The role of grain size in static and cyclic deformation behaviour of a laser reversion annealed metastable austenitic steel

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Abstract
Different grain sizes were created in a metastable 17Cr-7Mn-7Ni steel by martensite-to-austenite reversion at different temperatures using a laser beam. Two fully reverted material states obtained at 990°C and 780°C exhibited average grain sizes of 7.7 and 2.7 μm, respectively. The third microstructure (610°C) consisted of grains at different stages of recrystallization and deformed austenite. A hot-pressed, coarse-grained counterpart was studied for reference. The yield and tensile strengths increased with refined grain size, maintaining reasonable elongation except for the heterogeneous microstructure. Total strain-controlled fatigue tests revealed increasing initial stress amplitudes but decreasing cyclic hardening and fatigue-induced α′-martensite formation with decreasing grain size. Fatigue life was slightly improved for the 2.7-μm grain size. Contrary, the heterogeneous microstructure yielded an inferior lifetime, especially at high strain amplitudes. Examinations of the cyclically deformed microstructure showed that the characteristic deformation band structure was less pronounced in refined grains.

KEYWORDS
cyclic hardening, fatigue life, grain size, laser annealing, metastable austenitic stainless steel, reversion

1 | INTRODUCTION

Due to their mechanical properties in terms of excellent ductility and ultimate tensile strength accompanied by an outstanding hardening capability, metastable austenitic steels have attracted high interest in the last decades. In particular, the deformation mechanisms enabling these properties, namely the formation of α′-martensite and/or twins, have been investigated in detail.1–6 A requirement to trigger these mechanisms during deformation are stacking faults originating from dissociation of regular dislocations into two Shockley partials. Thus, the stacking-fault energy (SFE) of the material which is affecting the probability of dissociation and the width of the stacking faults is a dominant parameter regarding the mechanical properties and the deformation behaviour.4,5

In addition, the grain size is quite an important factor markedly affecting the properties of austenitic steels.7–11

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Besides an increasing yield strength with decreasing grain size, Kisko et al.\textsuperscript{18} reported a reduced hardening rate during tensile deformation at grain sizes down to 1.5 μm correlating with a reduced α’-martensite formation. Yoo et al\textsuperscript{12} have also found that α’-martensite fraction after tensile fracture decreased from 60% to 25% with decreasing grain size from 2 to 0.3 μm in a 10.30Cr-8.14Ni-7.47Mn steel. This trend of a less-pronounced α’-martensite formation is reported to be reverted in case of submicron austenitic steels.\textsuperscript{8,13-16} Though, the most plausible explanations for this behaviour are not linked to the smaller grain size itself but to carbide or nitride precipitation that occurred during heat treatment\textsuperscript{13,16} or to the presence of coarse-grained and deformed retained austenite.\textsuperscript{8} Furthermore, several works on the grain-size dependence of the fatigue behaviour of austenitic steels have been published, so far. Some of them were based on stress-controlled tests,\textsuperscript{15,17-20} others based on strain-controlled tests.\textsuperscript{21-25} In materials exhibiting pronounced cyclic hardening, the control mode is an essential factor regarding the fatigue behaviour and, in particular, the fatigue life. Nevertheless, only few of the works based on strain-controlled tests\textsuperscript{21,23} include considerations on fatigue life. The work of Droste et al\textsuperscript{21} investigated the influence of an average grain size below 1 μm, whereas the study of Chlupova et al\textsuperscript{21} suffers upon a lack of data points for a persuasive lifetime estimation inhibiting a comparison between the material states. Moreover, in some works, the range of applied strain amplitudes is limited to relatively high values of Δε/2 ≥ 0.6%\textsuperscript{22} or only two different strain amplitudes have been applied at all.\textsuperscript{24,25} Therefore, the present work aims to follow up on this research by investigating a broader range of total strain amplitudes and average grain sizes.

A common method for achieving different grain sizes in metastable austenitic steels is thermo-mechanically controlled processing.\textsuperscript{7,26-28} Recently, an extensive review on the process and the related properties has been published.\textsuperscript{29} The first step in the procedure is a cold deformation to initiate the formation of α’-martensite followed by so-called reversion annealing to trigger the reversion of α’-martensite back to austenite. The resultant grain size depends on different parameters such as the degree of deformation,\textsuperscript{30,31} the volume fraction of α’-martensite, annealing temperature,\textsuperscript{26,30,32-34} heating rate\textsuperscript{34,35} and annealing time.\textsuperscript{32,34} In general, two types of heat treatments can be distinguished: First, a conventional heat treatment at relatively low temperatures for several minutes for instance performed in a furnace, and, second, a short-time heat treatment at higher temperatures for a few seconds and at high heating rates using a Gleeble simulator, for example,\textsuperscript{16,32} induction coils,\textsuperscript{36} a laser\textsuperscript{37,38} or an electron beam (EB).\textsuperscript{39} All techniques exhibit their individual advantages and disadvantages. The laser and EB technologies, for example, offer a very local heat input, but on the other hand, the one-sided application of the heat source impedes a uniform microstructure and grain size over the cross section. Consequently, previous studies\textsuperscript{37-39} reported a difference between the top and bottom surface of the steel sheets. However, in case of Heinze et al\textsuperscript{39} investigating steel with similar chemical composition as the present work, the grain size gradient over the cross section was reduced using appropriate scan strategies and EB parameters.

The present study follows up on the latter work by investigating an austenitic 17Cr-7Mn-7Ni steel but replacing the EB by a laser beam due to its wider application in industry and the circumstance that no vacuum atmosphere is needed. The steel has been cold-rolled and subsequently heated to three different temperatures resulting from different linear speeds (LSs) of the laser beam. The influence of the resulting microstructure and grain size on the fatigue behaviour was investigated by total strain-controlled fatigue tests. The presented results include (i) the laser heat treatment and resultant microstructure, (ii) the tensile properties, (iii) the cyclic deformation behaviour and α’-martensite formation, (iv) the fatigue lives and (v) the microstructure after cyclic deformation. Furthermore, the reversion annealed states are compared with those of a hot-pressed reference material of the same steel.

## 2 | EXPERIMENTAL DETAILS

The starting material was N\textsubscript{2} gas-atomized steel powder (TLS, Bitterfeld, Germany) processed by the powder metallurgy technique of hot pressing (HP). An overview of the performed production steps is shown in Figure 1. The HP was conducted under vacuum at 1,250°C for 30 min at a pressure of 30 MPa (Fraunhofer IKTS, Dresden, Germany). The heating and cooling rates during the process were 10 and 5 K/min, respectively. The received compact circular discs exhibited a diameter of 150 mm, a height of approximately 19.5 mm and a chemical composition of 17.2 wt.% Cr, 6.9 wt.% Mn, 6.5 wt.% Ni, 0.03 wt.% C, 0.08 wt.% N and 0.2 wt.% Si. The C content was determined by the combustion infrared detection technique, N by inert gas fusion infrared and thermal conductivity detection and the other elements by spark spectrometry. The calculated M\textsubscript{d30} temperature (after Nohara et al\textsuperscript{40}) and SFE (after Noh et al\textsuperscript{41}) were 17°C and 18 mJ/m\textsuperscript{2}, respectively.

