INVESTIGATION OF $\gamma'/\gamma''$ LATTICE MISMATCH IN THE POLYCRYSTALLINE NICKEL-BASE SUPERALLOY IN738LC: INFLUENCE OF HEAT TREATMENT AND CREEP DEFORMATION

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Abstract—The $\gamma'/\gamma''$ lattice misfit in the polycrystalline nickel-base superalloy IN738LC has been determined by convergent beam electron diffraction (CBED) technique in a transmission electron microscope (TEM). The influence of heat treatment and creep deformation on the misfit was investigated. The chemical compositions of the $\gamma$ and $\gamma'$ phases were analysed with the help of Energy Dispersive Spectrometry (EDS). The influence of different heat treatments on the $\gamma'/\gamma''$ misfit can be explained in terms of the dependence of the chemical compositions of $\gamma$ and $\gamma'$ phases on the heat treatments. The creep deformation causes a tetragonal distortion in $\gamma'$ precipitates and leads to considerable changes in the misfit. The tetragonality in $\gamma''$ phase and the evolution of the misfit during creep process can be understood on the basis of a dislocation model. An estimation of the local internal stress in $\gamma'$ phase generated by $\gamma'/\gamma''$ interfacial dislocations formed during creep deformation shows it to be higher than the applied stress.

1. INTRODUCTION

Nickel base superalloys contain coherent precipitates of the ordered (LI$_2$) $\gamma'$ phase in the f.c.c. $\gamma$ matrix. The mismatch $\delta$ between the two lattices is defined as

$$\delta = 2(a_r - a_\gamma)/(a_r + a_\gamma)$$

where $a_r$ and $a_\gamma$ are lattice parameters of $\gamma$ matrix and $\gamma'$ phases, respectively.

The influence of $\delta$ on the growth and morphology (rafted structure) of the $\gamma'$ phase at elevated temperatures has been investigated both theoretically [1-3] and experimentally [4-8]. These investigations suggest that the magnitude and sign of $\delta$ play a major role in determining the morphology of this phase at elevated temperature under load. This parameter is therefore expected to influence the mechanical behaviour of the superalloys in the creep regime through its influence on the $\gamma'$ morphology. Such an influence has indeed been reported in the literature [9-16].

The sign and magnitude of $\delta$ is controlled by the chemical composition of the two phases [17-21] and their coefficients of thermal expansion [22-24]. In some recent investigations the influence of creep deformation on $\delta$ in the single crystal superalloy SRR99 (having negative misfit) has been reported [25-28].

The present paper is a part of an extensive research program in Berlin on gas turbine blade and disc materials [29-32]. The alloy IN738LC is a frequently used blade material in stationary gas turbines. In undeformed state it has a positive misfit.

In the present work we have measured $\delta$ in the alloy IN738LC by convergent beam electron diffraction (CBED) technique. The experiments have been conducted to study the influence of (a) heat treatment and (b) creep deformation on $\delta$.

2. EXPERIMENTAL

2.1. Materials and heat treatments

The alloy IN738LC (C-0.52, Al-7.16, Ti-4.12, Cr-17.42, Co-8.30, Ni-59.57, Ta-0.57, W-0.81, Nb-0.50, Mo-1.03 in at.%) essentially consists of coherent $\gamma'$ precipitates (ordered LI$_2$) in $\gamma$ matrix (f.c.c.). All the samples were solution treated (1473 K, 3 h/00) and then subjected to different aging treatments (Table 1) to study their influence on $\gamma'/\gamma''$ misfit. All these heat treatments produce a monomodal size distribution of $\gamma'$ precipitates. Depending on the heat treatments the mean diameter of $\gamma'$ precipitates varied between 350 and 600 nm. The volume fraction of the precipitates was around 40% [33]. No misfit dislocations at $\gamma'/\gamma''$ interfaces were observed in this alloy after any of the aging treatments.

