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The Growth of Fatigue Cracks in a Nickel Base Single Crystal

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ABSTRACT The growth of fatigue cracks in a precipitation strengthened single crystal superalloy has been investigated using a specimen design with a unique crystallographic orientation. Testing was limited to room temperature, with peak and mean stress being the primary variables.

In addition to providing essential data for component lifing purposes, the work has provided valuable information which can be used in modelling short fatigue crack growth in fcc materials which also occurs on specific slip planes. In particular it examines whether the rate of fatigue crack growth across a single grain can be characterized by conventional linear elastic fracture mechanics.

Introduction

Significant improvements in the creep strength of conventionally cast nickel based superalloys can be made by directional solidification. Further improvements in strength can be made by suitably modifying the alloy chemistry, and casting as the single crystal form. Components, such as turbine blades and nozzle guide vanes, designed in such high strength, high temperature materials, however, are not just loaded under conditions of static creep, and may often be subjected to significant fatigue loads during service.

The work described in this paper, is part of an ongoing programme aimed at gaining a fundamental understanding of fatigue crack growth phenomena in fcc (face centred cubic) single and poly crystalline materials, which are currently under consideration for several aero-engine applications. Load ratio and peak stress effects were explored as the primary variables.

In view of the crystallographic nature of fatigue cracking in fcc superalloy single crystals (**1**), it is pertinent to examine previous work on faceted fatigue crack growth in single and poly-crystals. In general, two classes of materials might be expected to support extensive crystallographic fatigue crack growth, namely (i) single phase materials of low stacking fault energy, and (ii) two phase materials with coherent precipitates.

The term 'persistent slip band' was first introduced by Thompson *et al.* (**2**) in studies of copper single-crystals oriented for single slip. In general, such slip bands and their co-planar fatigue cracks are limited in length. Two phase, precipitation strengthened materials support more extensive cracking, and it is

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in this area that the majority of research has taken place. Extensive faceted fatigue crack growth occurs as a consequence of either the dissolution of sheared precipitates or by the destruction of precipitate order on the operating slip planes. Materials which have been investigated include nickel based superalloys in equiaxed, columnar grained, and single crystal forms **(1)**(3)–**(10)**, aluminium–zinc–magnesium single crystals **(11)**–**(13)**, copper–cobalt **(14)**, copper **(15)**, and nickel–chrome austenitic stainless steel **(16)**. Fatigue crack growth has been observed predominantly on {111} in fcc materials, however, growth on {100} has also been reported by Sadananda and Shahinian **(4)(5)** in Udimet 700 in air and vacuum at room temperature and 850°C, and by Crompton and Martin **(10)** at 600°C. Vincent and Remy (7) showed the fracture plane to be temperature dependent, and King **(8)** indicated changes in growth mechanism at room temperature with applied stress intensity. King concluded that under conditions of predominantly planar slip (low temperature, restricted plasticity), octahedral slip on {111} was favoured and that more uniform deformation, as might be promoted by high temperature or extensive crack-tip plasticity, resulted in cube slip on {100}. Previous work **(1)** on the single crystal alloy under study here reported only {111} planar slip and fatigue crack growth.

The single crystal provides a means of examining crack growth within a grain, something very difficult experimentally in commercial polycrystalline materials. Furthermore crack closure effects which are known to affect long crack growth rates at low stress intensities were expected to be minimal due to the single, flat fracture path. This may be particularly informative in examining the difference between long and short crack behaviour which several authors have attributed solely to differences in closure.

The phenomenon of fatigue crack closure in relation to short cracks has been recently reviewed **(17)(18)**. It is evident in both reviews that closure and threshold determinations for short cracks, and in particular, crystallographic short cracks are very limited.

Hicks and Brown **(19)** in a study which compared the long and short crack growth behaviour of Astroloy and SRR99 suggested that crack propagation rates in single crystals may be the limiting maximum crack growth rate which may be observed for any particularly crystal system.

Experimental details

The nickel based superalloy used in this research had the following bulk chemical composition (per cent weight): 8.5 Cr, 5 Co, 5.5 Al, 2.2 Ti, 9.5 W, 2.8 Ta, 0.015 C and the balance Ni.

