Localized cyclic deformation and corresponding dislocation arrangements of polycrystalline Ni-base superalloys and pure Nickel in the VHCF regime

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Abstract

Knowledge about the fatigue behaviour of metallic materials in the range of very high cycle fatigue (VHCF, N > 10^7) is a key factor to improve the safety against failure of component parts. Even in the conventional field of application of nickel-base alloys (e.g. turbine blades) superimposed high frequency loading requires a save life at very high number of cycles. Thus, the fatigue behaviour of structural materials in the very high cycle regime has become an important and active subject of research. In this paper, polycrystalline nickel-base alloys (Nimonic 80A and Nimonic 75) and pure Nickel (wavy slip character) were investigated in the VHCF regime by varying the precipitation conditions (peak-aged, overaged and precipitation-free), dislocation slip behaviour and test frequency. In addition, mechanisms of fatigue failure in the VHCF regime were compared to conventional damage evolution (in the LCF and HCF region) on the basis of load-controlled low frequency tests. Surprisingly the overaged condition of Nimonic 80A shows a slightly higher fatigue strength in the VHCF regime as compared to the peak-aged condition. The results obtained
document that fatigue failure can still occur beyond $10^7$ cycles. Transmission electron microscopy (TEM) was carried out, in order to characterize the influence of microstructure, the resulting dislocation slip behaviour and the relevant dislocation particle interaction mechanism. Studies of the development of slip markings on surface grains were performed mainly applying scanning electron microscopy (SEM).

1. Introduction

Damage evolution during fatigue is mainly based on irreversible plastic deformation as a result of slip activities concentrated on active slip planes leading to the formation of intrusions and extrusions on the surface of a fatigued specimen [1]. In the very high cycle fatigue (VHCF) region ($N > 10^7$ cycles) crack initiation can be observed for ductile single phase metals, although from a macroscopic point of view specimens are subjected to a purely elastic strain. Hence, the damage evolution must be limited to locally accumulated deformation processes with a strong microstructural influence. A possible explanation was given by Mughrabi [2] by assuming, that the accumulation of slightly irreversible random slip can lead to surface roughening and subsequently persistent slip bands (PSBs) might form at the notch tips with the maximum stress concentration. This would explain the crack initiation in pure single phase ductile metals and has been recently examined by Stanzl-Tschegg et al. for pure polycrystalline copper [3]. PSB formation was assumed as consequence of well-pronounced extrusions and intrusions which were observed at surfaces of specimens after fatigue cycling at stress amplitudes below 56 MPa. Transmission electron microscopic studies revealed dislocation patterns, which slightly resembled the ladder structure of a PSB as well as elongated loop patches and cell structures.

In polycrystalline materials, such as the first generation wrought nickel-base alloy Nimonic 80A analyzed in the present research work, the formation of specific dislocation arrangements and as a consequence the fatigue behaviour as well as the crack initiation is strongly influenced by the
interaction between dislocations and obstacles, such as grain boundaries or precipitates, acting as motion barriers [4-6]. The initial dislocation density, e.g. resulting from a monotonic predeformation, as well as the size, shape and distribution of the $\gamma'$ precipitates have a strong influence on the dislocation rearrangement and mobility during fatigue. These effects have been extensively analysed for the low and high cycle fatigue regime [7-9]. In the case of an inhomogeneously distributed local plastic deformation, as is assumed for the VHCF range, the interaction between local microstructure and dislocation mobility gains further importance.

