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Scales of Metal Fatigue Cracking

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Abstract—A new paradigm is proposed for considering metal fatigue cracking based on the principles of synergetics and physical mesomechanics. Fatigue cracking is described as a three-stage process. Metal evolution is studied with stress growth from the micro- (ultrahigh cycle fatigue) to meso- (high cycle fatigue) and then macroscale (low cycle fatigue). The notion of two effective stress concentration factors on the metal surface and in its bulk is introduced; their variation pattern with stress growth is discussed. In the general case, the propagation of through-the-thickness cracks is shown to also occur in three stages—on the micro- (shear), meso- (rotation with the formation of triangular fatigue striations) and macroscale (rotation plus shear which lead to the formation of fatigue striations of complex shape), consecutively.

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1. INTRODUCTION

Synergetics, as the science of evolution of open systems, has led to the understanding that organic and inorganic nature shares common laws of evolution [1, 2]. The insight into the fact that metal behaves like a complex hierarchically organized system and evolves by the same laws as other natural objects was gained with the emergence of a new scientific field, i.e. physical mesomechanics [3, 4]. The conventional division of metal fatigue cracking into regions depending on the number of cycles to failure and types of external action (constant severe deformation of low frequency or constant load in a wide frequency range) proves to be related to the fundamental difference in metal behavior. The plastic deformation region (low cycle fatigue) around and above the yield stress and the macroscopic elastic deformation region (high cycle fatigue) below the yield stress [5] show a radically different behavior of the material under cyclic loading. It should be emphasized that high cycle fatigue corresponds to the metal behavior below the elastic limit while the transition to low cycle fatigue occurs in the stress interval above and below the yield stress of the material.

The physical mesomechanics ideas on metal fatigue resulted from the intensive study of processes of damage accumulation, which are primarily in high cycle fatigue regions. Consequently, low cycle fatigue should correspond to a higher scale of damage accumulation in metal [6]. In transition from one scale to the next, metal is expected to change a way of energy absorption. This change appears from the self-organized involvement of new mechanisms of evolution into damage accumulation, which are impossible on the previous scale. The transition to a different behavior on a new scale will plot as an irregularity of the fatigue curve, which alters the law used to describe metal evolution. The found irregularity was termed a discontinuity or knee of fatigue curves between low and high cycle fatigue regions [6, 7].

However, the question of existence of one more scale, namely, micro- or nano-, was still obscure. Any system undergoes changes in processes (mechanisms) of its evolution first on the micro-, meso-, and then on the macroscale. The system can omit one of the scales depending on conditions of influence but under steady-state loading its evolution occurs in the above sequence of scales, and not the reverse. Therefore, the question of existence of the microscale was crucial in justifying the fact that metal is a synergetic self-organized system and its evolution takes place on all scales, culminating in failure.

The existence of metal evolution on the microscale resulting in failure was proved with the rise of a new scientific field, namely, ultrahigh cycle (giga cycle) fatigue

[8, 9]. The material was found to fail after more than 10^8 cycles with the total strain amplitude per loading cycle several thousandths of a percent but the crack initiation site was under the surface.

However, numerous investigations of metal behavior could not provide a unified order and self-organization pattern without understanding its scale hierarchy. The study of propagation mechanisms of a crack in high cycle and low cycle fatigue regions, when initiated from the surface, enabled a representation of fracture as a three-stage, multiscale process [10]. It is this approach that makes possible a unified description of fatigue failure of metals not only at the stage of fatigue crack growth but also at their nucleation.

Below is a discussion of synergetic concepts of the scale hierarchy of metal fracture exhibiting order and self-organization in the sequence from the micro- (or nano-) to meso- and then macroscale. The proposed understanding of metal evolution is based on two scientific fields, i.e. physical mesomechanics [11–14] and synergetics [1, 10, 12].

2. BIFURCATION DIAGRAM

In the general case, the single-scale description of a fatigue process through the Wöhler curve employs the following formula [5]:

$$\sigma^{m_{\mathrm{f}}} N_{\mathrm{f}} = C_{\mathrm{f}}. \tag{1}$$

Formula (1) describes metal behavior as a single-scale evolution process, as a widely varying range of cyclic stress σ and fatigue life N_{f} with the only exponent m_{f} and the factor C_{f}.

As mentioned above, the study of the two fatigue failure regions (low cycle and high cycle fatigue) points to the need to relate differences in metal behavior to different scales, when fracture processes correspond to different mechanisms of damage accumulation. This approach to the fatigue failure analysis is necessary to answer the question why after 10^8 cycles the crack initiation site is deep under the surface. Within the entire range of fatigue life, including the ultrahigh cycle fatigue region, the specimen material is under through-the-thickness loading. Its surface layer and bulk sustain a simultaneous cyclic action and simultaneous damage accumulation. In one stress interval the surface accumulates more damages than the bulk and cracking starts from the surface; in another stress interval the bulk fractures earlier than the surface.

There exists a small stress interval between high cycle and ultrahigh cycle fatigue regions, in which at the same stress the crack initiation site can be surface or subsurface [8–10]. Either of the sites appears in a specimen with different probability independently of a stress level [15, 16]. In passing from one boundary of the discussed "transition" stress interval, called the bifurcation region [10], to the other, the probability P_i of one crack initiation site decreases while that of the other increases (Fig. 1). Therefore, fracture should be described by two equations of type (1) with different proportionality coefficients and respective exponents $(m_{\mathrm{f}})_{\mathrm{left}}$ and $(m_{\mathrm{f}})_{\mathrm{right}}$ for the left and right branches of the fatigue curve responsible for low cycle and high cycle fatigue (Fig. 1).

