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Experimental investigation and analytical description of the damage evolution in a Ni-based superalloy beyond 10^6 loading cycles

Anton Kolyshkin [a,*], Martina Zimmermann [b], Edgar Kaufmann [c], Hans-Jürgen Christ [a]

[a] Institut für Werkstofftechnik, Universität Siegen, 57068, Germany
[b] Institut für Werkstoffwissenschaft, TU Dresden, 01062, Germany
[c] Department Mathematik, Universität Siegen, 57068, Germany

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ABSTRACT

The fatigue life of metallic materials in the very high cycle fatigue (VHCF) regime is characterized by a large scatter of experimental results. The aim of the present work is to investigate the reasons for the scattering of the fatigue life as well as to model its impact statistically. For this purpose crack initiation in the Ni-base superalloy Nimonic 80A was investigated as a function of the number of cycles to failure at different stress amplitudes both in the HCF and the transitional HCF–VHCF regime. According to the experimental observations the fatigue crack initiation depends on the stress amplitude and occurs at twin boundaries and grain boundaries with a large misorientation angle. The application of two analytical approaches for the assessment of the local stress concentration at different grain boundaries is discussed. In particular in the VHCF regime, the misorientation factor developed by Blochwitz et al. (1997), which depends on the misorientation angle between two adjacent grains as well as on the orientation of their boundary with respect to the external load, was used to estimate the stress concentration at the grain boundaries. It was revealed, that the cracks initiate at the grain boundary of maximum misorientation factor calculated for each specimen. Furthermore, the calculated maximum misorientation factor per specimen show a direct relation to the number of cycles to failure and can be used as an additional microstructural information for statistically based fatigue life prediction models.

1. Introduction

The prediction of fatigue life in the very high cycle fatigue (VHCF) regime of quasi defect-free materials such as nickel-based superalloys is still a challenge. Recent studies, for example [2,3], show that a true durability does not exist for this group of materials and fatigue failures can occur even beyond 10^7 loading cycles. Moreover, the huge scatter band of experimental results at low stress amplitudes relates to a strong influence of the microstructural heterogeneities such as varying grain boundary character, distribution of grain size and microtextures, which together with the applied stress amplitude determine fatigue life. The influence of the microstructural heterogeneities has a probabilistic character and increases the scatter band width in the area of VHCF up to three decades for the SN-curve, thus impeding the application of a reliable fatigue life prediction concept by means of traditional statistical approaches and requiring a detailed investigation of the relevant fatigue damage mechanisms.

Although polycrystalline nickel-based superalloys are widely used in aerospace and power generation industries, only few systematical investigations describing the damage mechanisms during the cyclic loading in the VHCF regime are known. Investigations on fatigue crack initiation in the VHCF regime performed by Miao et al. [3] on René 88DT at elevated temperature reveal internal crack initiation from subsurface regions. The authors showed that the elastic incompatibility at twin boundaries of large grains causes a stress concentration and can lead to intercrystalline crack initiation far below the yield limit of the alloy. The room temperature investigations on WASPALOY [4] and René 88DT [5] in both HCF and transitional HCF–VHCF regime report a strong influence of so-called microstructural extremes on the crack initiation and fatigue life. Microcracks emerge at grains that are significantly larger than the average grain size of the investigated alloys and propagate through the adjacent grains only in those cases, where the misorientation between the adjacent and the already cracked grain is small. However, the damage mechanisms are different in both alloys. While in WASPALOY microcracks primarily initiate at persistent slip bands caused by the accumulation of irreversible cyclic

* Corresponding author.
E-mail address: anton.kolyshkin@uni-siegen.de (A. Kolyshkin).

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dislocation slip, the reason for crack initiation in René 88DT is the incompatibility stress at twin boundaries.

The important role of twin boundaries for fatigue failure in nickel-based superalloys is also reported by Chen et al. [6] and Gao et al. [7]. Chen et al. investigated small crack behavior of Inconel 718 in the VHCF-range and revealed an intensive slip band activity close to the twin boundaries that was assumed to be responsible for the fatigue crack initiation. Gao et al. investigated the small crack propagation in René 104 (ME3) at stress amplitudes corresponding to the HCF-range. The crack propagation preferably took place along twin boundaries. The investigations of Stöcker et al. [2] report a pronounced effect of the precipitation condition, prestrain and temperature on the fatigue behavior of Nimonic 80A and Nimonic 75 and form the main experimental basis for the study presented.

The mentioned investigations emphasize the importance of microstructural features with respect to the durability of nickel-based superalloys and describe their typical damage mechanisms beyond 10^5–10^6 numbers of cycles. However, there is a lack of knowledge about analytical approaches that allow an assessment of the local loading conditions at one or another crack initiation site and predict the corresponding number of cycles to failure in the VHCF-range. Such approaches will be proposed in the study presented. Their application will be discussed using the example of a nickel-based superalloy Nimonic 80A.

