Tailored deformation behavior of 304L stainless steel through control of the crystallographic texture with laser-powder bed fusion

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HIGHLIGHTS
- Manipulation of the crystallographic texture of 304L stainless steel with Laser-Bed Powder Fusion.
- Enhancement of the martensitic transformation and twinning effects by tailoring the crystallographic textures.
- Assessment of asymmetric deformation behavior between tension and compression, through the Effective Stacking Fault Energy.
- Revealing the effect of the evolving crystallographic texture on the deformation behavior.

ABSTRACT
Laser-powder bed fusion (L-PBF) has gained significant research interest, not only for its profound advantage of producing near-net shape complex geometries of metallic parts, but also for the possibility of producing tailored microstructures. Here we exploit the capability of manipulating the crystallographic texture by L-PBF to tailor the deformation behavior of austenitic stainless steels. In specific, by adjusting the laser power and the laser scanning speed, tailored crystallographic textures can be obtained, along the uniaxial loading direction in 304L stainless steel samples produced by L-PBF. In situ neutron diffraction and uniaxial tension and compression tests are undertaken to investigate the extent of the transformation induced plasticity effect and to correlate it with the tailored macrostructures. The influence of the initial and the evolving crystallographic texture on the deformation behavior is demonstrated and...
1. Introduction

The deformation behavior of austenitic stainless steels is a function of the stacking fault energy (SFE) [2–4]. The main factors affecting the SFE are the chemical composition of the alloy and the temperature in which the deformation takes place [3]. Existing literature [5–10] suggests that when the SFE is low enough, i.e. less than roughly 20 mJ m$^{-2}$, phase transformation occurs upon deformation, the so-called transformation-induced plasticity (TRIP) effect. As the value of the SFE increases (20 mJ m$^{-2} < $ SFE < 45 mJ m$^{-2}$), deformation twinning, known as the twinning induced plasticity (TWIP) effect, becomes the governing deformation mechanism, while with further increase of the SFE (i.e. > 45 mJ m$^{-2}$), only dislocation glide is observed. Intermediate values of the SFE, roughly between 13 and 18 mJ m$^{-2}$, have been reported to promote a combination of both TRIP and TWIP effects [11,12].

The TRIP effect results in a profound increase of the strength while retaining good ductility. This effect involves the transformation of the metastable austenite (known as γ-austenite) to either ε-martensite (with hexagonal close packed crystal structure – HCP) or to α′-martensite (with body centered cubic crystal structure – BCC). It has been shown that the formation of ε-martensite is strongly dependent on the crystallographic orientation of the parent grains with respect to the loading direction (LD) [7,14,14]. The nucleation mechanism of ε-martensite has been associated with the formation of stacking faults [15]. The formation and the width of the stacking faults is moreover dependent on the magnitude of the shear stress acting on the leading (LPD) and trailing (TPD) partial dislocations. When the LPD experiences higher resolved shear stress than the TPD, they separate and form an extensive stacking fault. The accumulation of stacking faults on every other (1 1 1)$\gamma$ plane leads to the formation of ε-martensite [15]. Moreover, the formation of deformation twins is governed by the same mechanism as that of ε-martensite. However, twinning requires the accumulation of stacking faults on every (1 1 1)$\gamma$ plane [3,7,16].

The orientation dependence of the martensitic transformation has been investigated in a recent study conducted on wrought austenitic stainless steel, where it was observed that under uniaxial tension, the grains with (1 1 1) and (1 1 0) crystallographic orientations, with respect to the LD, favor the transformation. In contrast, the transformation was suppressed in grains with (1 0 0) crystallographic orientation with respect to the LD [17].

In the context of producing metallic components by means of additive manufacturing (AM), laser-powder bed fusion (L-PBF) is one of the most widely used AM techniques. Recently, a lot of researchers have shifted their focus on the fabrication of parts with this method, not only for the profound advantage of designing and producing highly complex geometries but also for obtaining complex microstructures [18]. Austenitic stainless steels are within the materials that are often processed by L-PBF due to good processability and applicability [19]. It has only been recently that the community has steered its interest also towards the possibility of manipulating microstructures and crystallographic textures by altering the L-PBF processing parameters.

One of the pioneering studies was conducted by Niendorf et al. [20], where it was shown that by changing the laser power it is possible to obtain different crystallographic textures along the building direction. In particular, by employing a laser power of 1000 W they managed to obtain a strong (1 0 0) crystallographic texture along the building direction, in a single part of 316L stainless steel. On the other hand, a nearly random crystallographic texture was obtained when the same material was processed with a laser power of 400 W, while the scanning speed, hatch distance, layer thickness, scanning strategy were kept constant. Similar studies undertaken on Ni superalloys have demonstrated that the crystallographic texture can be altered by changing the laser power, the laser scanning speed and the layer thickness [21,22]. Another study conducted on pure molybdenum has demonstrated that by varying one process parameter at a time and by keeping the others constant (i.e. by varying either the laser power, or the laser scanning speed, or the hatching spacing, or the layer thickness) the formation of strong (1 0 0), (1 1 0) and even (1 1 1) crystallographic textures along the building direction is possible [23]. In the latter case it was found that by increasing the heat input, the crystallographic texture rotates towards (1 1 0), while relatively low heat input results in stronger (1 0 0) crystallographic texture. Finally relatively strong (1 1 1) crystallographic texture is obtained for increased values of laser power paired with low scanning speed. In more recent studies conducted on SS316L, it was shown that the cellular sub-cells inside the melt pool align with the direction of the heat flow, which close to the centerline of the melt pool was found to be vertical, while moving away from it, it tends to be inclined. As a result, the (1 0 0) directions of the FCC lattice align with the inclined heat flow and as a consequence the (1 1 0) crystallographic orientation aligns parallel to the building direction. In this way, it is possible to produce strong (1 1 0) crystallographic texture, almost single crystal-like, along the building direction [24–26]. In these studies, the process parameters affect the melt pool characteristics, which upon solidification, affect the crystallographic orientation and the morphology of the solidified grains. As a general observation, deep melt pools result in the formation of (1 1 0) crystalllographic textures, while shallow melt pools result in the formation of (1 0 0) crystallographic texture in the building direction [24].