The bifurcation region is a term used to characterize the region of metal evolution taken as a transition from one way of energy absorption (mechanism of crack initiation) to another. In Fig. 1 this region is represented by a transition from stress $(\sigma_{\mathrm{w2}})_{\mathrm{min}}$ to stress $(\sigma_{\mathrm{w2}})_{\mathrm{max}}$. For many alloys the transition under study corresponds to the life range of 10^6–10^8 loading cycles to failure.

A similar transition is observed at the life less than 10^3 loading cycles when passing from low cycle fatigue to repeated static fracture of the material [17]. It is found that at the fatigue life less than 100 cycles cracks are initiated under the surface while at higher life they nucleate at the surface (Fig. 2). The plastic strain scatter about 0.025 is most notable. Of six specimens tested at this strain and represented in the diagram, the three are fractured from surface crack initiation sites; the other three, from subsurface sites. Consequently, surface and subsurface fracture is equiprobable at this strain level.

This is obviously one more bifurcation region where with certain probability in the fixed strain range either of the mechanisms of evolution (achievement of the ulti-

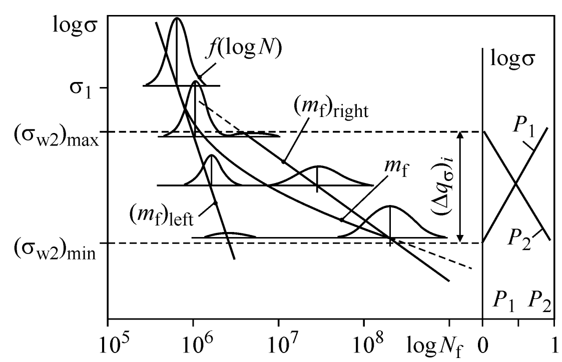

Fig. 1. Schematic of bimodal distribution of fatigue life, which is characterized in the bifurcation region $(\Delta q_{\sigma})_i$ by two fatigue curves with exponents $(m_{\mathrm{f}})_{\mathrm{left}}$ and $(m_{\mathrm{f}})_{\mathrm{right}}$ of equation (1) at different probability P_i of fatigue life distribution $f(\log N_{\mathrm{f}})$.

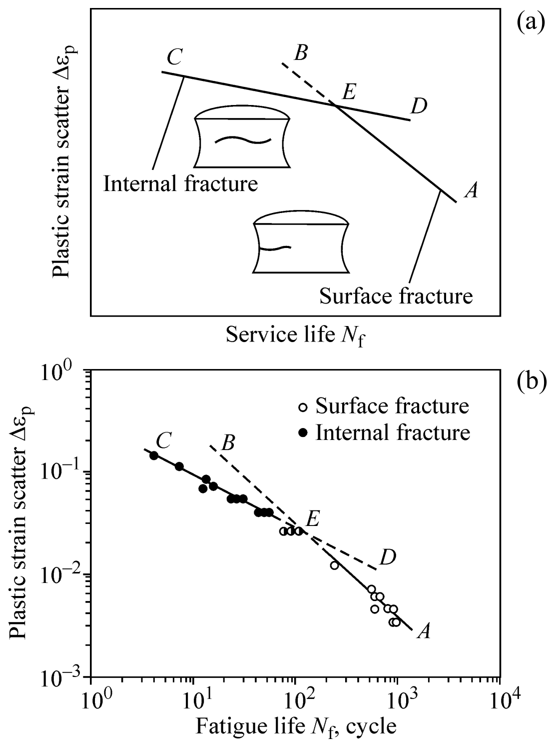

Fig. 2. Schematic (a) and experimental data (b) for low cycle fatigue tests on cast ferritic steel with a surface-initiated crack (*AB*) and for repeated static fracture tests with a subsurface-initiated crack (*CD*) [17]. A transition from one region to the other is marked by point *E*.

mate state due to crack initiation) appears as a self-organized process without changing external conditions of loading. In this case, only one level of strain is identified. In fact, metals demonstrate the strain interval in which the probability of one or the other fracture mechanism reverses with strain growth. The task of finding this interval has not been set earlier [17], so only the present test data are available.

The discussed transition from low cycle fatigue to repeated static fracture is poorly studied due to highly unstable behavior of the material under developed plastic deformation and due to a small number of cycles to failure.

The most intensive investigation is given to the bifurcation region associated with a discontinuity in fatigue curves, i.e. a transition from low cycle to high cycle fatigue [6]. This region registers a change in spatial orientation of the crack surface (flat fracture or at an angle to the surface) [7]; a fracture mechanism is also changed (intragranular fracture below the bifurcation region at low stresses and mixed intragranular plus intergranular fracture above this region) [6]. The stress range between low cycle and high cycle fatigue includes several pro-

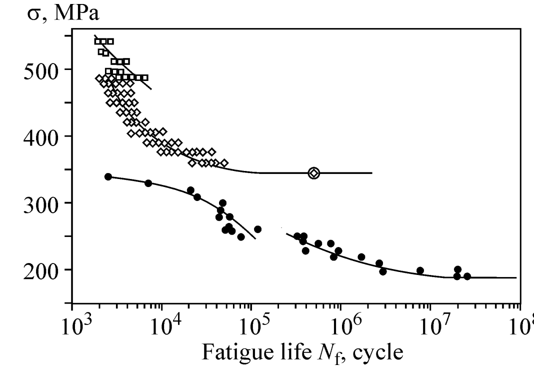

Fig. 3. Fatigue curves with a portion (marked by a circle) of transition from high cycle to low cycle fatigue, which is the bifurcation region referred earlier as a discontinuity of fatigue curves.

nounced levels, where one or the other fracture mechanism occurs with different probability for different specimens (Fig. 3).

