FATIGUE CRACK INITIATION AFTER DIFFERENT SURFACE TREATMENTS IN PRECIPITATION HARDENING ALLOYS

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ABSTRACT

The mechanism of transcrystalline fatigue crack initiation is discussed for $\gamma+\gamma'$-alloys (Nickel-Superalloy, austenitic precipitation hardening steel). Microscopic slip distribution affects highly the ease of crack formation. It can be varied in a wide range by the change of the microstructure by thermomechanical treatments. In addition, crack initiation is effected by surface treatments, where hard surface layers are less effective than surface deformation (by shot peening) to retard crack formation.

KEYWORDS

Fatigue; Crack Initiation; Precipitation Hardening; Surface Treatment; Extrusion and Intrusion; Shot Peening; Nitriding.

MICROSCOPIC DISTRIBUTION OF STRAIN IN $\gamma+\gamma'$-ALLOYS

Nickel based superalloys as well as austenitic precipitation hardened steels belong to the family of $\gamma+\gamma'$-alloys. Their microstructure consists of a fcc-matrix in which a high volume portion of coherent ordered γ'-phase is dispersed. A characteristic feature of such alloys is the microscopically inhomogeneous strain, which occurs under special microstructural conditions. On the other hand the microstructures can be modified in such a way that strain is almost homogeneous. This is the basis for the usual continuum-mechanical treatments of fracture. An understanding of microscopic inhomogeneity of strain is possible on the basis of the interaction of dislocations with dispersed particles and other obstacles such as sessile dislocations, which have been introduced by thermomechanical treatments. During an aging sequence the dislocation-particles interaction changes in such a way, that below a critical particle diameter the particles are sheared and above bypassed by dislocations as it was found by Gleiter and Hornbogen (1965). As a consequence the critical resolved shear stress in a particular slip plane will locally decrease ($d<d_c$), or increase for example due to the formation of Orowan-rings ($d>d_c$). Fig. 1 In addition a low stacking fault energy of the matrix will favor an inhomogeneous strain. Combination of the two factors, stacking fault energy and critical particle size, may produce a high degree of microscopic inhomogeneity of strain. The other extreme can be verified by an overaged alloy $d>d_c$ in which additional dislocations have been introduced by a thermomechanical treatment. In this case the degree of homogeneity may be such, that even by electronmicroscopical methods no discrete slip steps can be detected. Fig. 2.

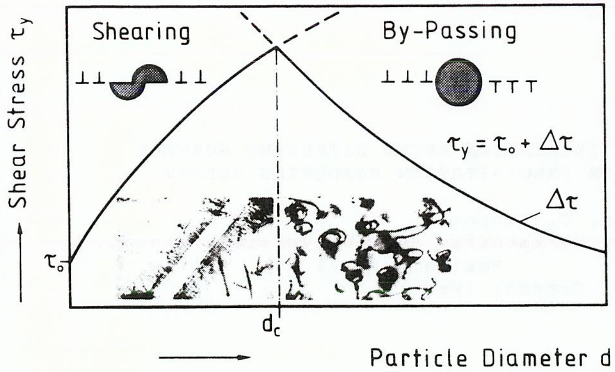


Fig. 1. Increase of c.r.s.s. due to particles and the transition from shearing to by-passing as function of particle diameter, schematic

A quantitative method has been developed by which the tendency of an alloy to deform inhomogeneously can be predicted. This method is based on the fact that the rate of decrease of critical resolved shear stress **per** number n of dislocations which pass a particle characterizes the tendency to deform inhomogeneously. For the superalloys this tendency is approximated well by the following equation:

$$\left| d\Delta\tau / dn \right| = - \frac{\gamma_{APB}}{d} \cdot f^{1/2}$$

which expresses that increasing volume portion f and decreasing diameter d of ordered particles which are associated with a certain energy of the antiphaseboundary γ_{APB}, favors inhomogeneous deformation, for $d<d_c$. This equation is based on the fact that the ordered particles are sheared by individual dislocations, which decrease the effective cross section of particle in a particular slip plane in which n dislocations have passed.

