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The significance of microstructural barriers during fatigue of a duplex steel in the high- and very-high-cycle-fatigue (HCF/VHCF) regime

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ABSTRACT

Austenitic–ferritic duplex steels are used for structural applications when high strength in combination with excellent corrosion resistance is required. Many of these applications imply cyclic loading and hence, fatigue damage needs to be considered for dimensioning. Most dimensioning strategies make use of the fatigue-limit concept. However, while the original fatigue-limit concept is based on the idea that existing slip bands or microcracks are blocked by microstructural barriers like grain or phase boundaries, more recent research work has shown that metallic structures may fail far below the conventional fatigue limit even at very high numbers of cycles to fracture. The present paper deals with the observation of local plasticity and fatigue damage in the VHCF regime by means of high-frequency fatigue testing in combination with scanning electron microscopy (SEM) and with electron back-scattered diffraction (EBSD). The results reveal that fatigue damage in the VHCF regime indeed causes the formation of slip bands followed by initiation and propagation of microstructurally short cracks in a very localized manner, manifesting itself by heat generation. The SEM observations and measurements of slip band geometries were correlated with calculations using a finite-element and a numerical short-crack model, which take the real two-phase microstructure and its elastic/plastic anisotropy into account and allow the prediction of both: (i) the fatigue crack initiation sites and (ii) microcrack propagation rates.

1. Introduction

Many structural-integrity concepts for dynamically high-loaded components imply mathematical approaches to the strain-based Wöhler diagram and the fatigue-limit concept, which is based on the assumption that below a certain stress value no cycle-dependent damage occurs. In the case of bcc carbon steels it is widely accepted that the fatigue limit corresponds with a number of cycles between 10^6 and 10^7 cycles. In the case of fcc materials, e.g. aluminum or copper alloys, a fatigue limit cannot be found, since their lower packing density causes a substantially lower critical shear stress to activate dislocation motion and hence, fatigue damage. However, in most practical cases the conventional fatigue-limit concept is applied for nearly all engineering metallic materials that experience more than 10^7 cycles during service.

More recently, it has been found that fatigue failure may occur at stresses well below the conventional fatigue limit at numbers of cycles beyond 10^7. The corresponding fatigue regime, which exhibits an increasing technical relevance, has been termed ultra-high-cycle-fatigue (UHCF) or very-high-cycle-fatigue (VHCF), respectively. Several mechanisms have been suggested to govern fatigue failure during VHCF (cf. [1]): Murakami and Endo [2] correlate the fatigue limit with the size and hardness of non-metallic inclusions. At very high numbers of cycles and correspondingly low stress amplitudes the significance of the plane-stress condition at the surface vanishes and cracks initiate in the interior of the material at inclusions under inert vacuum conditions. Frequently, an optically dark area (ODA) can be found on the fracture surface, forming together with the inclusion a so-called fish-eye, which is often correlated with hydrogen embrittlement during fatigue crack propagation [3]. Within the ODA a microscopically rough granular area appears around the inclusion (fine-granular area (FGA) or granular bright facet (GBF)) which can be attributed to the initiation and coalescence of many microcracks [4]. According to Murakami and Endo [2], the fatigue limit σ_{FL} can be calculated as a function of the cross section area of the inclusion A, the location of the inclusion, denoted by a constant C (C = 1.43, 1.56, 1.41 for an inclusion at the surface, underneath the surface or within the bulk, respectively), the Vickers hardness HV, a hardness-dependent exponent α ($\alpha = 0.226 + HV{\cdot}10^{-4}$) and the stress ratio R:

$$\sigma_{FL} = \frac{C(HV+120)}{\sqrt{(A)}^{1/6}}\left(\frac{1-R}{2}\right)^{\alpha}. \quad (1)$$

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In the absence of inclusions of a critical size, fatigue damage occurs by the formation of persistent slip bands (PSBs) as suggested by Mughrabi [5] (type I materials contrary to type II materials with crack initiation at internal inclusions). The conventional fatigue limit corresponds to the PSB limit. However, even below the PSB limit slip irreversibility occurs causing the nucleation of slip bands at phase or grain boundaries. This effect is due to the anisotropy of elastic deformation as it is shown in the present paper. According to Tanaka and Mura [6], slip irreversibility eventually causes initiation and propagation of microstructurally short fatigue cracks (cf. [7]). This kind of slip irreversibility can be attributed to vacancy-type annihilation of dislocations moving back and forth on neighbouring slip planes during cyclic loading (cf. [8]).

If the applied stress amplitude is not sufficient to make one of these crack nuclei overcoming the adjacent grain or phase boundary, the conventional fatigue limit is reached. As long as irreversible dislocation motion is not completely eliminated, very localized accumulation of plastic slip remains possible and fatigue failures at very high numbers of cycles ($N_f < 10^7$) can be observed. Thus, there should be an absolute fatigue limit, i.e., the irreversibility limit. This is schematically shown in Fig. 1.