Between ultrahigh and high cycle fatigue there is a rather wide bifurcation region for different metals with regard to different modes and conditions of their loading (Fig. 4). Ultrahigh cycle fatigue fracture is characterized by a further fatigue curve with new (different) parameters of equation (1).

It is evident that under cyclic loading metal behaves as a self-organized synergetic system, which changes energy dissipation mechanisms depending on the stress amplitude (under other constant conditions). Its stable behavior or evolution proceed on three scales: (1) microscale when a crack nucleates in the bulk (ultrahigh cycle fatigue region); (2) mesoscale when crack initiation is from a single surface site (high cycle fatigue region); and (3) macroscale when cracking starts from multiple surface sites. In transition to repeated static fracture, which

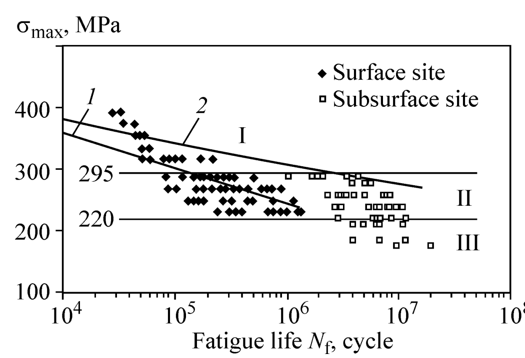

Fig. 4. Stress-life dependence for specimens from 2024-T3 aluminum alloy [8] and average values of the dependences for 2024-T351 alloy before (*1*) and after (*2*) laser surface hardening [10].

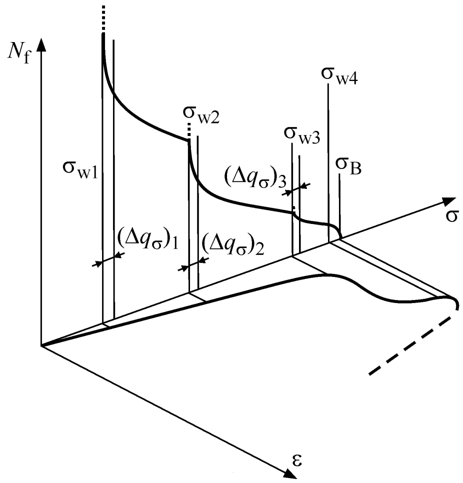

Fig. 5. N_f–σ bifurcation diagram of metal fatigue built with reference to the σ–ε tension diagram. Bifurcation regions $(\Delta q_\sigma)_i$ are distinguished in transition to the micro- or nano- $(\sigma_{w1} - \sigma_{w2})$, meso- $(\sigma_{w2} - \sigma_{w3})$, and macroscales $(\sigma_{w3} - \sigma_{w4})$.

is unstable, crack initiation sites are subsurface similarly to monotonic tension of the specimen to failure.

Generally under steady-state loading the material can endure long-term variable loads, which is shown in the bifurcation diagram (Fig. 5). Unlike the previous description of the metal behavior under cyclic loading, the diagram considers metal evolution in the direction of increasing stress, i.e. from the micro-, to meso- and then to macroscale.

On the microscale in the ultrahigh cycle fatigue region, metal behaves as a partially closed system as a crack is generated under the surface in the presence of external forces. The surface state can be arbitrary to some extent, including under aggressive environment, but crack nucleation is subsurface independently of the surface state [10]. The metal structure dictates its properties, including various stress concentrations at phase inhomogeneities, interfaces, inclusions, and others. Inhomogeneities of the stress state may appear as subsurface sites of fatigue cracks [18], not being material defects. The ultimate state depends on the statistics of distribution of inhomogeneities in the bulk, rather than damage accumulation in the surface layer. Therefore, an increase in fatigue strength on the micro- (or nano-) scale is possible only by changing the metal structure, making it purer, including due to a decreased percentage of residual gases or reduced size of inclusions [19]. A surface-initiate crack is possible on the microscale only when the surface itself is radically altered (deep corrosion pits, burn marks, mechanical deep scratches, and so on).

On the mesoscale in the high cycle fatigue region, the decisive role in metal behavior belongs to the surface layer and the surface itself [11, 12]. From the smooth surface without scratches and with relief of scarcely distorted geometry, dislocations move deep into the metal; the chessboard-like relief appears at the surface [20]. Curvature is of crucial importance here as resulting in high stress concentrations at the surface [14]. A crack generates at the surface of a specimen or a structural element at the instant all mentioned processes result in critical conditions of energy absorption under cycle loading due to a crack initiation site in the surface layer.

The external medium plays a crucial role in activating damage accumulation in the metal surface layer even when its composition does not lead to visible (identified) corrosion effects (Fig. 6).

