Deformation mechanisms of twinning-induced plasticity steels: In situ synchrotron characterization and modeling

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The plastic deformation behavior of twinning-induced plasticity steels of composition Fe–25Mn–3Si–3Al are investigated by means of in situ synchrotron high-energy X-ray diffraction and compared to self-consistent simulations. It is the first time the alternating interaction of {1 1 1} <1 1 0> slip and {1 1 1} <1 1 2> twinning have been directly observed in situ while undergoing uniaxial tension. The deformation texture is determined mainly by dislocation gliding, while deformation twinning impedes the reinforcement of texture.

It has been widely accepted that, due to its low stacking fault energy, twinning is an effective means of deformation in high manganese austenitic, so-called twinning-induced plasticity (TWIP) steel, which has been supported by transmission electron microscopy and electron backscatter diffraction (EBSD), showing that the volume fraction of twins increases concomitant with plastic deformation [1,2]. Specifically, one variant of the {1 1 1} <1 1 2> twin system is activated within grains at the early stages of plastic deformation, followed by the nucleation of twins of different variants of the {1 1 1} <1 1 2> twin system between the boundaries of the first set of twins, leading to a ladder like structure [3]. Recent work has further been performed by Barbier et al. [1], using texture component analysis to explore the fraction of dislocation slip and deformation twinning during plastic deformation. Nevertheless, the combination of various conventional surface microstructure analysis methods are limited in provision of bulk information about the evolving dislocation density, bulk texture, local strain, stacking faults and the deformation mechanisms operating in the total volume of the material.

In the present case, the evolution of grain statistics, which includes grain orientation distribution, grain rotation and elastic lattice strain, were determined by in situ high-energy X-ray diffraction. Self-consistent grain-scale models were then employed to predict the microscopic and macroscopic deformation behavior and the contributions of each deformation mechanism.

Tensile specimens were taken from a high manganese austenitic steel sheet with nominal composition of Fe–25Mn–3Si–3Al (wt.%). The material had been hot rolled from slab state to the final thickness of 2.5 mm, with average grain size of about 20 \( \mu \)m. Specimens with a gauge length of 15 mm and a width of 1 mm were wire-cut from the steel sheet. The tensile loading direction was applied perpendicular to the rolling direction of the plate.

Diffraction experiments were conducted at the high energy beam line ID15B of the European Synchrotron Radiation Facility [4,5], with an incident X-ray energy of 86.94 keV. The uniaxial tensile test was performed at ambient temperature at a constant strain rate of \( 1 \times 10^{-4} \) s\(^{-1} \) to a final true strain of 0.47. A Pixium 4700 area detector [6] recorded two-dimensional (2D) diffraction patterns at a frame rate of \( \sim 2 \) Hz. The X-ray beam was set to track the geometric center of the tension sample. Further details about the experimental setup can be found elsewhere [7]. The texture was measured before and after the tensile test on a dedicated goniometer on the same beamline.

Figure 1 presents the 2D diffraction patterns for the initial, undeformed state and after the tension test, and both reveal only reflections of the austenite face-cen-
tered cubic (fcc) $\gamma$-Fe phase. The received sample consisted of coarse grains oriented in arbitrary directions, which are indicated by discrete spots on the discontinuous Debye–Scherrer rings in Figure 1a. After tension, the diffraction patterns change into highly textured, smooth curves as seen in Figure 1b. The 1 1 1 reflection then shows maximum intensity along the tensile direction ($L$) and an intensity minimum in the transverse direction ($T$). Further side maxima occur in the directions of $L \pm 60^\circ$.

In order to explore the continuous evolution in time, the diffraction rings of a selected reflection were cut at the 6 o'clock direction, straightened and sequenced in time, to arrange them in an azimuthal-angle–time plot (AT plot) as shown in Figure 2. This representation shows the evolution of diffraction information along orientation and time and can be plotted for each $hkl$ reflection.

Figure 2 shows the AT plot for the 1 1 1 reflection which initially displays the spotlessness coming from a few coarse grains before the deformation commences. The progression of the deformation behavior is now explained in terms of distinct strain regions, as follows:

1. $\varepsilon \in 0 \ldots 0.00165$: Coincides with the elastic strain region, no new grains rotate onto, or reflecting grains rotate off the Ewald sphere. The initial reflection spots keep their original positions, azimuthal widths and intensities, without vanishing or gross azimuthal shift.