The correlation of microstructure and fatigue behaviour of Ni-base alloys at very high number of cycles has recently been investigated with regard to the influence of non-metallic inclusions and pores in the Ni-based superalloys (prepared by powder metallurgy techniques) René 88 DT [10] and N18 [11]. Kunz et al. [12] studied the influence of casting defects in the cast IN 713LC alloy on the fatigue behaviour at 800 °C. None of these research works revealed a deeper insight into the dislocation formation and movement during fatigue testing. Rather the studies focused on the correlation between microstructural defects (pores, non-metallic inclusions etc.) and VHCF behaviour. The interpretation of the damage evolution is solely based on S-N-data, fractographic analysis and analytical models. In contrast, the formation of extrusions and intrusions has been extensively studied for polycrystalline, single-phase pure copper [13] and aluminium [14]. For both metals the formation of slip planes on the surface of fatigue specimens was observed. Further investigations on single-phase copper showed that so-called VHCF-PSBs were formed at a stress/strain amplitude of $\Delta\sigma \sim 45$ MPa and $\Delta\varepsilon_{pl}/2 \sim 3 \times 10^{-6}$ after $2.7 \times 10^8$ cycles [15]. Although intrusions and small cracks could be detected, the stress/strain amplitudes needed to form a propagating “long” crack were about 100 % higher than those needed to form VHCF-PSBs. The study presented aims to contribute to a better understanding of the developed dislocation density during fatigue in the VHCF range and the corresponding damage evolution.
2. Experimental details

2.1 Materials

The study was carried out on the polycrystalline nickel-base alloys Nimonic 80A in the peak-aged (pa) and overaged (oa) condition, Nimonic 75 and pure Nickel. The chemical compositions are given in Table 1. The precipitation-hardening alloy Nimonic 80A contains titanium and aluminium that substantially improve the creep strength through the formation of coherent precipitates of the $\gamma'$ phase $\text{Ni}_3\text{(Ti,Al)}$. The Nimonic 80A specimens were solution heat treated for 30 minutes at 1080 °C and then quenched in water. This first heat treatment step was followed by a precipitation annealing of 16 hours at 710 °C for the peak-aged condition and 500 hours at 850 °C for the overaged condition. Peak-aged and overaged conditions were defined on the basis of the hardness course measured at room temperature (Fig. 1). Vickers hardness tests resulted in a value of 362 HV 30 for the peak-aged and 261 HV 30 for the overaged condition. The peak-aged condition is characterised by spherical $\gamma'$ precipitates of a mean particle diameter of 15-20 nm (STEM micrograph shown as insert in Fig. 1(a)). The over-ageing treatment led to a cuboidal $\gamma'$ size of 250-300 nm (see insert in Fig. 1(b)). According to [16] $\gamma'$ particle sizes exceeding 200 nm in Nimonic PE 16 (similar $\gamma'$ volume fraction as in Nimonic 80A) give rise to dislocation by-passing processes during room temperature deformation. Because of the positive-misfit in Nimonic 80A (+0.32 %) [17] $\gamma'$ begins to transform to the expected cuboidal shape with a (100) habitus with increasing ageing time (left STEM micrograph in Fig. 1(b)). The shape change is accompanied by alignment along the "soft" [100] directions, shown in the right STEM micrograph in Fig. 1(b). According to [18] the $\gamma'$ volume fraction of Nimonic 80A lies between 10 and 20%.

Nimonic 75 is a solid solution hardened alloy containing substitution elements such as chromium and iron. This material was analysed in order to represent the behaviour of the matrix of
Nimonic 80A (precipitation-free condition (pf)). In order to improve the comparability to Nimonic 80A, the mean grain size of Nimonic 75 was adapted to the one of Nimonic 80A by means of an annealing treatment that led to a mean grain size of around 60 µm. Pure Nickel was used as a reference material for wavy slip behaviour. Because of a grain size similar to the ones of the other materials, pure Nickel was tested as received (technical condition).

2.2 Testing procedures

Tensile tests were accomplished at ambient temperature for the different precipitation conditions (pa, oa and pf) and for pure Nickel. Table 2 gives an overview of the test results. The strong decrease of the yield strength of overaged Nimonic 80A in relation to the peak-aged condition is in very reasonable agreement with the decrease in hardness values.

