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TOTAL LIFE APPROACHES FOR METALLIC COMPONENTS FATIGUE BEHAVIOR

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A small crack can be thought of as a crack with a size on the order of the microstructure. In the early stage on fatigue behavior, the defects in internal structure of material such dislocation, vacancies and others present an very important role. The behavior of small cracks differs from the behavior of long cracks. The linear elastic crack growth is modeled using the Paris law. The criteria concerning the number of cicles to fatigue failure are presented in this paper. The total fatigue life is the sum of the life spent on nucleation, small crack growth and long crack propagation.

1. GENERAL FATIGUE

Fatigue is the name given to failure in response to alternating loads (as opposed to monotonic straining) and causes more than 90% of all catastrophic failures of structures.

This is manifested by material failure at fluctuating stress levels below those leading to fracture at static loading. The material is exposed to these fluctuating stress levels for long time, or rather, for many cycles.

A typical stress history during cycling loading is depicted below (fig. 1).

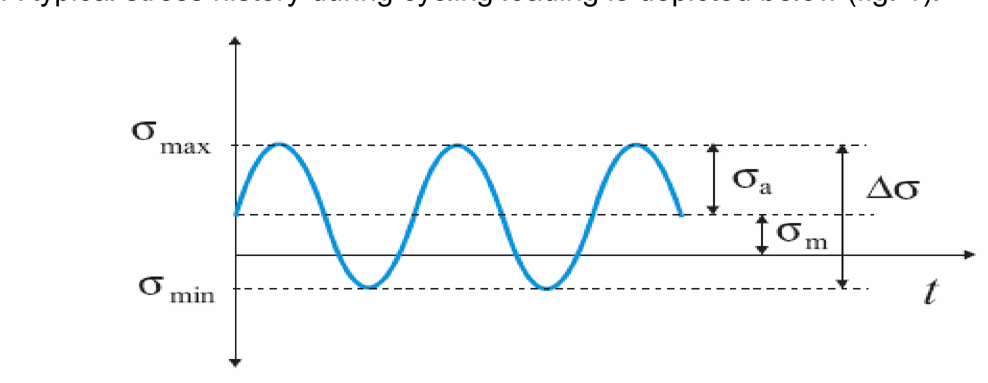


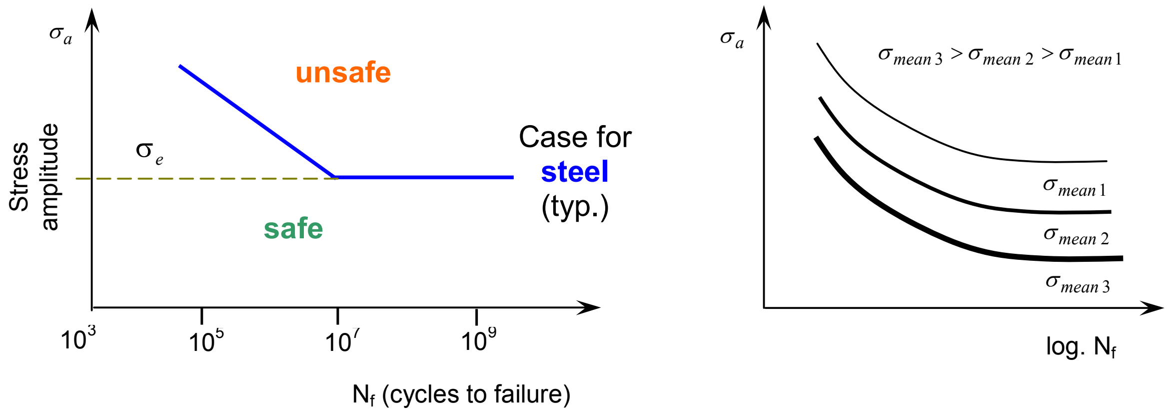
Fig. 1

The most important parameters that characterize a cycle loading are:

- **Stress Range**: $\Delta\sigma = \sigma_{\max} - \sigma_{\min}$;
- **Stress Amplitude**: $\sigma_a = \frac{1}{2}\cdot(\sigma_{\max} - \sigma_{\min})$;
- **Mean stress**: $\sigma_m = \frac{1}{2}\cdot(\sigma_{\max} + \sigma_{\min})$;
- **Load Ratio**: $R = \frac{\sigma_{\min}}{\sigma_{\max}}$.