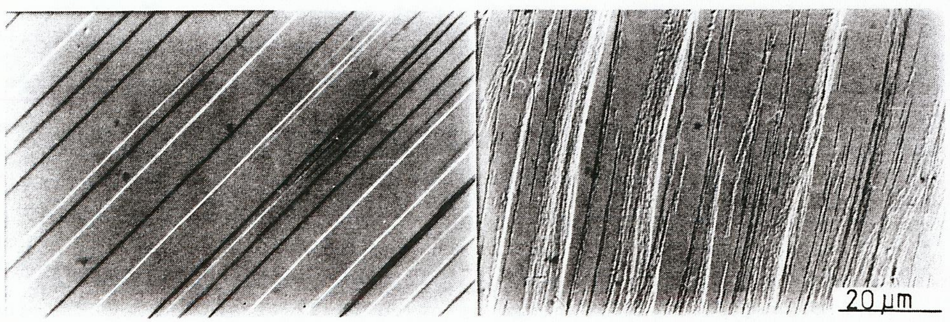


Fig. 2 Microscopic slip distribution (m.s.d.) for alloy B
a) Underaged, static strain ε = 1 %
b) Thermomechanical treated, static strain ε = 1 %

CRACK INITIATION AT SLIP STEPS

Using the fact that microscopic slip distribution in superalloys can be manipulated and varied in a very wide range, the consequences of this behavior on crack initiation will be discussed in the following investigation. It was conducted with two alloys, a nickel based superalloy (Nimonic 80 A, alloy A) and a precipitation hardened austenitic stainless steel (alloy B), with the following composition

TABLE 1

	Ni	Cr	Ti	Al	Fe	Si	Mn	C	B
ALLOY A	73,6	20,36	2,35	1,41	0,79	0,30	0,80	0,042	0,0036
ALLOY B	24,6	15,15	2,3	0,25	BAL	0,45	1,25	0,028	0,0040

The microstructure of these alloys was changed by thermomechanical treatments in order to modify the microscopic slip distribution (m.s.d.). (UA = Underaged, A = Aged, OA = Overaged, TMT = Thermomechanical Treatment). The first stage of the investigation always consisted in a characterization of the inhomogeneity of m.s.d. under static load. The consequent work was concerned with the consequences of m.s.d. on:

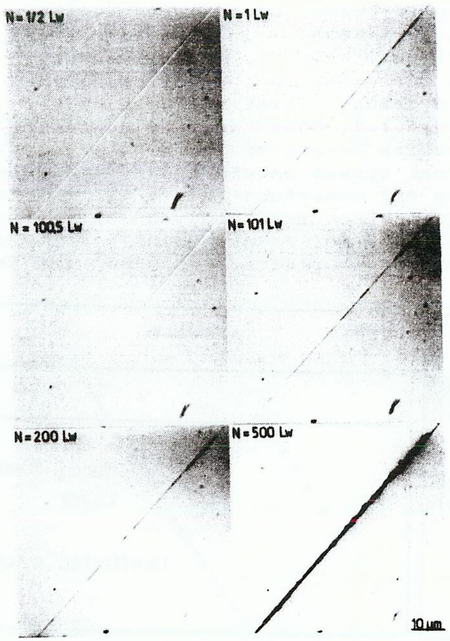
a) cyclic loading
b) the mechanisms of initiation of cracks at these steps
c) the effects of different surface treatments.

The effect of coarse m.s.d. caused by static loading may be important, if dislocations pile up in the interior of the material at embrittled grain boundaries or brittle particles at grain boundaries. However, **our** work was mainly concerned with the transcrystalline crack initiation at the surface, which is one predominant mechanism of fatigue crack initiation in this type of alloys. An investigation of m.s.d. for static and dynamic loading indicated to a good approximation a similar behavior for both types of loading. This suggests that a m.s.d. determined under static loading also can serve to predict the tendency of an alloy to initiate fatigue cracks as it was shown by Gräf and Hornbogen (1978).

In Fig. 3 the formation of a transcrystalline crack by localized deformation is shown. After the tension phase a slip step has formed which changes after the compression phase by sliding back on the active slip plane. At this stress amplitude the sliding is evidently reversible up to about 100 cycles. After 200 to 500 cycles extrusions start to grow at the site of the step.

In the underaged and aged condition high extrusions are formed (5-15 µm) at these slip steps. Their thickness can be determined by SEM (Fig. 4) by suitable tilting of their wavy sheet. Because the extrusions are only 35 nm thick they can be investigated directly by TEM applying an extractions-replica technique. The plane of the extrusion is {111}. Their surface contains rest-lines in a distance of 10-20 nm. This distance can be correlated with the stepwise growth of the extrusion per cycle.

Based on these observations a model can be derived which characterizes the tendency of an alloy to form extrusions. (Fig. 5) This tendency is determined by the number of reversible dislocations and by their tendency to spread into parallel planes. If all dislocations were reversible no permanent change of surface morphology would occur. (Case I)



In case II not all dislocations are reversible so that under pressure a thin sheet of material (Δx), the extrusion, is formed. Under tension separation can start between the cold worked extruded and the undeformed material, i. e. an extrusion can form.

