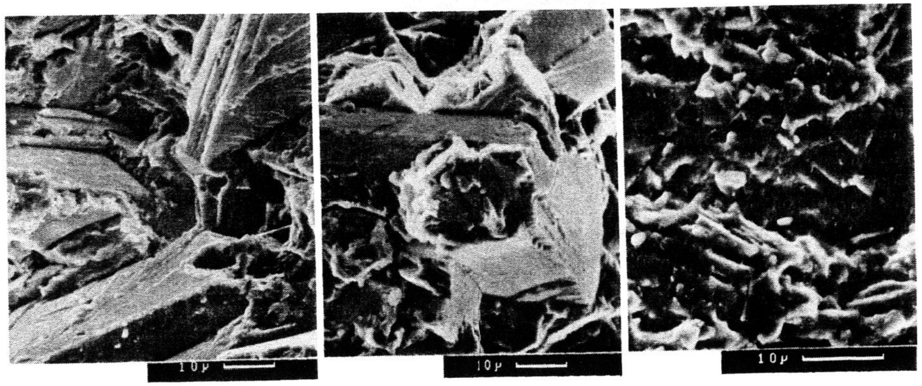


Fig. 7. Faceted crack growth in coarse grained material, at near threshold $\Delta K$ levels. In vacuum.
a) 20°C
b) 550°C
c) 650°C

### (2) Effect of Temperature on the Fine Grained, Direct Aged Material

Essentially very little change in crack growth rate was observed over the range 20°C to 650°C in the direct aged material. Fractography of the crack surfaces generated at the three temperatures revealed a ductile transgranular mode of crack growth in each case (Fig. 8). Using the plastic zone size criterion for predicting the position of a possible transition to faceted crack growth would suggest that this should occur at a $\Delta K$ level of between 6 and 10 $MPam^{1/2}$. In practice no clear transition was observed but in the near threshold regime a number of angular features were present. These were very fine and could not be classed as planar facets of a type similar to those formed in the coarse grained material. This is a reflection of the reduced slip planarity expected in the finer grained material (Antolovich and Jayaraman, 1983) which will reduce the tendency of the material to form facets.

Hence in the finer grained material the roughness induced closure effects which are important for the coarse grained material would be expected to be less significant.

Again due to the slow fall off of yield strength and elastic modulus with temperature the CTODs will only be increased by approximately 30-40 % between 20°C and 650°C and a similar increase in growth rates might be expected. In fact the data from the three temperatures fall within a single scatter band and there is little apparent temperature dependence over this range, even down to the long crack threshold level. This may reflect some other effect of temperature which counteracts that of the reduced yield strength and elastic modulus, such as a variation in the degree of slip reversibility with temperature. Alternatively, this may just be due to scatter in the data.

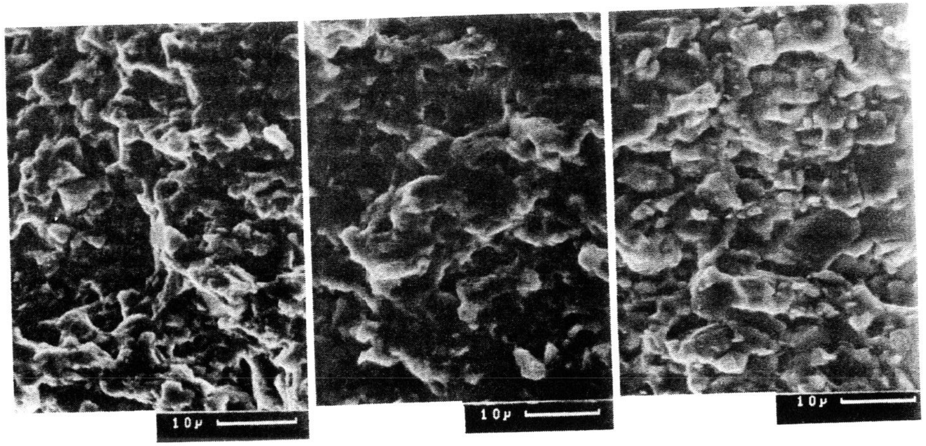


Fig. 8. Near threshold crack surfaces in fine grained material. In vacuum.
a) 20°C
b) 550°C
c) 650°C

### (3) Effect of Environment at 20°C

Both the fine and coarse grained material show increased crack growth rates and decreased threshold values in air when compared to corresponding data obtained under vacuum. They also show very similar results in air, unlike their behaviour under vacuum when the fine grained material showed the higher growth rates.