Single crystal bars of 12 mm diameter were cast, with the ⟨001⟩ crystal growth direction aligned along the axis of the bars. Heat treatment consisted of a 4 hour – 1300°C solution treatment, followed by a 16 hour – 870°C ageing treatment, which produced a microstructure of fine cuboidal γ', of size approximately 0.5 μm on edge, in a matrix of γ, as shown in Fig. 1. The volume fraction of γ' was

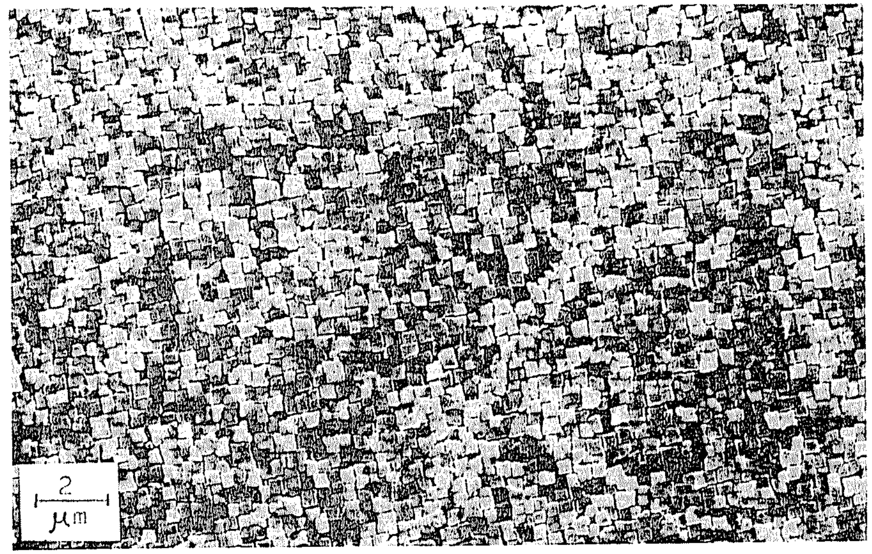

Fig 1 Alloy microstructure

80 per cent. The 0.2 per cent yield strength and ultimate tensile strength, measured in the ⟨001⟩ direction were 1000 and 1100 MPa, respectively.

A Laué back reflection technique was used to determine the precise crystallographic orientation of the single crystal bars, thus enabling specimens of specific orientation to be machined from them.

Fatigue crack growth rates were monitored in specimens of the corner notched form, shown in Fig. 2, using a standard replication technique **(20)**. The test pieces had been machined such that the four planar faces were of {001} form. The notch was ground to a depth of 0.25 mm. Prior to test the faces of the test pieces were mechanically polished to a 6 μm finish to faciliate crack length measurement.

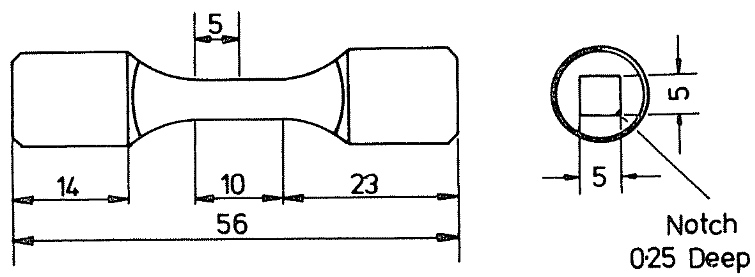

Fig 2 Specimen geometry (dimensions in mm)

Table 1 Test details

Test number	R	*Max. stress (MPa)*
1	0.1	500
2	0.3	500
3	0.5	500
4	0.8	500
5	0.8	750
6	0.1	250

Fatigue tests were performed using a closed loop servo-hydraulic load frame, at a frequency of 1 Hz. All tests were conducted in air at 20°C. Fatigue cracks were initiated at a load ratio of 0.1, i.e., until the first appearance of a fatigue crack, with subsequent crack propagation at the load ratio of interest. Cracks were monitored at intervals of at least 50 μm of crack extension. A constant peak stress was utilized for each test, during both the initiation and crack propagation phases, with the exception of test numbers 5 and 6 where a change in peak stress was necessary to produce growth rates over the required range of ΔK. A total of four load ratios (P min/P max) was investigated. Full test details are given in Table 1.

