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Short crack growth in polycrystalline materials

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Abstract

The early stages of fatigue damage in cyclic loading in several polycrystalline materials were studied. The initiation of fatigue cracks is influenced by a number of factors like the presence of inclusions, grain size and cyclic creep strain. The growth of short cracks was measured in symmetrical loading and in cycling with positive mean stress. The kinetics of short crack growth is affected by the presence of secondary cracks and can be approximated by an exponential law. The integrated crack growth law is equivalent to Coffin-Manson law allowing the interpretation of this law in terms of short crack growth.

Keywords: fatigue damage; short crack; crack growth; plastic strain; mean stress

1. Introduction

Fatigue damage of materials has been studied shortly after the identification of the fatigue phenomenon. Ewing and Humfrey [1] documented surface fatigue damage in polycrystalline Swedish iron cycled in rotating bending. Using optical microscope they observed pronounced surface markings that widened during cyclic loading and in which fatigue cracks developed later in fatigue life. Today the surface markings produced by cyclic loading are much better understood (see e.g. [2]). They are called persistent slip markings (PSMs) since they result from localization of the cyclic plastic strain into persistent slip bands (PSBs) emerging on the surface as PSMs. They typically consist of extrusions and intrusions and possibly also of unidirectional slip steps. They represent an antecedent stage of fatigue crack formation. Extrusions and more efficiently intrusions represent an effective stress and strain concentration and the irreversible sliding parallel to the plane along which PSB runs leads to crack initiation and early stage I crack growth.

The initiation of fatigue cracks is an extremely important phenomenon for the evolution of the fatigue damage, nevertheless, in majority cases it is not life-determining. Formation of fatigue cracks, which can be identified very early using modern experimental techniques, represents usually 1 to 20% of fatigue life of a smooth body. The majority of fatigue life is spent in the growth of short cracks. This is the reason why the growth of short fatigue cracks attracts such a high interest.

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Since the proposal of Paris and Erdogan in 1963 [3] the growth of macroscopic cracks has been correlated with the fracture mechanics parameters. Within the validity of linear fracture mechanics it is the stress intensity range and maximum stress intensity; for elasto-plastic loading it is the range of J-integral. This correlation is conceivable if the material structure is close enough to the isotropic continuum.

Already before the onset of fracture mechanics numerous studies of long crack growth were performed. Frost and Dudgale [4] studied experimentally the propagation of fatigue cracks in large thin panels with central notch serving as a crack starter made of steel, copper and aluminium alloy. They found that the crack growth rate was proportional to the crack length and to the third power of stress range. Numerous experimental studies of the crack growth in aeronautical structures reported by Molent et al. [5,6] also show the proportionality of the crack growth rate to the crack length in a wide interval of crack lengths. Nishitani et al. [7] and Murakami et al. [8] studied the growth of small cracks in smooth specimens or in specimens with a small artificial hole loaded with different stress amplitudes and found the crack growth rate proportional to the crack length and to the eighth power of the stress amplitude. Also theoretical analysis of the crack growth rate in elastoplastic cyclic loading by Tomkins [9] yields the proportionality between the crack growth rate and the crack length.

Previous experimental investigations of the short crack growth in specimens of duplex and ferritic steels [10,11] cycled in elasto-plastic loading revealed that the kinetics of short crack growth is determined by the applied plastic strain amplitude. In this contribution the experimental study of the fatigue crack initiation and short crack growth on several polycrystalline materials in different loading regimes is reported and summarized and the relation of the crack growth rate to the fatigue life curves in constant amplitude loading is established and discussed.

