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Damage evolution during fatigue in structural materials

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Abstract

Early stages of damage evolution in cyclic loading are described and discussed. The importance of the role of cyclic plastic strain in damage evolution is emphasized and the relation between stress and strain in cyclic straining is clarified. The principal stages of damage evolution in fatigued crystalline structural material are identified. The basic characteristic and theories of fatigue crack initiation are sketched and confronted with experimental observations. Early fatigue crack growth is characterized and quantitatively described. The relation between the growth of short cracks and fatigue life in the form of Manson-Coffin law is established.

Keywords: cyclic plasticity; crack initiation; crack growth; fatigue damage

1. Introduction

Selection of materials for structural application is a difficult task since multiple criteria are used to make an optimum choice. Since industrial structures undergo in service different loading conditions the most important properties of structural materials is the resistance to excessive deformation and the resistance to fracture. Resistance to external stresses in crystalline materials depends largely on the type of bonds between the atoms, on temperature and on the type of loading. However, even in cases when the resulting deformations from external stresses are low but happen frequently they can introduce damage which eventually could result in

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fracture. This type of fracture is usually sudden but is preceded by characteristic stages of damage evolution in which damage in material accumulates and results in final failure.

In this contribution the material response to cyclic loading is presented and principal stages of damage evolution are described and discussed.

2. Cyclic plastic straining of crystalline materials

Principal mechanism of cyclic plastic straining in crystalline materials represents the motion of dislocations. Even in annealed single crystals the initial dislocation density is high enough to allow the onset of plastic strain at low stresses. The movement of dislocation segments is practically not hindered by the stress field of other dislocations, the dislocation segments expand to dislocation loops and dislocation multiplication results in the increase of dislocation density. The shear stress necessary to move a dislocation in a crystal, Peierles Nabarro stress, represents the principal component of effective shear stress τ_{eff} necessary to overcome short range obstacles during dislocation motion. Effective shear stress depends on temperature T since overcoming of short range obstacles can be thermally activated. The plastic shear strain rate $\dot{\gamma}_p$ is proportional to the mobile dislocation density ρ_m and to the dislocation velocity v according to Orowan equation (Caillard and Martin, 2003)

$$\dot{\gamma}_p = b\rho_m v\,, \tag{1}$$

where b is Burgers vector. The motion of dislocation across short range obstacles under the action of the effective shear stress τ_{eff} is thermally activated and the velocity

$$v = v_0 \exp\left[-\frac{\Delta G_0 - \int_0^{\tau_{eff}} A\,d\tau}{kT}\right], \tag{2}$$

where v_0 is the frequency of thermal oscillations, ΔG_0 is the activation enthalpy of an obstacle at 0 K and A the activation area. Provided the activation area A does not depend on the effective stress eq. (2) can be written as

$$\dot{\gamma}_p = K_s \exp\left(\frac{\tau_{eff}}{\alpha_s}\right). \text{ and thus } \qquad \tau_{eff} = \alpha_s \ln\left(\frac{\dot{\gamma}_p}{K_s}\right) \tag{3}$$

where

$$\alpha_s = \frac{kT}{bA_{eff}} \quad \text{and } K_s = v_0 \exp\left(-\frac{\Delta G_0}{kT}\right) \tag{3a}$$

and A_{eff} is the effective activation area of an obstacle. The equation (3) shows that effective stress is a logarithmic function of the plastic shear strain rate provided mobile dislocation density is constant. In deformation of a single crystal or of a grain of a polycrystal with shear strain rate $\dot{\gamma}_p$ the effective shear stress τ_{eff} must be applied. In unidirectional straining the respective quantities are strain ε, plastic strain rate

$\dot{\varepsilon}_p$ and effective stress σ_{eff}.