The increased growth rates observed in air relative to vacuum are typical of most materials tested in air and vacuum when oxide induced closure effects are relatively unimportant. This is a consequence of adsorption of gas molecules at the crack-tip which leads to reduced slip reversibility on the active slip planes and hence causes increased crack growth rates in air (Kendall and Knott, 1984).

Fractographic examination showed that the fine grained material had an increased tendency to produce angular features at the lower $\Delta K$ levels in accordance with the predicted striated to faceted crack growth transition (occuring at a $\Delta K$ of between 6 and 10 $MPam^{1/2}$ (Fig. 10)). The crack surface was however very similar to, and no more rough than, those generated in vacuum.

The coarse grained material still showed highly planar facets (Fig. 9) but these were now much smaller than the grain size and it could be seen that in some grains crack growth had alternated between two or more slip planes leading to a very jagged type of crack growth. Apparently the air environment had encouraged crack growth to branch from one slip plane to another within a single grain rather than remain on one specific plane across a whole grain as had occured under vacuum. Although the crack surface was still highly faceted it was now faceted on a finer scale than when tested under vacuum. Therefore roughness induced closure effects would be expected to be reduced relative to those tests carried out under vacuum.

Although the crack surfaces from the two types of material appear very different they both correspond to transgranular crack propagation of a planar type and the very similar crack growth rates observed suggest that roughness induced closure effects are of similar magnitude for both grain sizes. Due to the fine scale of the crack surface features it is likely that the closure effects are small in both cases.

Fig. 9.

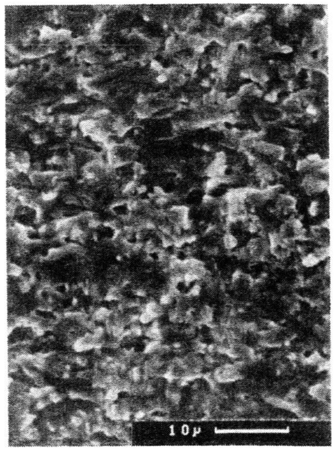


Fig. 10.

Fig. 9. Near threshold crack surfaces in the coarse grained material tested at 20°C, in air.

Fig. 10. Near threshold crack surfaces in the fine grained material tested at 20°C, in air.

## CONCLUSIONS

(1) The coarse grained, conventionally treated material showed increasing growth rates with increasing temperature between 20°C and 650°C under vacuum. Crack growth was via a facet formation mechanism. The increase of crack growth rates was linked to a reduction of roughness induced closure effects with increasing temperature.

(2) The fine grained, direct aged material showed very little effect of

temperature upon crack growth under vacuum. The crack surfaces had a more ductile appearance and were much smoother than those from the coarse grained material. Hence roughness induced closure effects were much less significant for this fine grained material.

(3) Both types of material showed faster crack growth rates in air than vacuum at 20°C. This is due to adsorption of gas molecules at the crack-tip leading to reduced slip reversibility and increased crack growth rates.

(4) The two grain sizes show very similar crack growth rates at 20°C in air. The coarse grained material showed finer scale crack surface features in air than in vacuum which may have lead to reduced roughness induced closure effects in air.

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## REFERENCES

Antolovich, S.D. and N. Jayaraman (1983). In: Fatigue: Environment and Temperature Effects (J.J.Burke and V.Weiss,eds), Proc. 27th Sagamore Army Materials Research Conf. (1980). Plenum Press.

Gell, M. and G.R. Leverant (1973). Fatigue at Elevated Temperatures. ASTM STP 520, 37-67.

Hicks, M.A. and J.E. King (1983). Int. J. Fatigue, 5, No. 2, 67-74.

Kendall, J.M. and J.F. Knott (1984), In: Fundamental Questions and Critical Experiments on Fatigue, Dallas, Texas.

King, J.E. (1982). Fatigue of Eng. Mater. Stru. 5, No. 2, 177-188.

King, J.E. (1986). In: Mechanical Behaviour of Superalloys , The Royal Society, London.

Pedron, J.P. and A. Pineau (1982). Materials Science and Eng., 56, 143-156.

Sadananda, K. and P. Shahinian (1981). Met. Trans. A, 12, 343-351.

Suresh, S. and R.O. Ritchie (1984). In: Fatigue Crack Growth Threshold: Concepts, (D.L.Davidson and S.Suresh, eds). TMS-AIME, Warrendale, PA.