Investigation of the damage mechanisms during very high cycle fatigue (VHCF) of a tempered carbon steel

A. Giertler*, U. Krupp
University of Applied Sciences Osnabrück, Faculty of Engineering and Computer Science, Albrechtstr. 30, 49076 Osnabrück, Germany

Abstract

The study is an overview of recent investigations dealing with the fatigue damage in the HCF and VHCF regime of 0.5C-1Cr martensitic steel (German designation: 1.7228) in a tempered condition. The experimental part is focussed on the microstructural characterization regarding crystallographic orientation, reconstruction and grain size distribution of prior austenite grains as well as metallographic investigations. Electrolytically polished cylindrical bulk specimens have been loaded under fully reversed (R=-1) loading condition using an ultrasonic fatigue testing machine (f=20000Hz) and an electro-mechanical resonance fatigue testing machine (f=95Hz). By means of a light microscope attached to the testing systems, fatigue crack initiation and propagation were correlated to the material microstructure and the specimens fatigue life. The application of a fast and high resolution thermography camera system in combination with the resonance testing system allows to use the heat dissipation determined on the surface of the fatigue specimen as a measure of specific fatigue damage. The tested specimens have been carefully investigated by means of scanning electron microscopy (SEM) in combination with automated electron back-scatter diffraction (EBSD) and energy dispersive X-ray (EDX) analysis. It was found that under HCF and VHCF loading conditions crack initiation is caused by slip band formation between the martensitic laths structure, which leads to microcrack initiation and propagation. The propagation of microcracks is sensitive to changing crystallographic orientations when crossing a grain boundary. In case of run-out specimens (10^9 cycles) microcracks of the length of several microns have been found that were blocked by prior austenite grain boundaries which act in this case as effective barriers against microcrack propagation.

* Corresponding author. Tel.: +49-541-969-3215; fax: +49-541-969-2958.
E-mail address: a.giertler@hs-osnabrueck.de

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1. Introduction

The fatigue life of a component is generally associated to the accumulated amount of cyclic plastic deformation. In the high cycle fatigue (HCF) and very high cycle fatigue (VHCF) regime cyclic plastic deformation is concentrated at local features within the microstructure, like grains with a favorable crystallographic orientation, non-metallic inclusions which act like stress raisers or zones with lower hardness [Kunio et al (1979), Kunio et al. (1981)] These features are leading to the development and growth of slip bands, crack initiation and eventually, fatigue crack propagation. However, microstructural features may also act as effective barriers against slip band transmission and fatigue crack propagation [Zhai et al. (2005)]. The investigated carbon steel is used for parts in the injection systems in modern Diesel engines which imply cyclic loading above $10^7$ number of cycles. Therefore, the knowledge about to fatigue crack propagation [Zhai et al. (2005)]. The investigated carbon steel is used for parts in the injection systems in modern Diesel engines which imply cyclic loading above $10^7$ number of cycles. Therefore, the knowledge about to fatigue behavior in the VHCF regimes makes the classical Wöhler approach inapplicable. Thus, a more sophisticated approach is needed to explain macroscopic failure by understanding the microscopic fatigue behavior. In-situ monitoring by high resolution thermographic allows the detection of fatigue damage in an early stage of VHCF life [Wagner et al. (2009)] on a microscopic scale. Even on this microscopic scale, most of the energy during plastic deformation is transferred into heat and only a small amount stored in cold work [Taylor and Quinney (1934)]. Therefore, heat dissipation is a favorable value to detect local plastic deformation.

2. Experimental Procedure

The material used is a low-alloy carbon steel 50CrMo4 (German designation: 1.7228). The composition of the steel and the heat treatment parameters are given in Table 1. The conducted heat treatment results in a fully tempered martensitic microstructure with a moderate hardness of 37HRC.

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Cr</th>
<th>Mo</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>50CrMo4</td>
<td>0.48</td>
<td>1.00</td>
<td>0.18</td>
<td>0.71</td>
<td>0.013</td>
<td>0.010</td>
<td>bal.</td>
</tr>
<tr>
<td>austenitizing: 850°C (0.5h) oil quench; temper heat treatment: 550°C (1h) air-cool</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

The crystallographic orientations, grain and phase distribution have been characterized by means of EBSD measurements. Figure 1a shows an inverse pole figure (IPF) mapping. The rolling direction is perpendicular to the image plane. One can easily distinguish between the individual martensitic needles. Figure 1b shows the result of an automated parent grain reconstruction applied to the EBSD measurement of Figure 1a [Cayron (2007)] using the software package ARPGE.

![Figure 1: (a) Inverse pole figure mapping of the 50CrMo4 steel in transversal direction; (b) Automated parent grain reconstruction](image-url)
The reconstruction given in Fig 1b is based on the orientation relationship of Kurdjumov-Sachs \{111\}_A||\{011\}_M \langle 110 \rangle_A \langle 111 \rangle_M and reveals the parent austenite grains. The martensitic microstructure is build up by a hierarchical setup [Kitahara et al. (2006)]: Blocks represent sets of martensitic laths with the same crystallographic orientation (variant). A packet includes several blocks, typically the blocks within one packet have the same \{111\}_γ plane in austenite. Several packets form one prior austenite grain. It is expected that on the boundary of these individual orientated areas compatibility stresses arise due to local elastic anisotropy.