For a surface-initiated crack, the surface layer as the main energy accumulator presents an independent subsystem and its role in crack initiation determines the fatigue life across the section at a given stress level. The

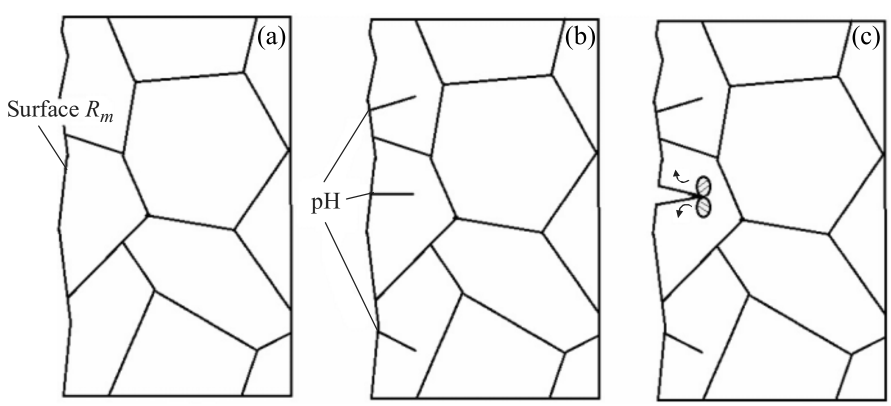

Fig. 6. Schematic of the surface layer state with different roughness R_m (a), with developed slip bands and under aggressive environment (pH—hydrogen ion exponent) (b), and with a crack along one of slip bands (c).

structure of the surface layer is governed, of course, by the material structure. However, its interaction with the environment makes it a place of accumulation of dislocation fluxes and cascades of slip bands whose certain density creates conditions for crack initiation at the surface [21, 22]. The environment is the object to which the metal surface exchanges continuously energy through chemical reactions, including in local heating under localized plastic deformation. Therefore, tests in vacuum even for corrosion titanium exemplify a substantial life increase at the fixed stress in the low cycle fatigue region [23]. The following is inferred from the study of short crack propagation at low stresses in various media in the electron microscope chamber for Ti–6Al–2Sn–4Zr–2Mo–0.1Si titanium alloy with the mixed structure (primary α-phase with an average size of 12.5 mm (+5.5 μm) and a volume fraction of 30% (+3) and mixed two-phase lamellar structure of α- and β-phases) [24]. A crack would propagate in air, which is known in advance; in transition to vacuum a crack is arrested and would not propagate till 10^7 loading cycles, growing afterwards at the surface from the already formed crack (Fig. 7).

Emphasize that a crack is initiated at the instant the critical molar volume is achieved. This is when the Gibbs potential becomes positive [13, 14] and crack edges (Fig. 6c) open due to a rotational mode of tension rather than a pure shear. At a reverse shear along the weakest slip plane due to decreased cohesion strength, cyclic deformation causes a plastic deformation zone to form around this weakening plane. As this takes place, the metal area near the formed stress concentrator opens along a slip plane and one boundary along the slip plane rotates in relation to the other.

On the macroscale in the low cycle fatigue region, the state of the metal bulk changes due to macroscopic plastic deformation sustained per loading cycle. Therefore, to increase the metal life in this deformation region both the metal bulk and surface layer should be affected. The simplest way to increase the fatigue life is through-the-thickness hardening, for example, by repeated forging, as with titanium alloy [25]. The material life grows by three times after through-the-thickness hardening as compared to unhardened specimens. The study of fracture surfaces reveals that the metal is in the state of a partially closed system, similarly to ultrahigh cycle fatigue, so that upon tension of specimens cut from forged pieces cracks are generated under the surface (Fig. 8) [10].

Thus, the metal behavior in a wide amplitude range of variable cyclic loading is characterized by the bifurcation fatigue diagram, which describes the metal behavior

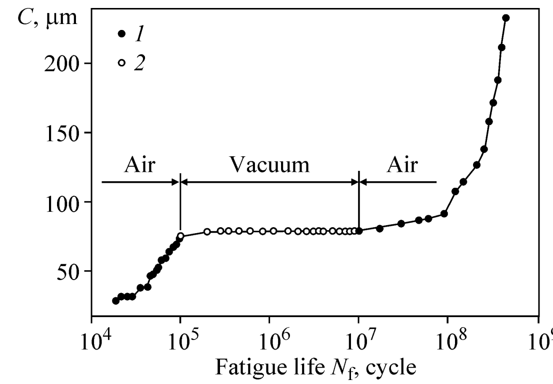

Fig. 7. Fatigue crack growth rate versus crack length at the surface of Ti–6Al–2Sn–4Zr–2Mo–0.1Si titanium alloy tested in air (*1*) and in vacuum (*2*) [24].

on the three scales and is illustrative of bifurcation transitions from scale to scale. Evolution proceeds with increasing scale; the multiscale metal evolution is generally described by the following system of equations:

$$N_f = \begin{Bmatrix} \dfrac{(C_f)_1}{\sigma^{(m_f)_1}} \\ \dfrac{(C_f)_2}{\sigma^{(m_f)_2}} \\ \dfrac{(C_f)_3}{\sigma^{(m_f)_3}} \end{Bmatrix} \begin{Bmatrix} \sigma_{w1} - \sigma_{w2} \\ \sigma_{w2} - \sigma_{w3} \\ \sigma_{w3} - \sigma_{w4} \end{Bmatrix} \begin{Bmatrix} 10^{12} < N_f < 10^8 \\ 10^8 < N_f < 5.10^4 \\ 5.10^4 < N_f < 500 \end{Bmatrix}. \quad (2)$$

Stress intervals which correspond to a particular scale of damage accumulation in metal are depicted in Fig. 5.