While the melt pool dynamics are very sensitive to the process parameters and they play a key role in the formation of crystallographic texture, the crystallography in cubic systems itself also allows for further manipulation of the in-plane crystallographic textures, when the crystallographic texture along the building direction is well-controlled. In a recent study by Sofinowski et al. [25] it was shown that strong (1 1 0) crystallographic texture can form along the building direction, and strong (1 0 0) can form along the scanning direction. By rotating the scanning direction at 55 degrees with respect to the sample frame, strong (1 1 1) in-plane crystallographic texture is obtained. On the other hand, Wang et al. [26] obtained similar results by employing appropriate values of power and speed to realize a strong (1 1 0) crystallographic texture along the building direction and strong (1 0 0) along the laser scanning direction. By then cutting tensile specimens at appropriate angles with respect to the building direction it is possible to obtain tensile specimens with very strong (1 0 0), (1 1 1) and (1 1 0) crystallographic textures along the LD.

Before the TRIP or TWIP effects are triggered, significant dislocation-based plasticity occurs in these materials which results in texture changes [27]. Therefore, despite the as-built texture can be controlled, it has not been explored how the evolving initial crystallographic texture affects the deformation behavior. By shedding light on the interplay between tailored crystallographic...
texture, evolution of deformation texture and preferred deformation mechanisms under different stress states, it will be possible to realize materials with superior mechanical properties.

The present study aims at conducting a comprehensive investigation on how the tailored initial crystallographic textures and their evolution with strain affect the deformation behavior of the L-PBF processed SS304L. Uniaxial tension and compression tests, paired with in situ neutron diffraction, allow following the evolution of the crystallographic texture along the LD and the deformation behavior.

2. Materials and experimental

For the L-PBF process, gas-atomized 304L powder with maximum particle size of 45 μm, was purchased from Carpenter Additive, UK. Table 1 shows the chemical composition of the powder. The L-PBF fabrication was performed using a Sisma MySint 100 with a spot size of 55 μm, under an argon atmosphere (<0.1% oxygen). The LPBF processing has been shown to result in N content variations in the range of approximately 100 ppm, between powders and samples with different process parameters [28–35], which is lower than the reported content variation that can affect the mechanical behavior of 304 steels [36]. Therefore, insignificant variations in N content due to potential outgassing during LPBF, and hence no variations in the mechanical behavior between different samples, are expected. The laser parameters and sample identification are summarized in Table 2. A “chess-board” scanning pattern (squares of 4 × 4 mm²) was employed, while each layer was rotated 90 degrees and shifted 1 mm in x and y with respect to the previous layer. Cylinders of 85 mm in length and 13 mm in diameter were fabricated for machining tensile specimens and cylinders of 10 mm in length and 7 mm in diameter for machining compression specimens. The effect of the residual stresses on the transformation behavior is not taken into consideration, since it has been found that they are relieved considerably during the machining process [37]. For the majority of the samples, the building direction is along the axial direction of the cylinders except for sample T3, which was built horizontally, as shown in Table 2.

Neutron diffraction experiments under uniaxial tension/compression were undertaken on the VULCAN instrument at the Spallation Neutron Source (SNS) of the Oak Ridge National Laboratory (ORNL), USA [38]. The 2-detector bank system of VULCAN was utilized to obtain diffraction information along the LD (detector bank 1) and the TD (detector bank 2). The in-situ measurements were undertaken continuously under load control with a loading rate of 0.26 N/s in the elastic regime and under strain control with a strain rate of 0.01 s⁻¹ in the plastic regime. The diffraction and mechanical data were binned every 60 s and the diffraction data were fitted using the single peak fitting routine of the VULCAN analysis software, VDRIVE [62], to obtain the integrated intensity of the different diffraction peaks, along with the peak position and the full width at half maximum. Rietveld refinement was undertaken using the open source software GSAS [39] utilizing the diffraction information from the transverse direction, along which the crystallographic texture is not as strong as in the axial direction. Initial tests showed that the martensite fraction obtained from the diffraction data from the axial and transverse directions do not vary significantly. Starting from the last histogram (i.e., at the highest strain), and after refining the lattice parameters of austenite and the α'-martensite, the scale factor of the phase fraction was refined. Then, a generalized spherical harmonic description for the measured complete neutron powder diffraction spectra, using a 12 and 9 harmonic order for the austenite and α'-martensite was used respectively, to account for the crystallographic texture of both phases. Four appropriate profile coefficients were used for modeling the peak shape in both phases (sigma-1, sigma-2, gamma and rsca). The backward sequential refinement (towards lower true strain) of all data was undertaken by refining the phase fraction, the peak profile coefficients, the texture, and hydrostatic/elastic strain to capture the lattice strain. The maximum number of refinement cycles was set to 10 and the refined parameters were propagating to the next histogram. The error of the phase fraction determination refers to the uncertainty in refining the phase fraction scale factor, while the χ² was below 3 for all the refined spectra, which indicates a satisfactory goodness of fit. Rietveld analysis was also utilized for estimating the inverse pole figures (IPFs) along the LD, assuming cylindrical symmetry. More information on the calculation of the IPFs can be found in [40]. The fitted texture parameters were then plotted in IPFs using MTEX [41].

