Fatigue of carbon and of alloy steels

Hans-Jürgen Christ (Prof. Dr.)
Institut für Werkstofftechnik
Universität Siegen
D- 57068 Siegen
Germany
3a Fatigue of Carbon and of Alloy Steels

Fatigue failure of components made of steel and used in vehicles and other technical systems revealed already in the middle of the last century that a detailed understanding of the processes taking place during cyclic loading is a prerequisite for reliability and safety in service. At that time, steels were the most important engineering materials, and this still holds true today as regards structural applications. Probably the most striking reason for the outstanding importance of steels is the possibility to adjust suitable mechanical properties within a wide range by means of changing the composition and the thermo-mechanical pretreatment. Thus, the term steel applies to a broad spectrum of materials with extreme variations in strength, composition and microstructure. As an example, the yield strength of steels at room temperature ranges from less than 100 MPa in the case of unalloyed low-carbon steels to 3.5 GPa of experimental maraging steels.

Mostly steels are divided into classes according to their crystal lattice and main microstructural constituent. The most common classes are austenitic (face-centered-cubic, fcc) steels, ferrite-pearlite or bainitic (body-centered-cubic, bcc) steels, martensitic (body-centered tetragonal, bct) steels, and duplex steels (fcc + bcc). There are many particular classes of steels from which the high-strength low-alloy (HSLA) steels and the maraging steels should be mentioned because of their steadily increasing importance as materials for structural parts of automobiles and aircrafts, respectively.

Within the scope of this article, it is naturally not possible to cover all the numerous and different aspects of the fatigue behavior of steels. Rather, some basic characteristics and general tendencies will be depicted. Furthermore, some effects which are typical of steels as compared to other classes of metallic materials will be dealt with. For this purpose, the article is organized into four major sections. First, the main fatigue characteristics of iron are briefly introduced, since these may define a kind of a reference behavior for unalloyed and low-alloy steels. Some
specific aspects of the cyclic stress-strain behavior are presented in section 2. In the third section some information on cyclic lifetime of steels is provided based on the concepts of the phenomenological and the defect tolerant approach, respectively. The last section is devoted to dynamic strain aging, since this phenomenon strongly affects the temperature dependence of the fatigue behavior of steels.

1. Fatigue Characteristics of \( \alpha \)-Iron

Pure \( \alpha \)-iron is not of importance as a structural material. However, \( \alpha \)-iron is the major constituent of many steels and often determines the fatigue behavior. For example, cyclic plastic deformation in ferrite-pearlite steels is almost completely restricted to the ferritic grains up to a carbon content of 0.45wt.% (Eifler 1985). Only at higher carbon concentrations, the ferritic plates of pearlite may contribute. \( \alpha \)-iron represents also a model material for the study of characterisitic fatigue mechanisms of bcc metals and alloys. From this viewpoint it appears surprising that \( \alpha \)-iron has received only little attention in fundamental studies yet.

The mechanical properties of pure bcc metals depend very strongly on temperature and strain rate and are sensitive to small amounts of interstitial impurities (such as carbon, nitrogen, and oxygen). In Fig. 1, a schematic representation of the temperature dependence of the flow stress \( \sigma_f \) typical of a bcc lattice is represented. The flow stress can generally be separated into an athermal component \( \sigma_G \) and a thermal component \( \sigma^* \) (Seeger 1954).

\[
\sigma_f = \sigma_G + \sigma^* (\dot{\varepsilon}_{pl}, T)
\]  

As indicated in Eq.1, \( \sigma^* \) depends on temperature \( T \) and plastic strain rate \( \dot{\varepsilon}_{pl} \) and is essentially a consequence of the extended core structure of the screw dislocations. The movement of a screw dislocation is a process which is thermally activated, i.e. the necessary energy can be brought up thermally or by a mechanical stress at low temperatures. At temperatures above the transition temperature \( T_0 \) the thermally provided energy is sufficient and hence \( \sigma^* \) disappears. Below \( T_0 \)
thermal activation is too small and the additional mechanical stress $\sigma^*$ is necessary to overcome the lattice friction stress for screw dislocations. $T_0$ depends on the plastic strain rate, since at low $\dot{\varepsilon}_{pl}$ more time is available to successfully perform thermally activated processes (i.e. $T_0$ decreases). The second flow stress component $\sigma_G$ depends only weakly on temperature. It arises from the elastic interaction of dislocations and is related to the dislocation density $\rho$ by the equation:

