Prehistory effects on the VHCF behaviour of engineering metallic materials with different strengthening mechanisms

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Abstract. Engineering materials often undergo a plastic deformation during manufacturing, hence the effect of a predeformation on the subsequent fatigue behaviour has to be considered. The effect of a prestrain on the microstructure is strongly influenced by the strengthening mechanism. Different mechanisms are relevant in the materials applied in this study: a solid-solution hardened and a precipitation-hardened nickel-base alloy and a martensite-forming metastable austenitic steel. Prehistory effects become very important, when fatigue failure at very high number of cycles (N > 10^7) is considered, since damage mechanisms occur different to those observed in the range of conventional fatigue limit. With the global strain amplitude being well below the static elastic limit, only inhomogeneously distributed local plastic deformation takes place in the very high cycle fatigue (VHCF) region. The dislocation motion during cyclic loading thus depends on the effective flow stress, which is defined by the global cyclic stress-strain relation and the local stress distribution as a consequence of the interaction between dislocations and precipitates, grain boundaries, martensite phases and micro-notches. As a consequence, no significant prehistory effect was observed for the VHCF behaviour of the solid-solution hardening alloy, while the precipitation-hardening alloy shows a perceptible prehistory dependence. In the case of the austenitic steel, strain-hardening and the volume fraction of the deformation-induced martensite dominate the fatigue behaviour.

1. Introduction
Many modern applications, such as in transportation, require fatigue lives beyond the classical fatigue limit of 10^7 cycles. Recent studies have revealed that crack initiation in the VHCF regime is initiated by other damage mechanisms than can be observed for low and high cycle fatigue behaviour (e.g. [1]). Fatigue life is dominated by crack initiation rather than crack growth [2] with local microstructural heterogeneities acting as crack nucleation sites. These heterogeneities can be defects like pores and inclusions [3] as well as precipitates or a second phase [4].

As classical damage mechanisms do not apply to the specific fatigue behaviour in the VHCF regime, the question arises whether acknowledged strengthening mechanisms, such as strain hardening, precipitation-hardening or martensite transformation, actually contribute to a cyclic strength enhancement in this regime. In the case of the nickel-base superalloy Nimonic 80A, it was demonstrated that the overaged condition shows a slightly better fatigue behaviour than the peak-aged condition [5], thus acting in contrast to its behaviour in the low cycle and high cycle fatigue regime. Moreover, those strengthening mechanisms are rarely used singularly but appear as combined processes throughout a manufacturing process, such as the prestraining of a metastable austenitic steel. In this case the pre-straining evokes a strength enhancement by means of a deformation-induced phase transformation.
from fcc austenite into bcc martensite [6] as well as strain hardening due to an increase of the dislocation density.

2. Material and experimental details
The material studied in the research work presented was chosen so as to typify different strengthening mechanisms: a solid-solution hardened nickel-base alloy Nimonic 75, a precipitation-hardened nickel-base alloy Nimonic 80A and a martensite-forming metastable austenitic steel AISI 304. Nimonic 80A was tested in the peak-aged and overaged condition (according to the heat treatments applied). Prestraining was carried out as monotonic tensile deformation by means of a servohydraulic testing machine at room temperature for the nickel-base alloys and at temperatures below room temperature for the austenitic steel.

The nickel-base alloys were prestrained up to a total strain of 8%. The martensite volume fractions (MVF) of 26% and 54% in AISI304 was gained by a plastic strain of 15% and deformation start temperatures of -20 and -70°C, thus adjusting the quasi-static mechanical properties. Prestrained specimens were electro-chemically polished after predeformation in the case of the austenitic steel and before prestraining for the nickel-base alloys. Some of the overaged specimens of Nimonic 80A were also electro-chemically polished after prestraining. Undefined micro-notches were introduced in some of the austenitic samples by soldering thermocouples to the specimens’ surfaces.

For the fatigue experiments a resonance pulsating high-frequency test system (f ~ 100 Hz), an ultrasonic fatigue testing machine (f ~ 20 kHz) and a servohydraulic test system (f ~ 760 Hz) were used. Tests were conducted under symmetrical tension-compression loading (R = -1) in ambient air. More details on the materials and the experimental techniques applied can be found in [5, 6].

3. Results and discussion
The quasi static strength of the analysed material was quantified by tensile tests. Table 1 gives an overview of the results for the yield strength and the fracture strain of the nickel-base alloys. The prestraining leads to a pronounced increase of the yield strength for the precipitation hardened nickel-base alloy. The enhancement is even more pronounced for the austenitic steel (from 284 MPa to 652 MPa for 26% MVF and even 752 MPa for 54% MVF). As can be seen from table 1 and figure 1 this strength increase comes along with a pronounced decrease of the ductility for those materials. In contrast the predeformation effect on the yield strength of the solid-solution hardening nickel-base alloy Nimonic 75 is rather small and though the ductility decreases slightly it is still quite high compared to Nimonic 80A.