The uniaxial VHCF tests were conducted in load-controlled or displacement-controlled constant-amplitude cycling in laboratory air at room temperature under symmetrical push-pull conditions (R=-1). Specimens with cylindrical gauge length of 3 mm up to 6 mm in diameter depending on the test system were machined from bars after the heat treatment. Test frequencies f varied according to the test system used: around 130 Hz for a resonance electromechanical device (gauge diameter: 6 mm), 760 Hz for a servohydraulic system (gauge diameter: 5 mm) and around 20 kHz for an ultrasonic fatigue test system (gauge diameter: 3 mm). Further informations to the high frequency testing systems used are available in the literature [e.g. 19]. The type of testing device also defined the number of cycles, which classified a sample as run-out. The fatigue tests were accomplished either up to final rupture or up to a number of cycles of $N_{\text{run-out}} = 2.8 \times 10^8$ (130 Hz), $N_{\text{run-out}} = 4 \times 10^8$ up to $7.2 \times 10^8$ (760 Hz) and $N_{\text{run-out}} = 8.7 \times 10^8$ up to $2.6 \times 10^{10}$ (20 kHz). The samples which show no appreciable crack or final rupture at the given load cycles were classified as run-outs. Before fatigue testing the gauge length of the samples were electrochemically polished in a solution of acetic acid and perchloric acid (10:1) at 14 °C after mechanically grinding. Heating of
the specimens due to the high loading frequencies was avoided by means of compressed-air cooling. The ultrasonic fatigue testing was executed in intermitting pulse and pause sequences. The pulsing and pausing periods were 200 and 600 msec, respectively.

The stress amplitudes for the S-N curve are defined on the basis of the measured load amplitude and the effective cross section of each specimen for the servohydraulic and the electromechanical resonance test system. The determination of the stress amplitude for the ultrasonic test, is based on the strain measurements being executed during initial calibration of the ultrasonic specimens and the Young’s modulus given in Table 2. Preliminary testing showed that transient behaviour in the amplitude range studied is negligible and global elastic behaviour according to Hookes law can be assumed during ultrasonic fatigue testing.

Besides the VHCF experiments, load-controlled low frequency tests (f=0.25 Hz up to 1 Hz) were carried out on a conventional servohydraulic test system, in order to compare dislocation arrangements, dislocation/γ’-precipitate interaction and the formation of slip markings in the LCF and VHCF regime. The specimen strain during the LCF tests was measured with a standard clip-on extensometer attached directly to the gauge length. In addition, load-controlled HCF tests (< 4 x 10⁵ cycles) were realized again on an electromechanical resonance device (f around 75 Hz) to complement the S-N curves.

Detailed microstructural studies were performed mainly by means of transmission electron microscopy (TEM) on samples cut parallel to the stress axis. The TEM studies were performed with a Hitachi H-8100 electron microscope at 200 kV operating voltage. Foil preparation included mechanical sawing by cutting thin slices from the gauge length, mechanically polishing on both sides and punching out small discs with a diameter of 3 mm. Finally the discs were jet thinned at 14 °C and a voltage of 40 V in a solution as used for electrochemically polishing.

In order to determine the size, form and arrangement of the γ’ particles in the γ matrix of the peak-aged and overaged Nimonic 80A, foils were prepared for a Zeiss Ultra 55 ultrahigh resolution field emission scanning transmission electron microscope (STEM) as described above. Standard
scanning electron microscopy (SEM) was used to study the formation of slip markings within the
gauge length of fatigued specimens in the LCF-HCF-VHCF regime and to examine the fracture
surfaces.

3. Results and discussion

3.1 Fatigue testing (S-N curves) and cyclic deformation behaviour in the LCF-HCF-VHCF regime

The results of fatigue tests performed on both precipitation conditions (pa and oa) are shown in
Fig. 2(a) for the LCF-HCF regime and in Fig. 2(b) for the HCF-VHCF regime. Failure occurred up
to a number of $N_f = 4.2 \times 10^8$ cycles for the peak-aged condition ($\Delta\sigma/2 = 268$ MPa) and
$N_f = 7.8 \times 10^7$ cycles for the overaged condition ($\Delta\sigma/2 = 330$ MPa). The overaged specimens
fatigue limit was found to be at 320 MPa. Run-out specimens are marked with arrows in Fig. 2(b).
Comparing the two different precipitation conditions, the overaged specimens surprisingly shows a
superior fatigue strength with a stress amplitude being 30-40 MPa higher in the VHCF regime.
Supplementary tests in the LCF regime show that the fatigue life behaviour of the peak-aged and
overaged condition is inverted to that in the VHCF range and thus correlates with the static strength,
as would have been expected. After a considerably higher fatigue strength for the peak-aged
condition in the LCF regime both S-N curves (pa and oa) converge in the HCF regime (Fig. 2(a)).
Already in the transition region from HCF to VHCF both S-N curves intersect leading to a higher
fatigue life for the overaged condition in the VHCF regime.