Earlier we pointed to the importance of two competitive processes of damage accumulation for metal, i.e. surface and subsurface ones. In this connection the notion of effective stress concentration factors should be considered.

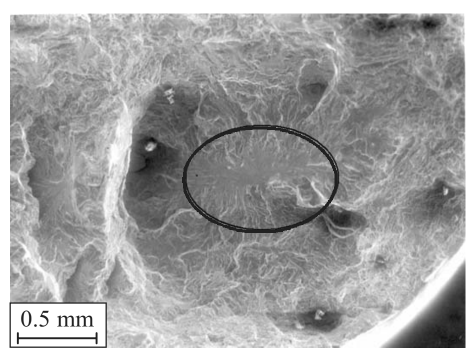

Fig. 8. Fracture surface appearance with the predominant facet relief and a subsurface fracture site (encircled) at the life 15 800 cycles and crack growth period about 3200 cycle [10].

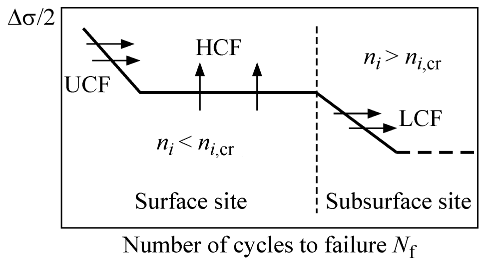

Fig. 9. Mughrabi's fatigue curve for different alloys [21, 22], n_i and n_{cr} are the current and critical numbers of bulk density of defects, respectively. UCF—ultrahigh cycle fatigue, HCF—high cycle fatigue, LCF—low cycle fatigue.

3. EFFECTIVE STRESS CONCENTRATION FACTORS

Crack initiation under the surface is explained on the basis of Mughrabi's model [21, 22], which reproduces metal evolution through the Wöhler diagram in the direction of decreasing cyclic loading amplitude (Fig. 9). The diagram describes metal evolution first on the macroscale (low cycle fatigue) and then on the meso- and macroscales. The mesoscale (high cycle fatigue) is represented by one stress level in the diagram, which means that it is missing. Such an approach to the description of metal behavior is governed by the fact that plastic deformation accompanies any level of metal loading that causes crack initiation and growth. Therefore, up to the fatigue life plateau the loaded metal exhibits a hysteresis loop, which corresponds to the low cycle fatigue region [22]. The loop closing is considered only at the transition boundary to the ultrahigh cycle fatigue region, i.e. in transition from the macro- to microscale. The model describes narrowing of the hysteresis loop as a single-scale process. The loop closes only in transition from low cycle to ultrahigh cycle fatigue; in so doing high cycle fatigue (the mesoscale) is omitted.

Note once again that the model describes metal behavior only on two scales: 1) presence of the hysteresis loop in each loading cycle and 2) absence of the hysteresis loop in each loading cycle. This approach eliminates the possibility of explaining the difference in the metal behavior concerning surface and subsurface sites of crack initiation because the hysteresis loop closing refers to the entire volume rather than to a certain metal area of fatigue crack nucleation. The approach is also incapable of justifying differences in metal behavior depending on single or multiple surface sites, which are however typical of metal cracking in high cycle and low cycle fatigue regions, respectively.

Why is the surface layer still able to relax energy without crack initiation in a certain stress interval while the bulk has already this possibility? To answer this question becomes more urgent since numerous investigations prove the generation of fatigue cracks in ultrahigh cycle fatigue due to a preliminary formed smooth facet, i.e. a fine-grained zone without any stress concentrator or material inhomogeneity [18].

The problem can be solved using approaches of physical mesomechanics, which takes the surface as an independent metal subsystem [11]. Let us consider metal evolution according to the above-stated hierarchy of scales and introduce the notion of effective stress concentration factors at the metal surface $\alpha_{\mathrm{eff}\,1}$ as an independent subsystem and in its bulk $\alpha_{\mathrm{eff}\,2}$ (Fig. 10).

With increasing stress the energy exchange with the environment becomes more active in the surface layer. The following processes occur simultaneously: (i) changing in the surface curvature as the main factor responsible for crack nucleation at the surface [13]; (ii) increasing of dislocation density and changing of its configuration; and (iii) increasing of density of slip bands [22] as geometric and physical stress concentrators which open at cracking due to the action of aggressive environment. Except for increased dislocation density and curvature of the dislocation relief, the material bulk in the hydrostatic stress state in the absence of exchange processes with the environment reveals no stress-dependent factors in the most weakened zone. Therefore, the effective stress concentration factor is constant in the metal bulk.