“Post-mortem” EBSD investigations were undertaken on pieces extracted from the gauge volume and pieces extracted from the sample grips, which remained plastically undeformed and they were used as reference of the as-built microstructure. The samples were ground with 1200 grit SiC paper and then electropolished for 10 s with a 16:3:1 (by volume) ethanol, glycerol and perchloric acid solution at 48 V. A field emission gun scanning electron microscope (FEG SEM) Zeiss ULTRA 55 equipped with EDAX Hikari Camera operated at 20 kV in high current mode with 120 μm aperture was used. For each of the IPFs shown later, several EBSD maps were stitched together to improve the grain statistics and the reliability of the texture analysis. The EBSD raw data was post-processed using the EDAX OIM Analysis 7.3 software or the open source software, MTEX [41].

3. Results

3.1. Initial microstructure & crystallographic texture

Fig. 1 shows EBSD orientation maps with IPF coloring of the as-built condition of the samples that were used for the tension tests, together with the corresponding IPFs, along the LD. The as fabricated material is fully austenitic, which is in agreement with the literature [37]. It should be noted that stacking faults, twins and α-martensite have been observed in the past in the as-built material by TEM [42] but their fraction and size is beyond the detection limit of EBSD. Sample T1 exhibits a strong (110) crystallographic texture along the LD. Sample T2 exhibits a relatively strong (110-) crystallographic texture along the LD too, but with a significantly higher fraction of (100) oriented grains along the LD, than sample T1. Sample T3 has a predominant (111) crystallographic texture along the LD. Moreover, samples T1 and T2 exhibit relatively small equiaxed grains, while sample T3 exhibits more elongated grains, which is associated with the different relation of the building direction to the loading direction of T3 (Table 2) as compared to samples T1 and T2 (Table 2). Since there is no significant difference in the size and grain morphology between the tensile samples, it is

| Table 1 | Chemical composition of the pre-alloyed gas atomized 304L powder used for manufacturing the samples. |
|---------|--------------------------------------------------|
|         | Fe | N  | O  | Cu | Ni | Cr | S  | P  | Mn | Si | C  |
| w/w %   |    |    |    |    |    |    |    |    |    |    |    |
|         | Bal. | 0.07 | 0.03 | <0.1 | 9.8 | 18.4 | 0.004 | 0.012 | 1.4 | 0.61 | 0.015 |
not expected to have any noteworthy effect on the mechanical and transformation behavior of the samples.

The L-PBF process parameters for the compression samples, were such to obtain a predominant $h_{110}$ crystallographic texture along the LD in both samples, while affecting the fraction of the $h_{100}$ oriented grains, which are favorable for TRIP/TWIP under compression. The orientation maps with IPF coloring with respect to the LD, of the two compression specimens are shown in Fig. 2, along with the corresponding IPFs parallel to the LD (Fig. 2c and e). Consistently with the tensile samples, the microstructure is fully austenitic. The crystallographic texture analysis with EBSD revealed that both samples exhibit strong $h_{110}$ crystallographic texture along the LD.

Despite the fact that several EBSD maps were stitched to improve the grain statistics and hence the reliability of the crystallographic texture analysis, the number of grains included in the area analyzed by EBSD is limited, when compared to neutron diffraction. Thus, there is a discrepancy between the IPF shown in Fig. 2c and 2d. As such, the IPFs in Fig. 2d and f, derived from analysis of the neutron diffraction data, show that sample C1 exhibits a slightly stronger $h_{100}$ crystallographic texture, along the loading direction, than sample C2. The impact of this difference on the TRIP effect will be discussed in the subsequent sections.

It should be noted that, despite using the same processing parameters for building the tension and compression samples, the resulting crystallographic texture differs significantly. This can be attributed to the smaller diameter of the compression samples, compared to the tension samples. The smaller diameter results in a smaller number of scanning tracks per layer, since the hatch distance is the same between the two samples, which ultimately leads to less energy input per layer. This is in accordance with results by Dong et al. [43], where the diameter of the L-PBF-processed specimens was seen to affect the microstructure. By increasing the diameter of the sample, the obtained microstructure exhibits more elongated grains, due to the relatively high heat input resulting from the higher number of scanning tracks. Moreover, other studies concluded that the crystallographic texture of L-PBF-processed SS316L shifts from random to $h_{100}$, along the building direction with reducing part diameter [44,45]. In addition to the significant size difference of the tension and compression samples, the fact that different base plates were used for the manufacturing of the samples (a smaller in diameter base plate was employed for the tension samples while the compression samples were built on a larger plate) and the different time delays between the layers, also contributed to the difference in the crystallographic texture.