Crack growth rates were determined by dividing successive surface increments of crack extension by the number of cycles taken for each increment. The stress intensity calibration was taken from Pickard (**21**), and was a three dimensional linear elastic finite element analysis which modelled the observed fracture geometries.

Results

The propagation rates, calculated from the raw data, are presented as a function of the applied stress intensity range, and the maximum stress intensity, in Fig. 3. The crack growth rates presented are those measured along the direction of propagation (rather than those resolved normal to the applied stress axis) and the mixed mode stress intensity is similarly that at the crack tip in the direction of crack growth on the specimen surface, i.e., the (⟨110⟩(111)) system. Full details for the calculation of K are given in reference (**21**).

The growth rates measured at load ratios of 0.1 and 0.3 are equivalent when plotted against K_{max}. Data collected from tests performed at a load ratio of 0.5 are marginally slower, whilst data from a test performed at a load ratio of 0.8 is slower still. If the data is plotted against the alternating stress intensity the trend is reversed, with the R of 0.8 data being the fastest with data from tests conducted at load ratios of 0.5, 0.3 and 0.1 being progressively slower.

The observed crack growth mechanism was in all cases crystallographic, occurring in the wake of intense shear bands on {111} planes (Fig. 4). Typical fracture surfaces are shown in Fig. 5, the views being parallel to the stress axis and normal to the plane of fracture in Fig. 5(a) and (b) respectively. The

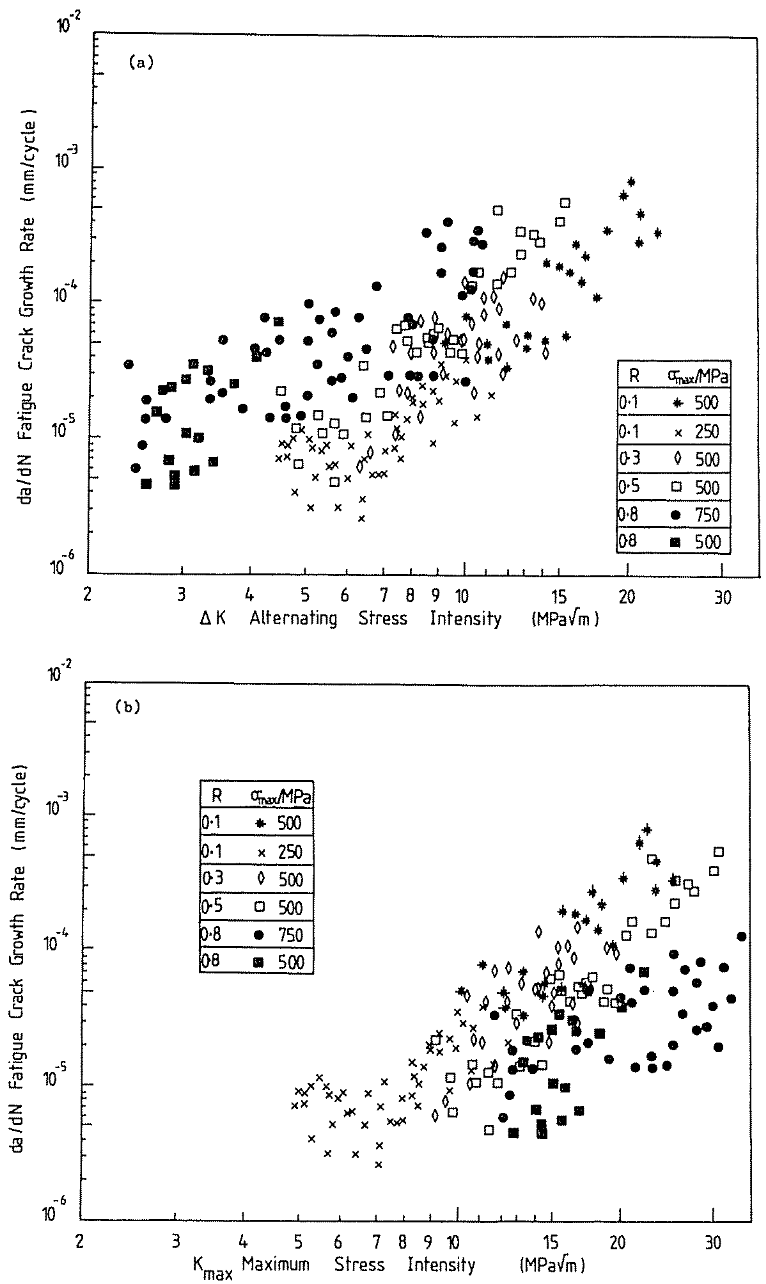

Fig 3 Fatigue crack growth rates
(a) as a function of alternating stress intensity (b) as a function of maximum stress intensity