2. Experimental

Austenitic-ferritic SAF 2507 type duplex stainless steel was supplied by Sandvik, Sweden as rods of 30 mm in diameter. The chemical composition (in wt. %) was: 0.02 C, 23.0 Cr, 7.0 Ni, 3.8 Mo and 0.27 N, the rest Fe. The structure of the steel is formed by the islands of austenite elongated in the rolling direction embedded in a ferritic matrix. The volume fraction of austenite is 53 %. The average cross section of the austenitic and ferritic islands in the plane perpendicular to the specimen axis is about 255 μm^2. 316L austenitic steel and ferritic X10CrAl24 stainless steel were supplied by Thyssen. The average grain size was 100 μm and 38 μm, respectively. Wrought aluminium EN-AW 6082/T6 alloy had fibre-like microstructure with secondary phases having the size up to 20 μm. Eurofer97 ferritic-martensitic steel was produced by Böhler Edelstahl Co. After austenitization and air-cooling it was tempered at 760°C. The prior austenitic grain size was around 15 μm.

Cylindrical specimens of the gauge length 12 mm and the diameter 8 mm were manufactured and their central part was ground to achieve a smooth surface. For the study of crack initiation and short crack growth a shallow notch in the central part of the cylindrical surface was made. The notch was produced by grinding a cylindrical surface 60 mm in diameter to a depth of 0.4 mm in the middle of the cylindrical gauge section of a smooth specimen. The central part of the notch could be observed during cycling using a long distance optical microscope. The stress and strain concentration in the centre of the notch can be evaluated from the nominal stress and the theoretical stress concentration factor $K_t = 1.15$ estimated using the finite element analysis. The stress is given only as the net nominal stress in the lowest cross section of the specimen. During cyclic creep the true stress increased considerably and can be evaluated from the net stress and the mean strain.

The probability of crack initiation in the central part of the notch was slightly increased due to stress and strain concentration but it was still close enough to the conditions of a smooth cylindrical specimen except for the lowest amplitudes. This is confirmed by the fact that at all amplitudes the cracks also initiated and started to grow outside the area of the notch. The cylindrical surface of the shallow notch was further mechanically and electrolytically polished. Fine marks were engraved on the area of the notch in order to facilitate the identification of the cracks.

Sinusoidal loading with constant nominal stress amplitude and constant nominal mean stress with a frequency 5 Hz was applied. In constant plastic strain amplitude loading specific loading procedure was used [10]. The strain was measured or controlled using an extensometer with 12 mm gauge length positioned in the central part of the specimen. The hysteresis loops were recorded and plastic strain amplitude and mean strain were evaluated in slightly notched specimens neglecting the presence of the shallow notch.

Cycling was interrupted in predefined number of cycles and the central area of the notch was photographed using Navitar microscope with a CCD MegaView IIIu digital camera. To document the surface damage evolution in the central area of the notch around 48 images were taken at each interruption of cycling. The growth of cracks was evaluated after termination of experiment starting from pictures taken at the highest number of cycles and continuing back to the nucleation. The history of all cracks contributing to the formation and growth of all macroscopic cracks could be assessed. In majority of cases only the cracks contributing to the growth of the principal crack and up to three largest secondary cracks were evaluated. The crack length *a* was evaluated as a half of the projected surface length to the plane perpendicular to the loading axis.

The initiation and the early growth of fatigue cracks was studied in selected specimens using scanning electron microscope (SEM) JEOL 6460 and high resolution field emission scanning electron microscope MIRA 3 FEG-SEM from Tescan Co.

3. Results

3.1. Crack initiation and early growth

Fatigue cracks in the shallow notch of specimens of 316L steel, ferritic steel, duplex steel and Eurofer 97 steel initiated from slip bands (persistent slip bands - PSBs). Persistent slip markings (PSMs) appear first on the specimen surface where PSBs emerge on the surface. They consist of extrusions, intrusions and in cycling with mean stress also from unidirectional slip steps. In the presence of mean strain PSMs preferentially nucleate at unidirectional slip steps produced during early stage of cyclic creep. PSMs in each material have some specific features. In materials with f.c.c. structure straight segments of extrusions often accompanied by parallel intrusions extending across the surface grain are produced [12,13]. In b.c.c. metals PSMs are often curved and in Eurofer 97 steel they extend up to the boundaries of prior austenitic grains. Later, presumably from intrusions, cracks initiate and start growing. Figure 1 shows the crack initiated in a ferritic island of duplex steel that starts to grow into neighbor austenitic grains. The further growth is facilitated by the simultaneous production of PSMs that transform in nucleated cracks in neighbor austenitic grains. The linkage with these cracks contributes to the early growth.