2.2. Creep tests

All the creep tests were performed at 1223 K on specimens subjected to the standard heat treatment (see Table 1). Different creep loads and strains were employed [see Fig. 7(a)]. One specimen (No. 9) was crept until rupture and the other two (Nos 7 and 8) were interrupted in stationary creep rate regions.
Table 1. The aging treatments given to the specimens

<table>
<thead>
<tr>
<th>Specimens</th>
<th>Aging treatments</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.</td>
<td>1273 K, 120 h cooled at 15 K/h</td>
</tr>
<tr>
<td></td>
<td>to 1123 K, 16 h AC*</td>
</tr>
<tr>
<td>2.</td>
<td>1223 K, 110 h WQ</td>
</tr>
<tr>
<td>3.</td>
<td>1273 K, 24 h WQ</td>
</tr>
<tr>
<td>4.</td>
<td>1273 K, 50 h WQ</td>
</tr>
<tr>
<td>5.</td>
<td>1273 K, 140 h AC</td>
</tr>
<tr>
<td>6.</td>
<td>1273 K, 140 h AC</td>
</tr>
</tbody>
</table>

*Standard heat treatment; WQ: Water quenched; AC: Air cooled.

2.3. Analytical transmission electron microscopy (TEM)

The thin foils for TEM analysis were prepared electrochemically [31]. The creep deformed samples were sectioned both transverse and parallel to the external stress axis.

The chemical compositions of γ matrix and γ' precipitates in different specimens have been determined by energy dispersive spectrometry (EDS) at an accelerating voltage of 120 kV with a lateral resolution of about 100 nm. The EDS analysis was based on the ratio method of Cliff and Lorimer [34]. The required factors of proportionality $K_{XNi}$ ($X$: Al, Ti, Cr, Ta, Mo) relating intensity to concentration were taken from literature [33]. Only the values of $K_{CoNi}$ (0.83) and $K_{WNi}$ (2.09) were determined experimentally in this study. The latter was obtained using sulfide standards [35] and the extrapolation method [36].

The EDS analyses of different specimens were performed under controlled conditions of electron beam size, take off angle, counts/s and specimen thickness, in order to be able to compare the results from different specimens. Every experimental value represents an average of at least 10 measurements each in γ and γ' phases.

The CBED technique, as described by Ecob et al. [37, 38], has been used to determine the γ/γ' mismatch at an accelerating voltage of 120 kV with a 100 nm spatial resolution. This method is based on an analysis of the High Order Laue Zone (HOLZ) lines patterns exhibited in the transmitted beam disk. The <111> zone axis was used to record the HOLZ lines patterns. To avoid large tilt angles only those grains with zone axis near <111> were analysed by tilting them to <111> orientation.

A computer program, based on kinematical approximation, was written to simulate the HOLZ patterns and to determine the relative lattice parameters of γ and γ' phases. This program can also evaluate a tetragonal distortion by introducing a tetragonality parameter $(c - a)/a$.

By matching the experimental and the simulated HOLZ patterns, the lattice constants $(a$ and $c$) of the γ and γ' phases and the γ/γ' mismatch as defined in Sections 1 and 3, can be determined (Tables 4 and 5). All the HOLZ patterns from a given specimen were obtained on the same day under controlled experimental conditions. Therefore small day to day fluctuations in accelerating voltage, which may affect absolute values of lattice parameter measurements [37, 38], do not affect the δ values. Since these measurements do not yield absolute values of $a$ or $c$ as against the X-ray measurements [25-27], any trend in $a$ or $c$ as a function of parameters like stress or strain may not be visible although the trend in δ will indeed be seen (see Section 3.2).

The values of lattice constants $a$, $a'$, and $c'$ reported in Tables 4 and 5 were obtained from the averages of HOLZ-measurements [e.g. ratio $(A + B)/A$, see Fig. 1(b)]. At least 5-6 measurements on different γ' particles and in γ were made. The corresponding standard deviations σ are also reported in Tables 4 and 5.

The angles between the external load axis and the c-axis of the tetragonal γ' lattice formed under certain creep conditions, see Section 3.2, have been determined in longitudinal specimens by marking the load direction on the periphery of TEM specimens and transferring the same to micrographs with the help of the observed diffraction geometry.

3. RESULTS AND DISCUSSIONS

3.1. γ/γ' mismatch in undeformed samples

The chemical composition of γ' and γ phases in specimens after the different aging treatments are given in Tables 2 and 3. The values of γ/γ' mismatch after these aging treatments are given in Table 4. Figure 1(a) shows typical experimental [111] HOLZ lines patterns (in the transmitted beam discs) from the γ' phase in specimen No. 1 after the
Table 3. Compositions in at.% of γ matrix after different aging treatments