Rectangular blocks with dimensions of 15 × 50 × 135 mm\textsuperscript{3} were milled out of the circular discs. They were multipass cold-rolled to a final thickness of
2.5 mm (≈83% thickness reduction) with reductions of 0.3 to 0.8 mm per pass. To avoid heating and to achieve a high amount of $\alpha'$-martensite, the sheets were cooled down to room temperature between the passes. According to Feritscope® measurements, the $\alpha'$-martensite fraction after cold rolling was about 45 ± 4 Fe-% (≈76 vol.% according to Talonen et al42).

The cold-rolled material was subsequently heat-treated by a diode pumped 4-kW Yb:YAG laser (Trumpf HDL 4002, Precitec YW50 welding head) mounted on a Motoman UP50N robot. For heating the steel sheets, a laser beam was oscillating along the rolling direction (RD) as schematically shown in Figure 2A. Variations of the maximum temperature and, thereby, the resultant grain size were achieved by using different LSs of the laser, that is, different velocities of the laser in RD: 4.5 mm/s, 6.0 mm/s and 7.0 mm/s, respectively. The other heating parameters were kept constant with a laser power set to 4 kW, a focus distance of 185.5 mm accompanied by a spot size of 19.8 mm (Gaussian beam intensity distribution) and a beam scan range of 32 mm. The different LSs were achieved by variations of the wavelength of oscillation while keeping the frequency roughly the same at about 2.1 Hz. The wavelength increased from about 2.2 mm to 2.8 mm and to 3.4 mm for LS of 4.5 mm/s, 6.0 mm/s and 7.0 mm/s, respectively. The temperature histories were measured by three K-type thermocouples attached to the bottom of the steel sheets.

After laser heat treatment, fatigue and tensile specimens were machined out of the sheets. The tensile specimens (at least two per material state) were oriented parallel to the RD (Figure 2A) and exhibited a gauge length of 25 mm at a cross section of about $6 \times 2.5 \text{ mm}^2$. Tensile tests were carried out at a constant strain rate of $5 \times 10^{-4} \text{ s}^{-1}$ using a Zwick100 test rig equipped with an extensometer. In contrast, the fatigue specimens were oriented parallel to the transverse direction, that is, perpendicular to the RD (Figure 2A), having a gauge length of 7 mm at a cross section of around $4.8 \times 2.4 \text{ mm}^2$ after polishing (Figure 2B). The polishing was done mechanically up to 3-μm polishing paste. In addition, cylindrical specimens (gauge length: ø6 mm × 14 mm) were machined out of the HPed material for reference purposes without performing the cold rolling and reversion annealing.

Single-step fatigue tests were performed in total strain-control with strain amplitudes in the range of $0.25\% \leq \Delta \varepsilon / 2 \leq 1.2\%$ using a servohydraulic MTS Landmark 250 testing system equipped with an MTS miniature extensometer exhibiting a gauge length of 5 mm. All tests were conducted under symmetrical
push-pull condition \((R_e = -1)\) following a triangular strain function with a constant strain rate of \(4 \times 10^{-3} \, \text{s}^{-1}\). Moreover, a Feritscope® (Fischerscope® MMS® PC) was attached to the specimens gauge length to record the \(\alpha’\)-martensite evolution in situ during cyclic deformation at a rate of 5 Hz. Due to the small cross section of the specimens as well as a calibration of the Feritscope® for \(\delta\)-ferrite, cf. Talonen et al,\(^42\) the readings were corrected to a value measured after failure by a MSAT magnetic saturation device equipped with a Lake Shore 480 fluxmeter.

For characterization of microstructures of the heat-treated as well as cyclically deformed specimens, examinations were performed by scanning electron microscopy (SEM) using a field-emission Tescan Mira 3 SEM operated at an acceleration voltage between 20 and 30 kV as well as by transmission electron microscopy (TEM) using a JEOL JEM-2200FS equipped with a field-emission gun operated at 200 kV. Therefore, compact specimens and TEM foils were prepared. The compact specimens for SEM were conventionally ground and polished with a final vibration polishing (SiO₂ suspension of 0.02-μm grade) for several hours. For TEM foils, thin slices were ground down to a thickness of approximately 150 μm and finally thinned by electrolytic polishing. The compact specimens were used for backscattered electron (BSE) imaging and electron backscatter diffraction (EBSD) measurements (EDAX, Ametec). Besides TEM, the foils were used for SEM investigations performed in transmission mode (t-SEM).

The EBSD measurements were used to calculate an area-weighted average grain diameter \(\bar{d}_A\) according to the following relationship:

\[
\bar{d}_A = \frac{\sum_{i=1}^{N} A_i \cdot d_i}{\sum_{i=1}^{N} A_i},
\]

where \(A_i\) and \(d_i\) are the area and diameter of the \(i\)th grain, respectively.

### 3 | RESULTS

#### 3.1 | Laser reversion heat treatment

The steel sheets cold-rolled up to 83% of thickness reduction contained a volume fraction of deformation-induced \(\alpha’\)-martensite of about 45 ± 4 Fe-% (≈76 vol.%) measured by Feritescope®. As shown in previous investigations on cold-rolled steels with comparable chemical composition,\(^7\) the microstructure is heavily deformed, contains shear bands and is characterized by a strong rolling texture typical for bcc materials (see Weidner et al\(^43\)).

The three different LSs led to significant differences in the peak temperatures reached during the laser reversion annealing. The temperature profiles recorded by three thermocouples attached to the bottom of the steel sheets are shown in Figure 3. The thermocouples were located at different positions (Figure 2A), which explains the time lag between the individual signals. However, all profiles exhibit a short heating-up stage before reaching the peak temperature within a few seconds. As the heat source was a continuously moving laser beam, no holding time was present. Thus, the cooling started right after reaching peak temperature. Cooling occurred under air, which led to a relatively slow cooling rate as evident from Figure 3.

Comparing the different laser heat treatments, the slowest LS with 4.5 mm/s as expected led to the highest peak temperatures (Figure 3A) and vice versa (Figure 3C). As all other beam parameters were kept constant, the heat input is the higher the slower the laser beam speed is along the steel sheet.\(^37\) The laser heat treatments at \(LS = 4.5 \, \text{mm/s}\) and \(LS = 6.0 \, \text{mm/s}\) exhibit average peak temperatures of 989°C and 779°C, respectively, both with a relatively small standard deviation of about 5°C (see Table S1 for listing of all recorded peak temperatures). In contrast, in the case of \(LS = 7.0 \, \text{mm/s}\), an average peak temperature of 608°C was recorded with a high standard deviation of about 44°C. This high scatter for the highest LS probably stems from a more sensitive influence of the exact position of the thermocouple in relation to the oscillation of the laser beam as

![Figure 3](https://wileyonlinelibrary.com/)

**Figure 3** Signals of the thermocouples attached to the bottom of the steel sheets for the laser heat treatments at the linear speeds (LSs) of (A) \(LS = 4.5 \, \text{mm/s}\), (B) \(LS = 6.0 \, \text{mm/s}\) and (C) \(LS = 7.0 \, \text{mm/s}\) [Colour figure can be viewed at wileyonlinelibrary.com]
the wavelength of the laser’s path on the material slightly increases from 2.8 mm for LS = 6.0 mm/s to 3.4 mm for LS = 7.0 mm/s. However, the different material states will be designated according to the type of heat treatment (LR: laser reversion) and the peak temperatures hereinafter in this paper: LR-990, LR-780 and LR-610.