With increasing dislocation density the stress fields of all dislocations in the crystal produce long range stress field. Arrangement of dislocations is not homogeneous and therefore the internal stress field that a dislocation must overcome in order to get into motion changes locally. Characteristic dislocation structures in cyclically deformed metals are vein or cell dislocation structures with dislocation rich and dislocation poor areas. Therefore the internal shear stress τ_i in a particular volume of the crystal depends on the local dislocation arrangement. The critical internal shear stress τ_{ic} of the individual volumes of the crystal has some typical distribution which can be derived from the dislocation arrangement. This distribution is characterized by the probability density function. Since individual grains of a polycrystal have different orientations in relation to the stress axis instead of probability density function of internal critical shear stresses $f(\tau_{ic})$ the probability density function of the internal critical tensile stresses $f(\sigma_{ic})$ can be introduced (Polák, 1991).

In an individual volume of the crystal the total critical stress necessary for deformation σ_{itc} is the sum of the effective stress and critical internal stress

$$\sigma_{itc} = \sigma_{eff} + \sigma_{ic} \tag{4}$$

Since in cyclic loading the strain in each cycle is low the hardening of individual volumes need not be considered and the individual volumes are deformed in such a way the they have the yield stress σ_{itc} identical in tension and in compression and beyond this yield stress they deform without hardening.

In a statistical approach (Polák, 1991) the individual volumes are deformed in parallel. The macroscopic stress σ in a polycrystal characterized by the probability density function of the internal critical stresses $f(\sigma_{ic})$ which was deformed to the strain ε is thus obtained by integration

$$\sigma = \int_0^{\varepsilon E} \sigma_{ic} f(\sigma_{ic}) \, d\sigma_{ic} + \varepsilon E \int_{\varepsilon E}^{\infty} f(\sigma_{ic}) \, d\sigma_{ic}. \tag{5}$$

E is the elasticity modulus

In cyclic loading with constant strain rate $\dot{\varepsilon}$ and total strain amplitude ε_a the hysteresis loop is separated in two half-loops. Using relative coordinates with the origin put into the maximum or minimum strain both half-loops are identical and can be treated using statistical theory. Relative stress of each half-loop σ_r is function of the relative strain ε_r analogically to eq. (5)

$$\sigma_r = 2 \int_0^{\varepsilon_r E/2} \sigma_{ic} f(\sigma_{ic}) \, d\sigma_{ic} + \varepsilon_r E \int_{\varepsilon_r E/2}^{\infty} f(\sigma_{ic}) \, d\sigma_{ic} \tag{6}$$

The probability density function $f(\sigma_{ic})$ can be thus obtained from the analysis of the loop shape, i.e. from the second derivative of the half-loop

$$f\left(\frac{\varepsilon_r E}{2} - \sigma_{es}\right) = -\frac{2}{E^2} \frac{\partial^2 \sigma_r}{\partial \varepsilon_r^2} \tag{7}$$