Results from fatigue tests performed on the precipitation-free condition (Nimonic 75) and on
pure Nickel are shown by means of the S-N data in Fig. 3. It is clearly seen that fatigue failure
occurs beyond $10^7$ cycles. The lowest fatigue resistance is shown by pure Nickel with a failed
specimen at around 150 MPa at $N_f = 8.9 \times 10^9$ cycles. The difference in the fatigue strength
between the pf condition (Nimonic 75) and pure Nickel is in the range of around 70-100 MPa.
(LCF-VHCF) and can be attributed to the solid-solution hardening of Nimonic 75. The precipitation-free matrix displays a fatigue limit for the VHCF range of around 225 MPa, thus being considerably lower than the precipitation-hardened alloy Nimonic 80A (pa and oa).

One of the characteristic features of fatigue testing in the VHCF range is the large scattering of the fatigue life results, e.g. the results for the peak-aged Nimomic 80A (Fig. 2(b)) are showing a scatter which comprises more than three decades of loading cycles. The plotted S-N data for the VHCF regime contains results from different test frequencies (130 Hz, 760 Hz and around 20 kHz). However, there is no significant and systematic influence of the frequency on the fatigue behaviour at RT.

In contrast to many investigations on the VHCF regime, which propose a transition of crack initiation from surface to interior [20] with increasing number of cycles until failure, in the study presented no correlation between crack initiation site (at the surface or in the bulk material) and the number of cycles could be detected. The fracture surfaces of the failed specimens were examined in SEM showing, that the exact location of fatigue crack initiation can hardly be determined in the tested materials. All specimens regardless of the heat treatment and the resulting strengthening mechanism revealed striations in the fracture surface and according to the more pronounced fracture paths the crack origin was always at or close to the surface, but the precise crack initiation site could not be identified unambiguously. Therefore, it must be assumed that crack initiation took place on the surface or just below the surface. This is not surprising, as the materials studied are practically defect-free and do not show any microstructural anomalies, such as very large twin grains [21]. Hence, the crack initiation will likely result from surface roughening during cyclic deformation as a consequence of dislocation patterns and/or localized plastic deformation due to strength mismatches between matrix and $\gamma'$ precipitates. TEM studies of the dislocation/particle interaction performed directly at the crack initiation sites are needed, in order to develop a better understanding of the fatigue mechanisms leading to the fatigue results presented.
Figure 4 illustrates the difference between fatigue tests in the LCF (a) and the VHCF regime (b) regarding cyclic deformation behaviour using the example of the overaged condition. Both precipitation conditions showed a cyclic hardening during the first cycles in the LCF regime that is caused by an increasing dislocation density. Because of the high stress amplitude ($\Delta\sigma/2 = 700$ MPa $> R_{p0.2}$) of LCF testing Nimonic 80A establishes a mean strain in the overaged condition (Fig. 4(a)), whereas in the peak-aged condition no mean strain was observed during LCF tests ($\Delta\sigma/2 = 820$ MPa $< R_{p0.2}$) [22]. The cyclic deformation behaviour of Nimonic 80A (pa and oa) is strongly governed by a strong dislocation/precipitate interaction. At RT cutting of peak-aged $\gamma'$ particles by dislocation pairs prevails [23], while in the oa-condition Orowan bypassing reduces the resistance against dislocation movement [16]. As will be shown in the next chapter, overaged $\gamma'$ particles in Nimonic 80A are surrounded by loops, giving evidence for the Orowan process as dominating mode of dislocation particle interaction in the LCF regime. Following the cyclic hardening process, the occurrence of bypassing in the overaged condition, the shearing in the peak-aged condition and an early microcrack formation led to a weak cyclic softening. On the contrary, in the VHCF regime the plastic strain amplitude (and the mean strain) is nearly zero because of the low stress amplitude (Fig. 4(b)). Under the low stress-amplitude conditions a solely elastic (reversible) deformation takes place during VHCF experiments. A cyclic deformation curve including hardening, saturation and softening does not exist in the VHCF regime. Because of the occurrence of only local plastic deformation, the cyclic deformation curve runs horizontally due to the fact that the curve simply reflects the integral deformation process of the gauge volume. Therefore, the crack initiation becomes more important for fatigue life in the VHCF regime, whereas in the LCF regime crack propagation is more relevant. The cyclic deformation curve in the VHCF regime (Fig. 4(b)) represents a reference fatigue test performed on a conventional servohydraulic test system ($f = 5$ Hz) up to $10^6$ cycles.
Prior to fatigue testing no appreciable dislocation density and pronounced dislocation arrangements were observed for both precipitation conditions. After cyclic deformation in the VHCF regime, Nimonic 80A shows planar dislocation arrangements in single grains with no interaction between the activated slip bands for the peak-aged (Fig. 5(a)) and the overaged condition (Fig. 5(b)). Direct shear processes of the $\gamma'$ particles were not found. However, it should be noticed that the precipitates in the peak-aged condition are very small and hence they might completely be dissolved by shearing. In contrast, there would be expected for the overaged condition that the Orowan mechanism takes place in the VHCF regime. However, Orowan loops around the overaged $\gamma'$ particles were not observed. Rather, the dislocation movement in the overaged condition is dominated by pile-ups at the larger and cuboidal-shaped particles [24]. The interaction of dislocations with precipitates in the VHCF regime is restricted and take place only in single grains of favourable grain orientations (with high Schmid factor). The schematic representation in Fig. 6 proposes that the overaged condition (Fig. 6(a)) exhibit a higher fatigue life in the VHCF regime because of the more homogeneous cyclic deformation as compared to the peak-aged condition (Fig. 6(b)). In single grains in the peak-aged condition single slip bands move through the whole grain and are only stopped by the barrier effect of the grain boundaries. The dislocation pile-up leads to a stress concentration at the grain boundaries and therefore to an earlier formation of microcracks at lower shear stresses than in the overaged condition. Typical of the VHCF microstructure is the coexistence of virtually dislocation-free grains and those with dislocation arrangements as described before. Please note that the cuboidal overaged particles in Fig. 6(b) are not located in the slip plane of those edge dislocations, which are drawn black.