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Al</th>
<th>Ti</th>
<th>Cr</th>
<th>Co</th>
<th>Ni</th>
<th>Ta</th>
<th>W</th>
<th>Nb</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.</td>
<td>4.62</td>
<td>1.24</td>
<td>28.58</td>
<td>1.09</td>
<td>0.77</td>
<td>0.58</td>
<td>1.79</td>
<td></td>
<td></td>
</tr>
<tr>
<td>2.</td>
<td>4.65</td>
<td>1.99</td>
<td>24.90</td>
<td>9.27</td>
<td>0.53</td>
<td>0.46</td>
<td>0.14</td>
<td></td>
<td></td>
</tr>
<tr>
<td>3.</td>
<td>4.62</td>
<td>1.24</td>
<td>28.58</td>
<td>1.09</td>
<td>0.77</td>
<td>0.58</td>
<td>1.79</td>
<td></td>
<td></td>
</tr>
<tr>
<td>4.</td>
<td>4.65</td>
<td>1.99</td>
<td>24.90</td>
<td>9.27</td>
<td>0.53</td>
<td>0.46</td>
<td>0.14</td>
<td></td>
<td></td>
</tr>
<tr>
<td>5.</td>
<td>4.62</td>
<td>1.24</td>
<td>28.58</td>
<td>1.09</td>
<td>0.77</td>
<td>0.58</td>
<td>1.79</td>
<td></td>
<td></td>
</tr>
<tr>
<td>6.</td>
<td>4.65</td>
<td>1.99</td>
<td>24.90</td>
<td>9.27</td>
<td>0.53</td>
<td>0.46</td>
<td>0.14</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

σ: standard deviation.

The 3m symmetry of the [111]-HOLZ pattern is a result of cubic crystal structure of γ phase. Similar HOLZ patterns with 3m symmetry were obtained from γ and γ phases under all conditions of heat treatment.

The values in Table 4 and Fig. 2 clearly show the influence of aging treatments on the γ/γ' misfit δ. By increasing the aging temperature from 1123 to 1223 K the γ/γ' misfit δ decreases from +0.08 to +0.04%, whereas between 1223 and 1273 K the change of δ is not very obvious. The influence of aging time on the misfit is shown in Fig. 3. At the shortest aging time (about 24 h) at 1273 K, the γ/γ' misfit is very small and negative (−0.01%). As the aging time increases, the misfit also increases and appears to reach a saturation value (about +0.04% to +0.06%).

The influence of aging treatments on γ/γ' misfit can be understood by considering the compositional change in γ and γ' phases caused by the treatments. The partition ratios \(C_{\gamma}/C_{\gamma'}\) of the concentrations \(C\) of γ and γ' phases of major elements obtained from Tables 2 and 3 are plotted as a function of aging temperature in Fig. 4. By increasing the aging temperature from 1123 to 1223 K the \(C_{\gamma}/C_{\gamma'}\) ratios of γ' forming elements Ti, Al (\(C_{\gamma}/C_{\gamma'} > 1\)) clearly decrease, but from 1223 to 1273 K the change is less pronounced. The \(C_{\gamma}/C_{\gamma'}\) ratios of γ forming elements Co, Cr (\(C_{\gamma}/C_{\gamma'} < 1\)) increase only slightly with increasing temperature. The \(C_{\gamma}/C_{\gamma'}\) ratio of Ni remains almost constant. This result is similar to that reported for another superalloy [21].

Due to the small concentrations of the elements Ta, W, Nb, Mo in the γ and γ' phases (most of them are

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**Table 4. Lattice constants and γ/γ' misfits in undeformed specimens**

<table>
<thead>
<tr>
<th>Specimens</th>
<th>(a) (nm)</th>
<th>(\gamma)</th>
<th>(\gamma')</th>
<th>(\delta) (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.</td>
<td>0.3556</td>
<td>0.3559</td>
<td>+0.08</td>
<td></td>
</tr>
<tr>
<td>2.</td>
<td>0.3553</td>
<td>0.3534</td>
<td>+0.04</td>
<td></td>
</tr>
<tr>
<td>3.</td>
<td>0.3553</td>
<td>0.3535</td>
<td>−0.01</td>
<td></td>
</tr>
<tr>
<td>4.</td>
<td>0.3554</td>
<td>0.3556</td>
<td>+0.06</td>
<td></td>
</tr>
<tr>
<td>5.</td>
<td>0.3554</td>
<td>0.3556</td>
<td>+0.06</td>
<td></td>
</tr>
<tr>
<td>6.</td>
<td>0.35575</td>
<td>0.3559</td>
<td>+0.04</td>
<td></td>
</tr>
</tbody>
</table>

The standard deviation of \(a\) and \(\alpha\) is \(±0.0005\ nm\), the corresponding standard deviation of \(\delta\) is about \(±0.01\%\).
The misfits of six undeformed samples (Nos 1-6) as a function of the Al content in the γ' phase are plotted in Fig. 5. As expected from the data in literature [39], the magnitude of misfit is roughly proportional to the Al content in the γ' phase. This shows that the observed behavior in Fig. 3 is a result of the increasing Al content in the γ' phase with the aging time.