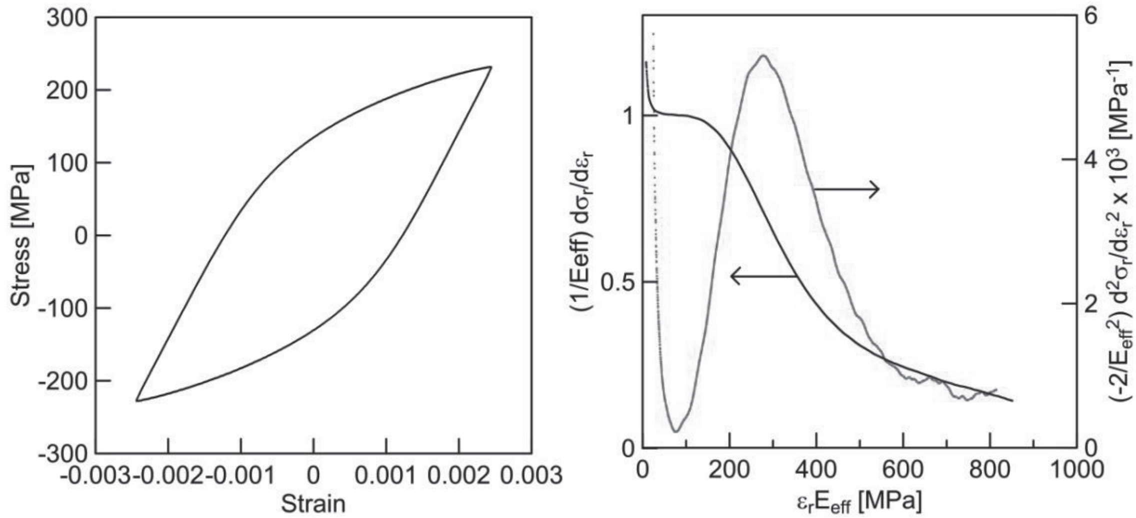


Fig. 1. (a) Hysteresis loop of 316L steel, $\varepsilon_a = 2.5\text{x}10^{-3}$. (b) the first and the second derivatives of the compression hysteresis loop.

Figure 1a shows saturated hysteresis loop of 316L steel cycled at room temperature with constant strain amplitude $2.5\text{x}10^{-3}$ and the strain rate $1\text{x}10^{-3}$ s^{-1}. The stress amplitude is equal to 229 MPa. Figure 1b shows the first and the second derivatives of the compression half-loop that are practically identical with those of the tensile half-loop. More detailed analysis shows that the effective saturated stress at room temperature σ_{es} = 50 MPa, i.e. 22 % of the stress amplitude. However, the internal critical stresses of the individual microvolumes are in the interval from 0 to 375 MPa with the maximum at 92 MPa.

If D is a critical distance between pinning point of a dislocation of a Frank-Read source that can emit

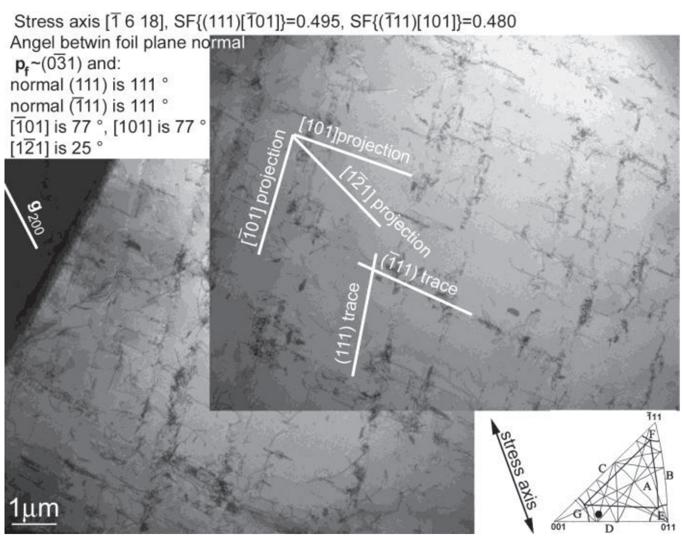


Fig. 2. Dislocation structure in a grain of 316L steel cycled with constant plastic strain amplitude $\varepsilon_{ap} = 1\text{x}10^{-3}$.

dislocations the local stress σ can be calculated from a simple relation (Argon, 2008)

$$\sigma = \frac{bE}{D} \tag{8}$$

Relation (8) allows evaluating the critical distances between pinning points in the interval from 0.15 μm to infinity with the most frequent value equal to 0.59 μm. This can be compared with electron microscopic observations of the internal dislocation structure of specimens after cyclic loading. Figure 2 shows the dislocation arrangement in a grain of 316L steel oriented for double slip and cycled with. Planar arrangements of dislocations from two slip systems intersect and determine the pinning points for dislocations of the other system. Characteristic distances are around 1 μm, which is in agreement with above presented estimations from the most frequent internal critical stresses.

3. Damage evolution

The evolution of fatigue damage in crystalline materials is tightly bound to the initiation and growth of fatigue cracks. Growth of a fatigue crack leads to the final fracture and eventually separation of the material in two pieces.

3.1. Fatigue crack initiation

The studies of the early stages of fatigue crack initiation in single phase alloys were reviewed recently by Man et al., 2009. An inhomogeneous cyclic straining of a material resulting in cyclic slip localization represents the pre-nucleation stage during which the surface relief is built up. Cyclic slip localization is characterized by the formation of an inhomogeneous internal dislocation structure and later by formation of a specific surface relief. Multiple investigations of the internal dislocation structure using transmission electron microscopy (TEM) and electron channeling contrast imaging (ECCI) revealed thin bands of ladder-like structures running parallel to primary slip plane in the grain.