Fatigue cracks in aluminium alloy start from the interface between the matrix and larger precipitates as shown in Fig. 2. The decohesion of the precipitate from the matrix results in early crack initiation and growth. The presence of non-metallic inclusions in duplex steel can also in some cases start the PSB and later the crack. The initiation period amounted 1% up to 50% of fatigue life in all materials studied but typical values were about 10 %.

In cycling of duplex steel with the mean stress cyclic creep produced appreciable mean strain in the initial stage of cycling. Figure 3 shows the surface of the specimen cycled with stress amplitude 459 MPa and mean stress 150 MPa for 8000 cycles (10% N_f). Cyclic creep produced mean strain ε_m = 0.05. The uniaxial slip markings are visible in austenitic and in ferritic grains; in some grains two systems of slip were active. Rarely some slip bands

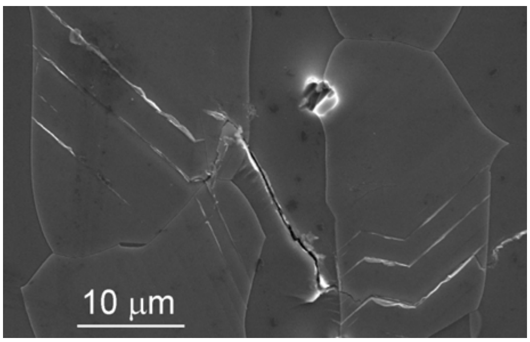

Fig. 1. Fatigue crack initiated in a ferritic grain of duplex steel cycled in symmetrical cycle with constant stress amplitude 432 MPa. Load axis vertical.

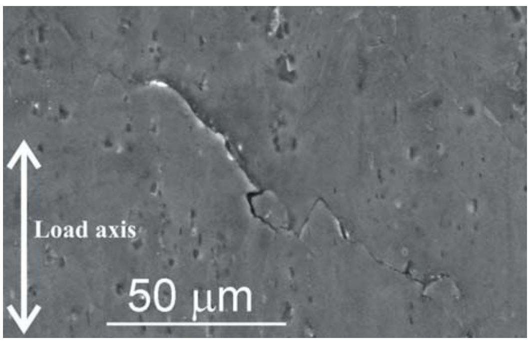

Fig. 2. Fatigue crack in aluminium alloy initiated by decohesion between a large particle and the matrix in cycling with stress amplitude 300 MPa.

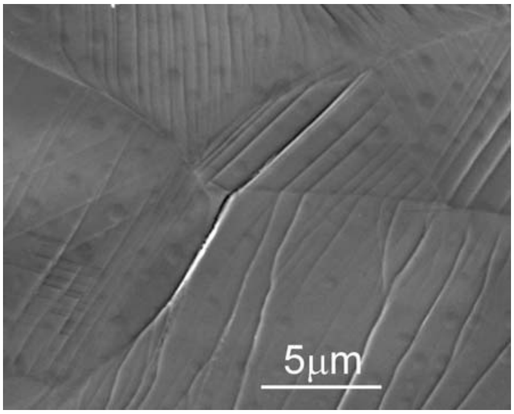

Fig. 3. Fatigue crack initiated at PSM of austenitic and ferritic grain in duplex steel; σ_a = 459 MPa, σ_m = 150 MPa, N = 0.1 N_f. Load axis horizontal.