2. Material and experimental details

2.1. Microstructural characterization

The material analyzed in this study is the polycrystalline nickel-based superalloy Nimonic 80A, whose nominal chemical composition (wt.%) is 19.91Cr, 2.69Ti, 1.69Al, 0.12Fe, 0.01Mn, <0.01Cu, 0.049C, 0.02Si and Ni (balance). The material was received after hot rolling, dynamic recrystallisation and subsequent grinding in form of 15 mm diameter rods and exposed to the following heat treatment:

1. Solution treatment at 1050 °C for 2 h followed by quenching in water
2. Precipitation hardening treatment at 710 °C for 16 h followed by air cooling.

The presented heat treatment was developed according to Ref. [2] and corresponds to the peak-aged condition of the investigated alloy. The microstructure of Nimonic 80A in the peak-aged precipitation condition consists of the γ matrix phase and the γ' phase. The size of the γ' phase was measured using transmission electron microscopy (TEM). TEM investigations were carried out using a FEI Talos F200A instrument. More details on the steps applied in TEM sample preparation can be found in Ref. [8]. A TEM micrograph with globular randomly distributed $Ni_3(AlTi)$ precipitates of a size of 15–20 nm is shown in Fig. 1a. The size of precipitates confirms the occurrence of the peak-aged structure [2].

The γ grains were quantitatively described on the basis of a large scan area using electron backscattering diffraction (EBSD) analysis. EBSD measurements were made by orientation mapping of metallographically prepared specimens in a Philips XL 30 scanning electron microscope (SEM) equipped with the OIM™ software from TSL, Inc (Draper, UT). The analysis of the EBSD data was performed using custom algorithms described in detail in [9,10]. Grain boundaries were characterized according to the angle/axis notation [11].

Fig. 2 confirms a homogeneous crystallographic orientation distribution after applying the heat treatment described. The orientation map of the alloy is shown in Fig. 3a. The grain boundary misorientation angle distribution depicted in Fig. 3b denotes a big fraction of twin boundaries, which have a special 60° misorientation angle.

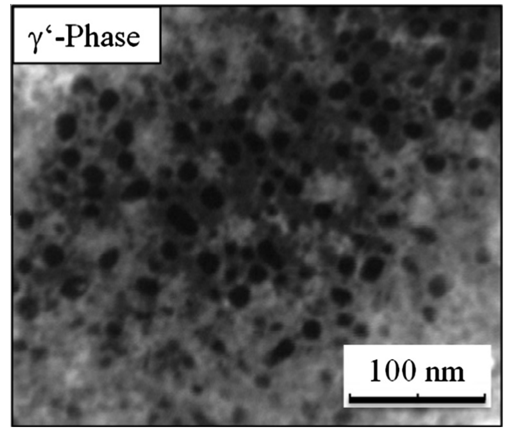

Fig. 1. TEM micrograph showing the size of precipitates in Nimonic 80A after heat treatment.

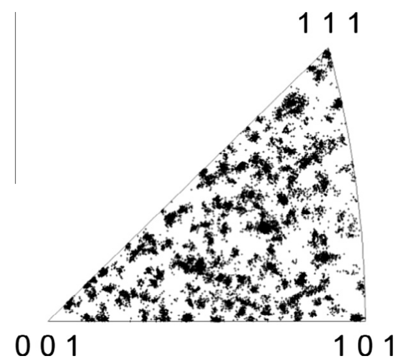

Fig. 2. Crystallographic orientation distribution of the γ grains measured using EBSD analysis.

The scan areas with a misorientation between individual scanning points smaller than the grain misorientation tolerance angle θ_t of 5° were defined as points belonging to the same grain. The calculated grain equivalent diameter distribution with the minimum grain size assumed to be 5 μm is shown in Fig. 3c. The average grain size of the alloy excluding twin grains is about 20 μm. In order to investigate the availability of the so-called microstructural extremes i.e. the groups of adjacent grains with small angle grain boundary character, the grain size was also measured for a 20° grain misorientation tolerance angle.

A comparison of the obtained grain equivalent diameter distribution depicted in Fig. 3c shows that both the average and the maximum grain size are not strongly affected by the change of the grain misorientation tolerance angle. This fact indicates a very low amount of small angle grain boundaries, which would be necessary to form such microstructural extremes as reported in [4,5].

2.2. Tensile properties

The quasi-static mechanical properties of Nimonic 80A that were achieved on the basis of the presented heat treatment are shown in Table 1. The tensile tests were carried out on cylindrical specimens with a gauge diameter of 6 mm on a servohydraulic testing system.