3.2 | Microstructure after reversion annealing

The microstructures resulting from the two laser reversion treatments LR-990 and LR-780, respectively, are shown in Figure 4 in terms of EBSD measurement results in comparison with the coarse-grained HP counterpart. The material LR-610 is not included due to an incompletely reverted microstructure still containing heavily deformed austenitic grains. Thus, Figure 4A-C shows the band contrast (BC) maps with high-angle grain boundaries (HAGBs) and twin boundaries marked by blue and red lines, respectively. Figure 4D-F represents the corresponding orientation maps shown in colour code of inverse pole figures related to the normal direction. The used sizes of area of interest as well as the step size for the EBSD measurements had to be adjusted to the different grain sizes resulting from the laser reversion treatment (please note the different scale bars). The corresponding grain size distributions and the misorientation distributions are plotted in Figures 4G,H, respectively. As indicated by the data points of the grain size distributions (Figure 4G) an equal number of 16 classes has been used for each material state. To cover the

![Figure 4](image-url)

**Figure 4** Electron backscatter diffraction (EBSD) band contrast (A-C) with high-angle grain boundaries (HAGBs) indicated by blue lines and twin boundaries (TBs) by red lines and corresponding orientation maps (D-F) of the reversion annealed states LR-990 (A and D) and LR-780 (B and E) and the hot pressing (HP) state (C and F). (G) Grain size distributions in terms of the area fraction plotted versus grain size and (H) misorientation distributions for misorientations in the range of 5° to 65° [Colour figure can be viewed at wileyonlinelibrary.com]
individual grain size ranges, the class width was varied from about 0.56 μm to 1.34 μm and to 11.7 μm for the LR-780, LR-990 and HP state, respectively.

First of all, the microstructures of LR-990 and LR-780 are fully austenitic without retained α'-martensite. It is obvious that the grain sizes are strongly dependent on the peak temperature during laser heat treatment. Thus, the average grain size of 2.7 μm obtained for LR-780 is significantly smaller compared with 7.7 μm for LR-990. Although grains finer than 3 μm dominate the microstructure of the LR-780 state, a few large grains with a diameter up to 8 μm are visible as well (Figures 4B,G). All grain-size distributions are close to a logarithmic normal distribution function (Figure 4G). Furthermore, for all material states, only few low-angle grain boundaries with misorientations of 5° to 15° were detected (Figure 4H). Instead, the majority of grain boundaries had a misorientation angle greater than 15°, exhibiting a maximum between 53° to 57° (Figure 4H). In addition, a significant number of twin boundaries (60°) was detected.

The determination of the average grain size of the LR-610 state is more complicated, since the microstructure was inhomogeneous, as shown in Figure 5 by t-SEM micrographs. Severely deformed areas (Figure 5A,B marked by DA) were observed alongside with mostly submicron grains (Figure 5D) and individual grains exhibiting diameters up to almost 2 μm (Figure 5C). As MSAT measurements of the LR-610 state revealed a ferromagnetic phase fraction below 1 vol.%, the different microstructural features were most probably of austenitic nature.

The largest average grain size of 62 μm was measured for the HP reference material which was not laser heat-treated (Figure 4G). An overview of all average grain sizes is provided by Table S2.

### 3.3 | Tensile properties

The engineering stress-strain curves for the three laser-treated states as well as for the reference material obtained by quasi-static tensile tests are plotted in Figure 6A. In addition, Table 1 provides an overview of the mechanical properties. Significant differences regarding strength and ductility were observed between the material states. With increasing LS during laser heat treatment, that is, with decreasing grain size, the yield strength increased from 359 MPa to 520 MPa and finally to 1,125 MPa for the LR-990, LR-780 and LR-610 states, respectively. In this study, the yield strength $\sigma_y$ was specified as the stress at 0.5% plastic strain due to early microplasticity effects in the material. The increasing yield strength with decreasing grain size is a well-known phenomenon and described by the Hall-Petch relationship

$$\sigma_y = \sigma_0 + k \frac{1}{\sqrt{d}},$$

(2)
where $\sigma_0$ is the friction stress, $k$ is the Hall-Petch constant and $d$ is the grain size, which in this study is equal to the area-weighted value $\bar{d}_A$. The values for the Hall-Petch parameters were determined for the present states LR-990, LR-780 and HP as well as additional data for other conditions of the same steel alloy taken from literature (Figure 6B). The fit reveals a Hall-Petch constant of $k = 0.599 \text{ MPa}\sqrt{\text{m}}$ and a friction stress of $\sigma_0 = 167 \text{ MPa}$ at a determination coefficient of $R^2 = 0.94$.

Figure 6A shows that the yield strength of the LR-610 state is by far the highest of all studied materials. Using the determined values of the parameters and the yield strength of the LR-610 state, an apparent average grain size of 0.4 $\mu$m can be estimated for the LR-610 state. However, it is evident that the grain size is not the only factor affecting the strength since deformed grains were also present, that is, dislocation strengthening contributes, too.

The HP material exhibited the lowest yield strength of 267 MPa due to its largest grain size, as seen from Figure 6B. Thus, cold rolling and subsequent laser reversion treatment is effective to enhance the yield strength significantly.

The reversed states LR-990 and LR-780 and the HP reference exhibited a pronounced work hardening enabling very high uniform elongation of 64% and 39%, respectively, showing that ductility remains good in spite of enhanced strength (Figure 6A). In contrast, the LR-610 state revealed very limited uniform elongation as necking started shortly after loading exceeded the yield strength.

### 3.4 Cyclic deformation behaviour

The cyclic deformation curves of the different material states are shown in Figure 7. Qualitatively, all material states except the LR-610 exhibited similar cyclic deformation curves. At low and medium strain amplitudes, an initial very slight cyclic hardening occurred followed by a stage of cyclic softening.

In the case of the LR-780 state, the cyclic softening continued until failure for the low strain amplitudes $\Delta\varepsilon_t/2 \leq 0.4\%$. At higher strain amplitudes as well as for all tests of the LR-990 and HP states, secondary cyclic hardening was observed after an incubation period dependent on material state and strain amplitude. Its onset of cyclic hardening was shifted to a lower number of cycles with increasing strain amplitude. In addition, its amount increased significantly with increasing total strain amplitude.