The stress varies with time and the key parameters are σ_a and σ_m.

If a plot is made of the applied stress amplitude σ_a versus the number of reversal to failure (S–N curve) the following behavior is typically observed (fig. 2a):



a) ***b)***

Fig. 2

From fig. 2.b results that if the number of cycles in the loading history is greater, the stress that material can withstand without failure is smaller.

The value σ_e represent the endurance limit or fatigue limit.

For most steels and copper alloys, it is considered that:

$$\sigma_e \approx 0.35 \cdot \sigma_{TS} \div 0.5 \cdot \sigma_{TS} \tag{1}$$

This is typically associated with the presence of a solute (carbon, nitrogen) that pines dislocations and prevents dislocation motion at small displacements or strains (which is apparent in an upper yield point).

If the material does not have a well defined σ_e, often σ_e is arbitrarily defined as the stress that gives $N_f = 10^7$ cycles.

The component has effectively *infinite life* if the stress is below σ_e.

Consider a structure in which a crack develops. Due to the application of repeated loads or due to a combination of loads and environmental attack this crack will growth in time. The longer the crack, the higher the stress concentration induced by it. This implies that the rate of crack propagation will increase with time.

There are two main types of fatigue. If the lifetime is limited to approximately 10^3–10^5 stress cycles, we are dealing with *Low Cycle Fatigue* (LCF). In this type of fatigue, the material is subjected to stress levels that produce macroscopic plastic strains. The standard engineering approach to solving LCF problems is to use Coffin–Manson relationship to calculate the fatigue life as a function of the number of applied strain cycles.

The other main type of fatigue is designated *High Cycle Fatigue* (HCF). Here, the material is subjected only to microscopic plastic strains and the number of cycles to fracture can be millions. In this case it is often not possible to use the same approach of modeling as in the case of low cycle fatigue.

Fatigue crack propagation predominate the fatigue life in LCF, while crack initiation predominates the life in HCF.

A broad field to fracture mechanics is presented in figure 3.