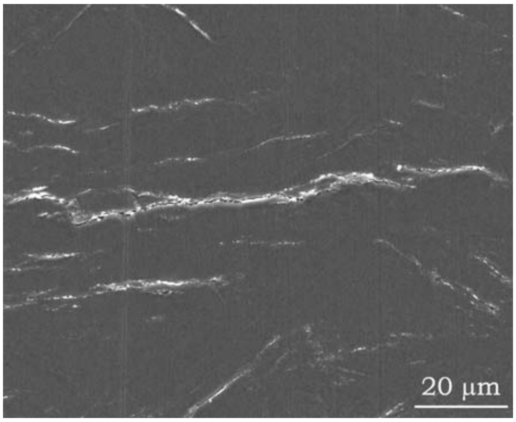

Fig. 4. Fatigue cracks initiated in Eurofer97 steel within in a prior austenitic grain; $\varepsilon_a = 4\text{x}10^{-3}$ ($\varepsilon_{ap} = 1.8\text{x}10^{-3}$ at half life). Load axis vertical.

producing unidirectional slip markings (slip steps) transformed into PSBs and PSMs in the form of extrusion-intrusion and later crack appeared on the surface. The fatigue crack initiation was thus accelerated since unidirectional slip steps served as the nuclei for early PSM formation from which the cracks started. The crack in Fig. 3 started simultaneously in ferritic and austenitic grains presumably since their mutual orientation was favorable [14].

Figure 4 shows the surface of the Eurofer97 steel cycled in symmetrical cycle with constant strain amplitude. Numerous PSMs consists of extrusions that are already cracked producing thus a shallow natural crack. Long PSM extends along the whole prior austenitic grain. The shallow crack will grow into the metal interior before extending its surface length.

The size of the initiated crack was different in individual materials. The typical size of a largest initiated crack is given by the grain size (Fig. 1) or by the size of a precipitate from which the crack started (Fig. 2). It can vary

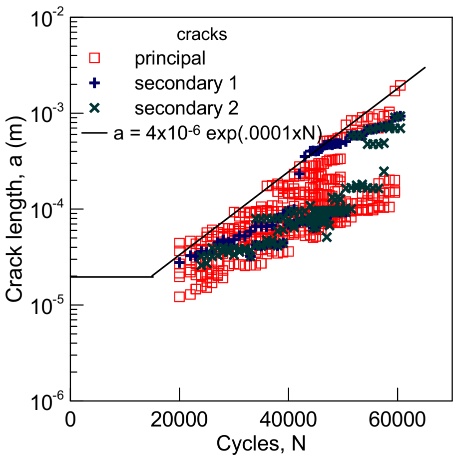

Fig. 5. Length of all cracks contributing to the growth of the principal crack and two secondary cracks. Duplex steel, σ_a = 459 MPa, σ_m =0 MPa.

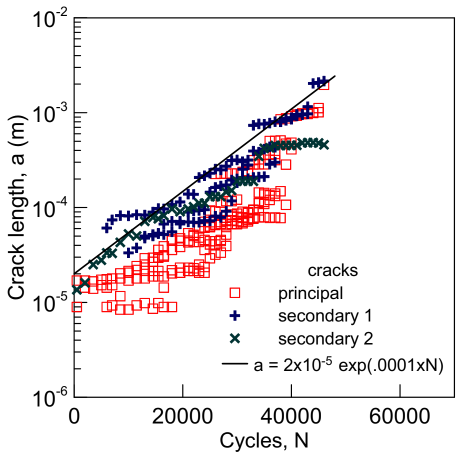

Fig. 6. Length of all cracks contributing to the growth of the principal crack and two secondary cracks. Duplex steel, σ_a = 459 MPa, σ_m = 150 MPa.

appreciably for different materials but only moderately for different specimens. In duplex steel it is given by the size of larger austenitic or ferritic islands and amounts to 10 µm up to 20 µm. In some cases it can extend in two neighbor grains. In ferritic steel and in austenitic 316L steel its size is around the half of the average grain size, i.e. around 20 µm and 50 µm respectively, in Eurofer 97 it is given by the size of prior austenitic grain, i.e. it is in the interval from 15 µm up to 50 µm and in aluminium alloy it is less than 10 µm. The cracks initiated in planes inclined close to 45 degrees to the loading axis. Initial growth was strongly dependent on the applied plastic strain amplitude. For low plastic strain amplitudes the crack advanced without linkage since the density of cracks was low. In cycling with high plastic strain amplitude the linkage with already initiated cracks contributed significantly to the crack growth.