3.2. Misfit in creep-deformed samples

The CBED analyses were performed on three creep deformed samples i.e. Nos 7-9 [see Fig. 7(a)] prepared from longitudinal sections with respect to the stress axis. The samples Nos 9 and 9* were taken from regions with different distances (1.5 and 6 mm) from the fracture surface (see Table 5) and are expected to correspond to different creep states [40] as shown schematically in the creep curve in Fig. 7(a).

Figure 6(a) shows the [111] HOLZ lines patterns taken from the γ' phase in a creep deformed sample. It is to be noted that the entire pattern (all the four patterns together) shows only 1m symmetry instead of the 3m symmetry always exhibited by (111) HOLZ patterns from the γ' phase in undeformed samples. This loss of 3m symmetry in γ' phase was observed in all the crept specimens except that with the smallest creep strain (sample No. 7). Ecob et al. [37] and Porter et al. [41] have discussed the different causes for the change of 3m to 1m symmetry and have shown that artifacts can also cause this change. Since in our study the loss of 3m symmetry was never observed in the undeformed specimens, it is assumed that the present effect is caused only by deformation induced symmetry change of the γ' structure and not by the artifacts discussed by Ecob et al. [37]. Keller et al. [28] have studied the influence of changes in lattice parameters (a, b, c, α, β, γ) on the simulated HOLZ lines patterns. According to these studies the [111] HOLZ lines pattern with 1m symmetry uniquely corresponds to a tetragonal lattice. The loss of 3m
LI and WAH: LATTICE MISMATCH IN IN738LC

Fig. 6. (a) The experimental [111] HOLZ lines patterns from the γ' phase in the creep deformed sample No. 9* (see Table 5). T = 1223 K, σ = 170 MPa, creep time = 2981 h, ε = 5.2%. The pattern shows only Im symmetry. (b) Computer simulation of (a) [120 kV, ~ -0.3550 nm, (c - a)/a = +0.0026].

symmetry in ⟨111⟩ HOLZ pattern was not observed in γ matrix of any creep deformed samples.

Figure 6(b) shows a simulated [111] HOLZ lines pattern [best match with Fig. 6(a)] with a tetragonality of (c - a)/a = +0.0026.

Table 5, Lattice constants and γ/γ' misfits in creep deformed samples measured in longitudinal section with respect to load axis

<table>
<thead>
<tr>
<th>Specimens</th>
<th>c (nm)</th>
<th>a (nm)</th>
<th>(c - a)/a</th>
<th>δ (γ) (%)</th>
<th>δ (c) (%)</th>
<th>δ (a) (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>7(σ)</td>
<td>0.3551 ± 0.00015</td>
<td>0.3552 ± 0.00001</td>
<td>—</td>
<td>+0.03 ± 0.03</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>8(σ)</td>
<td>0.35485 ± 0.00005</td>
<td>0.35485 ± 0.00005</td>
<td>0.35485 ± 0.00015</td>
<td>+0.0015 ± 0.0005</td>
<td>+0.09 ± 0.05</td>
<td>-0.06 ± 0.04</td>
</tr>
<tr>
<td>9(*)</td>
<td>0.35485 ± 0.00005</td>
<td>0.35485 ± 0.00005</td>
<td>0.35485 ± 0.00015</td>
<td>0.35485 ± 0.00015</td>
<td>+0.09 ± 0.05</td>
<td>-0.06 ± 0.04</td>
</tr>
<tr>
<td>9(σ)</td>
<td>0.35480 ± 0.00005</td>
<td>0.35480 ± 0.00005</td>
<td>0.35480 ± 0.00005</td>
<td>0.35480 ± 0.00005</td>
<td>+0.09 ± 0.05</td>
<td>-0.06 ± 0.04</td>
</tr>
</tbody>
</table>

Fig. 7. (a) A schematic illustration of creep curves. The discrete points with the creep deformed sample numbers on the curves indicate the creep strain ε and time t after which the misfit was determined. (b) The misfit δc, δa, δe in creep deformed samples as a function of creep time.

