Impact of hydrogen on the dislocation structure upon tensile loading in a high manganese twinning induced plasticity steel

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Abstract: High-manganese twinning-induced plasticity (TWIP) steels exhibit mechanical properties making them attractive for structural applications. However, they are prone to hydrogen embrittlement (HE), and may undergo a brittle, catastrophic failure upon hydrogen ingress in service conditions. We investigated the HE susceptibility of a Fe 28Mn 0.3C (wt.%) TWIP steel. We observed a reduction in engineering strain from 75% to 10% and a 20% increase in yield strength following the hydrogen-charging. The microstructural evolution was examined at three different, lower tensile strain levels, i.e., 3%, 5% and 7% by using correlative electron backscatter diffraction (EBSD) and electron channeling contrast imaging (ECCI). At 3% strain, dislocation cells form in the hydrogen-charged region, which can be attributed to the increased yield strength that promotes the slip activation on multiple slip systems. We introduce an approach for image processing to quantify the cell size. At 7% strain, we observed the formation of stacking faults and hexagonal close packed ε-martensite in the hydrogen-charged region due to the reduction in stacking fault energy (SFE) associated to the presence of hydrogen. In contrast, the microstructure of the uncharged samples was composed of planar dislocation arrangements at 3%, 5% and 7% strain, which is indicative of planar slip. We discuss the details of how hydrogen modifies the dislocation structures and the overall deformation mechanism, and the present study is expected to facilitate the future design of steels resistant to HE.

Keywords: TWIP steel, deformation mechanism, electron channeling contrast imaging, correlative microscopy, hydrogen embrittlement
1 Introduction
Hydrogen embrittlement (HE) encompasses a range of detrimental effects of hydrogen on the mechanical properties of metallic materials [1–3]. HE typically leads to a substantial loss in ductility and load-bearing capacity of metallic materials upon exposure to hydrogen-containing atmospheres [2,4]. HE has been known across the scientific community for more than a century, first reported by Johnson in 1875 [5]. Numerous models have been proposed to describe HE including hydrogen-enhanced decohesion (HEDE), hydrogen-enhanced localized plasticity (HELP), stress-induced hydride formation and cleavage and adsorption induced dislocation emission (AIDE) [1,4,6,7]. Yet, often, HE cannot be fully explained by a single mechanism [2,8]. Thus, despite decades of comprehensive studies, there is no consensus across the scientific community regarding the actual prevalent internal damage processes responsible for the macroscopic HE [1,4].

TWIP steels are a class of high manganese, austenitic steels which contain over 20 wt.% manganese and have a face centered cubic (FCC) crystal structure [9]. The deformation mechanism in TWIP steels is dislocation glide along with the continuous formation of mechanical twins during deformation [10]. This increase in twin density leads to a very high strain hardening rate which is often referred to as the ‘dynamic Hall-Petch effect’ [11]. TWIP steels hence have an excellent combination of high tensile strength and ductility, making them potentially suitable for a wide range of applications in the automotive industry, line pipe production and ship building for instance [12].

However, TWIP steels are highly susceptible to HE, as often demonstrated by hydrogen-induced delayed fracture in deep-drawn cups [13–15]. The degradation in mechanical properties due to hydrogen has been studied in an array of TWIP steel compositions including Fe-18Mn-0.6C (wt.%) [16], Fe-18Mn-1.2C (wt.%) [17], Fe-23Mn-0.5C (wt.%) [18] and Fe-22Mn-0.6C (wt.%) [19]. The plausible mechanisms proposed in these studies responsible for HE in TWIP steels include intergranular cracking associated to the promotion of deformation twinning [16–18] and martensitic transformation [16] by hydrogen.

*Ab initio* calculations suggest that the presence of the substitutional manganese significantly increases both the solubility and the mobility of hydrogen [20]. This is associated to a diffusion barrier for hydrogen along a connected chain of manganese atoms that is considerably lower than the hydrogen diffusion barrier along a connected chain of iron atoms [20]. Hence, the connected chains of adjacent manganese atoms, which are often referred to as the manganese percolation chains, act as efficient hydrogen diffusion channels, thereby providing a possible explanation for the high HE susceptibility of high manganese steels [20,21].

The low corrosion resistance of TWIP steels primarily leads to the hydrogen ingress during service conditions [12]. When a metal is exposed to the corrosive media, the corrosion reaction generates hydrogen which is introduced into the metal and leads to its embrittlement. Other studies also suggest that the HE susceptibility of high manganese TWIP steels particularly depends on its chemical composition, which further determines the material’s stacking fault energy, phase stability, tensile strength and hydrogen uptake behavior [22]. Although these factors influence the HE resistance of TWIP steels, the interaction of hydrogen with structural defects formed during the deformation of TWIP steels plays a decisive role [17,21]. The investigation of mutual interactions between hydrogen and structural defects requires the precise measurement and characterization of hydrogen, which is very challenging and hence
leads to many open questions [23–26]. Therefore, the resultant microstructural changes associated to the presence of hydrogen must be examined thoroughly.