Fig. 3 Slip step after a different number of cycles. Formation of slip steps.

The experiments showed that a decreasing number of cycles was sufficient to initiate a crack at these steps with increasing coarseness of slip line distribution. These cracks started at the stress concentration which were produced by the intrusions at the edges of these slip steps, Fig. 6. A quantitative evaluation of the statistical distribution of slip steps for an alloy after different treatments indicates that for a certain load a defined critical slip step height is required to initiate a crack. The frequency of occurence of this height h_c is higher for a coarse than for a fine slip distribution.

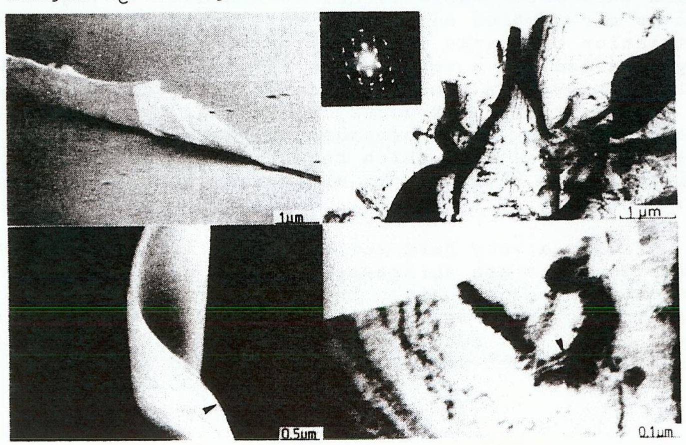


Fig. 4 Extrusion at a slip step (SEM) Extrusions are showing parallel traces (TEM)

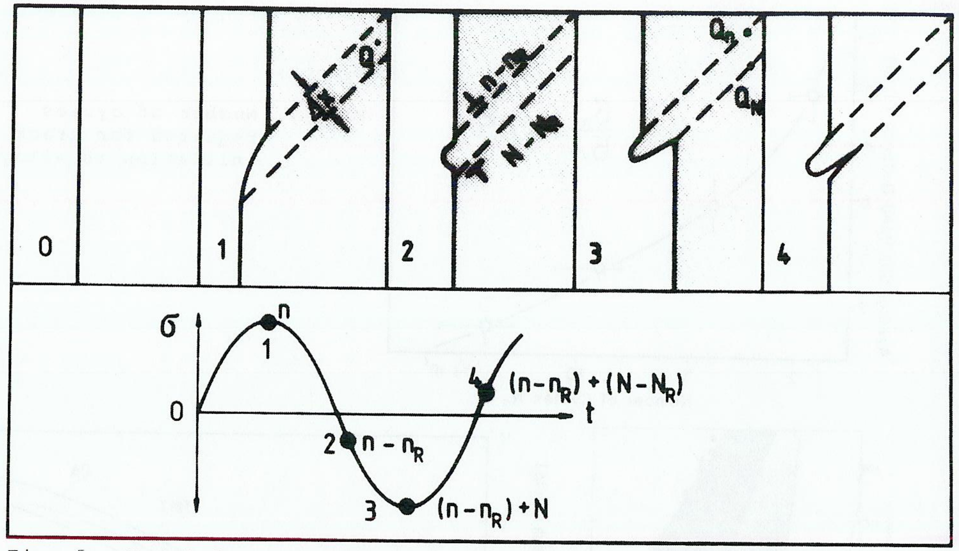


Fig. 5 Model for the formation of extrusions and cracks by cyclic load, n = dislocation form during tension, n_R = glide reversibility in compression, N = dislocations form during compression, N_R = glide reversibility in tension, $\Delta x > b$ diametre of slip band and is determined by the possibility of the dislocations to spread by cross slip from a slip plane determined by a original source Q.

However, the maxima of the distribution are usually not characterizing well this situation. An extrem-value-function (Weibull-distribution) should be used for a quantitative analysis of this situation as it was shown by Gräf and Hornbogen (1978).

It is well known from earlier investigations (Hornbogen and Zum Gahr (1976)) that the identical microstructural and micromechanical conditions which favor early formation of cracks lead to a relatively low propagation velocity and vice versa. This indicates that a strong effect on crack initiation should be expected just in the situation of coarse m.s.d, while for fine m.s.d. a surface treatment should be less effective.