$$\sigma_G = \alpha G b \sqrt{\rho}$$

where $b$ is the modulus of the Burgers vector, $G$ the shear modulus and $\alpha$ a geometrical constant. As a consequence of the low mobility of screw dislocations in bcc metals, cyclic plastic deformation at low temperature and small plastic strain amplitudes $\Delta\varepsilon_{pl}/2$ is carried mainly by edge dislocations in such a way that the gliding edge dislocations draw out long straight segments of sessile screw dislocations. Therefore, the dislocation multiplication rate by dislocation bowing is very low under these conditions. With increasing temperature the screw dislocations gain mobility until they become comparably mobile as edge dislocations.

Cyclic plastic deformation of pure $\alpha$-iron leads to the formation of a saturation dislocation arrangement in the form of a cell structure (Sommer et al. 1998 pp. 1527-1536). The cumulative hardening increment in tests with constant $\Delta\varepsilon_{pl}/2$ first increases with increasing temperature as a consequence of the rising dislocation multiplication rate and then decreases due to the increasing annihilation rate. The diameter of the dislocation cells formed which usually is expected to decrease continuously with increasing temperature (due to decreasing saturation stress amplitude $(\Delta\sigma/2)_S$) shows a minimum at intermediate temperature, since the difference of $(\Delta\sigma/2)_S$ and the friction stress is maximum there.

The pronounced strain rate dependence of the flow stress may strongly affect the cyclic stress-strain behavior of $\alpha$-iron. An example is given in Fig. 2 that shows the cyclic stress-strain curve (CSS-curve) for various temperatures determined in tests performed under plastic-strain control.
with constant absolute value of plastic strain rate $|\dot{\varepsilon}_{pl}|$. While at sufficiently high temperature (400 K) a “normal” behavior is observed, a minimum occurs at low temperatures and small $\Delta\varepsilon_{pl}/2$. The stress amplitude increases below this minimum with decreasing plastic strain amplitude. A comparison with the behavior of monocrystals documents that this inverse CSS-behavior is restricted to polycrystals. At low values of $\Delta\varepsilon_{pl}/2$ plastic deformation does not take place in a uniform manner in polycrystalline iron. Rather, some grains are plastically deformed, while others remain in a pure elastic condition. With decreasing $\Delta\varepsilon_{pl}/2$ the volume fraction of the plastically deforming grains decreases and under the test conditions applied the local plastic strain rate in these grains increases. As a consequence, the thermal stress component $\sigma^*$ increases, giving rise to higher stress amplitudes.

This behavior also affects the number of cycles to failure $N_f$ (Sommer et al. 1998 pp.1537-1546). Figure 3 presents the cyclic lifetime data that corresponds to Fig. 2 in a Coffin-Manson plot ($\Delta\varepsilon_{pl}/2$ vs. $N_f$). The unusual behavior in the CSS-curve at low temperatures and small value of $\Delta\varepsilon_{pl}/2$ manifests itself in an inverse dependence of $N_f$ on the plastic strain amplitude. Under the conditions mentioned above, cyclic life is shortened when $\Delta\varepsilon_{pl}/2$ is reduced. Hence, here the Coffin-Manson relation does not hold true. Rather, cyclic life is determined by the stress amplitude as can be shown by plotting the data in the form of a so-called derived Wöhler curve (resulting stress amplitude vs. $N_f$). In this representation all values can be described by a common curve irrespective of exhibiting an usual or an unusual behavior in the Coffin-Manson plot.

Carbon is the most important alloying element in iron, especially when considering the technical importance of the iron-carbon system. At room temperature, the carbon solubility of iron is very small. However, even a low concentration of these interstitial solute atoms change the fatigue behavior tremendously. In the low-temperature regime, the addition of some ppm carbon to iron leads to a solid solution softening as compared to the pure condition. This effect can be
attributed to an enhancement of the mobility of screw dislocations. At temperatures slightly above the transition temperature, small carbon contents strengthen the material and changes the dislocation arrangement completely. While in pure iron cell structures prevail, slightly carburized samples exhibit a strain localization in the form of persistent slip bands (PSBs) which show the structure of dislocation-poor channels and are similar to the PSBs found in low-carbon steels (Wilson and Tromans 1970, Abdel-Raouf and Plumtree 1971, Pohl et al. 1981).