The cyclic deformation behaviour of the LR-610 state shown in Figure 7C is completely different. First of all, the initial stress amplitudes are significantly higher, starting at about 540 MPa for the smallest strain
amplitude and reaching 1,180 MPa for $\Delta \varepsilon_t/2 = 0.8\%$. The highest strain amplitude of $\Delta \varepsilon_t/2 = 1.2\%$, on the other hand, exhibited a lower value than the latter. This is attributed to a relatively high scatter similar to the standard deviations of the yield strength (Table 1). This scatter probably originated from the inhomogeneous microstructure after laser heat treatment, cf. Section 3.2. Due to the relatively small volume of the gauge length, the position of the individual specimens in the steel sheet (Figure 2A) can be supposed to affect the mechanical properties measured. However, according to the stress-strain hysteresis loops, no necking occurred, neither at $\Delta \varepsilon_t/2 = 1.2\%$ nor at $\Delta \varepsilon_t/2 = 0.8\%$. Moreover, at high strain amplitudes $\Delta \varepsilon_t/2 \geq 0.6\%$, a cyclic softening occurred in the LR-610 state which is attributed to the deformed austenite found in the microstructure. The increasing plasticity enables these areas to soften in terms of annihilation of dislocations, causing a decrease in the dislocation density. Whereas the cyclic softening continued until failure at the strain amplitudes of $\Delta \varepsilon_t/2 = 0.6\%$ and $\Delta \varepsilon_t/2 = 0.8\%$, cyclic hardening began for $\Delta \varepsilon_t/2 = 1.2\%$ after about 30 cycles. In contrast, for medium and small strain amplitudes $\Delta \varepsilon_t/2 \leq 0.5\%$, a relatively stable cyclic behaviour in terms of constant stress amplitudes was observed.

It is commonly agreed that the secondary cyclic hardening, as observed for LR-990, LR-780 and HP, is caused by the fatigue-induced $\alpha'$-martensite formation. Alongside with the material state and applied strain amplitude, there is another parameter that affects the amount of $\alpha'$-martensite formed during cyclic loading—the cumulated plastic strain $\lambda_p$. This can be calculated by the following equation:

$$\lambda_p = 4 \cdot \sum_{i=1}^{N_i} (\Delta \varepsilon_{pl}/2)_i,$$

where $N_i$ is the $i$th cycle and $(\Delta \varepsilon_{pl}/2)_i$ is the plastic strain amplitude of the $i$th cycle.

The $\alpha'$-martensite evolution as a function of the cumulated plastic strain $\lambda_p$ is shown in Figure 8A for the two laser-treated material states LR-990 and LR-780 in comparison with the coarse-grained reference material (HP) for two total strain amplitudes of $\Delta \varepsilon_t/2 = 0.8\%$ and $\Delta \varepsilon_t/2 = 0.5\%$. At both amplitudes, the threshold value $\lambda_{p,th}$, that is, the value of cumulated plastic strain at which the $\alpha'$-martensite formation sets in, increases in the sequence HP, LR-990 and LR-780 states (although the signal of the latter at $\Delta \varepsilon_t/2 = 0.5\%$ is probably affected by a slight misalignment of the Feritscope).

Furthermore, a significant difference in the final volume fraction of fatigue-induced $\alpha'$-martensite fractions is observed for the three laser-treated material states as shown in Figure 8B. Thus, the largest amount of $\alpha'$-martensite has been developed in the coarse-grained microstructure of the HP reference state. The volume fractions of $\alpha'$-martensite of the three reversion-annealed
states are significantly smaller and decrease in the sequence LR-990, LR-780 and LR-610.

3.5 | Microstructure after cyclic deformation

The microstructure developed during cyclic loading shows significant differences depending on the austenitic grain size. Figure 9 shows results of EBSD measurements on specimens after cyclic loading at a total strain amplitude of Δε_t/2 = 0.3% for the HP reference state (A-C) and the two laser-treated states LR-990 (Figures 9D-F) and LR-780 (Figures 9G-I), respectively. Thus, Figure 9 shows for all three states the BC maps with indicated misorientations higher than 15° (blue) and twin boundaries (red) (Figure 9A,D,G), BC maps with overlayed phase maps of ε-martensite (yellow) and α'-martensite (blue) (Figure 9B,E,H) and, finally, the BC map overlayed with the crystallographic orientation map of the α'-martensite (Figure 9C,F,I).

The microstructure of the HP reference state is characterized by the formation of deformation bands on different slip systems consisting of ε-martensite and α'-martensite grains. According to the EBSD orientation map in Figure 9C, just a few α'-martensite variants had developed inside the grain despite a relatively coarse grain size. In some locations, the variants are arranged as twinned α'-martensite as indicated by ∑3 twin boundaries marked in red in Figure 9A. This microstructure is well known for coarse-grained microstructures of steels with comparable chemical composition after cyclic deformation.44,48

The microstructure of the LR-990 state exhibits a significantly smaller grain size compared with the HP state. A representative grain after cyclic deformation is shown in Figures 9D-F. The grain exhibits a high volume fraction of ε-martensite, which makes its band-like structure less obvious (Figure 9E). Though, taking also into account the arrangement of the α'-martensite islands formed within the grain, a pronounced band-like structure mostly limited to the primary slip system is still observable (Figure 9F). Like in the HP state, the islands preferentially arose in pairs or clusters exhibiting a misorientation of about 60° (Figure 9F; twin boundaries marked in red in Figure 9D) indicating twinned α'-martensite in order to minimizing the stress caused by the phase transformation inherent shear.

Figures 9G-I shows the microstructure of the LR-780 state after cyclic deformation. The depicted grain only contains a small fraction of austenite but is almost fully indexed as ε-martensite (Figure 9H). Neither the latter nor the fatigue-induced α'-martensite appear in a band-like structure. Though, the α'-martensite islands do not form apart and independently from one another but once again appear in pairs or clusters separated by ∑3 twin boundaries (Figure 9G, twin boundaries marked by red lines).