The Wöhlerdiagram, Fig. 8 indicates for coarse m.s.d. a early formation of slip steps, extrusions and cracks, but the growth rate indicated by the da/dN-curves of this condition is lowest of all heat treatments investigated.

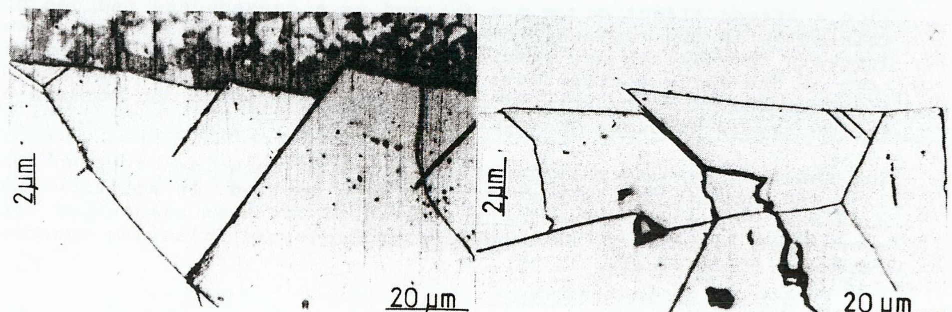


Fig. 6 Cracks are initiated at surface steps. Light microscopy, taper sectioning.

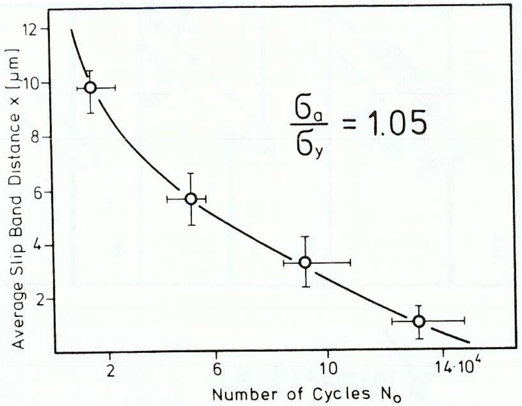


Fig. 7 Number of cycles required for crack initiation in alloy A

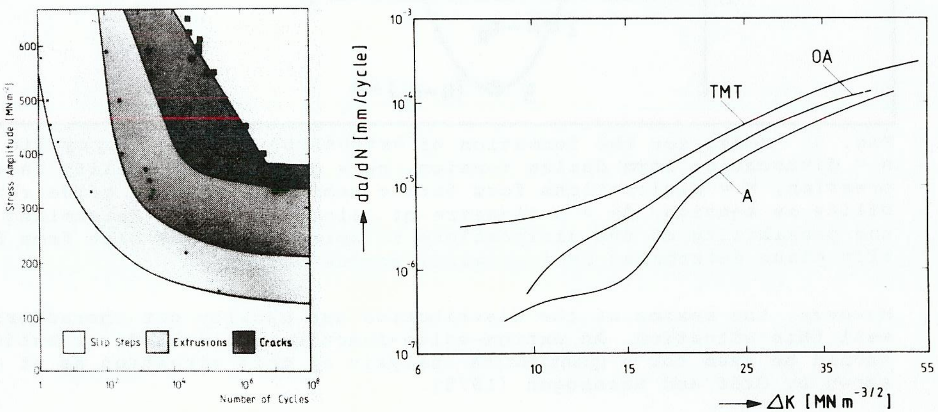


Fig. 8 a) Wöhler-diagram, also showing the formation of slip steps, extrusions and cracks, coarse m.s.d.
b) da/dN-curves indicating retardation of crack growth by coarse m.s.d.

EFFECT OF SURFACE TREATMENT

For alloy B slip distribution as well as crack initiation and crack propagation are well known. Therefore it was exposed to several additional surface treatments: nitriding, boriding, plasmaspraying and other surface coatings and shot peening. Some surface treatments produce a strong effect on the m.s.d., and as a consequence reduced crack initiation in the surface. However, new crack initiation processes underneath the surface can result in little improvement or even deterioration of the fatigue properties of the alloy. Emphasis is put on the discussion of surface work hardening by shot-peening and surface diffusion layers produced by nitriding.

For the shot-peening experiments an angle of coincidence of 90° was used and several particle velocities were applied. In addition the duration of the exposure of the surface to the beam was varied between 2 and 18 minutes. Shot-peening produces the following changes in the surface layer:

1. a dislocation density which increases towards the surface
2. a surface-zone of compressive stress
3. surface roughness and overlaps.