3.2. Crack growth kinetics

Crack growth was monitored during interruption of cycling by taking typically 48 pictures of the area of the shallow notch. Separate experiments on the crack shape using tinting procedure revealed that the shape of the growing crack in push-pull loading was close to the semicircle (see e.g. [15]). The crack length can be thus characterized by the radius of semicircle *a*. After completion of the experiment the length of the principal crack and of two or three largest secondary cracks was evaluated as a function of loading cycles. The initiation and the early crack growth usually followed slip plane which corresponds to the crystallographic plane of a grain with the largest Schmid factor [2]. However, shortly after the nucleation the crack growth took stage II growth i.e. the plane of the crack was on the average perpendicular to the loading axis. Crack length *a* was thus determined as the half of the projected surface crack length on the plane perpendicular to the loading axis.

Figs 5 and 6 show the crack length *a* vs. number of loading cycles *N* of duplex steel in cycling with the same stress amplitude and two levels of the mean stress. In symmetrical cycling (Fig. 5) cracks initiated early in the fatigue life (see sec. 3.1) but their length could be evaluated only later when they could be distinguished using optical microscope. All cracks contributing to the growth of the principal crack and two largest secondary cracks are shown in Fig. 5. Since saturated plastic strain amplitude was high ($\varepsilon_{ap,s} = 8 \times 10^{-4}$), the density of cracks was also high and crack linkage contributed to the growth of both the principal and the secondary cracks.

Cycling with the same stress amplitude but mean stress 150 MPa (Fig. 6) resulted in accelerated crack initiation. The unidirectional slip bands provoked early cyclic slip localization, production of PSMs and crack nucleation.

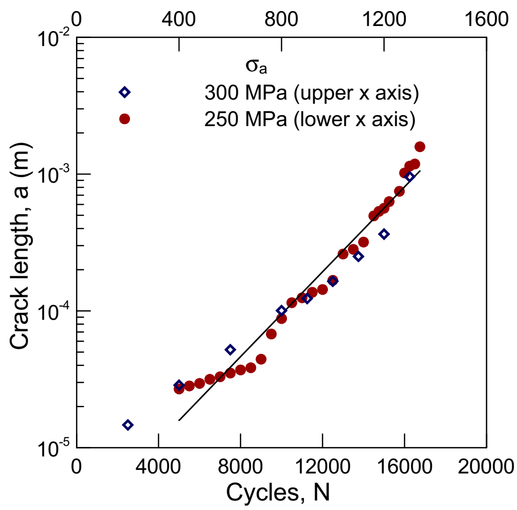

Fig. 7. The length of principal cracks in aluminium alloy vs. number of cycles in cycling with two stress amplitudes.

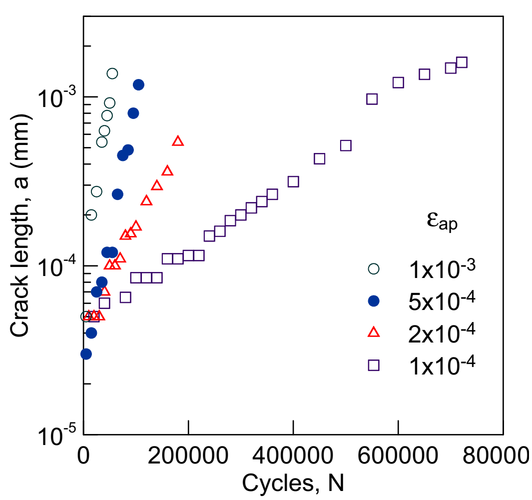

Fig. 8. The length of principal cracks in 316L steel cycled with constant plastic strain amplitudes.