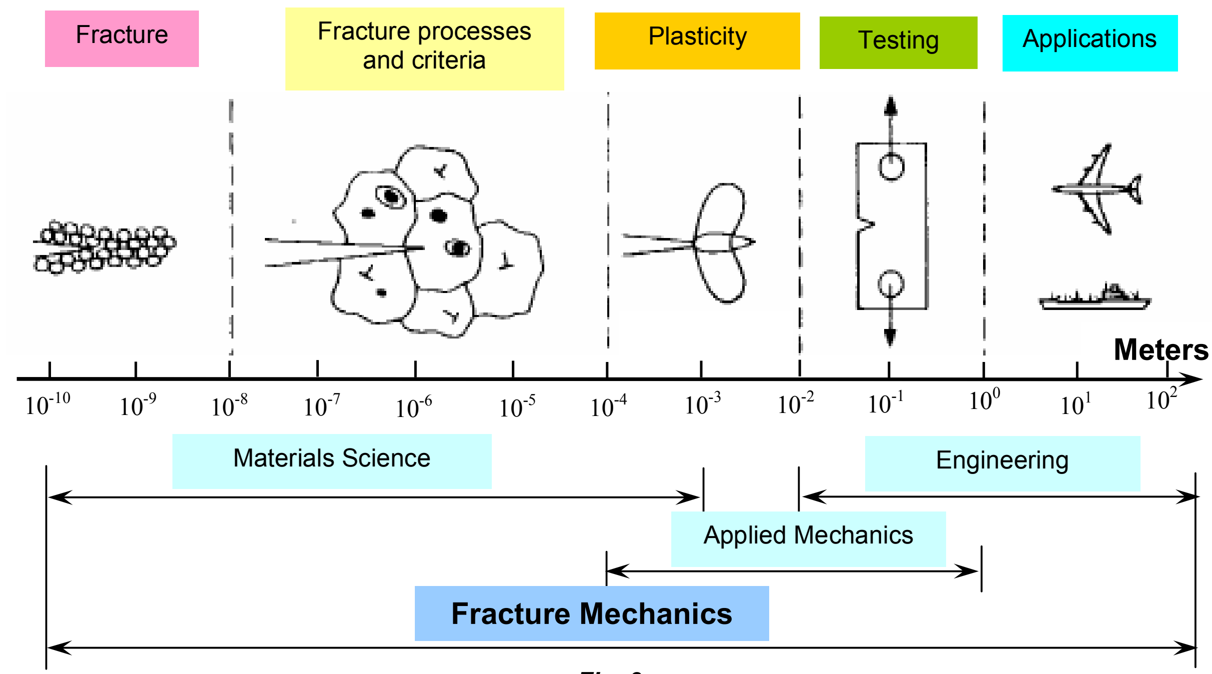


Fig. 3

2. FATIGUE CRACK STAGES

Current fatigue life prediction methods in metallic components consider three stages: crack initiation, long crack propagation and final fracture (fig. 4).

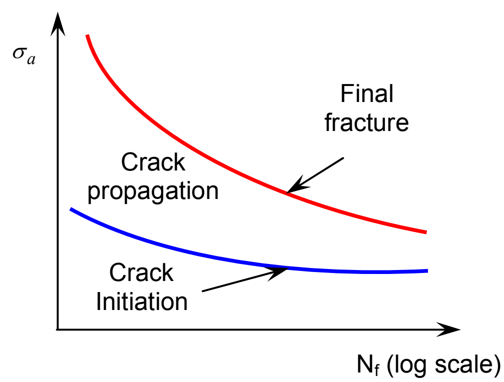


Fig. 4

Crack initiation is the early stage of damage accumulation where small cracks (cracks with depths less than several grain diameters) have been observed to deviate significantly from predicted long crack fracture mechanics behavior. The deviation is attributed to heterogeneous media in which small cracks evolve.

Long crack propagation and final fracture are the stages of damage accumulation that are well characterized using Linear–Elastic (L.E.F.M.) or Elastic–Plastic Fracture mechanics (E.P.F.M.).

Each stage is driven by different mechanisms and must be distinctly modeled. The stages must be quantitatively linked because the crack growth successively from one stage to the next. Failure of the macrostructure is defined by the first crack to nucleate and grow beyond the critical size.

The evolution to failure is presented in figure 5.

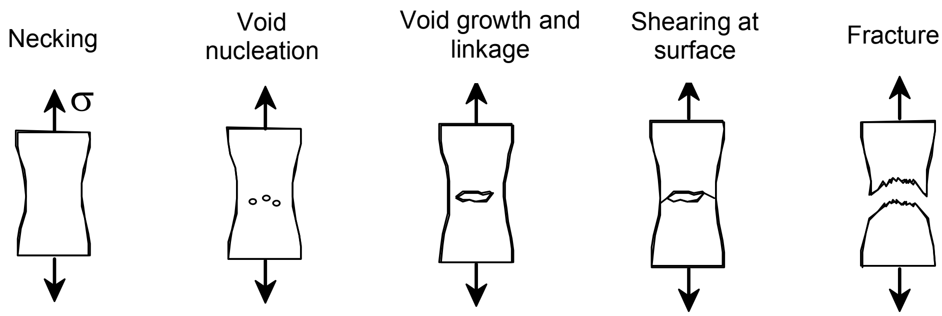
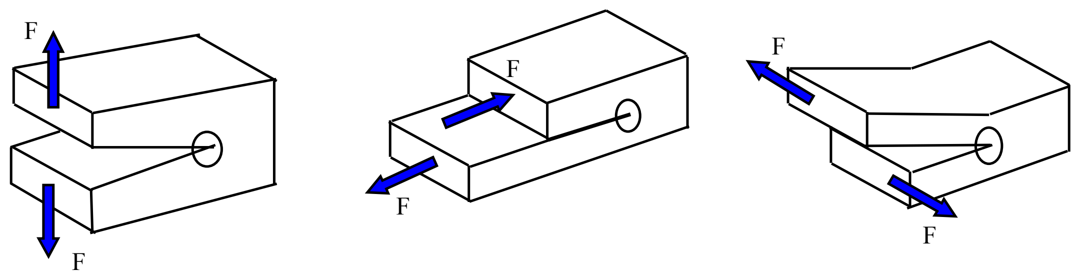