At elevated temperatures the effect of solute carbon is most pronounced. A strong interaction of carbon with dislocations takes place leading to an extremely enhanced stress amplitude. This so-called dynamic strain aging is dealt with in more detail later.

The type of fatigue crack initiation on the surface of pure $\alpha$-iron depends also very strongly on temperature. At low temperature a change of the shape of the grains gives rise to intergranular cracks as a consequence of the obviously incompatible plastic deformation. At higher temperatures a surface rumpling is reported and explained with the action of asymmetric glide planes in tension and compression. Cracks initiate within the valleys of this surface rumpling. The PSBs which are formed as a consequence of small carbon contents at intermediate temperatures can be found as slip bands on the surface. In this case, surface rumpling is suppressed due to the complete change in the dislocation arrangement. Cracks initiate transgranularly at the PSBs, but these cracks can be considered to be relatively harmless. Despite a high initiation rate their propagation occurs relatively slow and hence an extended fatigue life results as compared to pure $\alpha$-iron.

2. Cyclic Stress-Strain Behavior of Steels

One of the key behavioral responses to cyclic loading that is typical of ferritic-pearlitic steels in normalized condition is dynamic cyclic Lüders-band propagation. This effect is very closely related to the phenomenon of distinguished upper and lower yield stresses and the appearance of
discontinuous yielding in monotonic straining and can be attributed once again to the interaction of interstitial atoms (mainly carbon and nitrogen) with dislocations in the bcc lattice. The manner in which dynamic Lüders-band propagation affects the cyclic stress-strain response depends very sensitively on the testing mode applied. The effect is most striking under stress control (constant stress amplitude) in combination with a stress amplitude below the technical yield stress. Figure 4 (Eifler and Macherauch 1988) depicts the cyclic deformation curves obtained in a study on a normalized plain carbon steel containing 0.45% carbon (lower yield stress of about 440MPa). The plastic strain amplitude is plotted versus the number of cycles for a series of stress-controlled tests performed at various stress amplitudes.

In all cases shown in Fig. 4 an initial quasi-elastic incubation period is followed by a sudden and strong cyclic softening. Then a maximum is reached and finally cyclic hardening occurs till failure. The course of the curves can be described quantitatively on the basis of the microstructural processes in reasonable agreement with the experimental observations (see Klesnil and Lukáš 1992). Qualitatively, the shape of the curves can be explained as follows. During the first cycles, a small number of dislocations are unpinned from their surrounding clouds of interstitial atoms which have mainly formed during normalization. This process is repeated during the following cycles and leads to an increase of the number of mobile dislocation. Hence, the material is in an inhomogeneous condition, because besides the plastically deforming grains others behave still essentially elastic. Plastic deformation is concentrated in Lüders-bands which spread by triggering dislocation motion in their vicinity. The steady increase of the volume fraction of regions containing gliding dislocations causes the strong increase in plastic strain amplitude. Dislocation multiplication in these regions, however, leads to local cyclic hardening. Finally, this effect is no longer overcompensated by the formation of new plastically deforming regions and a macroscopic cyclic hardening prevails. The maximum in the cyclic deformation curve corresponds in a reasonable approximation to the
condition in which for the first time during cycling the entire gage length undergoes plastic deformation.

Low-alloy steels in the normalized condition exhibit a behavior very similar to that described above for a plain carbon steel. However, a martensitic microstructure usually shows no plastic strain at all during stress-controlled cyclic loading (e.g. Eifler 1985). Crack formation and failure occurs in a macroscopically brittle state. Quenched-and-tempered martensitic steels are prone to cyclic softening that manifests itself in a continuous increase of the plastic strain amplitude after a quasi-elastic incubation period. The cyclic softening that prevails till failure is extremely inhomogeneous and connected to the formation of fatigue zones to which plastic deformation is confined (Eifler and Macherauch 1990).