Fatigue life, corresponding to the formation of a macroscopic crack of a final length a_f = 2 mm, is shortened accordingly. The high density of fatigue cracks also resulted also in crack linkage that contributed to the crack growth. The inspection of the plots in Figs 5 and 6 shows that the different physical cracks were the instantaneous longest cracks. This is a consequence of the linkage of cracks. It is thus reasonable, in agreement with previous studies [10,11], to introduce the concept of an equivalent crack. It is the crack whose length is close to the longest crack present in an investigated area. If only one crack is growing the equivalent crack is identical with this physical crack.

Similar kinetics of short crack growth was obtained for aluminium alloy [16] cycled with constant stress amplitudes (Fig. 7) and in 316L austenitic steel cycled with constant plastic strain amplitudes (Fig. 8). In Figs 7 and 8 only principal cracks leading to fracture are plotted vs. number of cycles. Simultaneous measurement of the cyclic stress-strain response in aluminium alloy showed that plastic strain amplitude saturated and was approximately constant for most of the fatigue life.

The study of short crack growth in ferritic stainless steel [11] and in Eurofer97 ferritic-martensitic steel revealed similar kinetics, i.e. the logarithm of the length of the growing crack was proportional to the number of loading cycles.

The growth of the equivalent cracks in case of all materials studied can be approximated very well by the exponential growth, i.e.

$$a = a_i \exp\left(k_g N\right) \quad (1)$$

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 stress amplitude or at corresponding plastic strain amplitude. The extrapolated crack length a_i is smaller than the length of an initiated crack. The equation (1) is assumed to be valid in the interval from zero cycles where the hypothetical crack of a length a_i is present up to the final number of cycles N_f where the final length a_f is reached. In reality the period of crack initiation is substituted by the growth of an imaginary crack of length a_i present in the material at zero cycles to the length of a real initiated crack. Since initiation period represents only a small fraction of the fatigue life this approximation is reasonable. In quantitave evaluation of the fatigue life of smooth bodies this simplification allows to consider only the short crack growth (see sec. 4). The crack growth coefficient k_g could be evaluated from the slopes of straight lines in Figs 5 to 8 for each test performed for a particular stress amplitude and mean stress or for applied plastic strain amplitude.

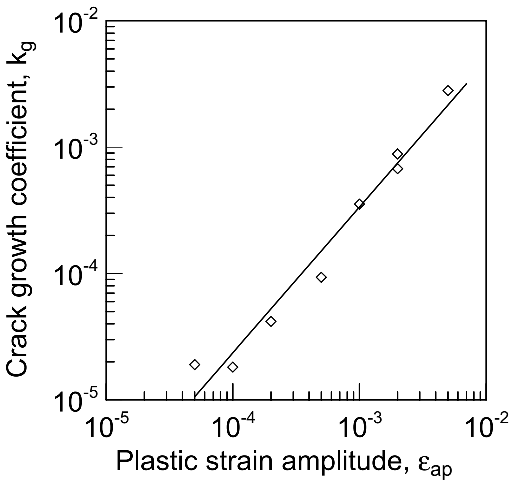

Fig. 9. Crack growth coefficient vs. plastic strain amplitude for duplex steel cycled in symmetrical cycling.

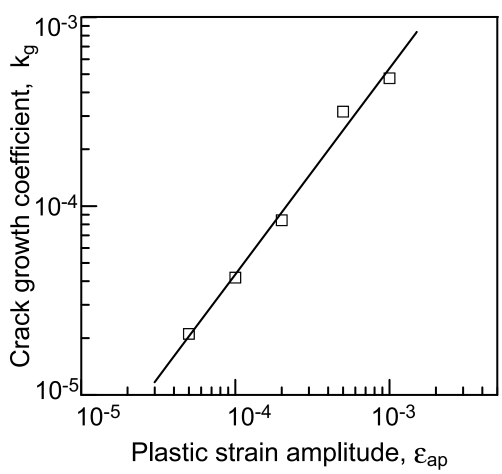

Fig. 10. Crack growth coefficient vs. plastic strain amplitude for ferritic steel cycled in symmetrical cycling.