Fig. 5

3. CRACK INITIATION STAGE

The crack initiation stage can be broken into two phases: *crack nucleation* and *small crack growth*. Crack nucleation is the locally complex process of crack formation on the microstructural scale. It is characterized by smooth fracture surfaces at angles inclined to the loading direction. This type of failure is indicative of shear stress Mode II (sliding mode) fracture, fig. 6b. Experimental evidence suggests that the nucleation size is on order of the grain size.

Small crack is characterized by fracture surface striations perpendicular to the loading direction. This type of failure is indicative of tensile Mode I (opening mode) fracture, fig. 6a.



a) ***b)*** ***c)***

Fig. 6

The behavior of small cracks tend to transition to linear or elastic–plastic fracture mechanics behavior when the crack depth reaches about ten mean grain diameters. Crack nucleation and small crack growth must be modeled separately because different mechanisms control each phase.

The relative importance of the crack nucleation stage on overall fatigue life depends on several factors. Materials which exhibit a strong preference for planar slip show a strong correlation between the crack causing final fracture and the earliest nucleated cracks. Materials which prefer cross slip showed almost no correlation between the crack causing final fracture and the earliest nucleated cracks.

The crack nucleation stage may also depend on the loading condition. If the loading is relatively low (High Cycle Fatigue), the majority of life will be spent in the nucleation of a crack. If the loading is high (Low Cycle Fatigue), cracks may nucleate early and spent the remainder of the fatigue life in the crack growth stages. However, high strength materials have been shown to spend the majority of fatigue life in the crack nucleation stage, even during low cycle fatigue.

Fatigue crack nucleation is a complex and obscure process. The mechanisms for crack nucleation change with material, loading, temperature and environment. One overriding observation is that cracks tend to nucleate near the free surface. For many loading conditions, the highest loads are at the surface. But even the nominal stress is constant throughout, crack tend to nucleate at the surface because deformation of each grain is allowed to concentrate on a crystallographic plane. In the interior, deformation on a single crystallographic plane is hampered by the constraints of the surrounding grains.

The cracks initiated at a depth from 0 to 2 μm are classified as ***surface cracks***. The selection of 2 μm is based on two considerations. First, in the experiments a large number of surface cracks did not reach a depth of more than 2 μm during the crack initiation stage. Secondly, for a crack initiated within a depth of 2 μm, the portion of life to propagate to the surface is so small that it has almost the same life as a surface initiated crack. The range for ***near surface crack*** initiation is from 2 μm to 0.4·B (where B is half Hertzian contact width). Since the boundary effects are very small at the depth larger than 0.4·B, damage outside of that range is in the regime for ***subsurface crack*** initiation.