2. $\varepsilon \in 0.00165 \ldots 0.044$: Existing diffraction spots start to spread along the azimuthal direction continuously with time. This indicates a subgrain formation resulting in a low-angle mosaic distribution consistent with plastic deformation by slip. A few new grains rotate into the reflection condition, for instance at $\psi = 40^\circ$ and $153^\circ$. From $\varepsilon = 0.0075$ onwards, many new diffraction spots appear distributed over all azimuthal angles, which immediately flare into larger mosaic spread before fading away. The observed reflections spread and rotate towards their closest preferred orientation. For example, the timeline at $\psi = 25^\circ$ is inclined towards the $L$ direction and $\psi = 40^\circ$ towards $L \pm 60^\circ$ [8]. This fading is attributed to the first occurrence of twinning, in which parts of the relatively large grains are reoriented. The timelines in the $L$ direction fade earlier than others. As we observe the 1 1 1 reflection, twinning is not activated on this plane; however, the other unit cell diagonal planes, such as (T 1 1), inclined by 70.5° to the reflecting (1 1 1) plane do suffer high shear stress, twinning 1 1 1 out of reflection.

3. $\varepsilon \in 0.044 \ldots 0.137$: At about $\varepsilon = 0.044$, many new timelines come into reflection, especially along the $L$ direction. These new born diffraction spots then strengthen in intensity and spread along the azimuth direction rotating towards the preferred orientation. The timeline along $\psi = 290^\circ$ is an example. At $\varepsilon = 0.082$, the whole diffraction intensity changes abruptly. From this time on, oscillations along azimuth direction and intensity can be seen over the whole orientation space. For instance, the fluctuation occurring in the timelines along $\psi = 130^\circ$ and $230^\circ$.

4. $\varepsilon \in 0.137 \ldots$ end: Gradually, the timeline distribution rearranges into the final texture while the oscillatory behavior continues. During the remainder of the deformation process there are ongoing oscillations due to twinning and reorientation.

Four diffraction phenomena need to be addressed here. Firstly, diffraction peak broadening occurs due to accumulation and nucleation of defects which are usually dislocations in fcc structure materials. This increase in azimuthal peak width results from the formation of substructure and subgrains from an increasing dislocation density. Secondly, the gradual progression of diffraction intensity along azimuthal angles reveals grain rotation driven by the integrated force from neighboring grains. Thirdly, sudden changes in diffraction intensities indicate the abrupt alteration of a diffraction volume’s orientation. Lastly, oscillation with time correlates with regular variation of the diffracting volume’s orientation within a certain oscillatory range. We can therefore conclude that mosaic broadening and the observed gradual rotation of grains must be attributed to slip as a deformation mechanism [8], while sudden changes in intensity and oscillations are due to large jumps in orientation space that are originated by twinning.

Besides the direct observation of the grain orientation behavior by the variation of timelines, quantitative analysis of the lattice strain also provides valuable information of how certain grain orientations interact under uniaxial tension. The lattice strain $\varepsilon = (d_{hkl} - d_{hkl}^{\text{ref}})/d_{hkl}^{\text{ref}}$ reflects the average elastic strain within a particular $hkl$ grain orientation. In the present study, the Debye–

![Figure 1. Half extracts from the 2D diffraction patterns of the as-received sample with isotropic orientation distribution (a); and the textured distribution after 47% uniaxial, tensile strain (b). Reflection indices from inside out are 1 1 1, 2 0 0, 2 2 0, 3 1 1, 2 2 2, 4 0 0 and 3 3 1. Polar reciprocal space coordinates $q$ and $\psi$ are indicated while the central cross marks the origin and scale of 1 Å$^{-1}$. The tensile direction $L$ is shown by arrows.](image)

![Figure 2. AT plot of the 1 1 1 reflection displaying the intensity maxima and minima in the $L$ and $T$ direction, respectively. True strain is proportional to time by $\varepsilon = 10^{-4}$ s$^{-1}$.](image)
Scherrer rings were divided into 720 sectors. The Bragg peaks as a function of the momentum transfer $q$ in each sector were fitted by a Gaussian function and used to obtain $d_{hkl} = 2\pi/q_{hkl}$ along the azimuthal direction. The stress-free lattice interplanar distance $d_{hkl}^0$ was evaluated as the lattice spacing prior to uniaxial tension averaged along all azimuth directions.

Simulation of the instantaneous change in lattice strain with uniaxial tension was achieved using an elasto-plastic self-consistent (EPSC) model [9]. The input parameters to the model were the initial texture, represented by 2000 discrete grain orientations weighted according to the texture and the single crystal elastic constants for austenitic steel. In this case only slip was permitted in the model on the \{1 1 1\} <1 1 0> system, omitting the effect of twinning in the model in order to illuminate the effect of twinning on lattice strains.