In cycling with different stress or plastic strain amplitudes the dependence of the crack growth coefficient can be established for a particular material. Since plastic strain represents the driving force for the growth of short cracks the dependence on the plastic strain amplitude was established. Crack growth coefficient in symmetrical cycling vs. plastic strain amplitude is plotted in Fig. 9 for duplex steel and in Fig. 10 for ferritic steel. In the double-logarithmic plot it can be approximated reasonably well by a straight line. The dependence of the crack growth coefficient on the plastic strain amplitude can be written as

$$k_g = k_{g0}\varepsilon_{ap}^{d} \tag{2}$$

where k_{g0} and d are parameters. The parameters of this dependence were evaluated using least-square method for all five structural materials and are listed in Table 1.

Table 1. Parameters of the power law dependence of the crack growth coefficient k_g vs. plastic strain amplitude obtained from testing specimens with a shallow notch.

Material	316L steel	ferritic steel	duplex steel	Eurofer97 steel	Aluminium alloy
k_{g0}	0.091	0.97	0.856	2.74	0.472
d	1.04	1.089	1.114	1.117	0.65

4. Discussion

The structure of material plays an important role in the early stages of the evolution of fatigue damage in crystalline materials. We have studied several polycrystalline structural materials and found that the phenomenon of plastic strain localization is common to all of them. In areas where the bands of localized slip emerge on the surface PSMs in the form of extrusions, intrusions and possibly also unidirectional slip steps appear and fatigue cracks nucleate here. Cyclic plastic strain localization and crack initiation can be enhanced by the presence of defects or inhomogeneities leading to earlier fatigue crack nucleation. The crack is nucleated as a shallow surface crack that starts propagating in the interior, usually along the plane which corresponds to low index crystallographic plane with high Schmid factor relative to external stress. This is early stage I growth within a grain in which the crack has been nucleated.

In spite of the importance of the crack initiation period in the process of fatigue damage evolution leading finally to fatigue fracture this period consumes usually only a low fraction of total fatigue life. Moreover, the identification of initiated crack and thus the determination of the length of initiation period depend strongly on the method used for the detection of the presence of early cracks. While optical microscopy can detect the crack presence at around 20% of fatigue life, the high resolution techniques like field emission SEM (FEG-SEM) can identify the nucleated cracks before 5% of fatigue life, i.e. still in the stage I growth. The length of initiation period thus depends on the method used for the identification of the nucleated crack.

The size of the nucleated crack is determined by the characteristic dimension of the material microstructure. The longest nucleated cracks arise in large grains or across two neighbor grains which have suitable mutual orientation. In case of duplex steel it could be two neighbor grains of austenite and ferrite which fulfill the Kurdjumov-Sachs relations [17]. The nucleated crack grows by creating cyclic plastic zone on its tip but a on the surface it can grow in more effectively by linkage with a neighbor crack provided it is available. The high density of initiated cracks thus promotes early crack growth.

Experimental data on the fatigue crack growth in all materials studied show clearly that it is the period of short crack growth that takes the highest fraction of fatigue life and therefore is life-determining. A thorough study of the growth of short fatigue cracks in different polycrystalline materials has shown that crack growth kinetics is similar in all materials in quite a wide interval of stress amplitudes and plastic strain amplitudes and is characterized by the exponential law (eq. (1)). In cycling with high plastic strain amplitudes or in cycling with stress amplitudes resulting in high saturated plastic strain amplitudes appreciable density of fatigue cracks is present and acceleration of the crack growth of the largest cracks by linkage with smaller secondary cracks present in the path of the growing crack is evident.

The exponential crack growth law (1) implies that the crack growth rate is proportional to the crack length a

$$\frac{da}{dN} = k_g a \, . \tag{3}$$

The exponential law is assumed to be valid starting from the hypothetical crack length a_i present in the material at zero cycles up to macroscopic final crack length a_f (around 2 mm) where fracture mechanics can be applied. In the initial interval (up to initiation of a real crack) the growth of hypothetical crack replaces the period of initiation. The growth of fatigue cracks according to the exponential law has been thoroughly documented in the early paper by Frost and Dudgale [4] using large specimens and in numerous cases of the growth of fatigue cracks in structures, namely in aeronautical structures [5,6].