By defining

\[ \delta_c = \frac{2(c_r - a_r)}{(c_r + a_r)} \]  (2)

\[ \delta_a = \frac{2(a_r - a_r)}{(a_r + a_r)} \]  (3)

where \( c_r \) is the lattice constant c of tetragonal γ' phase, the γ/γ' misfit in c and a directions can be determined.

Table 5 shows the misfit δc, δa, and δe in the creep-deformed samples. Figure 7(a) is a schematic illustration of the creep curve. The creep strains ε and times t after which the misfit was measured are shown as the discrete points with sample numbers corresponding to the creep conditions. The values of δc, δa, and δe in these deformed samples are shown in Fig. 7(b). The following points may be noted:

1. At small creep strain and time the γ/γ' misfit is small, positive and the γ' phase still has a cubic
symmetry. The value of $\delta$ is smaller than that in the initial undeformed state (see Table 4, specimen No. 1). It is assumed that this change is related to a change in the chemical composition of the $\gamma$ and $\gamma'$ phases in the early stage of creep caused by the difference in the final aging temperature 1123 K and the creep temperature (1223 K).

2. As creep time and creep strain increase, a small tetragonal distortion in the $\gamma'$ phase has occurred, leading to different values of misfit in $c$ and in $a$ directions.

3. The $\delta_c$ is always positive and increases with increasing creep time and strain and reaches a large value ($\sim +0.36\%$) at failure.

4. The $\delta_a$ decreases with increasing creep strain/time, changes sign from positive to negative at medium creep strain, and then gradually increases to a value at fracture which is about the same as the original $\delta$ value ($+ 0.06\%$).

The data points corresponding to 2% strain in Fig. 7(b) were obtained at a smaller stress of 110 MPa than the rest of the data points. The influence of different creep stresses on the misfit $\delta_c$ and $\delta_a$ has been investigated in a nickel base superalloy by Biermann et al. [26]. According to their study the difference between $\delta_c$ and $\delta_a$ increases with stress. This means that in our study at 2% strain if a stress $\sigma = 170$ MPa were to be employed, the value of $\delta_a$ would be higher and the value of $\delta_c$ would be lower than the values exhibited in Fig. 7(b). This will however not change appreciably the evolution trend of $\delta_a$ and $\delta_c$ with creep strain as derived in our study.

In the tetragonal $\gamma'$ phase the lattice constant $c$ is always larger than $a$ and the $c$ axis ([001] direction) has always the smallest angle $\theta$ to the load axis as compared with the two $a$ axes. The load axis in grains used for HOLZ-measurements was always either [113], [112] or [223].

3.3. $\gamma'/\gamma$ interfacial dislocation network and mesh size

During the creep deformation the $\gamma'/\gamma$ interfacial dislocation network and rafting of $\gamma'$ phase [see Fig. 8(a–c)] have developed. The $\gamma'$-precipitates show irregular shape and the $\gamma'/\gamma$ interfaces do not lie on well defined crystallographic planes. The TEM analyses of interfacial dislocations were performed on the samples sectioned both transverse and parallel to the load axis [42]. Only on those $\gamma'/\gamma$ interfaces which were approximately parallel to the surface of TEM-foils, the dislocations were analysed. These $\gamma'/\gamma$ interfaces have therefore been considered in the following as roughly perpendicular and parallel to the external stress and they will be referred to as $N$ and $P$ interfaces, respectively.

In the sample No. 7 [see Fig. 7(a)], corresponding to the small creep strain and cubic symmetry of $\gamma$ and $\gamma'$ phases, the dislocation network is not well developed on both $N$ and $P$ types of $\gamma'/\gamma$ interfaces. In the samples No. 8, interrupted after the medium creep strain, well developed dislocation networks have formed on both $N$ [Fig. 8(a)] and $P$ interfaces. The perfect dislocation network was maintained until fracture, as revealed by the sample No. 9 [see Fig. 8(b, c)].

On the basis of a large number of microstructural investigations of the dislocation networks at the $\gamma'/\gamma$ interfaces, it has been observed that the mesh size of dislocation network has undergone a change during
the creep process: (i) At intermediate creep strain (specimen No. 8) the dislocation networks have a relatively small mesh size on both N and P interfaces [e.g. Fig. 8(a)]. (ii) After the largest creep strain (specimen No. 9) the networks become distinctly coarser on N interfaces [compare Fig. 8(a) with (b)] and are even coarser than those on P interfaces [compare Fig. 8(b) with (c)]. In comparing Fig. 8(a) with (b) it is assumed that the slightly lower applied stress of 110 MPa (specimen No. 8) has only negligible influence on the dislocation spacing as reported by Gabb et al. [43].