While one can clearly distinguish between the VHCF behaviour of bcc materials and closed-packed fcc materials, the situation is more complex and not fully understood in the case of two-phase materials showing both lattice structures. In the present study, the VHCF behaviour of an austenitic–ferritic duplex steel is discussed with the focus on the mechanisms of fatigue crack initiation. Predominantly, cracks nucleate at slip bands at austenite/ferrite phase boundaries at the surface, only a few specimens showed sub-surface crack initiation at non-metallic inclusions. These observations are only partially in agreement with the work of Chai [9] on several duplex steels. He observed crack initiation exclusively underneath the surface mostly at non-metallic inclusions. However, it should be mentioned that the strength of both the austenite and the ferrite phase in duplex stainless steels can experience a strong variation by dissolved nitrogen in the fcc austenite phase or precipitation of the α phase leading to the so-called 475° embrittlement of the ferrite. Hence, the metallurgy of duplex steels can strongly alter the mechanical response of the two phase structure (cf. [10]).

2. Experimental details

The behaviour of microstructurally short fatigue cracks was studied by means of an austenitic (γ) – ferritic (α) duplex steel X2 CrNiMoN 22 5 3 (1.4462). The chemical composition of the material is given in Table 1.

After a 4 h-homogenization heat treatment at 1250 °C followed by slow-cooling to 1050 °C and quenching in water, the microstructure consists of approximately 50% austenite with a mean grain size of 46 μm and a Vickers hardness of 263 *HV* embedded in 50% ferrite with a mean grain size of 33 μm and a Vickers hardness of 280 *HV*. Fig. 2 shows the microstructure of the duplex steel after heat treatment.

Table 1
Chemical composition of the duplex steel used in this study (wt.%).

Fe	C	Cr	Ni	Mo	Mn	N	P	S
Bal.	0.02	21.9	5.6	3.1	1.8	0.19	0.023	0.002

Push–pull fatigue tests (stress control, stress ratio: $R = -1$ ($R = 0$)) were carried out on a high-frequency servohydraulic testing system (MTS 810, 1000 Hz) and a resonance testing system (Rumul Testtronic) using cylindrical specimens with a diameter of 5 mm and 7 mm, respectively. Prior to testing, the specimens were ground and electro-polished to eliminate any roughness effects on fatigue and to allow the application of the automated EBSD technique (electron back-scattered diffraction) and back-scattered electron channeling contrast imaging (ECCI) in the scanning electron microscope. The EBSD technique is used to analyze the distributions of the austenite and ferrite phase and the crystallographic orientations. In addition to SEM, slip activity in the different phases was studied by means of transmission electron microscopy (TEM).

3. Results

The results of the fatigue tests are summarized in the S/N diagram in Fig. 3. At a stress level between $\Delta\sigma/2 = 380$ MPa and $\Delta\sigma/2 = 385$ MPa three different situations were observed: (i) crack initiation at slip bands at the surface, (ii) crack initiation at a non-metallic inclusion (fish-eye), and (iii) no crack initiation (run-out specimens). Specimens were considered as run-out specimens when a number of cycles of $N = 10^8$ or 5×10^7 was exceeded. This limitation was necessary because of an excessive heat generation when loading the specimens at frequencies above $f = 300$ Hz. It is worth mentioning that even at stress amplitudes ($\Delta\sigma/2 = 350$ MPa, run-out $N > 10^8$ cycles) clearly below the macroscopic yield strength of $R_{p0.2} = 420$ MPa a substantial temperature increase was measured. This observation hints to the fact that plastic slip is not only active throughout the experiment but also by a considerable amount irreversible. In a recent paper, Wagner et al. [11] correlated the evolution of the specimen temperature with plasticity-induced energy-dissipation and crack initiation (strong temperature increase) during VHCF. A clear correlation between

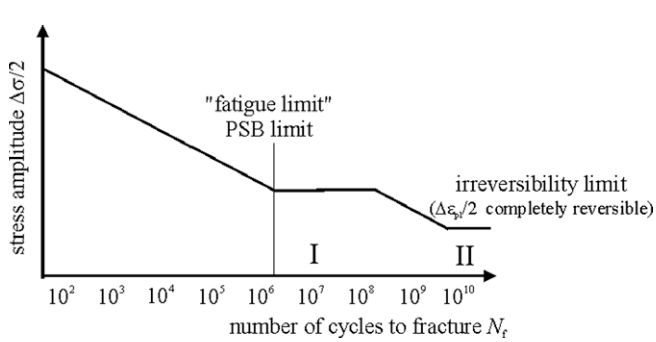

Fig. 1. Two-stage S/N diagram covering both the conventional fatigue limit and the irreversibility limit.

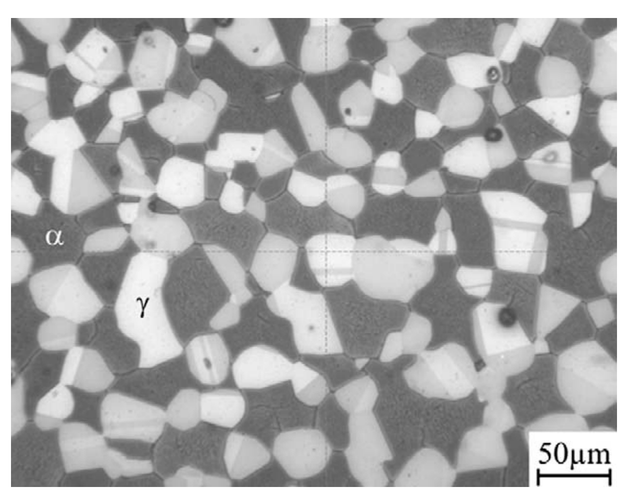

Fig. 2. γ-austenite and α-ferrite microstructure of the duplex steel used in this study.