Figure 3 shows these bands in fatigued ferritic stainless steel. Since cyclic slip was proved to be concentrated to these bands they were called persistent slip bands (PSBs). Figure 4 shows the typical surface

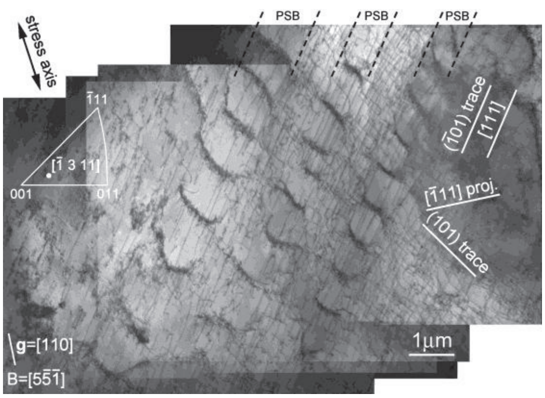


Fig. 3. Dislocation structure in a grain of ferritic stainless steel cycled with plastic strain amplitude $\varepsilon_{ap} = 2 \times 10^{-3}$; ladder-like structure of PSBs.

of fatigued ferritic stainless steel. At the locations, where PSB emerge on the surface, persistent slip markings (PSMs) are formed. PSMs do not resemble slip lines or slip bands produced during unidirectional straining. They consist of extrusions and intrusions and are result of localized intensive cyclic straining within PSB. In Fig. 4a the primary electron beam was perpendicular to the specimen surface and the intrusions were partly hidden by extrusions. In Fig. 4b the electron beam parallel to the primary slip plane allowed to see the intrusions better.

Three dimensional image of the PSMs can be obtained using atomic force microscope. Figure 5 shows the AFM image of the plastic replica of the surface of fatigued grain of 316L austenitic steel. PSBs are formed

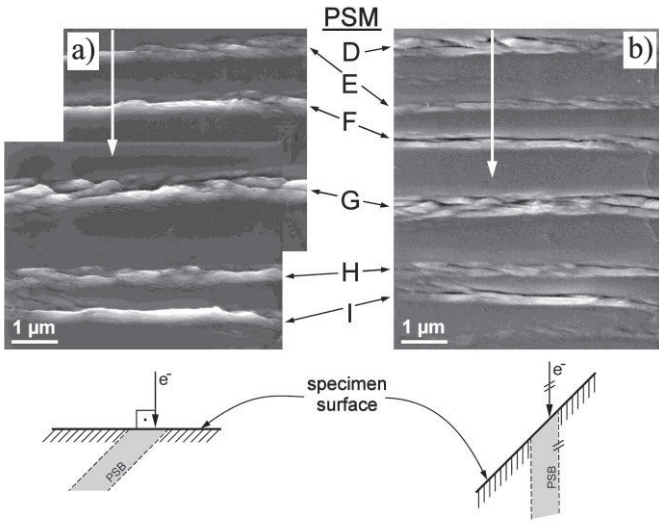


Fig. 4. Surface of the of ferritic stainless steel cycled with plastic strain amplitude $\varepsilon_{ap}= 1x10^{-3}$, SEM, a) electron beam perpendicular to the surface, b) electron beam parallel to the primary slip plane.