Since steels often do not show a pronounced state of cyclic saturation, the cyclic stress-strain curve (css curve) is often constructed from the stress and strain amplitude values observed at half life N/2. The resulting representations allow for direct comparison with monotonic tension curves in order to quickly determine whether a material undergoes cyclic hardening or softening. In Fig. 5 cyclic and monotonic stress-strain curves are compared for numerous representative steels (Landgraf 1996). Css curves are plotted as solid lines and corresponding monotonic curves are depicted as dashed lines. Ferritic-pearlitic steels, such as the low-carbon steel SAE 1010 and the microalloyed steel SAE 980X show cyclic softening at low strain amplitudes as a consequence of the dynamic Lüders-band propagation taking place below the yield point. At higher strains cyclic work hardening prevails upon the corresponding effect under monotonic conditions. All quenched-and-tempered steels shown in Fig. 5 (SAE 5160, SAE 52100, and SAE 5132) exhibit the significant cyclic softening that was already mentioned above. The 18% Ni maraging steel is primarily strengthened because of the formation of very fine precipitates in the relatively soft martensite during an suitable aging treatment. As seen in Fig. 5, also this material shows a strong tendency to cyclic softening. The only steel that shows cyclic hardening in the entire strain (amplitude) range represented is the ausformed H-11 tool steel. The
high stress level of this steel is obtained by severe hot working in the austenitic condition prior to quenching. The extremely fine microstructure which arises from this treatment and which can be maintained during tempering leads to very high monotonic stress levels. Since cyclic loading even increases this monotonic strength, a very extraordinarily high fatigue strength results. 

A comparison of the monotonic and the cyclic stress-strain curves for austenitic stainless steels documents in many cases a strong trend to cyclic hardening. This is due to the transformation of austenite to martensite in so-called metastable austenitic steels where this martensite formation may take place during cyclic loading. Usually a spontaneous martensitic transformation takes place at sufficiently low temperatures (below the martensite start temperature, $M_S$). This temperature limit can be raised by means of elastic stress or plastic deformation, which provide an additional driving force. Figure 6 illustrates the effect of cyclic-deformation-induced martensite formation on the mechanical hysteresis loop plotted as stress versus plastic strain (Maier et al. 1993, Christ 1996). The first 30 cycles of a test are shown that was carried out under symmetric push-pull conditions in plastic strain control at a plastic strain amplitude of $1.26 \times 10^{-2}$. Test temperature was 103 K in order to promote martensite formation which, however, can also be obtained at room temperature in the material studied. Obviously, the stress amplitude increases strongly, while the hardening rate decreases. Maximum stress amplitude is reached after about 40 cycles.

The transformation of austenite into martensite starts, if a threshold value of the plastic-strain amplitude is exceeded and a certain “incubation” cumulative plastic strain is reached that decreases with increasing amplitude (Mughrabi and Christ 1997). The increase in the stress amplitude shown in Fig. 6 can be attributed directly to the increase of the volume fraction of martensite. The negative mean stress $\sigma_m$ that develops during cycling results from the corresponding volume expansion that is due to the larger specific volume of martensite as compared to austenite. At low stress amplitudes the strain-induced martensitic transformation
may be restricted to the tip of a propagating fatigue crack. The dilation of the region undergoing transformation causes transformation-induced crack closure and generally reduces crack growth rate (Suresh 1998).

It is interesting to note that fatigue-induced martensitic transformation can be exploited in order to enhance the rather modest monotonic and cyclic strengths of these materials. While monotonic straining leads to a permanent shape change, the shape can be kept unchanged when cyclic deformation in strain control is applied. Moreover, an optimized low-temperature cyclic predeformation was shown to improve room-temperature fatigue life by a factor of almost 2 (Maier et al. 1993).