Besides the precipitation condition, the formation of microscopical notches (e.g. surface roughening from a monotonic predeformation) are of wide relevance for fatigue tests in the VHCF regime [25, 26]. The higher fatigue life of the overaged condition can consequently be attributed on
the one hand (as explained above) to a more homogeneous slip behaviour and on the other hand to
the higher ductility (less notch sensitivity). So the higher fracture strain (Table 2) has a positive
effect to the fatigue life in the VHCF regime for the overaged condition.

3.3 Dislocation arrangements versus slip character and stress amplitude

The slip character of a material is a basic parameter that determines the type of dislocation
arrangement formed during cyclic loading and therefore also the cyclic stress-strain response. The
tendency to form a three-dimensional network is described by the slip character. The extreme cases
in cyclic deformation processes are on the one hand a pure planar dislocation slip and on the other
hand a pure wavy dislocation slip behaviour [27]. The dependence of the dislocation arrangement
on the plastic strain amplitude or the number of cycles to failure and on the slip character was first
introduced in 1968 by Feltner and Laird [28] and later on confirmed quantitatively by Lukáš and
Klesnil [29]. However, it should be mentioned that a clear classification of materials according to
wavy or planar dislocation slip behaviour is only possible for single-crystalline metals. An exact
delimitation of the dislocation arrangement type is hardly possible in polycrystalline metals due to
the fact that by reason of the different grain orientations and the required compatibility of
deformations, the loading situation differs locally [30]. Nevertheless, the different types of
dislocation arrangements of the fatigued polycrystalline fcc materials of this study (pure Nickel,
Nimonic 80A and Nimonic 75) are mapped as a function of the number of cycles to failure (LCF
and VHCF regime, abscissa) and the slip character (ordinate) for clarification.