Tensile testing on cylindrical specimens with a gauge length diameter of 8mm reveals a yield strength of \( \sigma_y = 1000 \text{MPa} \) and a tensile strength of \( \sigma_t = 1095 \text{MPa} \) for this material. To describe the cyclic stress strain behaviour incremental step test have been conducted and a cyclic 0.01% yield stress of \( \sigma_{cy} = 400 \text{MPa} \) have been obtained. The fatigue tests were performed under fully reversed loading \( R = -1 \) on a RUMUL 100kN resonance testing machine with a test frequency of \( f = 95 \text{Hz} \) and an ultrasonic testing machine from BOKU Vienna with a test frequency of \( f = 20,000 \text{Hz} \). The technical drawings of the used specimens are given in Fig. 2 a and b. Both specimens are mechanically and electrolytically polished to remove any surface roughness. Additionally, a small shallow notch with a notch factor of 1.2 was fabricated in the middle section of the specimens. Fatigue tests with constant amplitude and load increase tests with a stepwise increase of 6MPa every 100000 cycles were performed. During the test the surface of the shallow notch area is observed with the aid of a light microscope and a thermography camera. The thermography camera ImageIR 8380 hp is equipped with two microscopic lenses that achieve a pixel size of 5µm in a field of view of 3200µm to 2600µm and 2µm pixel size with a field of view of 1200µm to 960µm, respectively. The fatigued specimens are investigated by means of analytical electron scanning microscopy (SEM) in combination with electron backscattered diffraction (EBSD) and focused ion beam (FIB).

3. Results

The fatigue life S/N diagram for the investigated material 50CrMo4 is given in Fig. 2c. Both test series were performed under fully reversed constant amplitude loading \( (R = -1) \) in laboratory air. During all tests fatigue cracks were shown to initiate at the surface of the specimens and no internal crack initiation was observed. The fatigue limit for the tests with 20kHz was \( \sigma_{FL} \approx 680 \text{MPa} \) and for the tests with 95Hz at \( \sigma_{FL} \approx 490 \text{MPa} \), respectively. The testing conditions for both series were kept constant, i.e., identical surface preparation, notch factor, specimen temperature and uniaxial tension-compression loading were chosen.

![Figure 2: geometry of the fatigue specimen for the (a) resonance fatigue testing and (b) for the ultrasonic fatigue testing equipment; (c) S-N diagram showing the different behaviour for 20kHz and 95Hz testing frequency.](image)

It is assumed that the pronounced difference in fatigue strength of \( \Delta \sigma \approx 200 \text{MPa} \) can be attributed to a frequency influence caused by a kind of strain rate effect. Since an increase in strain rate is equivalent to a decrease in temperature, it is assumed that at ultrasonic frequency an increase of the Peierls strength causes a high fatigue strength.
The observed behavior is typical of smooth specimens made of highly tempered materials. It is reported that, if the annealing temperature is reduced, the frequency effect on smooth specimens decreases because the athermal part of the critical shear strength exceed the Peierls stress contribution [Takeuchi et al. (2008)]. To prove this, it is planned to increase the strength of the material by reducing the annealing temperature and afterwards the same experiments will be repeated. The given results show the importance of microstructure, local plastic deformation and testing conditions on the fatigue behavior.

To understand the relationship between local plastic deformation and the macroscopic fatigue behavior, load increase tests have been carried out in the resonance fatigue testing machine, cf. Fig. 3 a. The load increase test starts with a stress amplitude of $\sigma_a = 360$MPa. This value is far below the yield strength of the material. The stress amplitude is stepwise increased during the test by 6MPa every 100,000 cycles until fracture of the specimen. The surface of the fatigue specimens has been observed by a thermographic camera and every thousand cycles a thermogram has been stored. Additionally, the change in resonance frequency as an indicating value for fatigue damage is plotted, cf. Fig. 3a.

The first significant change in resonance frequency can be observed after $N=2.4 \cdot 10^6$ cycles at a stress amplitude of 500MPa. The test was stopped until a change of frequency of $\Delta f = 0.1$Hz has been reached, which correlates with a surface crack of more than 3mm in length. In comparison to that, the temperature measured by the thermographic camera provides a more sensitive signal out of the microstructure. The first response of the temperature up to $N=100,000$ cycles can be attributed to a heating up process of the fatigue setup with the beginning of the test. Followed by a state of equilibrium, the continuous increase of stress amplitude leads to a negative change in specimen temperature. This behavior is due to the thermoelastic effect [Yang et al. (2004)], where a tension load applied to a metallic specimen causes a negative change in temperature. A continuous increase of specimen temperature can be observed from $N=700,000$ cycles by local plastic deformation. The corresponding stress amplitude at $N=700,000$ cycles was $\sigma_a = 396$MPa. This value corresponds well with the cyclic yield strength of $\sigma_{oy} = 400$MPa obtained from the incremental step test. The strong increase in temperature beginning at $2.4 \cdot 10^6$ cycles is caused by fatigue crack propagation.