In contrast, a smaller grain existing in the LR-780 state cyclically deformed at a strain amplitude of Δε_t/2 = 0.6% (Figure 10, marked by a white dotted line in Figure 10A) exhibited even bigger differences compared with a coarse-grained microstructure. Please note that the results reported next have been observed repeatedly in such small grains of the LR-780 state. First of all, according to the EBSD phase map of Figure 10B, no remaining austenite was indexed within the grain which therefore only contains ε- and α'-martensite. The number and the size of the α'-martensite islands were decreased. Furthermore, no twinned α'-martensite was observed. Although two adjacent laths are observed in the lower part of the grain (Figure 10C, marked by a dashed white circle); the α'-martensite islands are randomly distributed and seem to form independently from each other. However, the islands are supposed to grow rapidly until the whole grain is martensitic (cf.21), like it apparently happened in the grain in the upper right corner of Figures 10A-C. All observed islands in such small grains start or end at a grain boundary or other interfaces.
Due to the relatively large areas solely indexed as $\varepsilon$-martensite by EBSD, conventional and high-resolution (HR) TEM investigations were performed to have a closer look on the structure of the formed $\varepsilon$-martensite, cf. Figure 11. For a ‘perfect’ $\varepsilon$-martensite, the stacking faults have to be arranged on every second {1 1 1} plane, changing the stacking sequence ABCABC characteristic for the fcc crystal to ABAB for the hexagonal structure. As expected, no strictly periodic stacking was found for entire grains. Nevertheless, HR-TEM and corresponding fast Fourier transformation (FFT) images revealed relatively broad hexagonal $\varepsilon$-martensite bands (Figure 11A, pattern II) with a periodic ABAB stacking embedded in deformed austenite. In the present case, the hexagonal band has a thickness of about 28 nm. The austenite was present in form of highly faulted areas with a high density of stacking faults but in contrast to the $\varepsilon$-martensite band without a strictly periodic stacking on every second {1 1 1} plane (Figure 11A, patterns I and III). Figure 11B shows areas of alternating $\varepsilon$-martensite and faulted austenite at lower magnification. The stacking faults start/end at a twin boundary which is supposed to stem from reversion annealing. A selected area electron diffraction (SAED) pattern (Figure 11B, pattern VI) analysing an about 13-nm-thick $\varepsilon$-martensite band and the surrounding austenite reveals the Mangonon-Thomas orientation relationship between those two:
All ε-martensite bands underneath the twin seem to start/end at the twin boundary in Figure 11B. Above the upper twin boundaries, a markedly darker area due to a contrast change is observed. However, ε-martensite bands proceed through this area, suggesting that it is not α’-martensite. Instead, this contrast change is attributed to a change in the diffraction conditions due to stress fields at the twin boundary originating from cyclic deformation. The brightness variations and differences close to the twin boundary in Figure 11A may result from such stress fields as well.

3.6 | Fatigue life

The fatigue life with respect to the applied total strain amplitude was analysed according to the relationship of Basquin and Manson-Coffin

$$\frac{\Delta \varepsilon_t}{2} = \frac{\sigma_f'}{E} \left(2N_f\right)^b + \varepsilon_f' \left(2N_f\right)^c,$$

where $\sigma_f'$ represents the fatigue strength coefficient, $E$ the Young’s modulus, $b$ the fatigue strength exponent, $\varepsilon_f'$ the
fatigue ductility coefficient and $c$ the fatigue ductility exponent. The values for the parameters are given in Table S3.

It is seen from Figure 12 that the total strain-based fatigue life curves for the LR-990 and HP states are almost identical with negligible difference at the lowest applied strain amplitude. The fatigue life of the LR-780 state coincides with that of the former two states at high strain amplitudes but diverges at small strain amplitudes in terms of a slightly higher fatigue life. Thus, the LR-780 state exhibits the highest lifetime at $\Delta \varepsilon_t/2 = 0.3\%$. In contrast, the fatigue life of the LR-610 state is shorter than that of all other material states. The difference is most significant at $\Delta \varepsilon_t/2 = 1.2\%$ and decreases with decreasing strain amplitude. At $\Delta \varepsilon_t/2 = 0.3\%$, the lifetime curve even converges with the fatigue life of the LR-990 and HP states.

4 | DISCUSSION

4.1 | Influence of temperature on reversion behaviour

During cold rolling of metastable austenitic stainless steels at room temperature, a significant amount of deformation-induced $\alpha'$-martensite is introduced to the material depending on the degree of thickness reduction and the chemical composition. The deformation-induced $\alpha'$-martensite is known to appear in two different morphologies: (i) the lath-type martensite and (ii) the dislocation cell-type martensite.$^{30,31}$ Furthermore, it is known from literature that the lath-type $\alpha'$-martensite transforms to the dislocation cell-type $\alpha'$-martensite due to fractionizing and, consequently, refining of the laths$^{31}$ with an increase in cold rolling reduction. Although no TEM investigations were performed on the cold-rolled material in the present study, it is assumed that regarding the relatively high thickness reduction of 83% and the fine microstructure represented in Weidner et al.$^7$ for steel with comparable chemical composition, but still 24% of retained austenite, a mixture of both types of martensite is present in the microstructure.

Two mechanisms of reversion of deformation-induced $\alpha'$-martensite back into austenite are reported for metastable austenitic stainless steels$^{32}$: (i) the diffusional reversion mechanism and (ii) the shear reversion mechanism. Whereas diffusional reversion is characterized by a nucleation-growth process,$^{16,34}$ shear reversion includes three successive steps$^{32,34,51}$: (i) martensitic phase transformation of $\alpha'$-martensite to austenite of high dislocation density, (ii) formation of recovered austenite with defect-free subgrains and (iii) coalescence of subgrains to grains by continuous recrystallization. The most important factors determining the reversion type are probably the chemical composition, the annealing temperature and the heating rate.$^{34}$ For an entire reversion to austenite by the shear reversion mechanism, a critical driving force of the Gibbs free energy of $\Delta G^{\alpha'\rightarrow\gamma} = -500 \text{ J/mol}$ is required according to Tomimura et al.$^{34}$ For Cr-Ni steels, $\Delta G^{\alpha'\rightarrow\gamma}$ can be calculated as a function of the annealing temperature$^{34}$ as follows:

$$\Delta G^{\alpha'\rightarrow\gamma}(J/mol) = 10^{-2} \Delta G_{Fe}^{\alpha'\rightarrow\gamma}(100 - Cr - Ni) - 97.5 Cr + 2.02 Cr^2 - 108.8 Ni + 0.52 Ni^2 - 0.05 CrNi + 10^{-3} T(73.3 Cr - 0.67 Cr^2 + 50.2 N - 0.84 Ni^2 - 1.51 Cr Ni),$$

(5)

where $\Delta G_{Fe}^{\alpha'\rightarrow\gamma}$ is the free energy difference in pure iron, $T$ is the temperature in Kelvin and $Cr$ and $Ni$ are the chemical compositions of chromium and nickel in wt.%. For austenitic steels containing additional alloying elements such as carbon, nitrogen and manganese, Somani et al.$^{16}$ propose to replace $Cr$ and $Ni$ in equation 5 by their equivalents calculated according to the following:

$$Ni_{eq} = Ni + 0.6 Mn + 20 C + 4 N - 0.4 Si$$

(6)

and

$$Cr_{eq} = Cr + 4.5 Mo.$$  

(7)

The resulting diagram for the studied alloy is shown in Figure 13. The critical driving force of $-500 \text{ J/mol}$ is reached at 544°C. With increasing temperature, the free energy further proceeds towards a minimum of
finishing shear reversion is exceeded in heating above 522°C. All temperatures reached during laser reversion treatment were well above this temperature even for the fastest LS LR-610. Thus, it is supposed that the applied laser heating results in the shear reversion mechanism, which is also facilitated by the relatively high heating rates of approximately 200 K/s.34,35 Here, it has to be noted that in context with Misra et al32,51 that only the first step of the reversion—the phase transformation of α′-martensite to austenite of high dislocation density—is completed at 522°C during the fast heating. In the following stages, the recovery of austenite to defect-free subgrains and continuous recrystallization are dependent on time16,32 as well as on temperature.16,34