### 3.2. Phase transformation

[Fig. 3a](#) shows the true stress–strain curves of the L-PBF-processed SS304L tension specimens obtained during the in-situ neutron diffraction experiments. The coloring of the individual lines is based on the predominant crystallographic texture of the corresponding specimen, using the IPF coloring, where green corresponds to T1 (dominant $h_{110}$), red corresponds to T2 (dominant $h_{100}$) and blue corresponds to T3 (dominant $h_{111}$). It is seen that the tensile behavior of the specimens is dependent on the grain orientation along the LD. Particularly, T3 exhibits superior yield strength and ultimate tensile strength among the three samples, while T2 shows the lowest strength. As such, and in agree-
ment with [26], the sample with stronger (111) crystallographic texture has the highest yield point. Moreover sample T3 exhibits the highest Young’s modulus (192 GPa), compared to T1 and T2 which have similar Young’s modulus, i.e. approximately 165 GPa. This is expected from the single crystal mechanical properties in FCC materials, where {111} is the stiffest elastically family of planes. On the
other hand, the compression samples display very similar mechanical behavior, attributed to their similar texture and grain size. As a result of the difference in the crystallographic texture along the LD, the samples exhibit also different TRIP behaviors. Fig. 4a shows the evolution of the martensitic phase fraction with strain, for the tension samples. After approximately 20\% true strain, the formation of \( \alpha' \)-martensite starts, which agrees with previous observations on non-textured material [46]. It is seen that sample T3 contains the largest fraction of martensite as compared to samples T2 and T1, and with more martensite, more hardening is observed. It is also apparent that T2 exhibits the slowest formation of martensite, for the same plastic deformation amongst the three samples. Along with the formation of \( \alpha' \)-martensite in the microstructure, a small fraction of \( \varepsilon \)-martensite is detected, as shown in the neutron diffraction patterns shown in Figure S1, included in the supplementary material section.

Under compression, it is seen that a relatively high fraction of \( \varepsilon \)-martensite forms first and it is followed by the formation of \( \alpha' \)-martensite. Fig. 4b and c show the evolution of the \( \varepsilon \) and \( \alpha' \)-martensite fraction, respectively, with true strain for samples C1 and C2. It is observed that \( \varepsilon \)-martensite first appears at 5\% true strain for both samples while \( \alpha' \)-martensite starts to form at 10\% true strain. The growth of \( \varepsilon \)-martensite follows an increasing trend for both samples until 17.5\% true strain, after which, the formation of \( \varepsilon \)-martensite reaches a plateau, or even slightly decreases at higher strains. This observation is indicative that \( \varepsilon \)-martensite is consumed for the formation of \( \alpha' \)-martensite. It is well established that \( \varepsilon \)-martensite is a precursor of \( \alpha' \)-martensite and the later appears at the intercepts of the platelets of the former [15,37,46]. It is also observed, that sample C1 exhibits almost thrice the fraction of \( \varepsilon \)-martensite than sample C2. Similar applies for the \( \alpha' \)-martensite fraction, while its evolution appears to be steeper for sample C1.

### 3.3. Deformed microstructure

Fig. 5 shows the results of the “post-mortem” EBSD investigation for each tension sample accordingly. It is seen that all samples contain a significant amount of deformation austenite twins, besides \( \alpha' \)-martensite. Twins and martensite appear predominately in the grains with \( \langle 111 \rangle \) orientation along the LD, corresponding to the blue-colored grains (IPF coloring). In contrast, grains oriented with \( \langle 100 \rangle \) along the LD, i.e. the red-colored grains in IPF coloring, do not exhibit a significant number of planar defects such as twins and/or martensite. Fig. 5h shows the IPF along the LD for sample T3, after deformation, where the blue colored dots represent orientations of austenite grains, which contain twins and/or martensite, while the red dots represent orientations of austenite grains that are free of planar defects. The marked black dots represent the \( \langle 311 \rangle \) and \( \langle 210 \rangle \) crystallographic orientations, where the value of the Schmid factor of the LPD is equal with the value of the TPD. It should be noted that, based on the difference of the Schmid factor of the LPD and the TPD, the grains with orientation \( \langle 110 \rangle \) along the LD are considered favorable for transformation or twinning. However, it is observed that the remaining grains with \( \langle 110 \rangle \) orientation along the LD, do not exhibit many planar defects. Fig. 5g shows a high magnification orientation map with IPF coloring, along the LD, where deformation twins and deformation induced \( \alpha' \)-martensite can be observed. The deformation twins are apparent by their characteristic, about 60 degrees, misorientation with respect to the parent austenite grains. Such misorientation boundaries are highlighted with white in Fig. 5g. The crystallographic orientation of the twins appears to be \( \langle 100 \rangle \) along LD, forming in parent austenite grains with \( \langle 111 \rangle \) along the LD orientation.