The Burgers vectors and the line vectors of the network dislocations have been analysed [42]. All the dislocations were characterised as $a/2(110)$ type. In most cases the dislocation lines in networks lie along the directions parallel to the line of intersection between $\{111\}$ slip planes and $\gamma'/\gamma$ phase boundaries. No dislocations were found in the $\gamma'$ phase in the specimens Nos 7 and 8 but the dislocation pairs ($a/2(110)$ type) were observed in the $\gamma'$ phase in the fractured specimen No. 9 (see Fig. 10).

3.4. Dislocation model

In the following we present a simple dislocation model as shown in Fig. 9 to explain the above mentioned observations in creep deformed specimens. This concept essentially follows the ideas presented by Mughrabi and his co-workers in some earlier publications [25-28].

According to the results of misfit analyses reported in Section 3.2, the $c$ axis ($\langle001\rangle$ direction) has always the smallest angle to external load axis. In the following for the sake of simplicity, the load axis will be taken to be the $\langle001\rangle$ direction, as shown in Fig. 9.

At small applied loads the glide movement of dislocations in the $\gamma$ matrix channels is restricted by the $\gamma'$ precipitates, resulting in dislocation configuration shown in Fig. 9(a). The dislocation dipoles which are pressed against the $\gamma'/\gamma$ interfaces make an angle of 60° with the Burgers vector ($b = a/2[01\bar{1}]$). The other part of the dislocation which can glide within the $\gamma$ matrix channel has roughly screw character [40, 44, 45].

On N type interfaces the dislocation with 60° character have an edge component AE in $\langle 112 \rangle$ direction as shown in Fig. 9(a). This edge component can be further resolved into three $\langle 001 \rangle$ components AB, AC, AD, see Fig. 9(a). Those two $\langle 001 \rangle$ components AB, AC which are parallel to the N type interfaces influence the misfit $\delta_{e}$ and produce compressive stress components in the $\gamma'$ phase in these two $\langle 001 \rangle$ directions. Whereas the dislocations on the P type interfaces would produce tensile stress components in the $\gamma'$ phase along the $\langle 001 \rangle$ direction and influence the misfit $\delta_{c}$, see Fig. 9(b). This figure shows the 60° dislocations on parallel (111) planes presented in $[1\bar{1}0]$ projection. The internal stresses within the $\gamma'$ phase produced by the interfacial dislocations are also shown. The lengths of the arrows indicate schematically the relative magnitudes of the stresses.

The $\gamma'/\gamma$ misfit in the undeformed state of this alloy has a positive value (see Section 3.1, Table 4). This would result in compressive coherency stresses in the three $\langle 001 \rangle$ type directions in the $\gamma'$ phase. Figure 9(c) shows the stresses in the $\gamma'$ phase produced by coherency [Fig. 9(c1)] and by interfacial dislocations [Fig. 9(c2, 3)]. Superpositions of the stresses are shown in Fig. 9(c4, 5). The observed tetragonality of the $\gamma'$ phase in creep deformed specimens is a result of superposition of the stresses. The dislocations on N type interfaces increase the net compressive stress in the $\gamma'$ phase in directions perpendicular to the applied stress [Fig. 9(c4)]. These dislocations are therefore unstable. It is suggested that they migrate by glide and climb processes to adjacent P type interfaces where they relax the coherency stress. This process is shown schematically in Fig. 9(d) and is expected to be significant at large creep strains producing stress distributions shown in Fig. 9(c3, 5).

A consequence of the proposed dislocation migration is the dependence of mesh sizes of dislocation networks on the creep strain. At smaller creep strains the average mesh size on both types of interfaces should be about the same. With increasing creep strain the mesh size on N interfaces would increase and on P interfaces decrease. The TEM observations reported in Section 3.3 (see Fig. 8) are in agreement with this model.

A further consequence of the dislocation migration is that the net internal compressive stresses in the $\gamma'$ phase in directions both parallel and perpendicular to the applied stress would decrease, compare Fig. 9(c4) and (c5). With increasing dislocation density on P interfaces at large creep strain, the internal stress within the $\gamma'$ phase in the $\langle e \rangle$ direction can even change sign from initial compressive to tensile, see Fig. 9(c5). This would cause the misfit to increase in both $\langle e \rangle$ and $\langle a \rangle$ directions with increasing creep strain, as has been experimentally observed [Fig. 7(b)].