As a consequence of high-amplitude loading (LCF regime), a cell structure typical of multiple
slip in pure Nickel (wavy-slip material) is formed [31] (top left TEM micrograph in the diagram in
Fig. 7), whereas in the VHCF regime (low-amplitude loading) only dislocation dipole bundles were
observed in local regions (see top right TEM micrograph in the diagram in Fig. 7).
The dislocation arrangement in planar-slip type materials such as Nimonic 80A and Nimonic 75 is illustrated in the bottom TEM micrographs in Fig. 7 (left: LCF, overaged Nimonic 80A; right: VHCF, Nimonic 75). The dislocation density in the LCF regime is significantly increased and the dislocations/precipitates interaction provided the evidence of Orowan bowing in the overaged condition. In the VHCF regime, only a planar dislocation arrangement with low density restricted to isolated grains were found. However, it should be pointed out that the planar character of the dislocation arrangement for Nimonic 80A (pa and oa condition) and Nimonic 75 prevails in the whole number of cycles to failure range (LCF-VHCF) covered in this study. In the precipitation-free condition (Nimonic 75), only the grain boundaries act as a barrier for the dislocations, whereas in the precipitation hardened alloy Nimonic 80A the coherent $\gamma'$ particles impede the dislocation movement, since they are embedded as obstacles in the $\gamma$ matrix and therefore fatigue strength increase. Unlike the investigated planar slip alloys, pure nickel shows a more homogeneous dislocation distribution as a consequence of the wavy dislocation slip behaviour. Nevertheless, the fatigue strength of pure Nickel is the lowest in comparison to the other conditions (pa, oa and pf) of both nickel-base alloys.

3.4 Formation of slip markings in the LCF-HCF-VHCF regime

In addition to the TEM investigations, the formation of slip markings were studied by means of SEM to shed light on the predominant damage mechanisms of VHCF. It is clearly seen in Fig. 8((a) and (b)) using the example of fatigued pure Nickel that the density of slip markings decreases with decreasing applied stress amplitude in the LCF-HCF range. The direction of loading is horizontal in all micrographs shown in Fig. 8. For all three precipitation conditions (pa, oa and pf) investigated, the density of slip markings decreases from the LCF to the HCF range, too. Some grains show slip lines from secondary slip in addition to those from primary slip, but in most cases traces belonging to just one system are seen. At high stress amplitude, plastic deformation takes place in all surface
grains, whereas in the VHCF regime formation of slip markings appeared only locally on surface
grains with high Schmid factors [22]. The SEM investigations confirm the TEM results of inner
grains of the specimens that plastic deformation occurs heterogeneous distributed in the VHCF
regime. The corresponding slip steps at the surface favour transgranular cracking and leads to
failure beyond $10^7$ cycles. Although slip markings were visible on the surface of specimens fatigued
up to a number of cycles beyond $10^8$ cycles (e.g. Fig. 8(c)) and new slip lines are initiated beyond
$10^{10}$ cycles (Fig. 8(d)), this process does not necessarily result in a propagating fatal crack.

4. Conclusions

The main results obtained in this study can briefly be summarized as follows.

- The room temperature fatigue behaviour of nickel-base superalloys in three precipitation
  conditions (Nimonic 80A (peak-aged and overaged condition) and Nimonic 75
  (precipitation-free condition)) and of pure Nickel has been investigated in the LCF, HCF
  and VHCF regime.
- An increased monotonic strength established by peak-aging did not lead to a higher fatigue
  strength in the VHCF regime.
- The overaged condition showed an improved VHCF fatigue behaviour because of a more
  homogeneous dislocation slip arrangement and a less notch sensitivity as in the peak-aged
  condition.
- Nimonic 80A and Nimonic 75 showed planar dislocation arrangements in single grains,
  whereas in pure Nickel only dislocation dipoles were observed locally for the VHCF regime.
- Plastic deformation occurred only in single grains (in the bulk and at the surface) due to the
  low stress amplitudes applied in VHCF tests.
• SEM observations of surface grains reveal that the density of slip markings decreases with decreased stress amplitude applied (LCF-VHCF).