Here, we examined the HE mechanism in a model Fe 28Mn 0.3C (wt.%) TWIP steel. Hydrogen was introduced into the samples by cathodic hydrogen charging. We performed tensile tests and investigated the structural defects formed during the tensile deformation with and without hydrogen charging. Since the total elongation to fracture of the hydrogen-charged sample was obtained as only 10%, we chose three intermediate tensile strain levels, 3%, 5% and 7% for examining the microstructural changes with an increase in the tensile strain. Thermal desorption spectrometry (TDS) analysis was used for quantifying the total hydrogen content inside the hydrogen-charged samples. We mapped microstructural changes that result from the tensile deformation of the hydrogen-charged and the uncharged samples by using electron channeling contrast imaging (ECCI) that enables the precise identification of dislocation structures [27]. ECCI is applied to the surface of bulk samples, the sample preparation hence does not relax residual stresses which are, indeed, important to keep hydrogen at its sites for longer duration. Since ECCI allows to image a wider field of view than transmission electron microscopy, it enables the statistical investigation of structural defects in a microstructure [27]. We performed statistical analysis of structural defects in at least 20 randomly-selected grains from both the hydrogen-charged and the uncharged regions by using ECCI, supported by electron backscatter diffraction (EBSD). It was found that the dislocation structures were more strongly influenced by the presence of hydrogen than, for instance, by the Taylor factor. The Taylor factor determines the number of slip systems that need to be activated in order to accomplish a given external strain. An interactive supervised machine learning (ML) tool ilastik was used to estimate the size distribution of dislocation cells formed in the hydrogen-charged region of the 3% tensile strained sample. Such a statistical approach, enabled by correlative ECCI and EBSD microscopy, helps in understanding the HE mechanism in TWIP steels, which will guide the development of future, hydrogen-resistant structural materials.

2 Materials and Methods

We selected a Fe 28Mn 0.3C (wt.%) TWIP steel which was produced by strip casting [28] with a strip thickness of 1.6 mm. Its homogenization annealing was carried out at 1150 °C for 2 hours [28]. Subsequently, it was cold-rolled to achieve a 50% thickness reduction, and subjected to recrystallization annealing at 800 °C for 20 minutes, followed by water cooling to room temperature. The samples were cut into the desired geometry with electrical discharge machining (EDM). Next, the samples were mechanically ground by #600 and #1000 grinding paper, followed by mechanical polishing with 3 μm diamond paste for 25 minutes. Subsequently, a mechano-chemical polishing step was carried out with 50 nm colloidal silica suspension for 40 minutes. The samples were finally cleaned with ethanol and dried with compressed air.

The samples were cut into dog-bone tensile sample geometry with gauge dimensions of 4 x 2 x 1 mm³ by using EDM. All surfaces of the samples were thoroughly cleaned to remove corrosion prior to hydrogen charging. The side surfaces were hence finished with #400 grinding paper, while the top and bottom surfaces were prepared as described above, until the mechano-chemical polishing step, in order to permit the hydrogen ingress into the sample during subsequent cathodic hydrogen charging [29]. The cathodic hydrogen charging was conducted in an electrolytic cell in which an aqueous solution of 0.05 M H₂SO₄ was used as an electrolyte.
[30–32], and 1.4 g/l of thiourea (CH$_4$N$_2$S) was added as hydrogen recombination inhibitor so as to inhibit the hydrogen recombination, thereby enhancing the hydrogen ingress [33]. The charging solution was magnetically stirred at a speed of 10 rotations per minute throughout the charging process in order to maintain a constant rate of charging reaction. A DC voltage of 1.5V was applied to the cell where the sample was attached to the cathode and a Pt wire was attached to the anode [34,35]. The cathodic hydrogen charging set-up is shown schematically in Figure 1(a).

The hydrogen charging was carried out for 5 days at ambient temperature conditions. The sample surface was then cleaned by a 5 minutes polishing using 50 nm colloidal silica suspension and then ethanol. The tensile test with light-optical digital image correlation (DIC) technique was performed immediately following this cleaning. A DIC paste was applied on the sample surface in order to achieve contrast for light-optical observations, enabling the quantification of strain from the tensile test. The tensile tests were hence conducted within 10 minutes after the charging had finished at a constant strain rate of 10$^{-3}$ s$^{-1}$ at room temperature using a Deformation Devices System DDS-3 (Kammrath & Weiss GmbH). The ARAMIS v6.3.0 software was used for the data extraction and analysis of tensile tests.

![Figure 1.](image)

In order to measure the hydrogen content of the samples before and after charging, TDS was carried out using a Hiden TDS workstation equipped with an HAL 3F 510 PIC quadrupole mass spectrometer. The TDS sample of dimensions 10 x 15 x 1 mm$^3$ was first polished until the final mechano-chemical polishing step and then charged with hydrogen similarly by cathodic hydrogen charging for 5 days at ambient temperature conditions. The hydrogen-charged TDS sample was positioned in an ultra-high vacuum chamber and heated at a constant rate of 16 °C/min from room temperature to 800 °C. The heating leads to the desorption of hydrogen from the sample which was detected by a quadrupole mass spectrometer [36,37]. A curve of hydrogen desorption rate (wt. ppm/s) vs. temperature (°C) was hence plotted. The area...
underneath the curve was determined to provide the total hydrogen content of the sample [29,38,39].

The fracture surfaces of tensile deformed samples were examined using a Zeiss Sigma 500 scanning electron microscope (SEM) at a working distance of 10 mm, an accelerating voltage of 15 kV and a beam current of 9 nA.

ECCI studies were carried out using a Zeiss Merlin SEM, equipped with a retractable 4Q backscattered electrons detector. The microscope was operated at an accelerating voltage of 30 kV, a beam current of 2 nA and a working distance of 7 mm [27]. The dislocation structures of the tensile sample deformed to fracture were examined from the region delineated by the blue box in Figure 1(b).

Three hydrogen-charged samples were deformed to tensile strains of 3%, 5% and 7%, respectively. They were immediately vertically cut into two halves by using EDM, as shown schematically in Figure 1(c). The obtained cross sections were mounted with a mounting material at 180 °C for 6 minutes. Subsequently, they were polished until the final mechano-chemical polishing step on the same day and their dislocation structures were examined by ECCI within 12 hours of the tensile test. The hydrogen-charged region is represented by the region delineated by black box within 100 µm close to the side edges of sample surface, while the uncharged region is represented by the region delineated by red box across the cross-section, as shown in Figure 1(c). Further details related to the representation of the hydrogen-charged and the uncharged regions are discussed in the results section. Approximately 20 grains were randomly chosen from each of the two regions for examining the dislocation structures statistically. Four stage tilt angles were selected, i.e., ~0°, ~-1°, ~5° and ~10° to examine the grains with different orientations and at least 5 grains were examined at each of the tilt angles which were in two-beam diffraction conditions. In the two-beam diffraction condition, only one set of lattice planes is tilted into the diffraction condition [27].