The “post-mortem” EBSD analysis of the samples that underwent compression (shown in Fig. 6) was carried out at two different strains, at 14\% and at 28\% true strain. For both samples and at both strains, a significant fraction of deformation-induced martensite is observed. In particular, both samples exhibit a higher fraction of \( \varepsilon \)-martensite than \( \alpha' \)-martensite at 14\% true strain, in agreement with the Rietveld analysis shown in Fig. 4. Fig. 6g and h show high magnification EBSD maps of an austenite grain at 14\% true strain that exhibits \( \varepsilon \)-martensite nucleating at the center of the austenitic grain and extending to the grain boundary. The nucleation of \( \alpha' \)-martensite takes place inside the platelets of \( \varepsilon \)-martensite. As the strain increases, the \( \alpha' \)-martensite consumes a large fraction of \( \varepsilon \)-martensite, as shown in Fig. 6d which, together with the Rietveld analysis results, confirm that the transformation occurs with the sequence of \( \gamma \rightarrow \varepsilon \rightarrow \alpha' \) [15,47]. It is also observed that most of the grains that contain martensite have orientations close to \( \langle 100 \rangle \) along the LD, while a significant number of grains that exhibit \( \langle 110 \rangle \) orientation along the LD are free of martensite. At 28\% true strain, it is seen that even stronger \( \langle 110 \rangle \) crystallographic texture along the LD appears, while martensite appears also in grains that would be considered non-favorable for martensite formation, as shown in Fig. 6b, c and f. It remains unclear though, whether that martensite formed in an initially favorably oriented grain, while the retained austenite developed \( \langle 110 \rangle \) deformation texture with respect to the LD, which is no longer in favor for martensite formation. Besides this, grain interactions and stress concentrations at such high strains, may also locally alter the stress state and eventually allow the transformation in grains that are considered unfavorable with respect to the macroscopic stress state. It is also interesting to note, that unlike the tensile specimens, no indications of the existence of deformation twins under compression are observed. In summary, during tension it is seen that deformation twins and \( \alpha' \)-
Fig. 5. Orientation maps with IPF coloring along the loading direction (LD) after tension, (a) & (b) sample T1, (c) & (d) sample T2, (e) & (g) sample T3. In (g) the white lines represent twin grain boundaries and the black lines represent α'-martensite. (h) The orientations of twinned or transformed grains (blue dots) and of grains free of planar defects (red dots) for sample T3, plotted on the IPF along the LD. The IPF shows the orientation dependence of the difference between the Schmid factor of the LPD and TPD in grey color scale. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)
Fig. 6. Orientation maps with IPF coloring along the loading direction (LD) of sample C1 after compression at (a) 14% true strain and (b) 28% true strain. (c) is a high magnification orientation map with IPF coloring and (d) is a phase map of the same area, where austenite is represented with blue color, ε'-martensite with red color, α'-martensite with black color and the grain boundaries are represented with green color. Orientation maps with IPF coloring along the LD of sample C2 after compression at (e) 14% true strain and (f) 28% true strain. (g) is a high magnification orientation map with IPF coloring and (h) is a phase map of the same area, where austenite is represented with blue color, ε'-martensite with red color, α'-martensite with black color and the grain boundaries are represented with green color. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)
martensite co-exist, while under compression ε-martensite acts as an intermediate step of the \( \gamma \rightarrow \alpha' \) transformation. Additionally, EBSD analysis shows that \( \alpha' \)-martensite nucleates at the intersections of deformation twins (tension) and of ε'-martensite (compression). This behavior has been observed in the literature for the strain induced martensitic transformation [48].

3.4. Texture evolution

It is well documented that \( \langle 111 \rangle \) and \( \langle 100 \rangle \) textures develop along the LD under uniaxial tension [49]. Fig. 7 shows the evolution of the integrated neutron diffraction intensity, normalized with the proton beam intensity of the neutron spallation source, along the LD and increasing strain, for the \( 111_\gamma \), \( 200_\gamma \), \( 220_\gamma \) and \( 311_\gamma \) diffraction peaks for the austenitic phase, for each sample accordingly. It is observed that for all samples, the intensity of the \( 220_\gamma \) diffraction peak, which corresponds to the \( \langle 110 \rangle \) oriented grains along the LD, strongly decreases upon deformation. In case of FCC materials under tension, this is a typical behavior originating by the evolution of deformation texture. This decrease is accompanied by the increase in the diffraction intensity of all other three diffraction peaks. From the onset of the deformation, the integrated intensity of the \( 200_\gamma \) diffraction peak shows the highest increase for all tension specimens. The \( 111_\gamma \) and \( 311_\gamma \) diffraction peaks appear to follow an increasing trend, but with a significantly lower rate when compared to the \( 200_\gamma \). At higher strains, the \( 200_\gamma \) and \( 111_\gamma \) diffraction peaks still follow an increasing trend while the \( 311_\gamma \) appears to reach a plateau. It should be noted that above 20% true strain, when the transformation and twinning occur, the evolution of the integrated intensity of austenite cannot be interpreted that straightforwardly anymore.

By considering the Kurdjumov-Sachs (KS) orientation relationship, where \( \{111\}_\gamma // \{110\}_a \), the 24 possible martensite variants that form in a \( \langle 111 \rangle \)-oriented austenite grain along the LD exhibit a strong \( \langle 110 \rangle \) texture in the LD. Fig. 7d shows the evolution of the integrated intensity of the \( 110_\gamma \) diffraction peak of \( \alpha' \)-martensite for the three tensile specimens. Indeed, the evolution of the intensity with strain follows an increasing trend and it is stronger for sample T3, validating that the higher fraction of martensite in T3 is due to the higher fraction of \( \langle 111 \rangle \)-oriented austenite grains along the LD.

