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The behavior of short fatigue cracks during Very High Cycle (VHCF) Fatigue of duplex stainless steel

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ABSTRACT

During VHCF loading of duplex steels fatigue damage starts within the softer austenite phase. Depending on the spatial mismatch between the slip systems of neighboring grains, intergranular or transgranular fatigue cracks are initiated. If these cracks can propagate or are blocked depends on the interactions with the local microstructure. These interactions have been studied by ultrasonic fatigue testing in combination with electron microscopy and synchrotron diffraction experiments. It was found that the strength of the first microstructural barrier is decisive for VHCF life. The barrier strength as a function of the microstructural arrangement has been implemented in a micromechanical finite element approach that shall be used to predict the VHCF life or the existence of a fatigue limit for given microstructure parameters.

1. Introduction

Austenitic–ferritic duplex stainless steels (DSS) are increasingly used for structural applications that require a high resistance to stress- and fatigue–corrosion cracking in combination with reasonable fatigue strength. Examples of such applications are rotors in biogas plants, crank shafts for motor boats or welded reactors in chemical industry. Most of these applications imply fatigue loading up to very high numbers of cycles, i.e., $N_f > 10^7$. Therefore, fatigue design requires the knowledge of the fatigue limit of the material that was estimated to be in the range between $\sigma_{FL} = 330$ and $\sigma_{FL} = 350$ MPa (according to experimental work of this work and [1], $R = -1$). However, it has been reported in earlier work that even at stress amplitudes below the fatigue limit DSS is prone to fatigue damage manifesting itself by slip bands (extrusions and intrusions) and short cracks [2]. That led to the question wether metallic materials may exhibit infinite fatigue life in spite the occurrence of local plasticity or if the fatigue strength is generally decreasing with increasing number of load cycles.

According to Mughrabi [3], in VHCF fatigue failure may occur far beyond the "classical" fatigue limit experimentally determined for $N_f = 10^7$. He distinguishes between type I VHCF behavior, where accumulated plastic deformation causes a gradually increasing surface roughening and eventually the formation of persistent slip bands (PSB), and type II VHCF behavior, where the crack initiation site changes from the surface to the bulk, e.g., internal crack initiation at non-metallic

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Nomenclature

BE	boundary element
bcc	body-centered cubic
CLSM	confocal laser scanning microscopy
DCT	diffraction contrast tomography
DSS	duplex stainless steel
EBSD	electron back-scatter diffraction
fcc	face-centered cubic
FE	finite element
HCF	high-cycle fatigue
LCF	low-cycle fatigue
PCT	phase contrast tomography
PSB	persistent slip band
SEM	scanning electron microscopy
SFE	stacking fault energy
TEM	transmission electron microscopy
VHCF	very-high-cycle fatigue
a	crack length
C	cyclic irreversibility factor
d	grain size
E	Young's modulus
f	frequency
G	shear modulus
k^{cyc}	cyclic Hall–Petch constant
m	material-dependent exponent
N_f	number of cycles to fracture
N_i	number of cycles to crack initiation
R	stress ratio
$R_{p0.2}$	0.2% offset yield strength
R_m	(ultimate) tensile strength
T	temperature
U	electric voltage
t	time
n	Poisson ratio
σ_Y^{cyc}	cyclic yield strength
σ^{0cyc}	cyclic friction strength
τ^0	cyclic frictional shear stress
$\Delta CTSD$	cyclic crack tip shear displacement
$\Delta\tau$	cyclic shear stress interval
$\Delta\gamma$	cyclic shear strain interval

inclusions or agglomerations of grains with weak grain boundaries. In the case of VHCF of DSS, only very few specimens had shown internal crack initiation at Al_2O_3 inclusions [2]. These specimens failed at approximately $N_f = 10^7$ cycles. In general, fatigue behavior of DSS is determined by the difference in strength and ductility of the two phases ferrite and austenite, respectively. Due to the low critical shear stress, the fcc austenite grains tend to plastic deformation and slip band formation in the very first cycles of fatigue life. Depending on the applied stress amplitude, either all austenite grains are deformed plastically leading to immediate slip transfer to the ferrite grains (high stress amplitudes, LCF, cf. [5,6]), or only few austenite grains show slip bands emanating from grain boundaries where the local stress exceeds the critical shear stress while the remote stress amplitude is macroscopically elastic (low stress amplitudes, HCF and VHCF, cf. [6,7]). The accumulation of plastic slip within the fcc austenitic grains due to cyclic irreversibility (cf. [8,9]) leads to the formation of elastic incompatibility stresses (residual stresses) in the bcc ferrite grains. Once locally the incompatibility stresses overcome the critical shear stress of the ferrite grains, crack initiation can be observed either as (i) a consequence of slip transfer across the phase boundaries, or (ii) intergranular separation of phase ($\alpha\gamma$) and grain ($\alpha\alpha$) boundaries due to slip steps within the boundary [10]. Therefore, during fatigue life the load sharing between the initially softer austenite phase and the harder ferrite phase changes due to strain hardening of the austenite. This has been reported in [11] and was theoretically verified by a crystal plasticity model by Lillbacka et al. [12].

The initial strength of the two phases austenite and ferrite can be altered separately. Heat treatment at 475 °C results in a pronounced spinodal decomposition of the ferrite phase into a Fe-rich α matrix and small Cr-rich α' precipitates [4].