In order to obtain the orientations and morphology of the grains, EBSD orientation mapping was performed using a Zeiss Sigma 500 SEM equipped with an EDAX/TSL system with a Hikari camera at an accelerating voltage of 15 kV, a beam current of 9 nA, a scan step size of 0.5 µm, a specimen tilt angle of 70°, and a working distance of 14 mm [40]. EBSD data analyses were performed by using TSL OIM Analysis 7.0.

We used an ML tool ilastik [41] for estimating the size distribution of dislocation cells which were formed in the hydrogen-charged region of the 3% tensile strained sample. The manual calculation of the size of these cells or the use of a typical image analysis tool like water shed model was tedious. It could have led to errors due to the varying contrast between the cells and cell walls, making the distinction between the cells and cell walls ambiguous. Ilastik provides a graphical user interface (GUI) which enables to sparsely annotate regions belonging to different classes in a given image to create a training data-set. It hence allows to semantically segment a given micrograph during which a class is assigned to each pixel of a given image. We used the pixel classification workflow of ilastik for semantic segmentation. The workflow trains a random forest classifier using features obtained from user annotated training data to classify pixels into defined classes. The features used consider color, edges and texture at different scales in an image.

Two ECC micrographs were initially uploaded on the ilastik GUI as input data. The regions pertaining to the two classes, cell interior and cell walls, were sparsely annotated using brush
strokes. By this step, the incomplete walls were bridged such that the ML model learned to perform this task automatically on an unseen data. The regions with low contrast between cell and cell walls were also annotated to improve the model performance in such regions. After annotating, the ML model was trained and tested on the non-annotated regions of the images. More annotations were added when the results were not satisfactory. Once the model was trained, a batch processing tool was used to upload the other micrographs to segment them automatically.

One of the micrographs and the result after batch-processing are shown in Figure 2(a) and (b) respectively. The semantic segmentation hence increased the contrast between the cells and their walls and bridged the incomplete walls. After the semantic segmentation of a micrograph into cells and cell walls, a connected component analysis was performed to find the pixels existing as a part of a continuous segment forming a cell as shown in Figure 2(c). Pixels pertaining to each cell were identified and stored in an hdf5 file. The small pockets contained within the cell walls which could be confused as cells were removed by classifying only the continuous regions comprising of more than 100 pixels as separate cells. One pixel in each of our micrographs had an area of 4.5 x 4.5 nm² and as the number of pixels in each cell were known, we could hence calculate the area of each cell. The corresponding cell area distribution is hence shown in Figure 2(d) which provides the frequency of cells with a specific cell area (in µm²).

![Figure 2](image_url)

*Figure 2. (a) an ECC micrograph with dislocation cells; (b) the semantic segmentation of the same*
micrograph into cells and cell walls; (c) the formation of a continuous segment of cells; (d) the corresponding frequency distribution of the area of dislocation cells (µm²).

3 Results:

3.1 Initial microstructure
The initial microstructure of the recrystallized, uncharged Fe 28Mn 0.3C (wt.%) TWIP steel is displayed in Figure 3(a) by an EBSD-inverse pole figure (IPF) map with respect to the normal direction (ND). The random high angle grain boundaries are highlighted in red and Σ3 twin boundaries are highlighted in black. The material is fully austenitic and its grain size distribution is depicted in Figure 3(b).

Figure 3. (a) EBSD – ND inverse pole figure (IPF) map of the recrystallized Fe 28Mn 0.3C (wt.%) TWIP steel. TD – transverse direction; ND – normal direction out of the plane; RD – rolling direction; (b) grain size distribution.

3.2 Tensile test
Three hydrogen-charged and three uncharged samples were subjected to tensile tests and the mean value curves were plotted from the obtained engineering stress-strain curves, shown in Figure 4(a). The total elongation to fracture of the uncharged sample was ~75% (red curve) which was reduced to ~10% (black curve) after hydrogen charging. The yield strength of the uncharged sample was 250 MPa while it was 300 MPa for the hydrogen-charged sample, as highlighted in the close-up in the inset of Figure 4(a), i.e., the yield strength was hence increased by 20% after hydrogen charging. It is important to note that the hydrogen-charged sample shows a small plateau at the yield point following the yielding.

3.3 Thermal desorption spectrometry
The hydrogen desorption rate (wt. ppm/s) vs. temperature (°C) curve obtained from TDS analysis of the hydrogen-charged and the uncharged samples is shown in Figure 4(b). The total hydrogen content in the sample was determined from the area underneath the curve as 13.2 wt. ppm in the hydrogen-charged sample (black curve) against 1.3 wt. ppm in the uncharged sample (red curve). It must be noted that TDS experiments were performed on the recrystallized samples. The peak at ~120°C in Figure 4(b) shows that diffusible hydrogen was introduced into the sample by cathodic charging. The first peak hence corresponds to the hydrogen trapped in reversible trapping sites such as austenite matrix and grain boundaries [42]. The second peak at ~400°C is attributed to the non-diffusible hydrogen trapped in irreversible trapping sites [15,36] which is released at high temperatures.
Figure 4. (a) The engineering stress-strain curve of the uncharged (in red) and the hydrogen-charged (in black) samples (the inset shows the enlarged region delineated by black box); (b) The thermal desorption spectrometry curve for the uncharged and the hydrogen-charged recrystallized samples.