Under the application of compressive stress, \( \langle 110 \rangle \) crystallographic texture forms along the LD [50]. Since the two compression samples already have relatively strong initial \( \langle 110 \rangle \) crystallographic textures along the LD, the texture does not significantly evolve with strain. Fig. 8 shows the evolution of the integrated diffraction intensity of the \( 111_\gamma \), \( 200_\gamma \), and \( 220_\gamma \) diffraction peaks of the austenitic phase with respect to the LD and increasing strain for each sample accordingly. As it is expected, the intensity of the \( 220_\gamma \) diffraction peak is strong from the onset of the deformation and it increases.

![Fig. 7](image-url) Evolution of the normalized integrated intensity of austenite for several diffraction peaks upon tensile deformation for (a) T1, (b) T2 & (c) T3 samples. (d) Evolution of the integrated intensity of the \( 110_\gamma \) diffraction peak upon tensile deformation, for all three tension samples.
slightly during straining. On the other hand, the intensity of the 200\(_c\) diffraction peak decreases significantly for both samples, indicating the strong orientation dependency of the transformation as well as the evolving crystallographic texture.

4. Discussion

The present study highlights the significant impact that small initial crystallographic texture differences have on the deformation behavior of a SS304L. Samples with different crystallographic textures along the loading direction are manufactured with L-PBF by varying the process parameters. Combination of an increase of the laser power and the scanning speed with a decrease in the hatching spacing induces more (100)-oriented grains along the building direction of the samples. In contrast less power and lower scanning speed results in the formation of a stronger (1\1\1) crystallographic texture along the building direction. Similar results have been observed in the literature and are associated with the depth of the melt pool [23]. Under tension, sample T3 exhibits the highest fraction of martensite, while sample T2 exhibits the lowest fraction of martensite. This behavior is expected based on the theoretical analysis, i.e. for the grains with (1\1\1) orientation along the LD, the splitting of the partial dislocations is favored. In contrast, grains oriented with (100) along the LD do not favor the splitting of the partial dislocations. The EBSD results presented in Fig. 6 support this expectation and show that the (100) oriented grains with respect to the LD are free from planar defects. Under compression, the grains with (100) orientation along the LD are favorable for transformation. Despite the fact that sample C1 exhibits slightly stronger initial (100) crystallographic texture along the LD than sample C2, it exhibits a more pronounced formation of both \(\varepsilon\)- and \(\alpha'\)-martensite.

The favorably oriented grains under tension follow a different deformation behavior than the favorably oriented grains under compression. Under tension, the deformed microstructure consists of a small fraction of \(\varepsilon\)-martensite, as observed by neutron diffraction (Figure S1), deformation twins (Fig. 5) and \(\alpha'\)-martensite (Fig. 5). It should be noted that \(\varepsilon\)-martensite is not observed in the EBSD maps, and due to its very small quantity it is considered that the formation of \(\alpha'\)-martensite derives directly from the austenitic parent phase (\(\gamma \rightarrow \alpha'\)), without a strong contribution from \(\varepsilon\)-martensite. The formation of \(\alpha'\)-martensite is observed at relatively high strains (i.e., at 20% true strain) with twin grain boundary and twin intersections acting as nucleation sites, as evident by the post-mortem EBSD analysis (shown in Fig. 5b, d, f and g).

Deformation twins in (1\1\1)-oriented grains, appear to have orientations close to the (100) along the LD. As such, besides the formation of (100) deformation texture along the LD, the increase in the integrated intensity of the 200\(_{\gamma}\) diffraction peak, is partially also attributed to the formation of deformation twins. It is therefore observed that the (1\1\1)-oriented grains are partially consumed by the twinning/martensite formation. Thus, the combined effect of twinning/transformation of the austenite grains with (1\1\1) crystallographic orientation along the LD, while (1\1\1) deformation texture along the LD forms under tension, justifies the moderate increasing rate of the integrated intensity of the 111\(_{\gamma}\) diffraction peak shown in Fig. 7.

In a recent study conducted on a randomly textured 304L stainless steel, processed by L-PBF, it was also found that the formation of \(\alpha'\)-martensite in the microstructure starts only at 23% true strain [37]. The retardation of the martensitic transformation can be associated with the relatively high content of nitrogen (0.1 % approx.) in the powder used for this study. Nitrogen is known to be a strong austenite stabilizer, acting as an inhibitor of the formation of martensite upon deformation [51–53]. In addition, Polatidis et al. [28] observed the formation of \(\varepsilon\)-martensite upon tension, which was detectable in EBSD analysis, while no deformation twinning was observed. This behavior has to be attributed to the SFE difference between the two materials. The 304L steel of the present study has a SFE of 19.8 mJ m\(^{-2}\), calculated using the compositional relationship in [54], which is considered medium, and results in the formation of both deformation twins and \(\alpha'\)-martensite under tension. It has been reported that as the value of the SFE increases in Fe-Cr-Ni alloys, the deformation behavior shifts from the formation of \(\varepsilon\)-martensite and \(\alpha'\)-martensite, to the formation of deformation twinning and \(\alpha'\)-martensite, and to the formation of deformation twins alone [3,55].