3. Cyclic Lifetime of Steels

3.1 Phenomenological Approach

From an engineering point of view the most important information necessary for a reliable design of a component, that is subjected to cyclic load, is cyclic lifetime, expressed as number of cycles to failure $N_f$. Fatigue data handbooks (e.g. Boller and Seeger 1987, Bäumer and Seeger 1990, Buch 1998) are available for the practitioner and represent this data mostly either in the form of $S$-$N$-curves (stress amplitude versus log $N_f$) or in a double-logarithmic plot of the total strain amplitude versus $N_f$. In order to represent data obtained in tests with superimposed mean stresses, sometimes damage parameters that contain a combination of stress amplitude, strain amplitude and mean stress are used instead of solely stress amplitude or strain amplitude. The conversion of one type of fatigue life diagram into another is problematic since, as has been shown above, a pronounced cyclic saturation state, that defines a unique css-curve, cannot be assumed for most steel.

The basic idea of the strain-life plot is that total strain can be divided into a plastic and an elastic strain component and that simple power law equations apply connecting plastic strain amplitude as well as total strain amplitude to cyclic lifetime. In Fig. 7 fatigue life is plotted for two steels
and these respective power laws are represented as straight dashed lines (Landgraf 1996). The lines intersect at a value of $N_f$ which is commonly termed transient fatigue life $N_t$. The significance of $N_t$ results from the concept that it separates two regimes wherein particular mechanisms and behavioral patterns are operative. On the left of $N_t$, plastic deformation prevails and fatigue resistance is governed by the ductility of the material. Cracks form early in life, and (long) crack growth rate is considered to determine fatigue life. Conversely, on the right of $N_t$, elastic strain is dominant. Hence, crack initiation (or more precisely: the overcoming of microstructural barriers by short cracks) becomes increasingly important. As seen in Fig. 7a, the relatively soft low-carbon steel SAE 1010 is in its ductility-governed regime for most of the range of fatigue life. In the case of the 18% maraging steel (Fig. 7b), $N_t$ is essentially smaller and the regime of stress-governed behavior is larger. Moreover, a comparison of Fig. 7a and Fig. 7b reveals that it is not the absolute value of $N_f$ that determines which regime is relevant for a material. Rather, the relative level of plastic and elastic strain determines the behavior.

As a consequence of the dominant crack initiation stage in the stress-controlled regime, material defects, such as inclusions, as well as geometric defects, such as notches, play a very significant role in this regime and hence gain importance with increasing strength. In order to improve the fatigue resistance, cleanliness is an important issue. Particularly, a low sulfur level is necessary to obtain fewer nonmetallic inclusions, and calcium treatment further gives inclusion shape control.

The strain-life plots shown in Fig. 7 do not show a plateau at high value of $N_f$ which is characteristic for many steels and is more appreciable in the S-N representation. Then the fatigue limit appears as a stress amplitude below which the steel can endure an infinite number of cycles. In general, steels with a tensile strength below of about 1100 MPa exhibit a fatigue limit (Wilson 1996). The value of the fatigue limit depends to a large extent on the tensile strength. As a rule of thumb, the fatigue limit is about half the tensile strength. However, typical tensile strengths do not necessarily correlate with actual fatigue limits and there is large scatter.
particularly with the presence of inclusions. Duplex stainless steels are known to exhibit a relatively high ratio of cyclic to monotonic strength. For these steels the fatigue limit in air is almost equal to the yield strength. Nevertheless, the absolute values are not high because of the moderate tensile strength of duplex steels.

It should be noted that similar to the role of inclusions and notches, environmental effects are of particular significance in the stress-governed regime. For instance, the fatigue strength of a high-strength steel at $10^6$ cycles can be reduced in salt water to less than 10% of that in dry air and a fatigue limit does no longer exist. The role of the environment in this context is a corrosive attack of the smooth surface, creating local stress raisers that initiate fatigue cracks.

Figure 8 represents the transition fatigue life $N_t$ as a function of hardness for a variety of different steels. It is obvious, that $N_t$ correlates with the hardness in the sense that with increasing hardness $N_t$ decreases and therefore the fatigue behavior becomes more stress-governed and less plastic-strain-governed. Hardened steels show shorter transition fatigue lifes as compared to ferritic-pearlic steels. Hence, hardening gives rise to an increased sensitivity to notches and internal defects. On the other hand, the lack of a considerable cyclic plastic strain enables the efficient application of a surface-processing treatment, such as shot peening, in order to significantly improve fatigue resistance. The near-surface compressive residual stresses established are usually sufficiently stable, if the material is subjected to purely elastic strain.