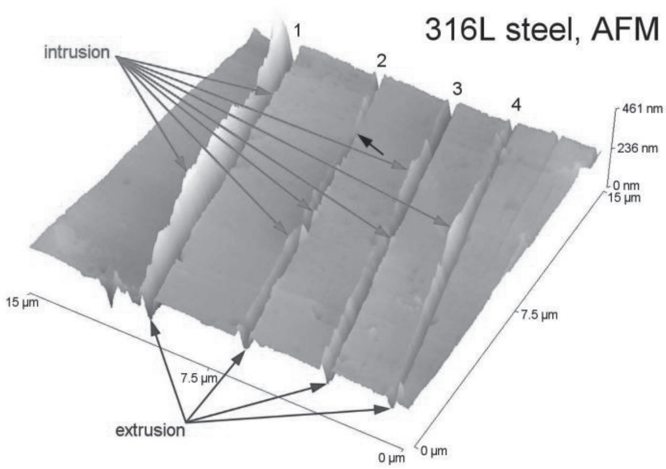


Fig. 5. AFM image of plastic replica of a grain of 316L steel cycled with constant plastic strain amplitude $\varepsilon_{ap} = 1x10^{-3}$.

mostly by extrusions. Extrusions are wide and extend across the whole width of the PSB. In plastic replica, extrusions look like depressions and intrusions like elevations of the surface. Figure 5 documents that intrusions arise on the boundary between a PSB and the matrix, they are very thin and deep. Since intrusions represent very sharp notch, it is supposed that early fatigue cracks start from the intrusions.

The relation between the internal dislocation structure and surface relief was obtained by covering the surface of fatigued 316L steel by platinum and preparing the surface foil using focused ion beam (FIB) technique. Figure 6 shows the image of the surface foil. Three PSBs are running parallel to the trace of primary slip plane. Two side PSBs produced extrusions and the middle PSB produced an intrusion on the surface. Due to two extrusions part of the surface close to extrusions was not covered by platinum. Figure 6 shows unequivocally the relation of the PSBs and PSMs.

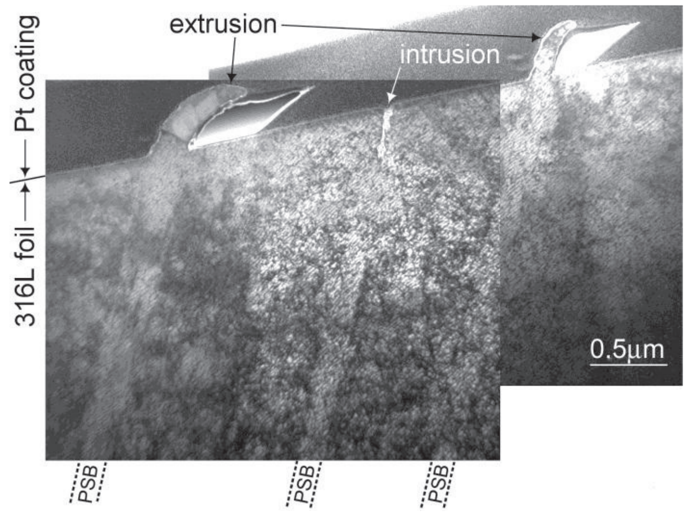


Fig. 6. Surface foil of 316L steel cycled with constant plastic strain amplitude $\varepsilon_{ap} = 1x10^{-3}$ The material was covered by platinum and the foil was prepared using FIB procedure.

