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Propagation mechanisms of microstructurally short cracks—Factors governing the transition from short- to long-crack behavior

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Abstract

Microstructural short fatigue cracks are known to exhibit an abnormal propagation behavior as compared to long cracks, which grow by a rate that can be described by, e.g., the Klesnil–Lukáš relationship. By means of carefully recording crack length versus number of cycles in the high-cycle fatigue regime in combination with a microtexture analysis using automated electron back-scattered diffraction, the parameters determining the scatter in microcrack propagation rates and the transition from short- to long-crack behavior were identified for an austenitic–ferritic duplex steel. The three-dimensional orientation relationship of the slip planes in grains involved in the crack propagation process turned out to be most significant. This relationship determines the barrier effect of grain and phase boundaries as well as the local crack propagation mechanisms, either operating crystallographically by single slip or perpendicularly to the applied load axis operating by double/multiple slip. To predict the propagation behavior of microstructurally short cracks in a mechanism-based way, a numerical model has been developed that accounts for local interactions between the crack tip and the microstructure as the current driving force for crack advance.

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1. Introduction

Particularly in the case of high-cycle-fatigue-loaded smooth components, the major part of the service life is determined by crack initiation and early crack growth. The propagation behavior of such microstructurally short fatigue cracks deviates considerably from that of long fatigue cracks. For the latter, the size of the plastic zone in the vicinity of the crack tip is negligibly small as compared to the crack length a, and the corresponding crack growth rate da/dN follows a characteristic relationship with the applied stress intensity factor ΔK, being described by the well-known Paris law [1], or more generally by the Klesnil–Lukáš relationship [2]:

$$\frac{\mathrm{d}a}{\mathrm{d}N} = \frac{C}{(1-R)^m}(\Delta K^m - \Delta K_{\mathrm{th}}^m), \quad (1)$$

accounting for the stress ratio R and the threshold value ΔK_{th} for initiation of "technical" cracks. The factor C and the exponent m are material constants. Microstructurally short cracks may initiate and grow well below the threshold for long cracks ΔK_{th}. They are of the same order of magnitude than the characteristic microstructure features [3] and exhibit an oscillating growth behavior [4]. Once a microstructural crack has been initiated, e.g., by accumulated irreversible slip or elastic anisotropy of the microstructure constituents, it usually grows along crystallographic slip bands driven by the cyclic shear displacement at the crack tip [5]. Depending on the crystallographic relationship between neighboring grains of the same or different phases and on the geometry and particular strength of the grain or phase boundaries, various propagation scenarios of microstructurally short cracks have been observed: (i) slip bands may cause slip steps in the boundary planes followed by intercrystalline cracking [6], (ii) the grain and phase boundaries may act as barriers against slip transmission and reduce the crack propagation rate [7], and (iii) depending on the orientation of the slip planes with respect to the local stress state and the lattice structure of the respective grains, the microcrack propagation mechanism can change from single slip to alternating multiple slip operating at the crack tip [8]. The latter mechanism is often referred to as stage II crack propagation driven by the nominal normal stress

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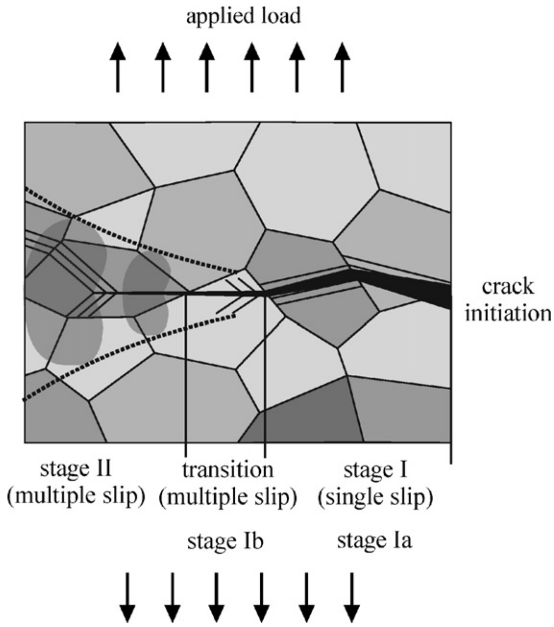

Fig. 1. Schematic representation of the different crack propagation regimes during fatigue damage of metals and alloys.

perpendicular to the direction of crack advance. However, even under these conditions the crack can still be very short, and the plastic zone is not necessary small compared to the crack length. Additionally, the microstructure may have an influence on the crack propagation, so the crack growth rate cannot directly be described by continuum mechanics, certainly not by linear elastic fracture mechanics. Hence, one should define a transition stage between stage I microcrack propagation along crystallographic slip bands and stage II crack propagation resulting in fatigue striations. This is shown schematically in Fig. 1, being valid for both, crack propagation along the specimen surface as well as into the bulk.

