INFLUENCE OF SURFACE TREATMENTS ON FATIGUE
CRACK INITIATION IN $\gamma+\gamma'$ -PRECIPITATION
HARDENING ALLOYS.

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The mechanism of transcrystalline fatigue crack initiation is discussed for $\gamma+\gamma'$ -alloys (Nickel-superalloy; austenitic, precipitation hardened steel). Microscopic slip distribution affects highly the ease of crack formation, and 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. Some supplementary results on the behavior of such surface zones at elevated temperatures are reported.

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 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 microstructure can be modified in such a way that strain is almost homogeneous. This is the basis for usual continuum mechanical calculations of mechanical behavior. A basic 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

Fig. 1: Transition from shearing to by-passing as function of particle diameter

interaction changes in such a way that below a critical particle diameter the particles are sheared and above by-passed by dislocations (1) Fig. 1. As a consequence the critical resolved shear stress in a particular slip plane will be locally decreased ($d < d_c$), or increased for example due to the formation of Orowan-rings ($d > d_c$).

In addition a low stacking fault energy of the matrix may 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 could be verified by an overaged alloy $d > d_c$ in which additional dislocations have been introduced by thermomechanical treatments. 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. 2: Microscopic slip distribution (m.s.d.) for alloy B
a) Underaged, static strain $\varepsilon = 1\%$
b) Thermomechanical treated, static strain $\varepsilon = 1\%$

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 is used as a measure to characterize 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 tells that increasing volume portion f and decreasing diameter d of ordered particles which are associated with a certain energy of the antiphase-boundary 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 this particular slip plane in which n dislocations have passed.

CRACK INITIATION AT SLIP STEPS

Starting from the fact that microscopic slip distribution in superalloys can be manipulated and varied in very wide ranges the consequences of this behavior on crack initiation should be discussed in the following investigation. This investigation 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: Tab 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 change the microscopic slip distribution (m.s.d.). (UA = Underaged, A = Aged, OA = Overaged, TMT = Thermomechanical Treated). The first stage of the investigation was always a characterisation of the inhomogeneity of m.s.d. under static load. The consequent work was concerned with the corresponding m.s.d.

- a) under cyclic loading
- b) the mechanisms of initiation of cracks at these steps
- c) the effects of different surface treatments
- d) the behavior of the surface zone at elevated temperatures.

The effect of coarse m.s.d. caused by static loading may be important if dislocation 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. Fig 3. A parallel investigation

Fig. 3: Extrusion at a fatigue slip step, alloy B, aged

of m.s.d. for static and dynamic loading indicated to a good approximation a similar behavior for both types of loading. (Compare Fig 2 and Fig 7) 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 (2).

The experimental results 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. 4

Fig. 4: Formation of slip steps, extrusions and cracks, cyclic loading, alloy B, aged.

a) in the surface b) taper sectioning, magnification by taper sectioning ~ 10 x

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 occurrence of this height h_c is higher for a coarse than for a fine

slip distribution. However, the maxima of the distribution are usually not characterizing well this situation as indicates Fig. 5 b. An extrem-value-function (Weibull-distribution) should be used for a quantitative analysis of this situation. (2)

Fig. 5: Number of cycles required for crack initiation in alloy A

Use of distribution functions to characterize the homogeneity of cyclic strain.

It is well known from earlier investigations (3) that the same microstructure and micromechanic conditions which favors 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.

Fig. 6 a) Wöhlerdiagram, indicating 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 the alloy B slip distribution as well as crack initiation and crack propagation are well known. In addition it was exposed to several surface treatments: nitriding, boriding, plasmaspraying 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, additional crack initiation processes underneath the surface resulted in little improvement or even deterioration of the fatigue properties of the alloy with exception of shot peening. Therefore further emphasis is put on the discussion of this particular method of surface treatment.

For the experiments an angle of coincidence of 90° was used and several particles 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. increasing dislocation density
2. zone of compressive stress
3. surface roughness and overlaps.

The effect 1) and 2) should retard crack initiation while 3) would be expected to favor it. In a coarse m.s.d. condition the dislocations 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 (see Fig. 5). The effect of the changed surface morphology on initiation of cracks is shown in Fig. 7. The number of cracks which are initiated in the shot peened condition is very much reduced compared with the 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. 8 a Wöhlerdiagram 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.

Fig. 7: Dispersion of originally coarse m.s.d. by dislocations in a shot peened condition
a) schematic b) untreated and shot peened; underaged.

Following this line an optimum treatment can be proposed which consists of 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. (4)

Fig. 8: Wöhlerdiagram, UA, showing the effect of shot peening.

Crack initiation at the shot peened surface.

SURFACE ZONE AT ELEVATED TEMPERATURES

This leaves for a discussion the question how such a surface layer behaves at elevated temperatures. Useful for the interpretation of the microstructural changes in the shot peened zone are investigations on recrystallization behavior of precipitation hardened alloys. (5) These investigations have shown that in such alloys recrystallization occurs only in an intermediate range of amounts of deformation. At very high amounts recrystallization is inhibited due to the nucleation effect which certain dislocation arrangements have on incoherent phases (η -particles). All microstructures shown in Fig. 9 should occur in a shot peened surface if the maximum amount of deformation is sufficiently high to produce the range III microstructure. Fig. 9 b. This microstructure is characterized by the fact that crystallization is inhibited and therefore the microstructure consists of a dispersion of incoherent particles and a dense network of subgrain boundaries.

Fig. 9: Types of microstructures obtained by heating after different amounts of deformation

This microstructure is very efficient to disperse strain localization and therefore the alloy should be protected up to rather high temperatures against transcrystalline crack initiation. It is well-known, that with increasing temperature an increasing tendency for intercrystalline crack initiation is to be expected. If grain boundaries and their environment are investigated in the shot peened zone it is found that a similar beneficial effect of the surface treatment applies to grain boundaries. Grain boundaries are highly curved and pinned by precipitates in the shot peened zone and as a consequence grain

boundary sliding should be highly reduced. It can be concluded, that a beneficial effect of shot peening on crack initiation should be pronounced as long as this type-III microstructure is preserved at the surface. If however, recrystallization occurs, i. e. at intermediate amounts of deformation (type II) or at very high temperatures, this effect will disappear. Therefore the shot peening conditions as well as the upper limit of temperatures and service times have to be controlled carefully for beneficial effects from this method during application of these materials at elevated temperatures. Practical experience has proven, that fatigue life of turbine blades (of alloy B) shows a pronounced increase after shot peening and use even above 550 °C.

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