A graphical approach to determine the fatigue limit is given in Figure 3b, by plotting the change in temperature vs. the corresponding stress amplitude. The range between 350MPa and 500MPa behaves almost linearly until above 500MPa the temperature increases exponentially. The linear temperature rise in the first region between 350MPa and 500MPa is dominated by local plastic deformation and the second range above 500MPa by fatigue crack propagation. Using a linear fit through each region gives an intersection of the two lines at a stress amplitude of $\sigma_a = 483$MPa. This value is in very good agreement with the fatigue strength of $\sigma_{FL} = 490$MPa, which has been determined by tests under constant amplitude loading, cf. Fig 2a.

An example of the resolution of the thermographic camera is given in Figure 4a. The label “C1“ corresponds to the crack initiation site. The respective temperature development vs. number of cycles at position C1 forms the data base for the diagram shown above in Fig. 3 a and b. In comparison, “C2“ marks an area in which no significant temperature increase was observed.
increase was detected during the test. Obviously, within the microstructure areas are located, where plastic deformation is concentrated (marked as “C1”), as well as areas in which there is no plastic deformation (marked as “C2”). The difference between the regions in terms of their temperature development is given in Figure 4b. The area “C2” shows a delayed reaction to the continuous increase in stress amplitude. The temperature increase can be attributed to the convection of heat from the areas with plastic deformation (“C1”) in the adjacent areas (“C2”). The thermogram in Figure 4a also shows a fatigue crack, whose origin is marked with “C1” in the center of a region which exhibits strong plastic deformation. During the load increase test, thermograms of the surface were saved periodically every 1000 cycles. An analysis of the thermograms regarding crack growth showed that only after 95% of the total lifetime fatigue crack growth sets in.

![Figure 4: (a) Thermogram at \( N = 3 \cdot 10^6 \) cycles. Point “C1” represents an area with strong local plastic deformation, “C2” marks an area with no plastic deformation; (b) Temperature profile for the points “C1” and “C2”.](image)

The ribbon-like structure that is developed, as a consequence of locally different plastic deformation as shown in Figure 4a is caused by microstructural banding. Microstructural bands have originated during solidification in continuous casting. Reason for this is the slow cooling in the core region of the block. This results in a dendritic solidification of the melt and the segregation of alloying elements cannot be prevented [Grange (1971)]. To verify this, EDX measurements have been performed on the respective samples in the area of crack initiation, to determine the distribution of alloying elements. It was found that the difference in the chromium content in the areas with local plastic deformation (“C1”) in comparison with areas without plastic deformation (“C2”) was at 0.2wt. % chromium.

![Figure 5: (a) Fatigue specimen surface at \( N = 2 \cdot 10^8 \) cycles loaded with a constant amplitude of \( \sigma_c = 490 \text{MPa} \) showing local plastic deformation; (b) Detail out of (a) showing a crack nucleus within the microstructure.](image)
[Giertler and Krupp (in press)]. The difference is particularly evident when looking at the respective zones by high resolution SEM, cf. Fig. 5 a.

Even at stress amplitudes in the range of the fatigue limit (≈\(N=10^8\) cycles), areas have formed with a strong development of local plastic deformation within the microstructure. On closer inspection of the surface of the specimen these areas have been identified as a crack nucleus, cf. Fig 5b.

4. Conclusion

For the investigated steel 50CrMo4 two values of the fatigue limits were identified for two different test frequencies of 20kHz and 95Hz. The difference can be attributed to a strong influence of the test frequency on the fatigue limit for the material in the tempered condition. Particular attention was paid in the development of local plastic deformation during the investigation of the fatigue mechanisms in the VHCF regime. Therefore, a test setup was developed, in which the damage evolution can be followed in-situ by high-resolution thermography. In a series of load increase tests a direct correlation between the stress amplitude and microscopic damage evolution was established. The first development of local plastic deformation was observed at a stress amplitude of \(\sigma_a=396\)MPa. This value corresponds well with the cyclic yield strength of \(\sigma_{cy}=400\)MPa determined the incremental step test. The presented test setup seems to be a promising alternative, to represent the damage processes within the microstructure and to allow simultaneous visual inspection. The direct comparison of the stress amplitude with the sample temperature allows to graphically determine the fatigue limit. The estimated value of \(\sigma_{FL}=483\)MPa corresponds with the in the experiments under constant amplitude loading determined value of \(\sigma_{FL}=490\)MPa. The combination of electromechanical testing machine and high-resolution thermal imaging enables the early detection of local plastic deformation during fatigue loading. In addition, surface analysis have shown by means of scanning electron microscopy have shown, that even after \(10^8\) cycles crack nucleation occurs. For a safe fatigue-resistant component design in the VHCF, regime it may be practicable to restrict the load during operation below the stress value of \(\sigma_a=396\)MPa as an initiation value for local plastic deformation during the load increase test.

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References


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