3.4 Fractography

The fracture surface of the uncharged sample was ductile and consisted of dimples, as shown by SEM imaging in Figure 5(a). The inset is a close-up of the dimples. The fracture surface of the hydrogen-charged sample exhibits three distinct regions, as visible in Figure 5(b). The outermost 40-µm region, close to the side edges, shows evidence of intergranular fracture, as revealed by the inset in Figure 5(b). Beyond, for the next 60-µm, Figure 5(b), the fracture surface shows features associated to quasi-cleavage [43]. The fracture mode in the center of the hydrogen-charged sample was dominated by dimples associated to ductile fracture.
Figure 5. (a) SEM image of the fracture surface of the uncharged sample, where the inset shows the enlarged region delineated by yellow box; (b) SEM image of the fracture surface of the hydrogen-charged sample, where the inset shows the enlarged region delineated by yellow box.

The brittle fracture due to hydrogen charging was hence limited to the side edges, i.e., close to the sample surface, which can be explained by the slow diffusivity of hydrogen within austenite, pointed out in Ref. [3,38,44] for instance. Therefore, the outermost region close to the sample surface was chosen for examining the dislocation structure of the hydrogen-charged region and the center region of the sample was chosen for examining the dislocation structure of the uncharged region, as indicated by the black and red boxes respectively in Figure 1(c).

3.5 Electron-channeling contrast imaging of the deformed samples
3.5.1 Samples deformed to fracture

The ECC micrographs of the samples deformed to tensile fracture were observed from the region delineated by a blue box in the schematic shown in Figure 1(b), approx. 500-µm away from the cleavage surface. Figure 6(a) shows an ECC micrograph of the hydrogen-charged sample deformed to tensile fracture. The plates of ε-martensite were observed along with dislocation tangles. It must be noted that the existence of ε-martensite is suggested here based on its morphology and on the suggestions of An et al. [45]. Similar microstructural features
were observed in the uncharged sample deformed to tensile fracture, as shown by a ECC micrograph in Figure 6(b). From these figures, it was very difficult to draw any conclusion about the HE mechanism due to the presence of an extremely high and heterogeneous tensile strain in both samples deformed to tensile fracture. The dislocation structures were hence examined at intermediate levels of tensile strain, i.e., 3%, 5% and 7% to better capture the formation mechanism of the deformed microstructure.

3.5.2 3% tensile strained sample
Figure 7(a) shows the ECC image of a grain from the hydrogen-charged region of the sample deformed up-to 3% tensile strain. The micrograph evidences the formation of dislocation cells, which is indicative of cross slip. In contrast, the ECC image of a grain from the uncharged region of the same sample was composed of planar dislocation arrangements, which indicates planar slip, as shown in Figure 7(b). The uncharged region does not develop dislocation cells.
Figure 7. (a) The ECC image of a grain from the hydrogen-charged region sample deformed up to 3% tensile strain depicting the formation of dislocation cells; (b) the ECC image of a grain from the uncharged region of the same sample which was composed of planar dislocation arrangements.

Ten grains with dislocation cells from the hydrogen-charged region were selected to calculate the cell area distribution in which the contrast between the cells and cell walls could be distinguished by using the software ilastik. The frequency distribution of dislocation cell area for 10 different micrographs, corresponding to these 10 grains was estimated by using the procedure described in Figure 2. The normalized frequency distributions for each of the ten micrographs are provided in the supplementary information (Figure S1). The densities in the normalized frequency distribution are summed to make the kernel density estimate, which is provided in Figure 8 for each of the ten grains. It indicates that the size of dislocation cells follows similar trends in all the investigated grains. The estimated dislocation cell sizes were in the range of 0.04 – 0.07 µm² with an average close to 0.05 µm².
3.5.3 5% tensile strained sample

Figures 9(a-b) show ECC images of two different grains from the hydrogen-charged region of the sample deformed up-to 5% tensile strain. The micrographs show the nucleation of stacking faults and the formation of dislocation loop dipoles. The features indicated by yellow arrows in Figure 9(a) are two closely-spaced dislocation partials bounding a stacking fault, indicating planar slip. The features indicated by red arrows in Figures 9(a-b) are likely dislocation loop dipoles, as there is no stacking fault visible between them. Such features are typically formed due to double cross slip. An ECC image of a grain from the uncharged, center region of the same sample in Figure 9(c), shows planar dislocation arrangements.
Figure 9. (a-b) The ECC images of grains from the hydrogen-charged region of the sample deformed up-to 5% tensile strain (the dislocation partials are indicated by yellow arrows and dislocation loop dipoles are indicated by red arrows); (c) the ECC image of a grain from the uncharged region of the same sample depicting the planar dislocation arrangements.

3.5.4 7% tensile strained sample
The ECC images in Figures 10(a-b) are from two different grains in the hydrogen-charged region of the sample deformed up-to 7% tensile strain. The features pointed by yellow arrows in Figure 10(a) indicate the formation of long, extended stacking faults, potentially along with ε-martensite, which can be ascribed to the lower stacking fault energy (SFE) due to the presence of hydrogen. The stacking faults can be distinguished from well-developed ε-martensite plates by the contrast difference in their ECC images. Stacking faults are only one-atomic layer thick. Thus, in ECCI, they appear as features that are marked by an intense bright line, where the fault intersects the sample surface followed by fading contrast to one side of the line where the stacking fault dives deeper into the material [27]. In contrast, an ε-martensite plate or a twin has an appreciable thickness and thus appears as a continuous bright band of constant brightness [27]. This is actually the case for the features marked by red arrows in Figure 10(b) which is hence considered to be nucleation of ε-martensite. The ECC image of a grain from the uncharged region of the same sample, Figure 10(c), shows planar dislocation arrangements and no ε-martensite phase formation.
Figure 10. (a-b) The ECC images of grains from the hydrogen-charged region of the sample deformed up to 7% tensile strain (the long, extended stacking faults are indicated by yellow arrows and the nucleation of ε-martensite phase is indicated by red arrows); (c) the ECC image of a grain from the uncharged region of the same sample showing the formation of planar dislocation arrangements.

3.5.5 Statistical analysis
A statistical analysis of the dislocation structures was performed by examining at least 20 randomly-selected grains from both the hydrogen-charged and the uncharged regions of each of the three samples, i.e., 3%, 5% and 7% tensile strained. The orientations of those grains were determined by using EBSD mapping.