On achieving the critical curvature of the surface, critical density of dislocations and slip bands plus the saturation limit for environment gases, which agrees with the synergetics of the ultimate state in the surface layer, the effective stress concentration factors become similar and

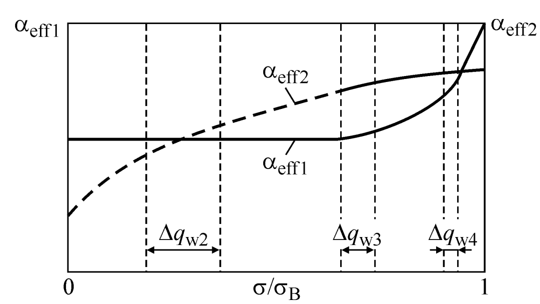

Fig. 10. Effective stress concentration factors $\alpha_{\mathrm{eff}\,1}$ and $\alpha_{\mathrm{eff}\,2}$ in the surface layer of metal and in its bulk, respectively, versus alternating stress devided by ultimate strength of the material.

surface and subsurface crack initiation sites are equiprobable. The material state varies strongly from specimen to specimen or from one structural element to another, which affects considerably the material ability to crack initiation at or under the surface. Therefore, with growing stress the probability of cracking from surface sites increases while that from subsurface sites decreases and vanishes.

In transition to the macroscale when plastic deformation proceeds through the thickness of a specimen or structural element, the material state changes in terms of a stress level and therefore the effective stress concentration factors begin to increase both in the bulk and in the surface layer. Since then the studied parameters vary with a different rate and in transition to repeated static fracture the bulk undergoes fracture due to highly constrained plastic deformation in the bulk stress state (starting with the internal volume). The probability of repeated static fracture increases and the probability of the low cycle fatigue mechanism decreases sharply with stress growth. In this regard the repeated static fracture region should be considered as a bifurcation region between fatigue and fracture processes in monotonic single tension of the material rather than another scale of fatigue.

The above notion of the influence of multiparameter action on a stress level, which corresponds to the transition from ultrahigh cycle to high cycle fatigue (Fig. 11), allows the following description of the upper boundary of the bifurcation region $(\sigma_{w2})_{max}$:

$$[(\sigma_{w2})_{max}]_e = [(\sigma_{w2})_{max}]_o F(X_1, X_2, ..., X_i). \quad (3)$$

In relation (3) values of $[(\sigma_{w2})_{max}]_e$ and $[(\sigma_{w2})_{max}]_o$ characterize the boundary of the bifurcation region under arbitrary loading conditions and under chosen standard (agreed in advance) loading conditions, respectively. The correction function $F(X_1, X_2, ..., X_i)$ takes into account the role of parameters of external action X_i on stress variation for the given boundary of the bifurcation region.

In the general case of the multiparameter action, relation (3) can be written as

$$[(\sigma_{w2})_{max}]_e = [(\sigma_{w2})_{max}]_o \times\left\{\sum_{i=1}^{k}\sum_{j=1}^{p}[A_p(X_i)^p]_i + \sum_{l=1}^{g} f[(X_i), (X_{i+1})]_l\right\}. \quad (4)$$

In relation (4) the sum $\sum_{i=1}^{k}\sum_{j=1}^{p}[A_p(X_i)^p]_i$ reflects a simultaneous change in several loading parameters. The correction term $\sum_{l=1}^{g} f[(X_i), (X_{i+1})]_l$ describes the synergetic effect of mutual influence of parameters of action. For

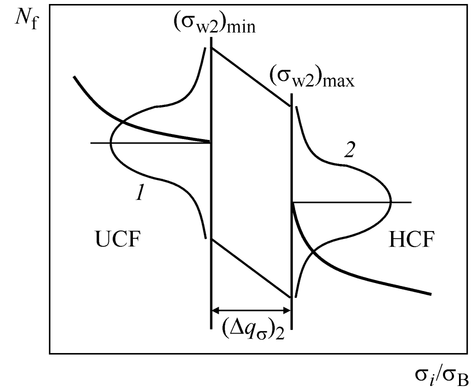

Fig. 11. Part of the bifurcation diagram, which renders a transition region from ultrahigh cycle to high cycle fatigue and possible motion of the region boundaries under external multiparameter action.

example, with a simultaneous change in temperature and loading frequency the influence of each parameter of action can be less than the effect of the only parameter on a stress level responsible for the bifurcation region boundary.

In the general case of one-parameter action, relation (4) can be rewritten as

$$[(\sigma_{w2})_{max}]_e = [(\sigma_{w2})_{max}]_o \times[A_0 + A_1X_i + A_2X_i^2 + A_3X_i^3]. \quad (5)$$

In relation (5) the coefficients A_i are derived experimentally from the found stress values for a variable parameter of action.

An example is the following functions for the studied transition boundary, which are derived experimentally for the stress ratio R and ambient temperature for high-strength steel SUJ2 with the ultimate strength 2316 MPa and air-hardened steel 1Cr–0.5Mo (SCMV 2), respectively [26, 27]:

$$[(\sigma_{w2})_{max}]_e = [(\sigma_{w2})_{max}]_o \times(354.06 - 255.47R - 121.84R^2 - 11.68R^3), \quad (6)$$

$$[(\sigma_{w2})_{max}]_e = [(\sigma_{w2})_{max}]_o \times(0.89 + 2.56\times10^{-4}t^0 + 1.3\times10^{-6}(t^0)^2). \quad (7)$$

We now turn our attention to the discussion of the scale hierarchy of metal fracture at the stage of crack growth.