It can be supposed that the driving force for the phase reversion transformation of deformation-induced α′-martensite to reverted austenite is independent of the degree of cold deformation.35 However, the degree of cold deformation has an influence on the amount and type of deformation-induced α′-martensite and presumably also on the kinetics of recovery and recrystallization and the resulting homogeneity and grain size of the reverted microstructure.7,16

4.2 Influence of temperature on the grain size and microstructure of reverted austenite

The different LSs result in different annealing temperatures, as shown in Section 3.1. Consequently, the microstructures of the reverted material states are significantly different regarding the grain size and the homogeneity. The obtained grain size decreases with decreasing annealing temperature, which is in good agreement with similar investigations on an AISI 301LN steel performed by Järvenpää et al.38 Thus, the state LR-990 exhibits a grain size of 7.7 μm in comparison with 2.7 μm for the state LR-780. The homogeneity of the microstructure over the entire cross section of the sheet was investigated as well (the corresponding EBSD maps are not shown here in detail). For the state LR-780 with the smallest grain size, scans were performed in the ND-RD plane close to the top and bottom surfaces and revealed at both places fully reverted austenitic grains with average grain sizes of 2.9 μm ± 1.6 μm and 2.6 μm ± 1.3 μm, respectively. Thus, neither nonreverted α′-martensite nor a significantly heterogeneous grain size distribution over the cross section was revealed. This is in contradiction to former reports on the microstructure after heat treatment by the one-sided application of either an EB39 or a laser beam.37,38 The investigations on EB39 reversion treatment were based on steel with comparable chemical

![Gibbs free energy ΔGα α γ plotted versus temperature. The plots are either modelled with Thermo-Calc or calculated after equation 5 using Cr and Ni equivalents after Somani et al16 [Colour figure can be viewed at wileyonlinelibrary.com]](image1.png)
composition and sheet thickness (3 mm) but a lower thickness reduction of 70% and an \( \alpha' \)-martensite fraction of 51 vol.% (cf.,\(^7\) same material as\(^{39}\)). The results revealed a heterogeneous microstructure over the cross section, which can be related to the completely different scan strategies used for the EB. Compared with the present investigations, Heinze et al\(^{39}\) observed mostly higher peak temperatures measured at the top surface by a pyrometer, but the cooling rates were also higher. Thus, the temperature time histories were considerably different compared with the present heat treatments (Figure 3), and the 'holding time' above a specific temperature, for example, 600°C, was significantly shorter. Accordingly, the time for the heat transfer from top to bottom was reduced probably causing a more pronounced temperature and, thereby, microstructure gradient over the cross section.

The dissenting results of a relatively homogeneous cross section of the present material in comparison with the other studies using a laser beam\(^{37,38}\) could be due to different thickness reductions, different \( \alpha' \)-martensite fractions after cold rolling and different beam parameters. However, the most important difference between investigations in previous studies\(^{37,38}\) seems to be the chemical composition of the studied steel resulting in different reversion mechanism. The reversion behaviour was investigated on steel AISI 301LN, which is known to differ from other results on shear reverted austenitic steels. In contrast to long-term conventional reversion heat treatment in a furnace, the one-sided application of high-energy heat sources such as an electron or laser beam needs an appropriate scan strategy to provide a homogeneous temperature field. In particular, similar peak temperatures are supposed to facilitate a homogeneous microstructure over the cross section. In the present study, the combination of beam scan range (32 mm), spot size with a Gaussian beam intensity distribution (19.8 mm), wavelength (2.2 mm and 2.8 mm) and LS (4.5 mm/s and 6.0 mm/s) apparently meets these conditions. The heat transfer from top to bottom over the sheet thickness of 2.5 mm is fast enough to enable a relatively homogenous temperature field within the reversion-relevant temperature range and, thereby, a completely reverted austenitic microstructure with a relatively homogenous grain size over the cross section.

However, in case of LR-610, a quite heterogeneous microstructure was obtained (cf. Figure 5) consisting of fully reverted austenitic grains with sizes of about 2 \( \mu \)m and heavily distorted austenitic grains marked by DA in Figure 5. These different microstructural features are most probably the results of different completion of individual steps of the shear reversion process. However, it has to be considered that after the completion of the annealing, it is difficult to distinguish between the shear-reverted austenite passed only step (i) of the reversion process and retained austenite not transformed to deformation-induced \( \alpha' \)-martensite during cold rolling. Both are characterized by a quite a high dislocation density, providing the driving force for subsequent recrystallization. The areas marked by DA in Figure 5 are supposed to mostly consist of retained deformed austenite which did not recrystallize due to the relatively low temperature during laser heat treatment at LS = 7.0 (state LR-610). However, the presence of some shear-reverted austenite which had just passed step (i) cannot be ruled out as well as steps (ii) and (iii) of shear reversion are time-dependent.\(^{16,32}\) The relatively small grains shown in Figure 5B,C marked by SG are ascribed to the formation of defect-free subgrains inside the reverted austenite, that is, step (ii) of the reversion process. Finally, the larger grains of about 2 \( \mu \)m (see Figure 5B,C) are assumed to be the result of the coalescence of such subgrains during step (iii). Furthermore, the presence of stacking faults observed in some grains (Figure 5C) are in accordance with other results on shear reverted austenitic steels reported in literature.\(^{32}\)

### 4.3 Influence of grain size of reverted austenite on fatigue behaviour

It has been shown in Section 3.4 that the cyclic deformation behaviour varies significantly among the reversion-treated material states with different LS as well as for the coarse-grained counterpart of the HP material. Thus, all material states except LR-610 exhibit a pronounced secondary hardening for total strain amplitudes above 0.5%. Among these three states, the coarse-grained HP material shows the lowest initial stress amplitudes caused by the larger grain size (62 \( \mu \)m) and the highest total volume fraction of fatigue-induced \( \alpha' \)-martensite after cyclic deformation. Contrary, the state LR-780 with the smallest mean grain size of 2.7 \( \mu \)m exhibits the highest initial stress amplitudes and the lowest volume fraction of \( \alpha' \)-martensite. Evidently, initial stress amplitudes increase whereas the total volume fraction of \( \alpha' \)-martensite decreases with a decrease in grain size. The exception is the state LR-610 revealing a very inhomogeneous microstructure. This microstructure yields the highest initial stress amplitudes, the lowest volume fraction of \( \alpha' \)-martensite and the highest scatter of data. The stress-strain curve reveals significantly higher yield strength and low ductility.
The tendency of decreasing α’-martensite formation with decreasing grain size was repeatedly reported before, for example, other studies.\textsuperscript{12,54–58} For medium and high Mn steels, some authors\textsuperscript{54–56} have proposed an increasing apparent SFE with decreasing grain size which would explain the less pronounced α’-martensite formation. Other works\textsuperscript{12,57,58} postulated an increasing austenite stability with decreasing grain size to explain the grain size-dependent α’-martensite evolution in tensile deformation. Contradicting results were reported by Matsuoka et al\textsuperscript{59} who observed that the mechanical stability of austenite is independent on the grain size. However, the present results confirm different α’-martensite evolutions (Figure 8A) as well as different α’-martensite fractions after cyclic deformation (Figure 8B) in dependence on the austenitic grain size, that is, a distinct grain-size dependence of the α’-martensite formation is observed.