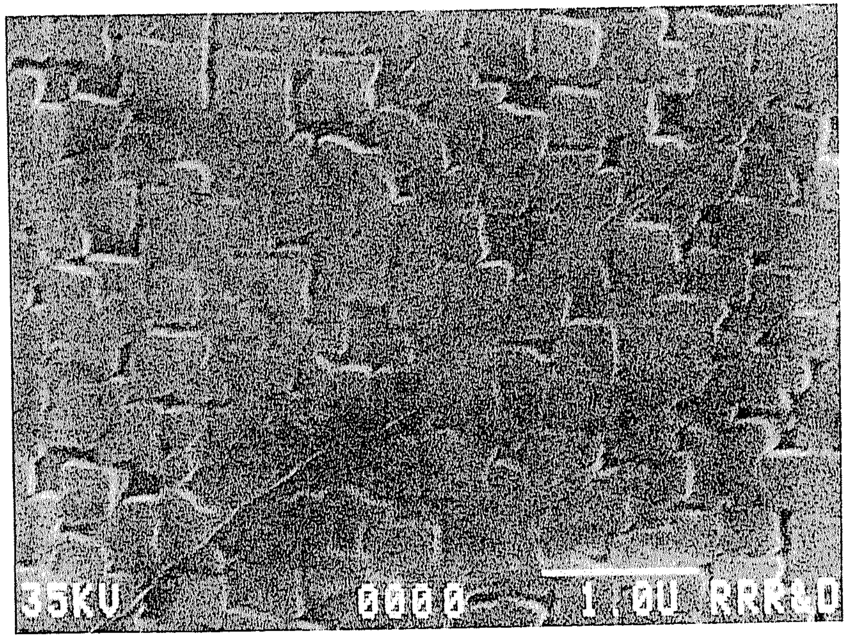

Fig 4 Deformation band ahead of the crack tip

fracture surfaces, were planar when observed macroscopically, but displayed a small degree of surface roughness when viewed by scanning electron microscopy. A similar study of high and low load ratio test pieces revealed no major difference in fracture morphology (Fig. 6).

Discussion

The propagation rate of fatigue cracks along co-planar slip bands has been monitored at four different load ratios. All have resulted in crack growth on {111} crystallographic planes. Fatigue crack growth rates have been plotted against a crack tip stress intensity factor which as been utilized previously (**1**), and which demonstrated a good correlation between the growth rates of similar crystallographic cracks growing in three separate specimen geometries, and at various stress levels between 500 and 1000 MPa.

The results show a sensitivity to mean stress which is similar to that reported for 'long' cracks in polycrystalline materials at low stress intensities. Comparison of this data with other work (**19**)(**22**) reveals the crack growth rates measured at $R = 0.1$ to be only a factor of two to three times faster than short crack growth rates measured in Astroloy, a Nickel base superalloy having a 50 μm grain size, and three to four times faster than those in ultra-fine grained Astroloy of 10 μm grain size, suggesting that within a single alloy system such as γ–γ' superalloys the growth rates measured within a single crystal represent