The Taylor factors for tensile loading and activity of slip systems \{111\}<0 1 -1> were evaluated from the orientation data. The EBSD-IPF and Taylor factor maps of each of the examined grains are shown in Figure 11. Figure 11(a) depicts the EBSD-IPF map of the hydrogen-charged region from the 3% tensile strained sample and the corresponding Taylor factor map of the investigated grains is shown in Figure 11(b). Similarly, Figures 11(c-f) displays the EBSD-IPF and Taylor factor maps from the hydrogen-charged region of the 5% and 7% tensile strained sample, while Figures 11(g-l) indicates the EBSD-IPF and Taylor factor maps from the uncharged region of each of the three samples. The Taylor factor values of all the grains lie in between 2.3 and 3.6 which indicates that the microstructural evolution observations were not biased by the local Taylor factor.

Figure 11. EBSD-IPF mapping and the Taylor factor maps of the grains examined by correlative ECCI from the hydrogen-charged and the uncharged regions of each of the three samples, i.e., 3% tensile strained-hydrogen charged (a-b), 5% tensile strained-hydrogen charged (c-d), 7% tensile strained-
hydrogen charged (e-f), 3% tensile strained-uncharged (g-h), 5% tensile strained-uncharged (i-j), 7% tensile strained-uncharged (k-l), respectively.

The inverse pole figures of the tensile direction were also drawn for each of the examined grains in order to determine their orientation in the tensile direction. It was revealed that there was no preferred orientation of grains and hence, the texture had no influence on the microstructural evolution observations, as indicated by the corresponding inverse pole figures for each of the examined grains included in the supplementary text (Figure S2).

4 Discussion
4.1 Engineering stress strain curves
The onset of plastic deformation occurs at the yield point with the onset of dislocation nucleation. The yield strength of the uncharged sample was 250 MPa, whereas it was 300 MPa for the hydrogen-charged sample. An increase in the yield strength due to hydrogen has earlier been reported [46,47]. A higher yield strength value in the hydrogen-charged sample suggests that hydrogen delays the dislocation nucleation, potentially by dragging of dislocation nuclei. The hindrance to the dislocation nucleation and motion by hydrogen atoms has been demonstrated experimentally [48] as well as by atomistic modelling [49,50].

4.2 Fracture surfaces
Studies based on atomistic modelling proposed that hydrogen leads to the formation of voids on the grain boundaries during tensile loading [51]. They predict that a higher hydrogen concentration leads to a higher density of voids, in particular at grain boundaries close to the sample surface. These voids coalesce into larger ones, thereby leading to the grain boundary decohesion. Thus, the fracture surface appears intergranular close to the sample surface, preserving the trace of the original grain boundary, which is consistent with our observation reported in the inset of Figure 5(b). Beyond the region of intergranular fracture, the fracture surface exhibits features associated to quasi-cleavage, where the hydrogen content was relatively lower due to the slow diffusivity of hydrogen in austenite [22]. At lower hydrogen contents, the density of voids at the grain boundary is lower and they require a significantly higher growth, tearing the grain boundary in order to cause grain boundary decohesion. Thus, the surface appears as for a quasi-cleavage fracture in which the traces of the original grain boundaries cannot be recognized [51].

4.3 Microstructural evolution: correlative ECCI with EBSD
At each of the three tensile strain levels, i.e., 3%, 5% and 7%, planar dislocation arrangements were observed in the uncharged region. This is in stark contrast with the hydrogen-charged regions: (i) dislocation cells were formed at 3%; (ii) dislocation partials and loop dipoles were formed at 5%; and (iii) the long-extended stacking faults with ε-martensite were formed at 7%.

We confirmed that this microstructural evolution was due to the impact of hydrogen, and not associated to the texture or the Taylor factor, as was revealed by careful examination of several grains from both the hydrogen-charged and the uncharged regions of each of the three samples. The dislocation structures in all the grains were examined again after approx. 2 months and none of the observed dislocations, stacking faults, dislocation partials and loop dipoles had changed. The ECC micrographs of two different dates are provided in the supplementary information (Figure S3) for the hydrogen-charged region of the 3% tensile strained sample. It implies that the deformation mechanism in the studied TWIP steel was modified by the
presence of hydrogen during the deformation, and is stable even if the hydrogen progressively leaves the sample.

We observed an increase in the yield strength due to hydrogen which indicates that hydrogen delays the dislocation nucleation and leads to solid solution strengthening [46]. These factors activate the slip on several slip systems at 3% strain in order to cause any deformation at the applied load; in other words, a higher stress is required to cause any deformation [52]. Subsequently, dislocation slip is activated on multiple slip systems simultaneously [52], and dislocations interact to form the dislocation cells as observed in the hydrogen-charged region of the 3% strained sample and also schematically shown in Figure 12(a). Dislocation cells had been observed in other TWIP steels, for instance, in Fe 22Mn 0.6C (wt.%) TWIP steel studied by Gutierrez-Urrutia et al., without hydrogen charging [53]. Yet comparison remains difficult because of different values of SFE arising from different alloy compositions. The SFE of the present alloy is 27 mJ/m² [54] without hydrogen, while that of the Fe 22Mn 0.6C (wt.%) TWIP steel is 22 mJ/m² [53,55] in which the dislocation cell formation was observed. This could be an indication that the presence of hydrogen modifies the SFE to make it locally closer to 22 mJ/m² in the studied TWIP steel.

The distribution of dislocation cell size was quantified by our image processing approach aided by ML. This method still requires substantial user input, thus only a limited yet representative number of grains could be processed. Within the set of grains observed, no specific correlation of the cell size with grain orientation or Taylor factor, or their depth with respect to the surface, was found. In contrast to the previous reports under cyclic deformation where a strong influence of dislocation pattern on grain orientation was found [45], here we do not observe this across the 20 grains in which the dislocation pattern was examined. The reason for this is probably that at the lower tensile strains studied here, no characteristic orientation-dependent patterns are formed yet. Instead, the effect of hydrogen on the microstructure is more pronouncedly visible.