In contrast to the tension samples, a significant fraction of \(\varepsilon\)-martensite forms under compression (Fig. 4b and 6). This aberration in the deformation behavior between tension and compression has been observed in the past in wrought austenitic stainless steels [55,56]. The suppression of the formation of \(\varepsilon\)-martensite under tension and the facilitation of \(\varepsilon\)-martensite formation under compression has been attributed to the volume change that accompanies the \(\gamma \rightarrow \alpha'\) and the \(\gamma \rightarrow \varepsilon\) transformations, where the \(\gamma \rightarrow \alpha'\) transformation is accompanied by a 2.6% volume increase, while the \(\gamma \rightarrow \varepsilon\) is accompanied by a small volume.

![Fig. 8. Evolution of the normalized integrated intensity for several austenite diffraction peaks upon compressive deformation for (a) C1 & (b) C2 samples.](image-url)
decrease [56]. Hence, the \(e\) formation is apparently facilitated under compression.

The suppression of twinning under compression in a TRIP/TWIP 304L steel has been observed in the literature [56], without elaborating the mechanisms behind this behavior. As such, martensite formation is favored under compression, while twinning under tension, in the material studied in this work. The term of the effective stacking fault energy (\(\gamma_{\text{eff}}\)) has been previously used to describe then non-equilibrium separation of the stacking faults under the application of external load and to highlight the orientation-dependent character of deformation twinning [14,26,59–59]. The \(\gamma_{\text{eff}}\) is calculated by using the following formula: [14,26,59,59].

\[
\gamma_{\text{eff}} = \gamma \pm \frac{(m_{\text{trailing}} - m_{\text{leading}})}{2} \sigma b_p
\]

where \(\gamma\) is the SFE of the material calculated as in [54], \(m_{\text{trailing}}\) and \(m_{\text{leading}}\) are the Schmid factors of the TPD and the LPD accordingly. For this calculation it is assumed that single slip system with the highest Schmid factor (and the corresponding pair of partials) are activated, as previously observed [27,60]. \(\sigma\) is the external applied stress and \(b_p\) is the Burger's vector which is calculated as \(\alpha/6\), where \(\alpha\) is the austenite lattice parameter, deduced from the neutron diffraction data. It is noted that the LPD under tension becomes the TPD under compression and vice versa for the TPD under tension. Fig. 9 shows the evolution of the \(\gamma_{\text{eff}}\) for the most favorable orientations, i.e. (111) along the LD for tension, and (100) along the LD for compression, as a function of the applied stress. It is seen that from the onset of deformation, the \(\gamma_{\text{eff}}\) remains lower under uniaxial compression than under uniaxial tension. The lower value of the \(\gamma_{\text{eff}}\) promotes the formation of wider stacking faults and hence the formation of \(e\)-martensite, under compression. In contrast, the slightly higher \(\gamma_{\text{eff}}\) under tension, suppresses the formation of \(e\)-martensite, hence twinning appears.

As already discussed, (111) and (110) orientations along the LD are considered to be both favorable for transformation, since the Schmid factor of the LPD is higher than the Schmid of the TPD. Moreover, sample T3 (with (111) crystallographic texture) exhibits more martensite than sample T1 (with (110) crystallographic texture). This predominantly happens because of the evolution of the crystallographic texture, before the martensitic transformation is initiated. Fig. 10a shows the evolution of the crystallographic texture, in the form of IPFs along the LD of the sample T1, as calculated from the Rietveld analysis of the neutron diffraction data. It is observed that the initial strong (110) crystallographic texture gradually evolves towards a (311) orientation along the LD with strain up to 20% true strain. This crystallographic orientation evolution is not favorable for TRIP/TWIP, i.e. the Schmid factor of the LPD is nearly the same as for the TPD. Upon further straining, the grains rotate from (311) crystallographic texture towards (111) and (100), along the LD, as also observed in the literature [61], forming the commonly observed deformation texture under tension of FCC metals. Interestingly, in a similar investigation on nearly single crystal-like SS316L samples produced by L-PBF, the (110) crystallographic texture along the LD evolves towards (111) along the LD, while remaining in the favorable regime of orientations in the entire course of deformation [26]. However, this is not the case for the less strongly textured materials of the present study and hence, the initial (111) crystallographic texture along the LD, is found to be the most favorable initial crystallographic texture for TRIP/TWIP under tension.