Not only the phase fractions but also the microstructures and, thereby, the nucleation and arrangement of α’-martensite islands change with decreasing grain size. Previous investigations on coarse-grained material of comparable chemical composition have shown that fatigue-induced α’-martensite is formed inside deformation bands crossing the austenitic grains.\textsuperscript{44,60} In situ investigations by Weidner et al\textsuperscript{60} revealed the formation of deformation bands containing high density of stacking faults in quite an early stage of cyclic deformation. These bands grew both in number as well as in length and thickness, respectively, with an increase in number of cycles. With ongoing cyclic deformation, the increasing number of deformation bands leads to a fragmentation of the austenitic grains causing a reduction of the mean free path of perfect and partial dislocations. Within the deformation bands, the formation of α’-martensite grains also yields a reduction of the mean free path for the movement of dislocations. This, in particular, significantly contributes to the pronounced secondary hardening, as shown in previous studies.\textsuperscript{3,61} The reduction in grain size is accompanied by the reduction of the mean free path of dislocations even without the presence of stacking faults, the formation of deformation bands or α’-martensite. Although the mean free path L was not experimentally evaluated in this study, the following theoretical experiment can be carried out, cf. Figure 14. Assuming that the mean free path for dislocations in this kind of steel is $L = 2.7 \, \mu m$, it is in the same dimension as the average grain size of state LR-780, whereas for the state LR-990, the grain size $d_A = 7.7 \, \mu m$ is about three times larger, and for the HP state with $d_A = 62 \, \mu m$, even more than 20 times larger. This would be even more pronounced for smaller mean free path $L$. This, finally, will result in change of the microstructure developing during cyclic loading from coarse/fine-grained (100 µm-10 µm) to microcrystalline (10 µm-100 µm) and ultrafine (<1 µm) microstructures. This is what is observed and well-known, for example, for fcc materials such as nickel.\textsuperscript{52} Whereas coarse/fine-grained nickel during cyclic loading develops characteristic dislocation arrangements such as vein structure and persistent slips bands (dependent on the grain orientation), the microcrystalline nickel still shows dislocation patterning but with a less pronounced orientation-dependence. In contrast, the submicrocrystalline nickel does not exhibit any dislocation patterning.\textsuperscript{62}

The change in the microstructural characteristics after cyclic deformation in dependence of austenitic grain size is schematically illustrated in Figure 14. In case of the fine-grained LR-990, the cyclically deformed microstructure even in smaller grains still consists of certain amount of deformation bands containing α’-martensite islands. Though, deformation bands on secondary glide systems form less frequently. With a further decrease in grain size, the deformation band structure is less obvious and it becomes more difficult to distinguish individual deformation bands (cf. LR-780 in Figure 9G-I). In submicron grains (<1 µm, cf. Figure 10), no well-defined deformation bands are observed at all, as already reported for stress-controlled fatigue tests.\textsuperscript{15} Instead, whole grains contain a high density of defects in form of ε-martensite and highly faulted austenite (Figure 11), that is, the whole interior of ultrafine grains resembles the structure.
of deformation bands of the coarse-grained counterpart. Furthermore, the α'-martensite islands are arranged randomly within the grain and nucleate at grain boundaries or lattice defects within these ultrafine grains (cf. Droste et al.21). The same tendency was observed by some other authors.8,14 who reported a change in the nucleation mechanism under tensile loading from a preferred nucleation in deformation bands in coarse grains to a nucleation at grain and twin boundaries in ultrafine grains.

A further special feature in case of the submicron grains after cyclic deformation is the absence of the twinning relationship between the adjacent α'-martensite islands (cf. Figure 10), which is mainly attributed to the smaller size of the islands. The shear due to the phase transformation and the accompanying stresses of coarse increase with increasing size of the α'-martensite nuclei. Thus, the requirement of compensation by means of twinning decreases with decreasing size. Accordingly, for this fine austenitic grain size, it seems likely to conclude that the α'-martensite islands reach a grain or twin boundary before they exceed a critical size which requires for compensation by twinning.

After discussing the effect of grain size on the α'-martensite fractions and the microstructure after cyclic deformation, the reasons for the less pronounced cyclic hardening in case of small austenitic grain size should be considered. Besides the reduced α'-martensite formation regarded as the major factor, a further contribution is attributed to the different mechanisms responsible for the α'-martensite-related hardening, namely (i) slightly higher indentation hardness of the α'-martensite (see nanoindentation experiments61,63), (ii) fine grain size of α'-martensite in comparison to that of the austenite and (iii) reduced mean free path for dislocations in the austenite due to the α'-martensite nuclei.3,61 Whereas mechanisms (i) and (iii) are reported to be independent on grain size for the present grain size ranges (contrary to the UFG alloy, cf. Droste et al.21), the hardening effect of mechanism (ii) vanishes at small austenitic grain size. The grain size difference between the individual grains of austenite and α'-martensite and, thereby, the relative difference in strength is reduced with decreasing austenite grain size. Thus, this aspect is assumed to further contribute to the smaller cyclic hardening of the LR-780 state besides the lower α'-martensite fraction.

Contrary to the differences discussed so far, a feature all material states of the present alloy have in common is the requirement for plasticity before forming fatigue-induced α'-martensite. It has been presented for coarse grains that the martensitic transformation is triggered by partial dislocations on two different slip systems.1,64 Thus, a specific threshold value of the cumulated plastic strain λ_{p,th} has to be exceeded to form the first α'-martensite nuclei.46,65-67 Even though no deformation bands and a change in the nucleation mechanism have been observed in the fine grains of the LR-780 material state (cf. Section 3.5), an incubation period for the formation of α'-martensite is revealed as well. As prior to α'-martensite formation, no cyclic hardening occurs for medium and small strain amplitudes (Figure 7); it is not an increasing stress which triggers the phase transformation. Instead, the increasing cumulated plastic strain facilitates the nucleation of α'-martensite, that is, plasticity is required to create appropriate nucleation sites. This threshold value λ_{p,th} decreases with increasing strain amplitude (Figure 8A), causing the shorter incubation period for onset of α'-martensite formation and corresponding cyclic hardening. Likewise to the coarse-grained material, the reduction in the threshold value is attributed to an increase in lattice defect density with increasing strain amplitude, that is, an increase in potential nucleation sites.65,66