3.2 Damage Tolerant Approach

Besides the use of fatigue life data for design purposes, the existence of irregularities and cracklike imperfections in steel components often demands the application of the concepts of fracture mechanics in order to safely prevent fatigue failure. The basic idea, which is more closely described in the articles of this Encyclopedia on fatigue cracks, is that cyclic life is equal to the number of cycles which grow a crack from an initial length to a final length corresponding to failure.
Figure 9 shows the general behavior of fatigue crack propagation in the commonly used double-logarithmic plot of the crack growth rate da/dN as a function of the stress-intensity factor range ∆K=K_{max}-K_{min}. Typically three regions, which extend from an almost non-propagating crack to fast fracture, describe the growth behavior of long cracks. By means of an integration of such curves, an assessment of the expected fatigue life of components with given flaw sizes is possible, or the maximum tolerable initial crack size that allows a part to reach its design life can be estimated.

Region 1 (Fig. 9) corresponds to the stress-intensity range in which crack growth rate is small and below which cracks are considered to propagate negligibly slow. The threshold value ∆K_{th} can be as low as one-twentieth of the fracture toughness that characterizes the resistance to crack propagation under monotonic loading. Typical values of ∆K_{th} for steels range between 3 and 20 MPa\sqrt{m}. Besides mean stress and environment, microstructure is known to have a very strong effect on ∆K_{th}. Often order-of-magnitude differences in growth rates are observed in the near-threshold region that are caused by variations in pretreatment such as tempering (Richie 1979).

Since the fatigue crack growth threshold is an important design parameter for such applications which involve high-frequency, low-stress cyclic loading (e.g. rotation shafts), microstructural control of steels is of significance, and a substantial improvement in fatigue crack growth resistance near the threshold can be obtained by microstructure optimization. In low-strength steels, as an example, the value of ∆K_{th} increases with increases grain size. This effect, which is attributed to a microstructurally sensitive crack path, can be applied to drastically expand the ∆K range of non-propagating fatigue cracks. However, as already discussed above, it must be taken into account that a large grain size is normally connected with a low yield stress and low resistance to crack initiation.

At intermediate values of ∆K (region 2 in Fig. 9) a straight line is usually found in the logarithmic fatigue crack growth curve. This is described by the well-known Paris law
\[ \frac{da}{dN} = C \Delta K^n \]  

where C and n are constants which are experimentally determined for the steel considered under suitable test conditions. The Paris equation is generally valid within the \( \Delta K \) range of 10 to 60 MPa\( \sqrt{m} \). Extensive fatigue crack growth rate data for steels is reported in the literature and indicate that in this region 2 mean stress, frequency of cyclic loading and wave form do not strongly affect the rate of crack propagation. More important, the effect of microstructure is much less than in region 1. The variations in region 1 due to microstructural modifications may extend to region 2, but this usually gives rise to changes in \( da/dN \) in region 2 of not more than a factor of two or three.

The relative insensitivity of the crack growth behavior in region 2 has led to the development of simple equations which describe the upper bound of scatter bands that result when crack propagation data of various steels of the same class are plotted together (Rolfe and Barsom 1977).

\[ \frac{da}{dN} = 1.35 \cdot 10^{-10} \left( \Delta K [\text{MPa}\sqrt{\text{m}}] \right)^{2.25} \left[ \frac{\text{m}}{\text{cycle}} \right] \]  

[4]

\[ \frac{da}{dN} = 6.9 \cdot 10^{-12} \left( \Delta K [\text{MPa}\sqrt{\text{m}}] \right)^{3.0} \left[ \frac{\text{m}}{\text{cycle}} \right] \]  

[5]

\[ \frac{da}{dN} = 5.6 \cdot 10^{-12} \left( \Delta K [\text{MPa}\sqrt{\text{m}}] \right)^{3.25} \left[ \frac{\text{m}}{\text{cycle}} \right] \]  

[6]

Equation 4 corresponds to martensitic steels, while Eqn. 5 describes ferritic-pearlitic steels and Eqn. 7 relates to austenitic stainless steels.