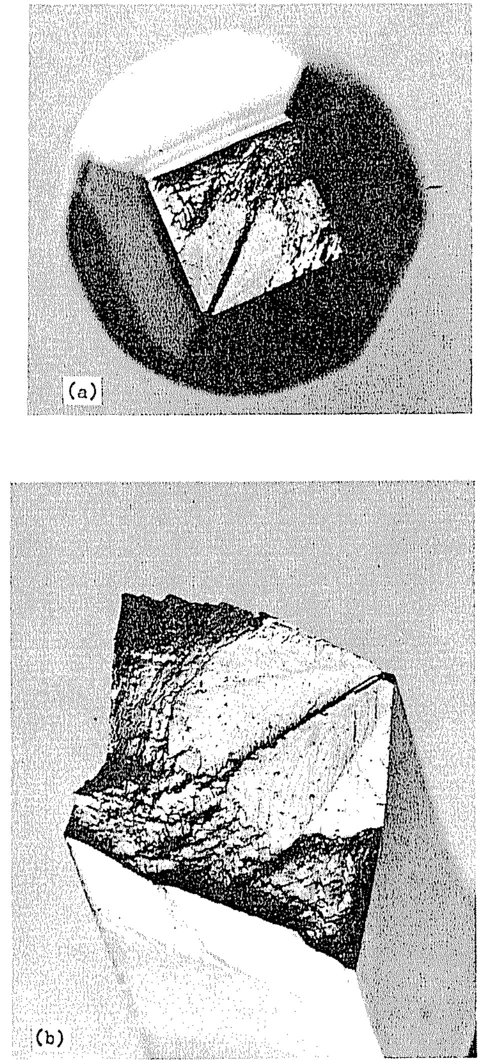

Fig 5 Fatigue fracture surface at $R = 0.1$ and maximum stress 500 MPa
(a) view parallel to applied stress axis
(b) view normal to the fracture surface

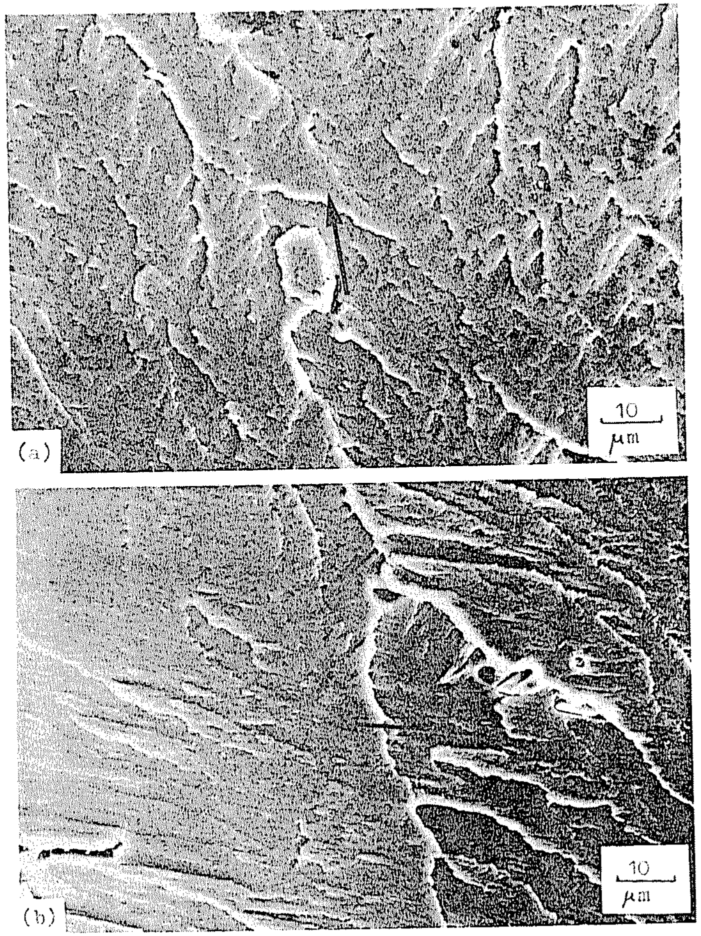

Fig 6 Fracture surface details for a maximum stress of 500 MPa; the arrow shows the direction of crack propagation
(a) $R = 0.8$
(b) $R = 0.1$

the maximum growth rate, with growth rates being reduced as the number of grain boundaries is increased. It should be pointed out, however, that there are fundamental differences in the analysis methods employed. In the case of the single crystal, growth rates and stress intensities have been suitably modified so that they are truely representative of the observed fracture path, whereas polycrystalline materials are usually treated in a manner such that standard Mode I stress intensity solutions are used, even where significant crack path tortuosity exists, and growth rates are derived from resolved crack lengths. Both of these latter factors are sources of error in the analysis of short crack growth phenomena.