Prediction of the crack propagation rate during the early stage of fatigue damage requires a model that accounts for alterations in the local resistance to crack advance. Navarro and de los Rios [9] and Taira et al. [10] developed analytical models where such alterations are attributed to a dislocation pile up between the crack tip and the grain boundary. Once the pile-up stress acting on a dislocation source in the adjacent grain exceeds a critical stress, plastic slip sets in giving rise to an intermittent increase in the crack propagation rate.

While the analytic model of Navarro and de los Rios considers the relationship between local crystallographic orientation and the variation in crack propagation only in one dimension by an average orientation factor, Section 4 of the present paper introduces a numerical two-dimensional short crack model that is capable to treat single- and two-phase microstructures with measured variation in the grain size and crystallographic orientation distribution.

2. Experimental details

Microcrack propagation was studied on electro-polished, shallow-notched specimens (Fig. 2) of an austenitic–ferritic

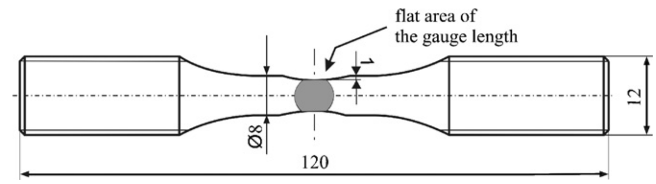

Fig. 2. Geometry of shallow-notched fatigue specimens.

Table 1
Nominal chemical composition of the stainless steels used in this study (wt.%)

	Fe	Cr	Ni	Mo	Mn	Si	Nb	N	C
1.4462	Bal.	21.9	5.6	3.1	1.8	0.5	–	0.1871	0.020
1.4404	Bal.	16.6	11.1	–	1.3	0.6	0.01	0.0296	0.018
1.4511	Bal.	16.3	–	–	0.7	0.5	0.25	–	0.012

(γ–α) duplex steel 1.4462 (ASTM A182 F51) with an average grain size of $d(\alpha) = 46\ \mu m$ and $d(\gamma) = 33\ \mu m$, respectively, and an α/γ-volume ratio of approximately 0.5. The barrier strength of $\gamma\gamma$ and $\alpha\alpha$ grain boundaries as well as of the $\alpha\gamma$ phase boundaries was quantified by means of a cyclic Hall–Petch analysis, applying incremental step tests on ferritic steel 1.4511 and austenitic steel 1.4404 (AISI 316L) reference materials with various grain sizes. Details on this investigation can be found elsewhere [11]. The chemical compositions of all the materials studied are given in Table 1.

Fatigue experiments were carried out in a MTS810 servohydraulic testing machine under fully reversed ($R = -1$), stress-controlled push-pull loading conditions. The specimens were periodically removed from the testing machine to evaluate microstructurally short cracks within the electropolished shallow-notched gauge length by means of scanning electron microscopy (SEM). The crystallographic orientations of the grains and phase patches involved in the crack propagation process were determined by means of automated electron back-scattered diffraction (TSL OIM^{TM}, orientation imaging microscopy).

Information about three-dimensional effects of microcrack propagation was obtained: (i) by stepwise polishing the surface in the vicinity of selected microcracks and (ii) by applying ISDG (interferometric strain/displacement gage) measurements to monitor the development of crack-mouth opening displacement (CMOD).

3. Results and discussion

In most cases, cracks in the $\gamma\alpha$ duplex steel are initiated in the vicinity of grain or phase boundaries. This might be attributed to the interdependence between the elastic constants and the crystallographic orientation of the two-phase microstructure constituents (elastic anisotropy). As an example, Fig. 3 shows a surface microcrack initiated at a γ-austenite twin-boundary, and further growing from (1) by alternate operating $(1\,1\,1)\ \langle 1\,1\,0\rangle$-slip systems in a similar way as it was recently observed and discussed by Blochwitz and Tirschler [12]. At the left-hand side of the cracked twin boundary, microcrack propagation resumes from a triple point (2) into an adjacent bcc α-ferrite grain gov-

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