The factor of proportionality between the crack growth rate and crack length is the crack growth coefficient k_g. The dependence of the crack growth coefficient on the plastic strain amplitude can be very well approximated by relation (2). Provided the cyclic stress-strain curve of the material can be represented by Ramberg-Osgood relation

$$\sigma_a = K' \varepsilon_{ap}^{n'} \tag{4}$$

where K' and n' are parameters the dependence of the crack growth coefficient on the stress amplitude is also given by the power law relation

$$k_g = k_{g0} \left(\frac{\sigma_a}{K'} \right)^{\frac{d}{n'}} . \tag{5}$$

Defining the number of cycles to fracture N_f as the number of cycles necessary to reach the final crack length a_f the fatigue life N_f is primarily given by the crack growth coefficient k_g

$$N_f = \frac{1}{k_g} \ln \left(\frac{a_f}{a_i} \right) \tag{6}$$

while the crack growth coefficient depends on the plastic strain amplitude according to eq. (2). The comparison of the eqs (2) and (6) with the Coffin-Manson law in the form

$$\varepsilon_{ap} = \varepsilon_f' \left(2N_f \right)^c \tag{7}$$

allows obtaining the parameters of the Coffin-Manson law from the parameters of the crack growth law and vice versa. The parameters of the short crack growth law can be calculated provided standard low cycle fatigue test is performed and the parameters of the Coffin-Manson law and of the cyclic stress-strain curve are evaluated

$$d = -1/c \quad , \quad k_{g0} = 2\varepsilon_f'^{\,1/c} \ln \left(\frac{a_f}{a_i} \right) . \tag{8}$$

Parameters of the crack growth law shown in Table 1 for five materials were obtained by direct measurement of short crack growth on slightly notched cylindrical specimens. Though the effect of the notch is small the plastic strain amplitude in the notch root can be appreciably higher than in the smooth cylindrical specimen. It was proved by comparison of the measured Coffin-Manson curve of the slightly notched and smooth specimens [11]. Therefore relations (8) offer a more exact and far less laborious way how to determine the parameters d and k_{g0} of the short crack growth law in comparison with the direct measurement using notched specimens.

The relations (2) and (3) are in good agreement with the Coffin-Manson law that is generally valid in low cycle fatigue domain. However, the validity of these relations extends for several materials [4-6, 10. 11] also into high cycle domain and in the domain of long crack growth. Provided exponential law of crack growth is valid the crack growth rate is proportional to the second power of the stress intensity range, which is in many cases in disagreement with reported results of long crack growth. However, the phenomena of crack closure, the threshold conditions and also the acceleration of crack growth rate for high stress ranges beyond that predicted by Paris law can account for some of the indicated differences.

The short crack growth law can be used for life prediction and residual life estimation of smooth bodies cycled in the domain where the crack is initiated and grows under the effect of cyclic plastic strain. The stress and plastic strain distribution in a body must be evaluated, e.g. using finite element method, and the integration of short crack growth law yields fatigue life.

5. Summary

Study of the initiation and the growth of short cracks in several polycrystalline structural materials lead to following conclusions:

(i) In areas of cyclic slip localization PSMs appear during cycling in the form of extrusions/intrusions. Shallow fatigue cracks nucleate within PSMs or close to heterogeneities and defects that can amplify the plastic strain localization.

(ii) Fatigue cracks grow originally in stage I later with the help of crack linkage in stage II in a plane perpendicular to the stress axis and take approximately the shape of a semicircle. Crack length is characterized by the radius of the semicircle.

(iii) Kinetics of short crack growth follows the exponential law. Parameters of this law were established for five materials using cylindrical specimens with a shallow notch.

(iv) Comparison of the integrated crack growth law with Coffin-Manson law gives efficient way how to evaluate the parameters of the short crack growth law from the results of low cycle fatigue test.

(v) Short crack growth law can be used for life prediction and residual life estimation.

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