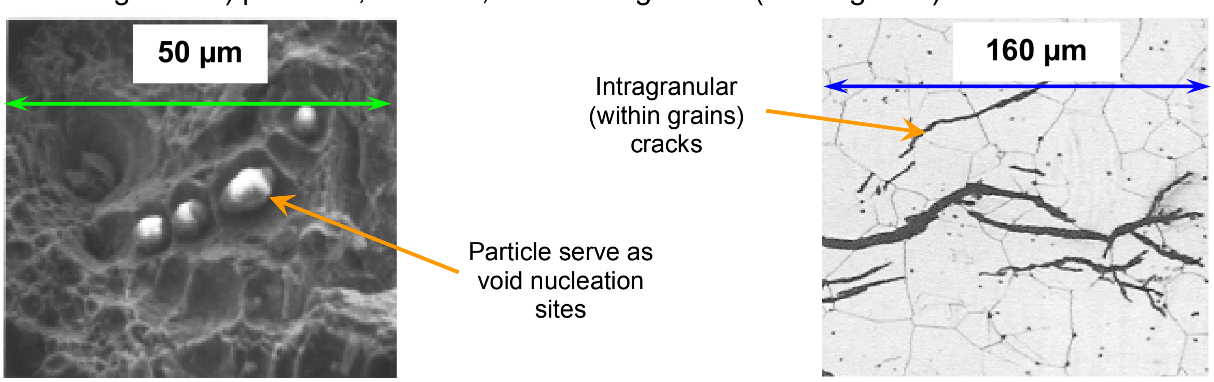
Experimental evidence clearly shows that defects in the material can cause fatigue crack nucleation by acting as stress concentrations and the cracks tend to nucleate along the preferred slip planes. The cracks are believed to be the combined result of vacancy creation, repulsive dislocation stresses and surface roughening stress concentrations.

There are two fundamentally types of slip–band–induced crack nucleation. The first consider that a very small crack (much smaller than the grain size) nucleates along the slip plane very early in life. A crack is evident from crack opening displacement when a static load is applied. The size of plastic zone is relatively small, being equal to, or less than the crack size. The crack propagates in Mode II until it reaches an obstacle, often the grain boundary.

The more prevalent though less recognized slip band induced crack nucleation is sudden crack nucleation. In sudden crack nucleation, a slip band which stretches across the grains forms very early in life but no crack is formed. Upon continued cycling, the slip band is blocked by the grain boundary and does not grow in length. The depth and the width of the slip band increase slightly until, suddenly, a crack is form. This slip band crack nucleation is observed in many alloys including steel.

Examples of defects include pores, ceramics inclusions, second phase particles and microcracks (fig.7 a).

Figure 7 b) presents, for steel, small intragranular (within grains) cracks.



a) ***b)***

Fig. 7

Few analytical or semiempirical models regarding the initiation stages of steel cracks are presented below.

Microstructural models which predict crack nucleation life and crack nucleation size have been proposed by many researchers.

Tanaka–Mura's model is applicable to metallic components for which crack nucleation takes place by transgranular shear stress fracture. The dislocations pile up at the grain boundary which acts as an obstacle to dislocation movement. The dislocation movement is assumed to be irreversible such that when the reverse load is applied, dislocations of the opposite sign pile up on a closely spaced plane. Since the residual load from the back stress of the positive dislocations act in the same direction as the reverse applied load, unloading will cause negative dislocation movement. During each of the subsequent load cycles, the number of dislocations monotonically increases.

Crack nucleation takes place when the total stored energy after N_n cycles is equal to the fracture energy of the grain. Based on this fact, Tanaka–Mura have proposed next relationship between the grain size and specific fracture energy (eq. 2)