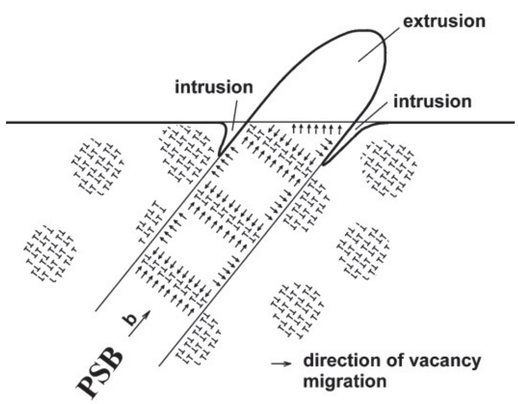
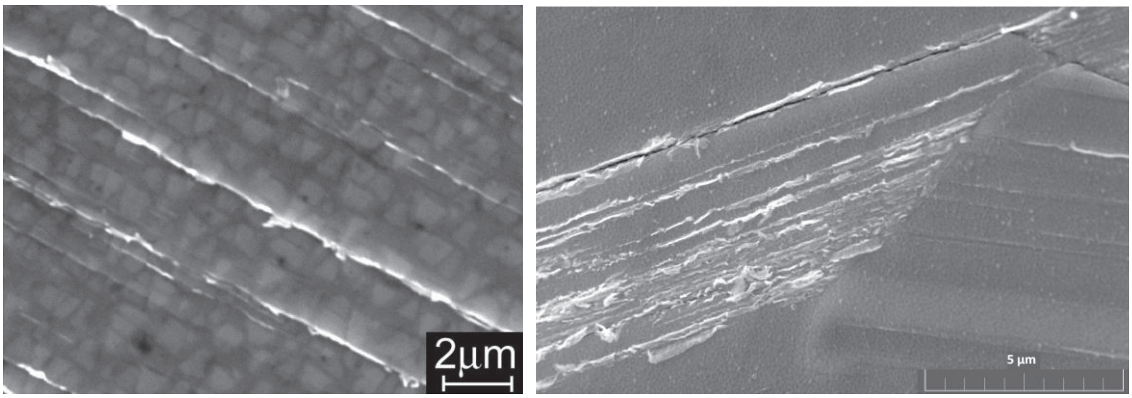


Fig. 7. Schematic section through the grain showing the dislocation arrangement in the PSB and in the matrix and resulting extrusion and two parallel intrusions..

The mechanisms of surface relief formation and primary fatigue crack initiation are based on the dislocation models of the PSBs and the properties of lattice defects. The early model of Essmann et al., 1981 considered mostly the formation of static extrusions due to accumulation of vacancies produced by the mutual annihilation of edge dislocations or by a non-conservative motion of the jogs on screw dislocations. The model predicts the saturation of the extrusion height simultaneously with the saturation of vacancy concentration. Extension of the model by Essmann et al. is the model introduced by Polák, 1987 that considers not only the production but also the migration of vacancy type defects in the PSB and out of the PSM to the matrix. As a result of systematic flow of excess vacancies from PSB to the matrix the mass is accumulated in PSB and resulting compression stress is relaxed by moving dislocations. An extrusion and two thin intrusions on the boundary with the matrix are produced as schematically depicted in Fig. 7.



(a) (b)

Fig. 8. PSMs in a grain of (a) 738-LC superalloy and (b) TiAl intermetallic.

Though the width and the dislocation structure of the localized bands in fatigued materials (PSBs) in different materials vary considerably, the PSMs produced on the surface by localized cyclic straining within these bands are similar. Figure 8 shows the surface of the grain of fatigued nickel superalloy and of TiAl intermetallic. The extrusions are thinner than in austenitic steels and fatigue cracks in nickel superalloy cyt both the matrix and the Ni_3Al precipitates. In TiAl intermetallic a crack already started from a largest parallel intrusion.

3.2. Short crack growth

The early stages of the growth of fatigue cracks initiated in polycrystalline materials differ considerably from the long fatigue cracks. Since plastic strain was necessary for the initiation of the fatigue crack also the growth of short crack progress under general yield conditions, though the plastic strain amplitude is low. As a result the interaction with existing PSBs and with other initiated crack is important and the crack growth rate could be accelerated. On the other hand grain boundaries represent an obstacle to the growth of a crack due to different orientation of neighbor grain since the direction of the crack growth is changed. It usually follows the primary slip system of a neighbor grain in which PSB is initialized first and the crack follows the PSB. Though local direction of the plane of the crack is inclined under certain angle to the stress axis, the crack grows on the average perpendicular to the direction of the stress axis.

Figure 9 shows the short crack on the surface of SAF 2207 duplex steel cycled with constant plastic strain amplitude $\varepsilon_{ap} = 5x10^{-5}$. Shortly after the initiation the crack acquires the shape close to the shape of a semicircle (Balbi et al., 2009). Its size can be characterized by a radius of the semicircle *a*, which is derived from the half of the surface length.