The experimental observations [HOLZ pattern symmetry and behaviour of $\delta(e)$] agree with the predictions of the present dislocation model even for specimens with $\langle 223 \rangle$ stress axis. Therefore, the approximation with respect to the direction of the load axis in this model appears to be justified. The reason for this compatible behaviour of specimens with different external stress directions is perhaps the fact that in all of them the largest component of stress was along only one $\langle 100 \rangle$ direction, the $\langle 001 \rangle$ direction ($c$ axis).

According to this model, in an alloy with an initial negative $\gamma'/\gamma$ misfit the dislocation mesh size on N interfaces after the creep deformation should be smaller than that on P interfaces. This has been confirmed by Feller-Kniepmeyer and Link [40, 46] in the alloy SRR99. Further, in this alloy the change in the sign of misfit under creep deformation should
take place in the direction parallel to the applied stress, and not perpendicular to the applied stress as is observed in the present alloy with positive initial misfit. This expected misfit behaviour has indeed been reported in SRR99 [25-27].

In this study the tetragonal distortion was not observed in the $\gamma$ matrix. The deformation induced internal distortions in the $\gamma'$ phase, because of its larger volume fraction (~60%), are obviously so small that they are not measurable with the present
In other alloy systems [25–28] the volume fraction of \( \gamma' \) (\( \sim 70\% \)) is higher than that of \( \gamma \) so that the tetragonal distortion is essentially observed in the \( \gamma \) phase.

4. ESTIMATION OF INTERNAL STRESSES GENERATED BY THE \( \gamma' / \gamma' \) INTERFACIAL DISLOCATIONS

The tetragonal distortion in the \( \gamma' \) phase and the variations of \( \gamma / \gamma' \) misfit which have occurred during creep deformation indicate that local internal stresses have developed. The magnitudes of these stresses have been evaluated for some single crystal nickel base superalloys [25–27, 47–49]. In the following we give an estimation of the local internal stresses in the polycrystalline IN738LC studied in the present work.

The state of stress in a solid is given by six components of stress tensor. In view of the observed 3m and 1m symmetries of <111> HOLZ patterns (corresponding to cubic and tetragonal structures) obtained from all the deformed specimens in the present study, we assume that the shear components of the stress tensor are negligible [28]. The axial components of the internal stresses can be derived by estimating the axial strains. Total axial strain in [001] direction \( (\epsilon_{[001]}^\prime) \) is the sum of the magnitudes of strains in \( \gamma \) and \( \gamma' \) phases in this direction and is related to the misfit as follows [25]

\[
\epsilon_{[001]}^\prime = \epsilon_{[001]}^\gamma + \epsilon_{[001]}^\gamma' = \Delta \delta_{[001]} = \Delta \delta_x. \tag{4}
\]

In the present study the value of \( \Delta \delta \) in [001] direction is maximum [Fig. 7(b)], therefore in the following we estimate the axial stress only in [001] direction. The values of misfit in undeformed and deformed (at largest creep strain) specimens are taken from Table 4, specimen No. 1 and Table 5, specimen No. 9

\[
\epsilon_{[001]} = \Delta \delta_x = +0.36\% - (+0.08\%) = +0.28\%. \tag{5}
\]

Similarly

\[
\epsilon_{[010]} = \Delta \delta_y = +0.06\% - (+0.08\%) = -0.02\%. \tag{6}
\]

According to the theory of elasticity [50], the deformation induced stress in [001] direction \( \Delta \sigma_{[001]} \) in isotropic materials is given by

\[
\Delta \sigma_{[001]} = \frac{E}{(1 + v)} (\epsilon_{[001]} + \epsilon_{[010]} + \epsilon_{[101]}) \tag{7}
\]

where \( E \) is the elastic modulus and \( v \) the Poisson's ratio.

This stress \( \Delta \sigma_{[001]} \) in equation (7) is the sum of the absolute values of deformation induced internal stresses \( \Delta \sigma_{[010]} \) and \( \Delta \sigma_{[101]} \) in the \( \gamma \) and \( \gamma' \) phases in [001] direction, respectively. The stress equilibrium in the unloaded state requires that \( \Delta \sigma \) and \( \Delta \sigma' \) obey the law

\[
\Delta \sigma \cdot f_\gamma + \Delta \sigma' \cdot f_\gamma' = 0 \tag{8}
\]

where \( f_\gamma \) and \( f_\gamma' \) (\( = 0.4 \)) are the volume fractions of \( \gamma \) and \( \gamma' \) phases, respectively.