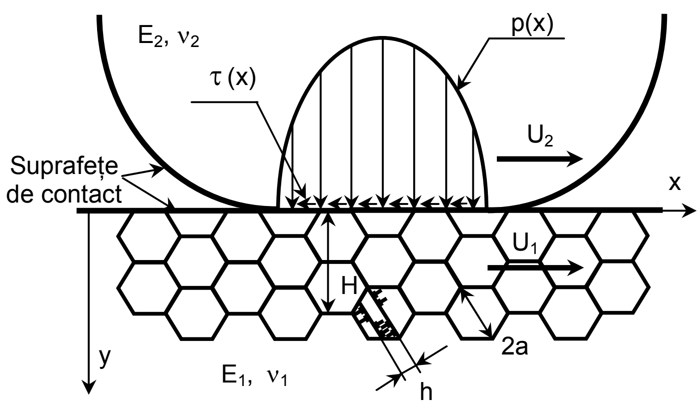
$$N_n = \frac{4GW_s}{(\Delta\tau - 2k)^2 \pi (1-\nu) d} \tag{2}$$

In equation 2, d represents the grain diameter, W_s is the fracture energy per unit area, k is the frictional stress which must be overcome to move dislocations, G is the shear modulus, v is Poisson's ratio and $\Delta\tau$ can be calculate using eq. 3:

$$\Delta\tau = 2k + \left[\frac{4GW_s}{\pi(1-\nu)}\right]^{\frac{1}{2}} d^{-\frac{1}{2}} \tag{3}$$

The behavior of small cracks differs from the behavior of long cracks. Small crack growth rates vary widely, from several orders of magnitude greater than that predicted by continuum based ΔK to complete arrest. A small crack can be thought of as a crack with a size on the order of the microstructure. The anomalous growth of small cracks has been attributed to two competing factors: high growth rates due to lack of closure and plane stress at the surface and growth retardation due to microstructural obstacles.

Cheng's model is based on boundary condition for the two–body contact system, which will influence the stress field of dislocation and its pileup (fig. 8).



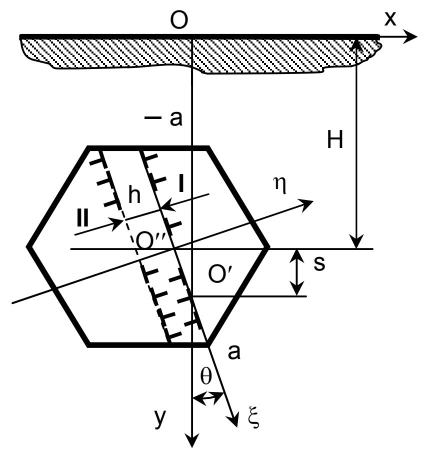


Fig. 8

The more general equation for crack initiation either on the surface or subsurface is:

$$N_i = C^* \frac{2}{(\Delta\tau - 2\tau_f)^{1+q}} = C_1 C_2 \frac{2}{C_3(\Delta\tau - 2\tau_f)^{1+q}} \tag{4}$$

with $C_1 = \frac{\gamma}{h\left[\ln\left(\frac{8a}{h}\right) - \frac{3}{2}\right]}$; $C_2 = C_2(\theta, H)$, where γ is the surface energy, $\Delta\tau$ – critical shearing stress amplitude resolved on the slip layer, *2a* – dislocation pileup length (or grain size), θ – slip band inclination angle, *h* – width of slip band (dislocation dipole), *H* – depth of crack initiation position from the contact surface.

4. FATIGUE CRACK GROWTH

The fracture mechanics approach is based on the assumption that the crack tip conditions are uniquely defined by a single loading parameter, e. g. the ***stress intensity factor***, ***K****.* In the case of propagation of cracks, the range of the stress cycle is used. Also, the mid value of stress will have an influence (for instance due to the effect of hydrostatic stress acting on the crack) and the stress history may have an interest. Under these assumptions, the crack propagation can be characterized by the relationship (5):

$$\frac{da}{dN} = f(\Delta K, R., H) \tag{5}$$

where $\frac{da}{dN}$ is the crack growth per cycle, $\Delta K = K_{\max} - K_{\min}$, $R = \frac{K_{\min}}{K_{\max}}$ and H is a history term, which may have influence if $K_{\max}$ or $K_{\min}$ varie during the load history.