Table 1. Static mechanical properties before and after prestraining (pa = peak-aged, oa = overaged).

|                  | Nimonic 80A pa | Nimonic 80A oa | Nimonic 75 |
|------------------|----------------|----------------|------------|
| prestrain 0%     | Yield Strength |                |            |
| [MPa]            | 976            | 537            | 302        |
| A [%]            | 26             | 34             | 48         |
| prestrain 8%     | Yield Strength |                |            |
| [MPa]            | 1201           | 811            | 436        |
| A [%]            | 18,5           | 26,3           | 40,2       |

* True yield strength.

Figure 1. Stress strain curves and MVF for AISI304 at different testing temperatures.
Figure 2a and 2b shows a comparison of the dislocation arrangement found in overaged Nimonic 80A in the annealed starting condition without plastic deformation and after a prestraining of 8%. The prestraining leads to a considerable increase of the dislocation density. A comparable effect of a high dislocation density due to prestraining can likewise be assumed for the other materials. In addition to strain hardening, the austenitic steel undergoes a phase transformation from austenite into ε-martensite and α’-martensite as can be seen in figure 2c. Therefore, the strengthening effect in AISI304 is a combination of strain hardening and phase transformation.

![Figure 2](image_url)

**Figure 2.** Microstructure before a) and after b) prestraining with a significant increase of dislocation density in overaged Nimonic 80A and c) martensite transformation for AISI304 after prestraining.

The prestraining not only leads to changes in the quasi-static mechanical properties but also induces a surface roughening in the tensile test specimens. For the nickel-base alloys this resulted in a change of surface roughness from $R_z = 1.4 \mu m$ to $R_z = 6.6 \mu m$. As this surface condition is representative for the major application examples, where no further surface treatment ensues the original forming process, it was maintained for the majority of fatigue specimens except for the austenitic steel and some overaged specimens of Nimonic 80A. This allows a distinction between the effect of strengthening mechanism alone and superimposed micro-notch effects on the VHCF behaviour.

Figure 3 depicts the fatigue testing results for the analysed materials without and with predeformation. The most intense influence of a prestrain can be found for the austenitic steel (Fig. 3a). The combination of strain hardening and martensite formation induces a fatigue strength enhancement of more than 200 MPa with an astonishingly small discrepancy between 26% and 54% martensite volume fraction. Hence, the major strengthening seems to be caused by strain hardening rather than phase transformation (as the change in MVF is mainly generated by different deformation starting temperatures and 15% plastic strain). A comparison of the results for the austenitic steel reveals that although the undeformed material has a much lower fatigue strength, it possesses a true fatigue limit which so far could not be proven for the specimens with 54% MVF. Moreover with an increasing amount of MVF the scatter in the fatigue life results seems to increase tremendously from 26% to 54% MVF.

The nickel-base alloys are lesser affected by a prestraining prior to fatigue testing (Fig. 3b). Nonetheless for the precipitation-hardening Nimonic 80A the fatigue strength decreases by 30-40 MPa compared to its original strength in the VHCF range for the peak-aged (pa) and overaged condition (oa). On the contrary no significant influence of prestraining can be found for the solid-solution hardening alloy Nimonic 75 as well as for those specimens of overaged Nimonic 80A which had undergone a second electro-chemical polishing.
A comparison of the different fatigue results presented in the given study shows that the strengthening mechanism alone does not define the fatigue behaviour of the materials in the VHCF range. An increase in strength does not necessarily lead to a higher fatigue limit in the VHCF regime and whether an increased dislocation density enhances the fatigue limit due to an alteration of the dislocation structure to a more stable state is also not predictable on the basis of the quasi static behaviour. What seems to underlie all of the fatigue results is a change in ductility after prestraining. Keeping in mind that the global strain amplitude during VHCF testing ranges far below the elastic limit, local stress raisers like micro-notches gain importance. This evokes the question whether not the strengthening mechanism itself than rather its effect on the notch sensitivity dominates the fatigue behaviour in the VHCF range.

The given assumption is underlined by the fact that surface polishing of the Nimonic 80A specimens after predeformation erases the prestraining effect on the fatigue behaviour. This would also explain why Nimonic 75, with its rather high ductility even after prestraining, does not show any prehistory dependence of its VHCF behaviour. Even in the case of the austenitic steel the missing true fatigue limit can be explained with an increased notch sensitivity, as some specimens with 54% martensite volume fraction failed due to crack initiation around a “stress raising” inclusion starting from the interior of the specimen, where the brittle martensite phase prevails. Moreover, tests with soldered thermocouples confirmed the high notch sensitivity of the prestrained austenitic steel and its influence on the VHCF behaviour.

4. Conclusions
Cyclic strength enhancement by means of acknowledged strengthening mechanisms does not necessarily lead to a beneficial VHCF behaviour. With the applied global strain well below the elastic limit and local plastic strain only inhomogeneously distributed the influence of micro-notches acting as local stress raisers gain importance. Hence, prehistory effects depend above all on subsequent changes in the ductility as well as the introduction of surface roughening due to forming processes.

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