The dislocation cells were formed only in the hydrogen-charged region at 3% and not in the uncharged region. It implies that the dislocation cells are stable only in the presence of hydrogen. The formation of dislocation cells creates two regions within the grains. One where hydrogen is trapped, and hence more highly concentrated, and another in which hydrogen is relatively depleted. It has been shown that hydrogen accumulates at dislocations [56,57] and can be assumed to be trapped by the dislocation cell walls. The surrounding matrix is hence depleted of hydrogen. Hydrogen lowers the stacking fault energy in austenitic steels, as pointed in ref. [31], and this segregation hence create regions of high and low stacking-fault energy. Studies also suggest that the nanoscale compositional variations can lead to the local changes in stacking-fault energy [58]. With a further increase in the applied load, at 5% strain, as suggested by multiple studies [57,59], hydrogen is dragged by the dislocations breaking the dislocation cell walls. These regions with a low SFE, where solute hydrogen is present in higher concentration, develop into dissociated dislocations moving by planar slip, as depicted by mechanism (i) in Figure 12(b). This explains why the cells are no longer visible at this stage of the deformation. The regions depleted of hydrogen, in contrast, have a higher SFE allowing cross slip of individual dislocations from the surrounding cell walls which is visible by the formation of dislocation loop dipoles through double cross slip, as explained schematically in Figure 12(b) by mechanism (ii).
A further increase in the applied load at 7% strain leads to the redistribution of hydrogen in the material while lowering the overall SFE of the TWIP steel, which is in agreement with previous reports in austenitic steels [31,60]. This leads to the formation of long, extended stacking faults which stack together to nucleate ε-martensite as also schematically shown in Figure 12(c). The microstructure of the 5% tensile strained sample hence acts as the transition region where the deformation mechanism changes from the formation of dislocation cells, indicating cross slip, to the formation of stacking faults which indicates planar slip. We hence observe the features associated to both mechanisms at 5% strain, dislocation loop dipoles associated to cross slip and dislocation partials associated to planar glide. Our experimental results illustrate that hydrogen interacts with the dislocation structures in TWIP steels, as already reported in TWIP [39] and other austenitic stainless steels [61,62].

5 Conclusion
We examined the tensile properties and microstructure evolution of a Fe 28Mn 0.3C (wt.%) TWIP steel with and without hydrogen charging. We observed a drop in ductility from 75% to 10% (as engineering strain to fracture) and a 20% increase in the yield strength after hydrogen charging. The dislocation structures were examined at three different tensile strain levels. It was found that the deformation structures were independent of Taylor factor and orientation of the principal strain axis. In the hydrogen-charged region, grains developed dislocation cells at 3% tensile strain, while stacking faults with ε-martensite were formed at 7% tensile strain. In contrast, the uncharged region was composed of planar dislocation arrangements at each of the tensile strain levels. We hence conclude that the observed evolution of dislocation structures from the dislocation cell formation, requiring cross slip, to the ε-martensite phase formation, requiring planar slip, with an increase in the tensile strain was due to hydrogen. Our study hence emphasizes the role of hydrogen in modifying the deformation mechanism in TWIP steels.

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References:

[1] P. Sofronis, I.M. Robertson, Viable mechanisms of hydrogen embrittlement - A review, AIP Conf. Proc. 837 (2006) 64–70. https://doi.org/10.1063/1.2213060.

[2] P. Novak, R. Yuan, B.P. Somerday, P. Sofronis, R.O. Ritchie, A statistical, physical-based, micromechanical model of hydrogen-induced intergranular fracture in steel, J. Mech. Phys. Solids. 58 (2010) 206–226. https://doi.org/10.1016/j.jmps.2009.10.005.

[3] C. Zhang, H. Zhi, S. Antonov, L. Chen, Y. Su, Hydrogen-enhanced densified twinning (HEDT) in a twinning-induced plasticity (TWIP) steel, Scr. Mater. 190 (2021) 108–112. https://doi.org/10.1016/j.scriptamat.2020.08.047.

[4] O. Barrera, D. Bombac, Y. Chen, T.D. Daff, E. Galindo-Nava, P. Gong, D. Haley, R. Horton, I. Katzarov, J.R. Kermode, C. Liverani, M. Stopher, F. Sweeney, Understanding and mitigating hydrogen embrittlement of steels: a review of experimental, modelling and design progress from atomistic to continuum, J. Mater. Sci. 53 (2018) 6251–6290. https://doi.org/10.1007/s10853-017-1978-5.

[5] W.H. Johnson, On some remarkable changes produced in Iron and Steel by the action of hydrogen and acids, (n.d.). https://www.jstor.org/stable/113285.

[6] H.K.B. and P. Sofronis, P.Sofronis, H.K. Birnbaum, H.K.B. and P. Sofronis, P.Sofronis, H.K. Birnbaum, Hydrogen-enhanced localized plasticity--a mechanism for hydrogen-related fracture, Mater. Sci. Eng. A. 176 (1994) 191–202.

[7] S.K. Dwivedi, M. Vishwakarma, Hydrogen embrittlement in different materials: A review, Int. J. Hydrogen Energy. 43 (2018) 21603–21616. https://doi.org/10.1016/j.ijhydene.2018.09.201.

[8] A. Barnoush, H. Vehoff, Recent developments in the study of hydrogen embrittlement: Hydrogen effect on dislocation nucleation, Acta Mater. 58 (2010) 5274–5285. https://doi.org/10.1016/j.actamat.2010.05.057.

[9] J. Min, J. Lin, B. Sun, Effect of strain rate on spatio-temporal behavior of Portevin-Le Châtelier bands in a twinning induced plasticity steel, Mech. Mater. 68 (2014) 164–175. https://doi.org/10.1016/j.mechmat.2013.09.002.