### 4.4 Influence of grain size of reverted austenite on fatigue life

The damage occurring during fatigue is always linked to plasticity of the material. For total strain-controlled tests, the plastic strain amplitude of the i-th cycle Δε_{pl/2i} can be estimated using the following relationship:

\[
Δε_{pl/2i} = Δε_{i}/2 = Δε_{el}/2 = Δε_{i}/2 - \frac{Δσ/2i}{E},
\]

where \(Δε_{el}/2\) is the elastic strain amplitude of the i-th cycle. Accordingly, the plastic strain during cyclic deformation is linked to the stress amplitude, that is, the strength of the material. At a given total strain amplitude, an increasing strength is accompanied by a decreasing plastic strain. The evolution of the plastic strain amplitude during cyclic deformation shown in Figure 15 for the present material states illustrates this relationship. The plastic strain amplitude is the highest for the HP material and decreases in the sequence LR-990, LR-780 and LR-610. The difference between the fully reverted states in absolute values does not change significantly from high (\(Δε/2 = 1.2\%\)) to low (\(Δε/2 = 0.4\%\)) total strain amplitude. The relative difference, that is, the ratio, between the HP and LR-780 state, on the other hand, is much more pronounced at the smaller total strain amplitude of 0.4% (Figure 15). Thus, the strength of the material plays a major role for the fatigue life (see the Basquin law). In contrast, at the high total strain amplitude of 1.2% (Figure 15), the relative difference between the fully reverted states is substantially reduced. Thus, the
The influence of the material strength on the fatigue life becomes negligible. Instead, the ductility is the dominating factor determining the lifetime of the material, as demonstrated by the Manson-Coffin law.

Accordingly, the similar fatigue life of the LR-990 and HP states at high strain amplitudes (Figure 12) can be attributed to the comparable ductility (Table 1). At low strain amplitudes, the lower strength of the HP state is probably compensated by the more pronounced fatigue-induced α'-martensite formation leading to cyclic hardening and the corresponding decrease in plastic strain amplitude (Figure 15). This finding is in agreement with results reported earlier, for example, other studies, indicating a beneficial effect of the α'-martensite formation for the fatigue life at low strain amplitudes. A further contribution compensating for the lower strength may be provided by the cylindrical specimen geometry of the HP state yielding a more homogenous stress distribution over the cross section.

The LR-780 state, on the other hand, exhibits a slightly higher fatigue life at low strain amplitudes (Figure 12) due to its higher strength. Despite a less pronounced α'-martensite formation, the stress amplitudes exceed the ones of the LR-990 and HP reference material (Figures 7a-c). Consequently, the plastic strain amplitudes remain lower during all stages of cyclic deformation (Figure 15, Δε/2 = 0.4%, dashed green line), enhancing the fatigue life. A different situation is observed at high strain amplitudes where the fatigue life of the LR-780 state coincides with that of the other two fully reverted states despite its lower ductility (Table 1). Two reasons are supposed to be responsible. Likely, the most important one is the less-pronounced fatigue-induced α'-martensite formation, which is known to be detrimental for the fatigue life at high strain amplitudes due to decreasing ductility. In addition, a slightly more homogeneous strain partitioning between α'-martensite and austenite probably contributes to the comparable lifetime. As discussed above, the strength difference between the two phases decreases with decreasing austenite grain size. Thus, it seems reasonable to suggest that the resistance against plastic deformation of the two phases converges with decreasing austenitic grain size, that is, in case of the LR-780 state, the α'-martensite is forced to bear a higher portion of the applied total strain compared with the α'-martensite in coarse grained austenitic steels, cf. Droste et al. Thus, the plasticity in the austenite is reduced and the damage is delayed.

The fatigue life of the LR-610 state is shorter than that of all other states (Figure 12). The very low ductility of the LR-610 state is the reason for a markedly inferior fatigue life at high strain amplitudes. Even at low strain amplitudes, the higher strength and significantly lower plastic strain amplitudes (Figure 15, Δε/2 = 0.4%, blue dotted line) can hardly compensate for the poor ductility. Only at the lowest total strain amplitude of Δε/2 = 0.3%, the fatigue life is comparable with the LR-990 and HP materials.

5 | SUMMARY

The reversion behaviour and the grain-size dependency cyclic properties of a metastable 18Cr-7Mn-7Ni austenitic stainless steel were analysed based on three material states. Furthermore, a hot-pressed counterpart of the same steel was studied for reference. The results can be summarized as follows:

- Laser reversion annealing was successfully applied to achieve reverted austenitic microstructures. The two highest peak temperatures (989°C and 779°C) yielded grain sizes of 7.7 and 2.7 μm, whereas a more heterogeneous microstructure consisting of submicron-sized grains and deformed austenite as well as individual grains with diameters up to almost 2 μm was produced by the lowest peak temperature (608°C).
- The yield strength doubled with decreasing grain size from 62 μm (reference material) to 2.7 μm (LR-780 state), while the uniform elongation decreased from 63% to 39%. The LR-610 state exhibited highest strength but showed necking after short plastic yielding.
- The initial cyclic stress amplitudes increased with decreasing grain size, but the cyclic hardening and fatigue-induced α'-martensite formation were reduced.
• Excluding the LR-610 state, the fatigue lives based on total strain amplitude were comparable with a minor advantage for the fine-grained LR-780 state at low strain amplitudes corresponding to its higher strength. In contrast, the LR-610 state exhibited an inferior fatigue life particularly at high strain amplitudes due to its significantly lower ductility.
• With refining grain size, deformation bands became less pronounced. Instead, complete grains exhibited a similar structure as deformation bands. Accordingly, a high ε'-martensite fraction was observed in submicron reverted grains which can contain more than one α'-martensite variant.

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NOMENCLATURE

- $d_A$: area weighted grain size
- $E$: Young’s modulus
- $k$: Hall-Petch constant
- $L$: mean free path for dislocations
- $N$: number of cycles
- $T$: temperature
- $ΔG$: Gibbs free energy
- $Δε_t/2$: total strain amplitude
- $Δε_{pl}/2$: plastic strain amplitude
- $λ_p$: cumulated plastic strain
- $λ_{p, th}$: threshold value of the cumulated plastic strain at which α'-martensite formation sets in
- $ξ_{fin}$: final α'-martensite fraction
- $σ_0$: friction stress
- $σ_y$: yield strength

ABBREVIATIONS

- BSE: backscattered electron
- DA: deformed austenite
- EB: electron beam
- EBSD: electron backscatter diffraction
- HAGB: high angle grain boundaries
- HP: hot pressing
- LR: laser reversion
- LS: linear speed
- ND: normal direction
- RD: rolling direction
- SEM: scanning electron microscopy
- SFE: stacking fault energy
- TEM: transmission electron microscopy
- t-SEM: scanning electron microscopy in transmission mode

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Additional supporting information may be found online in the Supporting Information section at the end of this article.

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