• Although slip density increased in single surface grains with increasing number of loading cycles in the VHCF regime even at $10^{10}$ cycles, no crack propagation took place (run-out samples).

Acknowledgement

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References


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Fig. 1. Hardness values for peak-aged (a) and overaged (b) Nimonic 80A depending on annealing time and γ’ precipitation size.

Fig.2. Fatigue results for peak-aged and overaged Nimonic 80A (arrows indicate run-out specimens) in the LCF-HCF regime (a) and HCF-VHCF regime (b).
Fig.3. Fatigue results for Nimonic 75 and pure Nickel (arrows indicate run-out specimens) in the LCF-VHCF regime.

Fig.4. Cyclic deformation behaviour of the overaged condition of Nimonic 80A in the LCF (a) and in the VHCF regime (b).

Fig.5. TEM micrographs of the microstructure of peak-aged (a) and overaged (b) condition of Nimonic 80A in the VHCF regime.

Fig.6. Schematic representation of the dislocation/precipitate interaction of Nimonic 80A for the peak-aged (a) and overaged (b) condition in the VHCF regime.

Fig.7. Dislocation arrangement at RT as a function of plastic strain amplitude or the number of cycles of failure and of the slip character.

Fig.8. SEM micrographs showing the density of slip markings in surface grains of pure Nickel after cycling loading at
   a.) \( \Delta \sigma/2=280 \) MPa, \( N_f=2.53 \times 10^4 \), b.) \( \Delta \sigma/2=220 \) MPa, \( N_f=3.20 \times 10^5 \),
   c.) \( \Delta \sigma/2=150 \) MPa, \( N=6.00 \times 10^8 \) and d.) \( \Delta \sigma/2=150 \) MPa, \( N_{\text{run-out}}=1.08 \times 10^{10} \) cycles.

Tables

Table 1. Chemical composition [mass-%] of the materials studied.

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<th></th>
<th>Ni</th>
<th>Cr</th>
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<th>Al</th>
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<th>Mn</th>
<th>Cu</th>
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Table 2. Mechanical properties of Nimonic 80A in the peak-aged (pa) and overaged condition (oa), Nimonic 75 and pure Nickel.

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<th>Condition</th>
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<th>Tensile strength [MPa]</th>
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Figures

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Fig. 3. Fatigue results for Nimonic 75 and pure Nickel (arrows indicate run-out specimens) in the LCF-VHCF regime.

(a) Nimonic 80A (oa) \[ \Delta \varepsilon_{pl}/2, \Delta \varepsilon_{tot}/2, \Delta \varepsilon_{pl,m} \] [\%] \( N_f = 1.57 \times 10^3 \) \( \Delta \sigma/2 = 700 \text{ MPa} \)

(b) Nimonic 80A (oa) \[ \Delta \varepsilon_{pl}/2, \Delta \varepsilon_{tot}/2 \] [\%] \( N_{\text{run-out}} = 1.00 \times 10^6, f = 5 \text{ Hz} \) \( \Delta \sigma/2 = 325 \text{ MPa} \)

Fig. 4. Cyclic deformation behaviour of the overaged condition of Nimonic 80A in the LCF (a) and in the VHCF regime (b).

Fig. 5. TEM micrographs of the microstructure of peak-aged (a) and overaged (b) condition of Nimonic 80A in the VHCF regime.
Fig. 6. Schematic representation of the dislocation/precipitate interaction of Nimonic 80A for the peak-aged (a) and overaged (b) condition in the VHCF regime.

Fig. 7. Dislocation arrangement at RT as a function of plastic strain amplitude or the number of cycles to failure and of the slip character.

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a.) $\Delta\sigma/2=280$ MPa, $N_f=2.53\times10^4$

b.) $\Delta\sigma/2=220$ MPa, $N_f=3.20\times10^5$

c.) $\Delta\sigma/2=150$ MPa, $N=6.00\times10^8$ and

d.) $\Delta\sigma/2=150$ MPa, $N_{\text{run-out}}=1.08\times10^{10}$ cycles.