One reason frequently used to account for the anomalously fast propagation rate of short cracks at a given level of ΔK is the breakdown of continuum mechanics when the crack is contained within the relevant microstructural dimension. In such cases the crack tip plastic zone is likely to be distorted away from the classical shape in favour of a shape which reflects the nature of the available deformation. The deformation zone shown in Fig. 4, and those discussed more fully by Vogel *et al.* (**11**)(**12**) for Al–Zn–Mg single crystals are clearly planar in nature, with material only a few microns either side of the deformation band being totally undisturbed. Previous work on the growth of cracks in nickel base single crystals (**1**)(**3**)–(**10**) suggested that the influence of crack length in single crystal material is very small because of the uniformity of yield strength throughout the test piece which is unaffected by weak grain orientations and grain boundaries. Hence, plasticity, although not of classical shape, will be confined to a zone which is well defined by K. It will only be at very short crack lengths, which appoach the scale of crack tip plasticity that differences in crack growth rate might be seen. This is borne out by the demonstrated independence of growth rate, at constant R and K, of maximum stress (Fig. 3).

A number of theories have been put forward to explain the effect of mean stress on crack propagation rates, the most widely accepted being that of closure. Current models of crack closure rely upon either plastic wake effects, roughness or oxidation of the fracture surfaces. By far the most commonly observed of these mechanisms, particularly at room temperature, is roughness induced closure (**23**). Contact of opposing surface asperities would inevitably result in frettage damage of the fracture faces, especially in cases where significant in plane shear is occurring. However, scanning electron microscopy of the high (0.8) and low (0.1) load ratio fatigue test pieces showed no such damage. Fine fracture surface detail from the two tests are shown for comparison in Fig. 6. The absence of frettage damage on the low load ratio tested specimen, although not conclusive proof of the absence of closure, does suggest that closure effects are not responsible for the results shown in Fig. 3. Although the growth mechanism involves a large shear component there are insufficient surface asperities to influence the behaviour of the crack tip; also one would have expected to see a better agreement with the $R = 0.5$ and $R = 0.8$ crack growth rates (Fig. 3) which, in view of the high minimum loads, should be free from closure.

The only differences in fracture morphology between the high and low load ratio tests which support a closure argument come from macroscopic observations. The $R = 0.1$ test piece shows markings (Fig. 5(b)), possibly caused by frettage along directions corresponding to ⟨112⟩ on the fracture surface, which are absent on the $R = 0.8$ test piece shown in Fig. 7.

The precise mechanism by which the accelerated growth rates at high load ratios are generated, such as crack opening response, or dislocation/particle interactions is not clear. Further work is required to examine these effects in detail, and quantify crack opening behaviour.

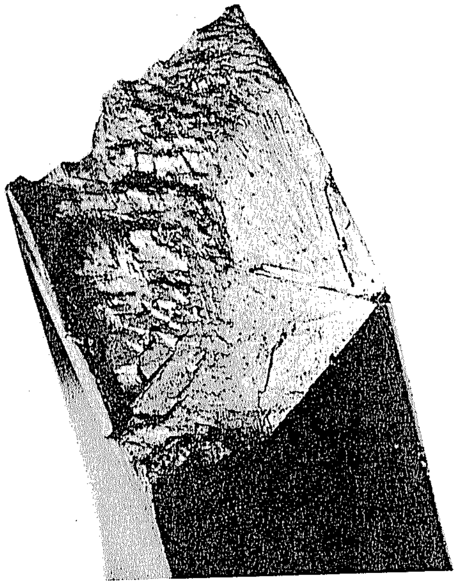

Fig 7 Normal view of the fracture surface for the $R = 0.8$ test at 500 MPa maximum stress

Conclusions

(1) Fatigue crack growth at room temperature was observed to be exclusively crystallographic, occurring on {111} planes.

(2) Presentation of the data in terms of the alternating stress intensity, ΔK, revealed a simple progression with data at $R = 0.8$ being fastest, and $R = 0.1$ being slowest.

(3) Although the observed trends in the data suggest the operation of a simple closure mechanism, the inherent smoothness of the fracture surfaces tends not to support this concept.

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