Region 3 in Fig. 9 defines the transition condition from subcritical crack extension by fatigue to unstable fracture behavior as a consequence of a monotonic overload. Therefore, the fracture properties of the steel considered are of importance in this region meaning that microstructure plays a significant role. As regard to loading conditions, the maximum stress-intensity factor is
the key parameter. Since its value depends not only on $\Delta K$ but also on the stress ratio $R$, a strong influence of mean stress exists.

In Fig. 10 the threshold stress range of 300M ultrahigh-strength steel tempered at temperatures of 373K (curve A), 573K (B), 743K (C), and 923K (D) is plotted as a function of crack length. The curves define the boundary between non-propagating (below) and propagation (above) fatigue cracks. The diagram documents in a schematic way that there is a strong need to distinguish between resistance to short and long fatigue crack growth. High strength obtained by a low tempering temperature is on the one hand beneficial to the fatigue limit and the prevention of short crack growth. However, on the other hand long fatigue crack propagation is promoted. Since low strength (high tempering temperature) acts exactly the other way around, the curves in Fig. 10 intersect. Consequently, an optimization of the microstructure depends on whether a high resistance to crack initiation or to crack propagation from pre-existing flaws is strived for.

4. Dynamic Strain Aging

Numerous applications of steels involve elevated temperature so that the fatigue behavior at intermediate and high temperature is of interest. A description of all the relevant aspects in this context would be far beyond the scope of this article, however the phenomenon of dynamic strain aging will briefly be introduced, since it occurs very pronounced in steels and affects high-temperature fatigue behavior significantly.

Studies on the cyclic softening and hardening behavior of ferritic steels document a very marked temperature dependence and reveal a temperature range around 600K in which an anomalously strong cyclic strength exists. This is the regime of maximum dynamic strain aging in which the diffusion rate of the interstitially dissolved atoms (mainly carbon) is similar to the dislocation velocity leading to a strong hindrance to dislocation motion. The effect, which has been studied very thoroughly on plain-carbon steels (Eifler and Macherauch 1988), leads to very small plastic
strain amplitudes in stress-controlled tests. Hence, mostly an increase in fatigue life is observed in the temperature regime of dynamic strain aging. In a plastic-strain-controlled test, however, enhanced cyclic hardening manifests itself in a strong increase of the stress amplitude. This, in turn, can lead to a reduction in fatigue life.

Figure 11a depicts the plastic strain amplitudes at half life of stress-controlled tests and the stress amplitudes of plastic-strain-controlled tests observed on the normalized steel SAE 1045 over a range of temperatures from room temperature to 400°C (673K). The plastic strain amplitude and the stress amplitude exhibit a minimum and a maximum, respectively, at around 300°C (573K), i.e. in the range of maximum dynamic strain aging. The resulting consequence for the cyclic lifetime is shown in Fig. 11b. A minimum of the plastic strain amplitude at constant stress amplitude leads to a maximum of fatigue life. A maximum stress amplitude at constant plastic strain amplitude coincides with a marked valley in the course of the number of cycles to fracture versus temperature. This illustrates that the effect of dynamic strain aging depends sensitively on the mode of testing. Furthermore, dynamic strain aging is also affected by the cyclic strain rate (frequency), since it is a thermally activated process.

In the case of austenitic stainless steels, dynamic strain aging is believed to be caused by the dynamic interaction of dislocations with carbon-carbon pairs, carbon-vacancy pairs, clusters of carbon atoms, and with mobile substitutional atoms such as chromium. As a consequence, the effect is most pronounced at higher temperatures as compared to ferritic steels. In the temperature range between about 250°C (523K) and 550°C (723K) the stress amplitude at constant plastic strain amplitude first increases with increasing temperature and subsequently decreases again. Further indications of dynamic strain aging are serrated yielding and a negative strain-rate sensitivity. In addition, a marked change in cyclic slip mode occurs which leads to a clearly appreciable change in the dislocation arrangement. Figure 12 shows three characteristic transmission electron micrographs which have been taken after cyclic loading at 250°C, 400°C and 650°C. Below and above the temperature regime in which dynamic strain aging processes
are noticeable, cell structures are formed, while within this temperature regime the impeded dislocation mobility leads to a disorderly dislocation arrangement resembling that of planar-slip materials.