To further evaluate the effect of the evolving crystallographic texture on the TRIP effect, simulations using the Taylor model implemented in the MTEX tool box [41] were carried out. It should be noted that the simulations were run until 20% true strain, following the texture evolution from the initial crystallographic texture of the T1, T2 and T3 samples, without considering any texture change due to possible initiation of (mild) deformation twinning and only considering the macroscopic applied strain being uniform in all grains. The Taylor model, thus, simulates well the crystallographic rotations caused by the initiation of dislocation slip in the microstructures. Fig. 10b shows the simulated texture evolution, \(\frac{\Delta I}{I_0}\) where \(\Delta I = I-I_0\) is the change of the IPF value, \(I\), evolving with strain with respect to the initial IPF value, \(I_0\), integrated within 15 degrees with respect to the (111) pole of the IPF. From the experimental data, \(\Delta I/I_0\) corresponds the change to the integrated intensity, \(\Delta I\), with respect to the initial integrated intensity, \(I_0\), of the 111\(\alpha\) diffraction peak. At the VULCAN instrument, the vertical coverage of the detector is approximately +/- 15 degrees, and therefore, the averaging of the IPF is done in such way to emulate the detector coverage of VULCAN. It is observed that the intensity of the 111\(\alpha\), diffraction peak increases with strain, which is attributed to the development of deformation texture. The evolution of the (111) crystallographic texture along the LD can be well captured by the Taylor model, which is seen to be steeper in sample T2 and T1, compared to T3. In absolute values though (Fig. 10c), T3 has significantly stronger (111) texture, followed by T1 and the T2. Fig. 10d shows the evolution of the favorable fraction of orientations with respect to the LD for TWIP/TRIP, based on the initial crystallographic texture and the Taylor simulation up to 50% total strain. The initial favorable fraction of favorable crystallographic orientations is calculated from the IPF along the LD, which is obtained from the Rietveld analysis. The initial favorable fraction of orientations shows that T1 and T3 have equally favorable microstructure for TWIP/TRIP effects, while T2 slightly lower. By applying the Taylor model, though, it is seen that the fraction of favorable orientations slightly increases in sample T3, while it decreases for samples T1 and T2, upon deformation. This simulation, thus, can very well simulate the behavior that is observed experimentally: not only the initial but also the evolving crystallographic texture of sample T3 is the most favorable, amongst the samples, followed by sample T1 and finally sample T2. The simulation results agree well with the evolution of martensite formation in Fig. 4a. Therefore, both experiments and simulations suggest that an initially strong (111) crystallographic texture along the LD is the most favorable for TRIP/TWIP, since the grains will remain in the favorable regime throughout the deformation process. To further support this claim one more sample manufactured with...
L-PBF was examined under tension, sample ST, presented in the supplementary material section. Figure S2a shows the initial microstructure of sample ST, as captured from the EBSD analysis, along with the IPF along the loading direction. It is evident that sample ST has a strong \( \{110\} \) crystallographic texture along the LD, very similar to sample T1. However analysis of the diffraction data at the initial state, showed that sample T1 contains a higher fraction of \( \{111\} \)-oriented grains along the LD, while sample ST contains a higher fraction of \( \{100\} \)-oriented grains along the LD (highlighted in Figure S2b). Figure S3a shows the evolution of the normalized integrated intensity of the \( 111 \) diffraction peak, and figure S3b shows the evolution of \( \alpha' \)-martensite with true strain, deduced from Rietveld analysis. From the onset of deformation it is seen that sample T1 exhibits a higher fraction of \( \{111\} \)-oriented grains, which further increases, as the deformation progresses due to the formation of \( \{111\} \) deformation texture. At 20% true strain (where \( \alpha' \)-martensite first forms) sample T1 exhibits a significantly higher fraction of \( \{111\} \)-oriented grains along the LD, compared with sample ST. The difference in the fraction of \( \{111\} \)-oriented grains along the LD, between samples T1 and ST, which have otherwise very similar crystallographic textures along the LD, validates our hypothesis that the \( \{111\} \) crystallographic texture along the LD is the most favorable initial crystallographic texture for the enhancement of the TRIP/TWIP effects. Finally, under compression the mechanism is simpler since the only favorable orientation for transformation is the \( \{100\} \) along the LD, while the deformation texture evolves towards the \( \{110\} \) along the LD orientation. As such, under compression, the most favorable crystallographic texture for enhanced TWIP/TRIP effects is a \( \{100\} \) crystallographic texture along the LD.

**5. Conclusions**

The effect that both the initial and the evolving crystallographic texture, along the loading direction, have on the deformation behavior of 304L stainless steel samples produced by laser-powder bed fusion is systematically investigated using neutron diffraction and EBSD. Careful selection of the L-PBF process parameters, enables the control of the crystallographic texture of the samples along the loading direction. The relatively mild variations between the samples highlight the strong effect of the crystallo-
graphic texture on the deformation behavior of 304L TRIP/TWIP steel. The sample with the strongest initial (111) crystallographic texture along the loading direction displays the highest fraction of α’-martensite upon deformation while the sample with the strongest (100) crystallographic texture along the LD, displays the least fraction of α’-martensite, under tension. In contrast, the sample with the strongest initial (100) crystallographic texture along the LD displays a significantly higher fraction of both martensitic phases upon deformation. Under tension, deformation twins coexist with α’-martensite. On the other hand, under compression a significant amount of ε-martensite prevails the formation of α’-martensite. Despite the theory based expectation of a favorable behavior of the (110) crystallographic orientation, the corresponding material did not display as much transformation, upon tensile deformation, as the sample with relatively strong (111) texture along the LD. It is observed that at the strain at which the martensitic formation starts, the majority of the (110) oriented grains already rotate towards non-favorable orientations, which restrict the TRIP/TWIP effects. As such, the (111) crystallographic orientation along the LD is the most favorable for the facilitation of the TRIP/TWIP effects under tension. Finally, by utilizing the Taylor model and by assuming the macroscopic stress state on the grain level, the evolution of the (111) crystallographic texture upon deformation is simulated and it can sufficiently determine which sample is more favorable for transformation. It is concluded that the texture control through L-PBF processing and the simulation of texture evolution allows for designing materials with controlled deformation behavior.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Data availability

The raw data required to reproduce these findings are available to download from: https://doi.org/10.5281/zenodo.5779418

Appendix A. Supplementary data

Supplementary data to this article can be found online at https://doi.org/10.1016/j.matdes.2022.110789.

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