The effect 1) and 2) should retard crack initiation while 3) are expected to favor it. In a coarse m.s.d.-condition the dislocation should be dispersed by the dislocation forest of the shot peened zone. In addition the compressive internal stress should reduce the effective tensile stress in the surface and therefore reduce the critical step height. The effect of changed surface morphology on initiation of cracks is shown in Fig. 9. The number of cracks which are initiated in the shot peened condition is very much reduced compared with the

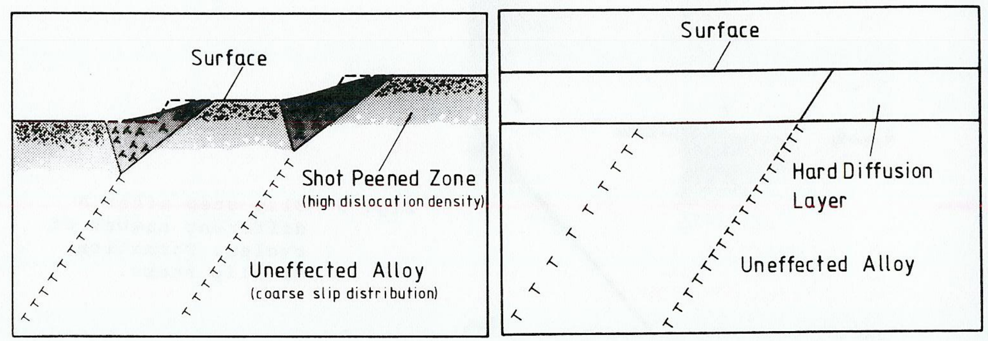


Fig. 9 Dispersion of originally coarse m.s.d. by dislocations in a shot peened condition.

Reduction of fatigue steps in the surface by a hard diffusion layer.

untreated material. The predominant sites of initiation are the areas in which overlap of material has been caused by intensive shot peening. As a consequence of this effect there exists an optimum duration and intensity at which this unfavorable effect is minimum. In Fig. 10 a Wöhler-diagram is shown which summarizes this behavior. It is quite evident that initiation of cracks in the alloy with originally coarse m.s.d. is highly retarded. As a consequence specimen life is increased. Following this line an optimum treatment can be proposed which consists in a heat treatment which produces coarse m.s.d. and a small initiation period, and a low crack propagation rate, in addition to an optimum shot-peening treatment which requires a large number of cycles until initiation of these cracks as it was shown by Gräf and Verpoort (1979).

By nitriding or boriding a very hard surface layer is obtained, so that the high fatigue steps are surpressed. But if a sufficiently high number of dislocations pile up against the hard surface layer a stress concentration originates which is sufficient to crack the layer and thus initiate fracture of the specimen. The improvement of the fatigue life is less than for shot-peening as it is shown in the Wöhler-diagram. Fig. 11.

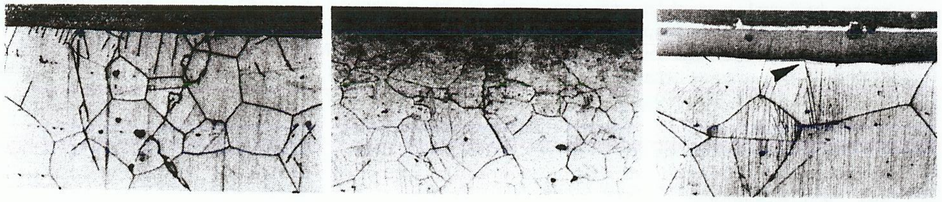


Fig. 10 Reaction of slip bands with the surface:
a) untreated b) shot peened c) nitriding

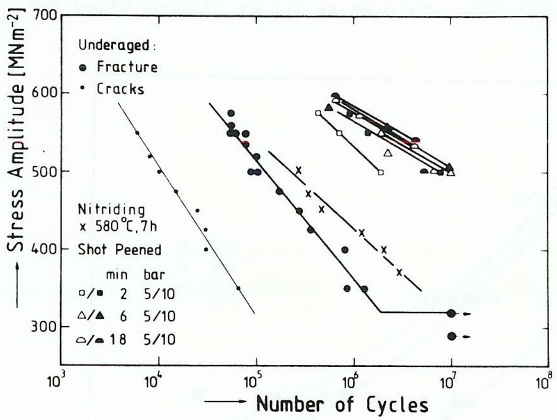


Fig. 11 Wöhler-diagram, underaged, showing the effect of shot-peening and nitriding on the fatigue life.

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