[10] B.C. De Cooman, O. Kwon, K.G. Chin, State-of-the-knowledge on TWIP steel, Mater. Sci. Technol. 28 (2012) 513–527. https://doi.org/10.1179/1743284711Y.0000000095.

[11] G. Dini, R. Ueji, A. Najafizadeh, S.M. Monir-Vaghefi, Flow stress analysis of TWIP steel via the XRD measurement of dislocation density, Mater. Sci. Eng. A. 527 (2010) 2759–2763. https://doi.org/10.1016/j.msea.2010.01.033.

[12] D.M. Bastidas, J. Ress, J. Bosch, U. Martin, Corrosion mechanisms of high-mn twinning-induced plasticity (Twip) steels: A critical review, Metals (Basel). 11 (2021) 1–45. https://doi.org/10.3390/met11020287.

[13] B.C. De Cooman, K. Chin, J. Kim, High Mn TWIP Steels for Automotive Applications, New Trends Dev. Automot. Syst. Eng. (2011). https://doi.org/10.5772/14086.

[14] X. Guo, A. Schwedt, S. Richter, W. Bleck, Effects of Al on delayed fracture in TWIP steels – discussion from the aspects of structure homogeneity, hydrogen traps and corrosion resistance, SteelyHydrogen2014 Conf. Proc. (2014) 5–7.

[15] H.K.D.H. Bhadeshia, Prevention of Hydrogen Embrittlement in Steels, ISIJ Int. 56 (2016) 24–36. https://doi.org/10.2355/isijinternational.ISIJINT-2015-430.
[16] M. Koyama, E. Akiyama, K. Tsuzaki, Hydrogen embrittlement in a Fe-Mn-C ternary twinning-induced plasticity steel, Corros. Sci. 54 (2012) 1–4. https://doi.org/10.1016/j.corsci.2011.09.022.

[17] M. Koyama, E. Akiyama, K. Tsuzaki, D. Raabe, Hydrogen-assisted failure in a twinning-induced plasticity steel studied under in situ hydrogen charging by electron channeling contrast imaging, Acta Mater. 61 (2013) 4607–4618. https://doi.org/10.1016/j.actamat.2013.04.030.

[18] B. Bal, M. Koyama, G. Gerstein, H.J. Maier, K. Tsuzaki, Effect of strain rate on hydrogen embrittlement susceptibility of twinning-induced plasticity steel pre-charged with high-pressure hydrogen gas, Int. J. Hydrogen Energy. 41 (2016) 15362–15372. https://doi.org/10.1016/j.ijhydene.2016.06.259.

[19] M. Koyama, E. Akiyama, K. Tsuzaki, Hydrogen-induced delayed fracture of a Fe-22Mn-0.6C steel pre-strained at different strain rates, Scr. Mater. 66 (2012) 947–950. https://doi.org/10.1016/j.scriptamat.2012.02.040.

[20] L. Ismer, T. Hickel, J. Neugebauer, Ab initio study of the solubility and kinetics of hydrogen in austenitic high Mn steels, Phys. Rev. B - Condens. Matter Mater. Phys. 81 (2010) 20–23. https://doi.org/10.1103/PhysRevB.81.094111.

[21] T. Hickel, R. Nazarov, E.J. McEniry, G. Leyson, B. Grabowski, J. Neugebauer, Ab initio based understanding of the segregation and diffusion mechanisms of hydrogen in steels, Jom. 66 (2014) 1399–1405. https://doi.org/10.1007/s11837-014-01055-3.

[22] M. Koyama, E. Akiyama, Y.K. Lee, D. Raabe, K. Tsuzaki, Overview of hydrogen embrittlement in high-Mn steels, Int. J. Hydrogen Energy. 42 (2017) 12706–12723. https://doi.org/10.1016/j.ijhydene.2017.02.214.

[23] B. Sun, D. Wang, X. Lu, D. Wan, D. Ponge, X. Zhang, Current Challenges and Opportunities Toward Understanding Hydrogen Embrittlement Mechanisms in Advanced High-Strength Steels: A Review, Acta Metall. Sin. (English Lett. 34 (2021) 741–754. https://doi.org/10.1007/s40195-021-01233-1.

[24] T. Neeraj, R. Srinivasan, J. Li, Hydrogen embrittlement of ferritic steels: Observations on deformation microstructure, nanoscale dimples and failure by nanovoiding, Acta Mater. 60 (2012) 5160–5171. https://doi.org/10.1016/j.actamat.2012.06.014.

[25] X. Lu, D. Wang, D. Wan, Z.B. Zhang, N. Kheradmand, A. Barnoush, Effect of electrochemical charging on the hydrogen embrittlement susceptibility of alloy 718, Acta Mater. 179 (2019) 36–48. https://doi.org/10.1016/j.actamat.2019.08.020.

[26] M. Koyama, S.M. Taheri-Mousavi, S.M. Taheri-Mousavi, H. Yan, J. Kim, B.C. Cameron, S.S. Moine-ardakani, J. Li, J. Li, C.C. Tasan, Origin of micrometer-scale dislocation motion during hydrogen desorption, Sci. Adv. 6 (2020). https://doi.org/10.1126/sciadv.aaz1187.

[27] S. Zaefferer, N.-N. Elhami, Theory and application of electron channelling contrast imaging under controlled diffraction conditions, Acta Mater. 75 (2014) 20–50. https://doi.org/10.1016/j.actamat.2014.04.018.

[28] M. Daamen, S. Richter, G. Hirt, Microstructure analysis of high-manganese TWIP steels produced via strip casting, Key Eng. Mater. 554–557 (2013) 553–561. https://doi.org/10.4028/www.scientific.net/KEM.554-557.553.