4. STAGE OF THROUGH-THE-THICKNESS CRACK GROWTH

It is agreed that through-the-thickness cracks grow in three stages [5, 10]: (i) shear instability which causes the formation of specific fracture relief with stringers, pseudostriations and others; (ii) mode I or tensile opening of crack edges resulting in fatigue striations for many alloys; and (iii) unstable crack growth when the fracture surface demonstrates simultaneously the relief of the second fracture phase and pitted relief inherent in the final fracture stage after a half-cycle of loading in the ascending branch. The given three scales correlate with the physical mesomechanics ideas about three scales of deformation processes—micro, meso, and macro [13].

However, the scale hierarchy of fatigue (through-the-thickness) crack growth is differently understood at present [28]. Only two scales are distinguished—micro and macro. The transition boundary from one scale to the other has no physical substantiation. Moreover, the near-threshold region of crack growth (relative to the threshold stress intensity factor), which does exist, is not identified as a separate stage of crack propagation and is not considered. It is obvious that this approach does not reproduce the existing sequence of fracture mechanisms and contradicts the repeatedly discussed three-stage sequence in formation of various elements of the fracture surface [5, 10].

As a result of numerous studies of metal fracture mechanisms at the stage of through-the-thickness crack growth, their systematic behavior is proposed to describe by the following cascade of equations [10]:

$$\frac{da}{dN} = \begin{Bmatrix} C_1 \\ C_{Is} \\ C_2 \end{Bmatrix} \times K_e^{m_p} \dots \begin{Bmatrix} m_p = 4 \\ m_p = 2 \\ m_p = 4 \end{Bmatrix}$$

$$\dots \begin{Bmatrix} \delta_q < da/dN < \delta_{12} \\ \delta_{12} < da/dN < 2.1\times10^{-7}\ \text{m/cycle} \\ 2.1\times10^{-7}\ \text{m/cycle} < da/dN < \delta_{23} \end{Bmatrix}, \quad (8)$$

$$C_{Is} = (1-\nu^2)/8\, \bar{D}^2_{mean} \pi\sigma_{0.2}E, \quad (9)$$

$$C_1 = C_{Is}/\Delta^2 K^2_{Is}, \quad (10)$$

$$C_2 = C_{Is}\Delta^2/K^2_{Is}. \quad (11)$$

Equations (8) employ average rates and fracture relief parameters in the form of a fracture quantum δ_q, fatigue striation spacing δ_{12} and δ_{23} for transition boundaries between fracture mechanisms. In equation (9) the fractal dimension D_{mean} is introduced for the fracture relief formed at the second stage. Equations (10) and (11) represent the universal fracture constant based on Δ for alloys, which is proposed by Ivanova for the metal fatigue description [1].

However, the already conventional idea of the three-stage, multiscale, rather than two-scale, growth of fatigue cracks should be specified in view of the new data on the evolution of metal as an open synergetic system.

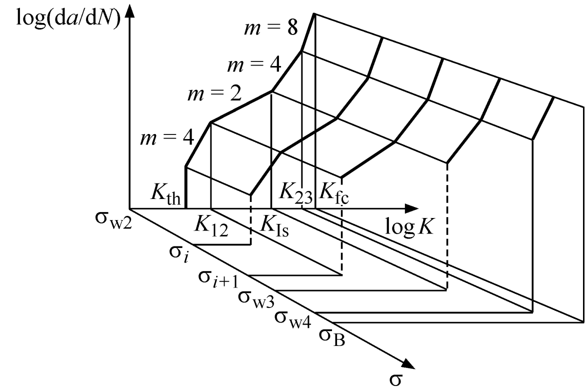

Fig. 12. Generalized surface of fatigue kinetic curves of crack growth, which is built with reference to the bifurcation diagram in Fig. 5.

1. The near-threshold region of crack growth rates near the stress intensity factor K_{th} (Fig. 12). An inconsiderable change in the current stress intensity factor causes an abrupt increase in rates by more than three orders of magnitude, which should be attributed to the region of unstable behavior of the material, i.e. to the region of the bifurcation transition from a solid material to a material with incipient cracks and their unstable growth. Crack growth is impossible to control in this region; any fluctuation exerts a pronounced effect on the crack growth period.

2. Transition to the microscale (or nanoscale). It usually occurs on achieving the average crack growth rate, which is determined experimentally and approaches a fracture quantum, being several lattice parameters [10]. On the microscale, a stable crack starts growing by a shear mechanism. It dominates until a transition to the second fracture stage when rotational plastic deformation and fracture become pronounced in the crack tip.

3. Transition to the crack growth rate at which rotational deformation and fracture are possible in the material. This is a bifurcation transition to the opening mode (in cyclic tension), which corresponds to the mesoscale. Mesoscopic cracks grow with the rate several tens of nanometers per loading cycle. In so doing, the fracture surface demonstrates a stable formation of fatigue striations in the descending branch of the cycle due to an isolated dislocation crack generated initially in front of the crack tip (Fig. 13). The kinetic curve for crack growth illustrates the first knee associated with a sharp decrease in crack acceleration. It is significant that the fracture surface fractography reveals both fracture mechanisms