Figure 3 shows the experimental and calculated lattice strains along the longitudinal direction for the 1 1 1, 2 0 0, 2 2 0 and 3 1 1 reflections. As expected the model deviates strongly from the experiment since only slip is considered in the model. The advent of nonlinear elastic lattice strain is an indication of a grain orientation commencing plastic deformation. In this case, the model clearly indicates the 2 0 0 orientation is the first to plastically deform at $\sim$375 MPa. That is, micro-plasticity occurs in these grains before the gross yield stress of 450 MPa. Slip is most likely in these grains because firstly, they are the most populated according to the texture (Fig. 4a), and secondly, they have comparatively the second highest Schmid factor of 0.408 for slip. As 2 0 0 slip commences, the EPSC model indicates load is transferred to plastically harder grain orientations, especially the 2 2 0 reflection. In reality we do not see this in the experimental curve. This is because \{1 1 1\} <1 1 0> twinning has the highest relative Schmid factor of 0.471 in this orientation. So we see that the load is not borne by the 2 2 0 and some twinning occurs in this orientation, according to the texture has a relatively low population resulting in a low volume fraction of twins. Deformation slip commences next for the plastically soft 3 1 1 orientation (Schmid factor = 0.446) and finally on the most plastically hard 1 1 1 reflection. Both slip and twinning are relatively difficult for grains in the 1 1 1 orientation with calculated Schmid factors of 0.348 and 0.368, respectively.

Another obvious feature of the experimental lattice strains is oscillation with deformation; this coincides with the variation of diffraction intensities observed for each reflection under deformation. The severe fluctuation of reflection 2 2 0 is due to huge error bars caused by the weakness of diffraction intensity along the longitudinal direction, coinciding with its texture minimum. Comparison with the modeled lattice strains, which operated on the basis that only dislocation slip occurred, shows the moment when gross deformation twinning was activated. For instance, the deviation between calculated and experimental lattice strain for 1 1 1 happened at $\sigma = 550$ MPa, which is consistent with the sudden change of intensity and orientation at the moment of $\varepsilon = 0.044$ in AT plot.

Just as the lattice strain was characterized by experiment and an EPSC model, the texture evolution can be interpreted using a visco-plastic self-consistent (VPSC) model [10]. In this model, the influences of both \{1 1 1\} <1 1 0> slip and \{1 1 1\} <1 1 2> twinning were considered. The results of the measurements and simulation are shown in Figure 4. The initial texture (Fig. 4a) is random in the $T$ directions, as borne out by the initial diffraction pattern in Figure 1a. There is though a stronger \{0 0 1\} fiber (3x random) oriented in the $L$ direction. The initial texture was input to the VPSC model and slip and twinning applied in different proportions and hardening rates until a match was achieved with the final measured texture (Fig. 4b), while also matching the constitutive macroscopic stress and plastic strain behavior (Fig. 4d).

As deformation progresses the initial inverse pole figure in the $L$ direction shows a sharp increase in texture index from 1.3 to 2.1 at the final 47% plastic strain. The texture strengthening is dominated by the contribution of slip which progressively strengthens the \{1 1 1\} poles in the direction of uniaxial tension. In comparison, the VPSC model appears to slightly over-predict the contribution of twinning (Fig. 4b and c), which is why the final

![Figure 3](image-url)  
Figure 3. Experimental and modeled lattice strain–stress curves for selected reflections.

![Figure 4](image-url)  
Figure 4. Comparison of experimental and VPSC model results: (a–c) show the inverse pole figures in the two $T$ and one $L$ direction for the initial, final and VPSC simulated texture, respectively. The constitutive stress–strain behavior given in (d) contains experiment (continuous), VPSC simulation (dotted) and modeled fraction of twin activity (dashed).
the broadening and rotation of the timeline moving towards the final preferred orientation, while the sudden change of intensity and orientation does not produce timelines in other major directions. The twinning angle is 70.5°, which is not too far from the cube texture maximum lying 60° apart and contributes mostly to a broadening and reduction of the pole density of the preferred orientation (Fig. 2). Similarly in the modeled texture, the final texture is achieved by a relatively low activity of twinning in the order of 10% (Fig. 4b–d). This proportion of activity concurs with previous measurements by EBSD in TWIP steel [1].

In conclusion, by use of high-energy synchrotron X-ray diffraction and self-consistent models, it is the first time the alternating interaction of {1 1 1} <1 1 0> slip and {1 1 1} <1 1 2> twinning has been directly observed in situ. The unique properties of TWIP steel of high ductility and work hardening rate can be explained by the interaction of slip and twinning. The deformation texture is determined mainly by dislocation gliding, while deformation twinning impedes the reinforcement of texture.

This new diffraction technique adds greater understanding of the progressive development of microstructure, internal stress and texture. In future it would be desirable to conduct tests for high manganese steels with different compositions and mechanical process.

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