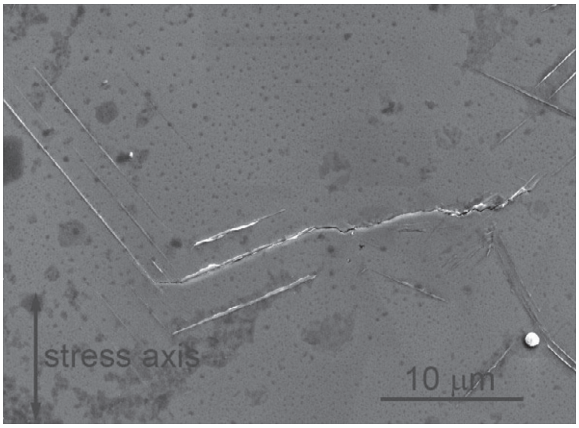


Fig. 9. Short crack in duplex steel cycled with constant plastic strain amplitude $\varepsilon_{ap} = 5x10^{-5}$.

The kinetics of short crack growth has been measured on cylindrical specimens with a shallow notch. Polák and Zezulka, 2005 reported the growth of the principal cracks in duplex steel cycled with constant plastic strain amplitudes leading to the fracture. They reported the exponential growth with number of cycles in the form

$$a = a_i \exp\left(k_g N\right) \tag{9}$$

where a_i is the extrapolated crack length to zero cycles and k_g is the crack growth coefficient which characterizes the growth rate of a crack at the applied plastic strain amplitude. The dependence of the crack

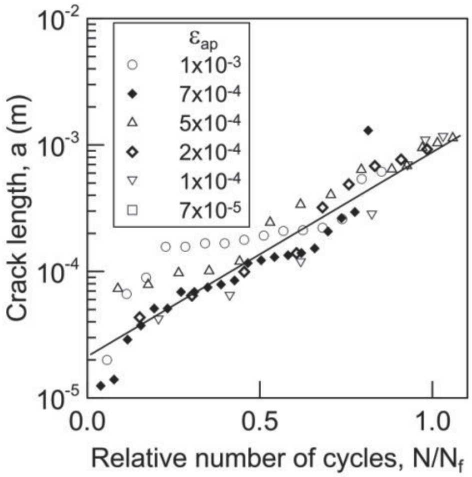


Fig. 10. Crack lengths of principal cracks in 42CrMo4 ferritic-pearlitic-bainitic steel cycled with different plastic strain amplitudes.

growth coefficient k_g on the plastic strain amplitude has the form of power law

$$k_g = k_{g0}\varepsilon_{ap}^d \quad (10)$$

Similar kinetics of growth of short cracks was observed in other materials, austenitic stainless steel, ferritic stainless steel, aluminium alloy and Eurofer97 steel (Polák et al., 2010). Figure 10 shows the crack length vs. the relative number of cycles (number of cycles divided number of cycle to reach crack length equal to 1mm) in 42CrMo4 ferritic-pearlitic-bainitic steel cycled with different plastic strain amplitudes. Figure 10 demonstrates that in this steel the majority of the fatigue life is spent in the period of short crack growth and the exponential law (9) is valid also for this steel. Moreover, the crack growth coefficient k_g is also a power law function the plastic strain amplitude ε_{ap} like the fatigue life N_f . Therefore the Manson-Coffin law can be derived by integrating the short crack growth law represented by eqs. (9) and (10).

4. Conclusions

- The damage evolution in the early stages of the fatigue process in crystalline materials can be now elucidated. The principal role is played by the cyclic plastic strain. Therefore the relation between cyclic stress and cyclic plastic strain amplitude is of primary importance.
- Irreversible cyclic plastic strain becomes soon concentrated into bands of easy cyclic slip- persistent slip bands, Specific surface relief (PSMs) consisting of extrusions and intrusions is formed. The sharp surface defects lead to the initiation of surface crack. Multiple surface cracks are formed and their number increases with increasing plastic strain amplitude.
- Initiated crack grows due to stress and strain concentration in the tip of the crack and also by interaction and linking with other cracks. The kinetics of short crack growth can be expressed using exponential law. The crack growth coefficient depends on the plastic strain amplitude and its power law dependence results in the Manson-Coffin law

Acknowledgements

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