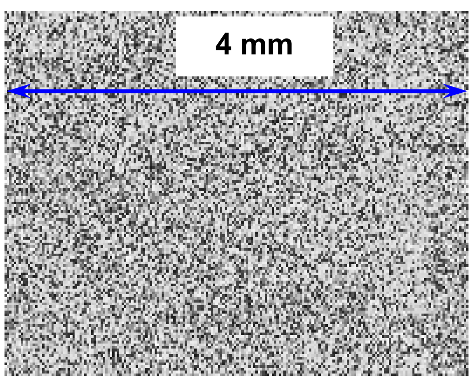
The linear elastic crack growth is modeled using the Paris law representation of a surface crack in a semi infinite body subjected to a constant stress cycle.

$$\frac{da}{dN} = C \cdot \Delta K^n \tag{6}$$

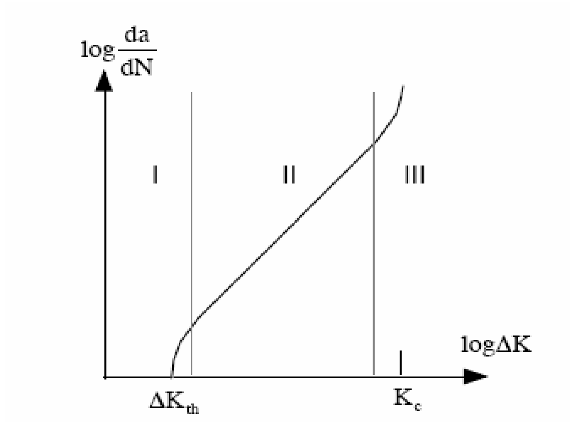
where $\Delta K = \beta \cdot \Delta s_{xx}\sqrt{a}$, a is the crack length, N is number of cycles, ΔK is the stress intensity factor range, Δs_{xx} is the stress range, β is a geometric constant ($1.12\sqrt{\pi}$), C and n are material properties. Condition for crack propagation is:

$$K \geq K_C \tag{7}$$

where K depends on load and geometry and K_C is the *fracture toughness* (depends on the material, temperature, environment and rate of loading). Figure 8 a) presents intergranular (between grains) cracks.



a)



b)

Fig. 9

The fatigue crack growth behavior in metals is described by the crack growth rate (da/dN) vs. the width of the stress intensity factor during one loading cycle (ΔK), fig. 9 b).

The number of cycles needed for the crack to grow to failure, N_g, can be determined with equation 8:

$$N_g = \frac{a_i^{1-\frac{n}{2}} - a_f^{1-\frac{n}{2}}}{C \cdot \Delta s_{xx}^n \cdot \beta^n \left(\frac{n}{2}-1\right)}, \quad n \neq 2 \tag{8}$$

where a_i is the initial crack size and a_f is the failure crack size.

If $n > 2$ and $a_i << a_f$, then $a_i^{1-\frac{n}{2}} >> a_f^{1-\frac{n}{2}}$, equation 8 can be written as:

$$N_g = \frac{a_i^{1-\frac{n}{2}}}{C \cdot \Delta s_{xx}^n \cdot \beta^n \left(\frac{n}{2}-1\right)} \tag{9}$$

5. CONCLUSIONS

The fatigue process is divided into three phases. The first phase is the crack nucleation. Dislocations pile up at the grain boundaries with each load cycle. When the energy associated with the dislocation pile up exceeds a critical value, a crack forms along the slip band of the grain.

The second phase is the small crack growth phase and the crack growth rate is modeled as a function of the crack tip opening displacement. The local microstructural variables considered random are: grain size, grain orientation, micro-stress and frictional stress. The variables are common to both the crack nucleation and small crack growth models.

The third phase is the long crack growth phase. The long crack growth rate is modeled using Paris law.

The distribution of number of cycles spent in the nucleation, small crack growth and long crack growth was presented.

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