Using the reported values of elastic modules \( E = 92 \text{ GPa} \) for a monocrystalline nickel base superalloy at 1223 K [51], and taking \( v = 0.4 \) [25], the internal stresses acting in \( \gamma \) and \( \gamma' \) phases in [001] direction can be evaluated from equations (5)–(8) as follows

\[
\Delta \sigma_{\gamma[001]} = -200 \text{ MPa}
\]

\[
\Delta \sigma_{\gamma'[001]} = +300 \text{ MPa}.
\]

This estimate is based on strains in [100] directions derived from measurement in individual grains of the polycrystalline material. In this material these strains include contributions from neighbouring grains. The estimated stress values therefore also contain the mean residual stresses arising from polycrystallinity of the present material.

Considering the large statistical deviation in \( \Delta \delta \) [see Fig. 7(b)] even higher local stresses may exist.

This simplified estimate shows that the creep deformation induced internal tensile stress within the \( \gamma' \) phase in [001] direction can reach a value of about 1.7 times the applied stress (170 MPa) at large creep strains. The effective stress (applied + internal) within the \( \gamma' \) phase in [001] direction is therefore much higher than the applied stress.

The internal stress gradually evolves during the creep process and influences both the creep rate and the deformation mechanism. As a result the creep rate should increase with strain. The substantial increase in effective stress at large creep strains appears to be at least partly responsible for the following observations made in this and similar alloys:

1. Increasing creep rate with strain until fracture in the region of large creep strain.
2. Shearing of \( \gamma' \) precipitates by dislocation pairs only at the large creep strain (in sample No. 9), as shown in Fig. 10.

Fig. 10. Dislocation pairs shearing the \( \gamma' \) precipitate at the large creep strain in sample No. 9 (\( \sigma = 170 \text{ MPa}, \text{ creep time} = 2981 \text{ h}, \epsilon = 5.2\% \)).
Pollok and Argon [47, 48] and Müller et al. [49] have made finite element analyses and TEM investigations in the single crystal superalloy CMSX-3 and SRR99. It has been shown (theoretically and experimentally) that during creep deformation the accumulated internal stress acting in non-deforming \( \gamma' \) phase can be high enough in the later stages of creep to cause shearing of the \( \gamma' \) particles by dislocation pairs.

5. CONCLUSIONS

1. The \( \gamma'/\gamma \) lattice mismatch \( \delta \) in the undeformed polycrystalline nickel-base superalloy IN738LC was determined to be between +0.04% and +0.08% depending upon the heat treatment.

2. The influence of different heat treatments on \( \delta \) is a result of the change in chemical compositions of \( \gamma \) matrix and \( \gamma' \) phase caused by the heat treatments. The \( \delta_{\text{metal}} \) is roughly proportional to the Al content in the \( \gamma' \) phase. The temperature dependence of \( \delta_{\text{metal}} \) is directly related to the temperature dependence of partition ratios \( C_{\gamma'}/C_{\gamma} \) of the major \( \gamma' \)-forming elements Al and Ti.

3. Creep deformation leads to a tetragonal distortion of the initially cubic \( \gamma' \) phase.

4. The \( c \) axis \((c/a > 1)\) in the tetragonal \( \gamma' \) phase is always in the \((001)\) direction which makes the smallest angle with the external stress axis. The misfit \( \delta \) increases with creep strain and has a positive value.

5. In the other two \((001)\) directions \((a\)-directions\) the misfit \( \delta_a \) changes sign twice with increasing creep strain.

6. The observed tetragonal distortion of the \( \gamma' \) phase and the evolution of misfit \( \delta \) during creep deformation can be interpreted with the help of a dislocation model.

7. The internal stress within the non-deforming \( \gamma' \) phase produced by \( \gamma'/\gamma \) interfacial dislocations at the largest creep strain is estimated to be approx. 1.7 times the external stress.

8. The consequences of the increasing local internal stress with strain are the shearing of the \( \gamma' \) phase and a continuously increasing creep rate at large strains. Both are observed experimentally.

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REFERENCES

42. J. Li and R. P. Wahl, to be published.
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