5. Closing Remarks

The outstanding significance of steels as engineering materials in mechanical engineering has triggered an exceptionally large amount of investigations on the fatigue behavior of steels in order to produce knowledge and data useful as basis for fatigue failure prevention. This article provides some insight into the variety and complexity of cyclic deformation, fatigue crack growth and fatigue failure behavior of steels and tries to expound the underlying microstructural mechanisms. The examples illustrate characteristic behavioral patterns, underline the differences in the behavior of different steels and document the sometimes striking dependence on loading conditions. Despite the continuous interest in the fatigue behavior of steels for more than one and a half century, a detailed understanding of the role of microstructure is just developed to some extent. This situation, which might be a consequence of the broad range of microstructures existing in steels, demands further thorough work directed to fatigue of steels with close relation to microstructure.
Abdel-Raouf H, Plumtree A 1971 On the steady state cyclic deformation of iron. *Metal. Trans.* 2, 1251-1254


Eifler D 1985 Zusammenhang zwischen Mikrostruktur und Schwingfestigkeit bei Stählen (Relation between microstructure and fatigue strength of steels). In: Munz D (ed.) *Ermüdungsverhalten metallischer Werkstoffe* (*Fatigue Behavior of Metallic Materials*). DGM Informationsgesellschaft, Oberursel, Germany, pp.73-105


Eifler D, Macherauch E 1990 Microstructure and cyclic deformation behaviour of plain carbon and low-alloyed steels. *Int. J. Fatigue* 12, 165-174


Mughrabi H, Christ H-J 1997 Cyclic deformation and fatigue of selected ferritic and austenitic steels: Specific aspects. *ISIJ International* 37, 1154-1169
Pohl K, Mayr P, Macherauch E 1981 Cyclic deformation behavior of a low carbon steel in the temperature range between room temperature and 800K. *Int. J. Fract.* **17**, 221-233


H.-J. Christ

[University of Siegen, Siegen, Germany]
Figure Captions

Figure 1
Schematic representation of the dependence of flow stress on temperature for the bcc crystal structure (following Seeger 1954)

Figure 2
Unusual course of the cyclic stress-strain curve of decarburized α-iron at low temperature and small plastic strain amplitudes (from Sommer et al. 1998 pp. 1527-1536)

Figure 3
Manson-Coffin representation of fatigue life of decarburized α-iron under plastic-strain control (from Sommer et al. 1998 pp. 1573-1544)

Figure 4
Cyclic deformation curves of the normalized ferritic steel SAE 1045 at room temperature for different stress amplitudes (with courtesy of Eifler and Macherauch 1988)

Figure 5
Comparison of monotonic and cyclic stress-strain curves of representative ferrous alloys (reprinted with permission from Landgraf 1996 © ASM International)

Figure 6
Stress-strain path of the first 30 cycles at 103K of a metastable austenitic stainless steel under plastic-strain control (reprinted with permission from Maier 1993 © Carl Hanser Verlag)

Figure 7
Strain-life curves for two representative steels which differ in strength (following Landgraf 1996)

Figure 8
Transition fatigue life as a function of hardness for various ferrous alloys (reprinted with permission from Landgraf 1996 © ASM International)
**Figure 9**
Fatigue crack growth behavior of a ferritic steel with a yield strength of 470 MPa at room temperature (reprinted with permission from ASM Handbook Vol. 19 © ASM International)

**Figure 10**
Threshold stress range at R=0 as a function of crack size. The predicted lines are based on data for a 300M ultra-high strength steel that was tempered at temperatures of 100°C (A), 300°C (B), 470°C (C) and 650°C (D) to adjust different tensile strengths (reprinted with permission from Ritchie 1979 © ASM International)

**Figure 11**
Cyclic deformation behavior (a) and corresponding cyclic life (b) of the normalized plain-carbon steel SAE 1045 under plastic-strain control and stress control, respectively (from Weisse et al. 1993)

**Figure 12**
TEM micrographs illustrating the dependence of dislocation arrangement of a cyclically deformed austenitic stainless steel on temperature (from Zauter et al. 1992)