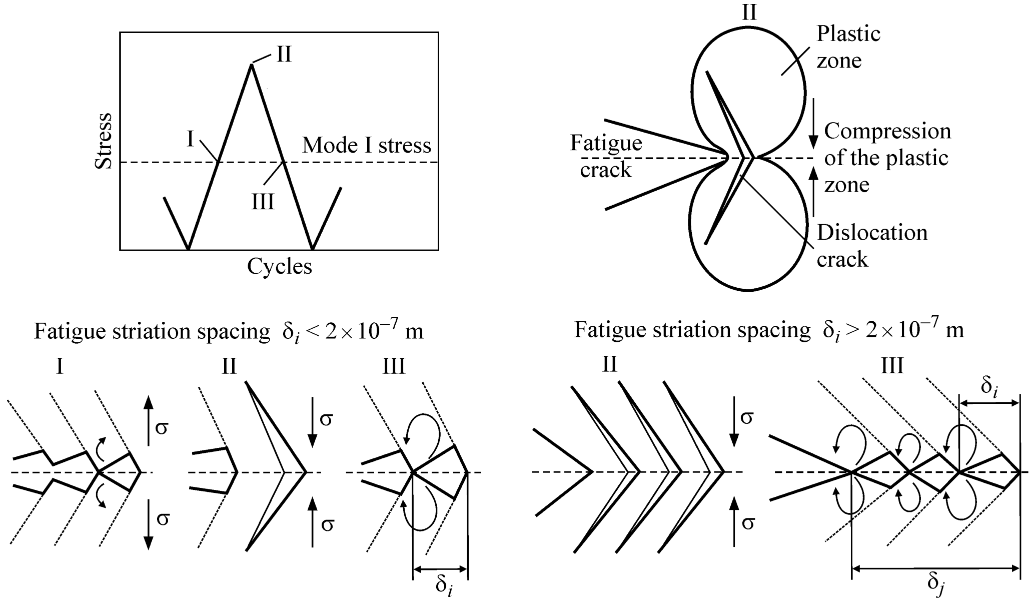

Fig. 13. Schematic of formation of the fatigue striation profile before and after achieving the spacing about $\delta_i \sim 2\times10^{-7}$ m/cycle [27].

(in different areas along the crack front) in transition from the micro- to mesoscale. Along the crack front the stress state varies in its intensity—from a three-dimensional state in the middle of the front to a two-dimensional one at crack edges near the specimen surface or structural element. A scale is changed to another when the crack front passes the bifurcation region.

4. The bifurcation transition to the macroscale, which occurs at the average crack growth rate $\sim 2\times10^{-7}$ m/cycle. The mechanism of formation of fatigue striations is dramatically different here [29]. In the loading cycle the striation profile is formed in the descending branch of loading due to an initially formed cascade of dislocation cracks. This process follows immediately the fatigue crack initiation in the low cycle fatigue region. Therefore, as previously indicated by Miller [30], the described stage of macroscopic crack growth in the high cycle fatigue region coincides with the material fracture in low cycle fatigue region. Cracks propagate to a small depth because their initial rate is high on account of the high stress intensity factor (see Fig. 12 for stresses more than σ_{w3}).

5. Transition to unstable crack growth, when a crack demonstrates signs of fracture inherent in static sliding, which is typical of the material fracture under monotonic tension; it might be considered as a bifurcation region. The material itself shows unstable behavior. Inconsiderable fluctuations lead to a radical increase in crack growth rate. This region reveals groups of fatigue striations, whose fraction reduces sharply along the growing crack while the pitted relief covers an increasing area of the fractured metal. Activation of deformation and fracture is also pronounced in that the striation relief appears simultaneously in both loading branches [29]. The ascending branch corresponds to plastic blunting of the crack tip, which manifests itself in the formation of the striation front shaped into the so-called extension zone. The descending branch corresponds to the final formation of the fatigue striation profile, as in the case of the macroscale.

The aforesaid suggests that with the stress growth causing crack nucleation and initiation, first the bifurcation region is missing (transition to the first stage occurs immediately without the near-threshold region). At a further (see Fig. 12) increase in stress the stage of near-threshold crack growth and the microscale are omitted. On achieving the stress of low cycle fatigue, small cracks start growing on the macroscale. The physical sequence of fracture mechanisms and their operation at any stress level remain unchanged.

Mechanisms of crack initiation and growth from internal sites were detailed elsewhere [18] and, therefore, are not discussed here.

5. CONCLUSION

Metal evolution under cyclic loading corresponds to the behavior of a synergetic system, which occurs as a self-organized process on three scales in the direction of increasing scale.

At the stage of crack initiation and growth, fatigue fracture of metals presents a cascade of consecutive evolution mechanisms due to damage accumulation on three scales and transitions from the micro- to meso- and further to macroscale through unstable behavior regions, which present bifurcation points.

In the bifurcation region the fatigue life demonstrates a bimodal distribution where with increasing stress the probability of a damage accumulation mechanism on the previous scale decreases and its probability on the next scale increases up to 100%.

At the crack growth stage in the bifurcation region, fracture along the crack front proceeds simultaneously by mechanisms of the previous and next scales with different probability, which is seen in the formation of a mixed fracture relief increasing along a growing crack, as with fracture on a higher scale.

Near-threshold crack growth, transition from low cycle fatigue through repeated static fracture to final failure, which corresponds to the stage of unstable crack growth or to transition from the macroscale to fracture of the specimen under single tension, should be attributed to an instability region or bifurcation region.

With increasing stress the material passes not all fracture stages inherent in it; it omits stages characteristic of lower values of crack initiation stress or stress intensity factor at the crack growth stage. New or different fracture mechanisms do not arise and do not manifest themselves at a higher stress level.

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