[29] D. Pérez Escobar, L. Duprez, A. Atrens, K. Verbeken, Influence of experimental parameters on thermal desorption spectroscopy measurements during evaluation of hydrogen trapping, J. Nucl. Mater. 450 (2014) 32–41. https://doi.org/10.1016/j.jnucmat.2013.07.006.
X. Zhu, W. Li, T.Y. Hsu, S. Zhou, L. Wang, X. Jin, Improved resistance to hydrogen embrittlement in a high-strength steel by quenching-partitioning-tempering treatment, Scr. Mater. 97 (2015) 21–24. https://doi.org/10.1016/j.scriptamat.2014.10.030.

N. Narita, C.J. Alstetter, H.K. Birnbaum, Hydrogen-Related Phase Transformations in Austenitic Stainless Steels., Metall. Trans. A, Phys. Metall. Mater. Sci. 13 A (1982) 1355–1365. https://doi.org/10.1007/BF02642872.

X. Guo, S. Zaefferer, F. Archie, W. Bleck, Hydrogen effect on the mechanical behaviour and microstructural features of a Fe-Mn-C twinning induced plasticity steel, Int. J. Miner. Metall. Mater. 28 (2021) 835–846. https://doi.org/10.1007/s12613-021-2284-4.

N. Zan, H. Ding, X. Guo, Z. Tang, W. Bleck, Effects of grain size on hydrogen embrittlement in a Fe-22Mn-0.6C TWIP steel, Int. J. Hydrogen Energy. 40 (2015) 10687–10696. https://doi.org/10.1016/j.ijhydene.2015.06.112.

H. Khanchandani, S.-H. Kim, R.S. Varanasi, T. Prithiv, L.T. Stephenson, B. Gault, Hydrogen and deuterium charging of site-specific specimen for atom probe tomography, Open Res. Eur. 1 (2021) 122. https://doi.org/10.12688/openreseurope.14176.1.

Y.S. Chen, H. Lu, J. Liang, A. Rosenthal, H. Liu, G. Sneddon, L. McCarroll, Z. Zhao, W. Li, A. Guo, J.M. Cairney, Observation of hydrogen trapping at dislocations, grain boundaries, and precipitates, Science (80-. ). 367 (2020) 171–175. https://doi.org/10.1126/science.aaz0122.

S. V. Merziklin, S. Borodin, D. Vogel, M. Rohwerder, Ultra high vacuum high precision low background setup with temperature control for thermal desorption mass spectroscopy (TDA-MS) of hydrogen in metals, Talanta. 136 (2015) 108–113. https://doi.org/10.1016/j.talanta.2015.01.014.

B. Sun, W. Krieger, M. Rohwerder, D. Ponge, D. Raabe, Dependence of hydrogen embrittlement mechanisms on microstructure-driven hydrogen distribution in medium Mn steels, Acta Mater. 183 (2020) 313–328. https://doi.org/10.1016/j.actamat.2019.11.029.

M. Koyama, S. Okazaki, T. Sawaguchi, K. Tsuzaki, Hydrogen Embrittlement Susceptibility of Fe-Mn Binary Alloys with High Mn Content: Effects of Stable and Metastable ε-Martensite, and Mn Concentration, Metall. Trans. A. Phys. Metall. Mater. Sci. 47 (2016) 2656–2673. https://doi.org/10.1007/s11661-016-3431-9.

Y.S. Chun, K.T. Park, C.S. Lee, Delayed static failure of twinning-induced plasticity steels, Scr. Mater. 66 (2012) 960–965. https://doi.org/10.1016/j.scriptamat.2012.02.038.

S. Zaefferer, G. Habler, Scanning electron microscopy and electron backscatter diffraction, Eur. Mineral. Union Notes Mineral. 16 (2017) 37–95. https://doi.org/10.1180/EMU-notes.16.3.

S. Berg, D. Kutra, T. Kroeger, C.N. Straehle, B.X. Kausler, C. Haubold, M. Schiegg, J. Ales, T. Beier, M. Rudy, K. Eren, J.I. Cervantes, B. Xu, F. Beuttenmueller, A. Wolny, C. Zhang, U. Koethe, F.A. Hamprecht, A. Kreshuk, Ilastik: Interactive Machine Learning for (Bio)Image Analysis, Nat. Methods. 16 (2019) 1226–1232. https://doi.org/10.1038/s41592-019-0582-9.

H.J. Kim, Effects of Prior Austenite Grain Size on Hydrogen Delayed Fracture of Hot-Stamped Boron Martensitic Steel, Metall. Mater. Trans. A. Phys. Metall. Mater. Sci. 51 (2020) 237–251. https://doi.org/10.1007/s11661-019-05523-3.

T. Michler, J. Naumann, Microstructural aspects upon hydrogen environment embrittlement of various bcc steels, Int. J. Hydrogen Energy. 35 (2010) 821–832. https://doi.org/10.1016/j.ijhydene.2009.10.092.
T. Kanezaki, C. Narazaki, Y. Mine, S. Matsuoka, Y. Murakami, Effects of hydrogen on fatigue crack growth behavior of austenitic stainless steels, Int. J. Hydrogen Energy. 33 (2008) 2604–2619. https://doi.org/10.1016/j.ijhydene.2008.02.067.

D. An, X. Zhang, S. Zaefferer, The combined and interactive effects of orientation, strain amplitude, cycle number, stacking fault energy and hydrogen doping on microstructure evolution of polycrystalline high-manganese steels under low-cycle fatigue, Int. J. Plast. 134 (2020) 102803. https://doi.org/10.1016/j.ijplas.2020.102803.

R. Silverstein, D. Eliezer, B. Glam, Hydrogen Effect on Duplex Stainless Steels at Very High Strain Rates, Energy Procedia. 107 (2017) 199–204. https://doi.org/10.1016/j.egypro.2016.12.172.

E.G. Astafurova, G.G. Zakharova, H.J. Maier, Hydrogen-induced twinning in $\langle 0 0 1 \rangle$ Hadfield steel single crystals, Scr. Mater. 63 (2010) 1189–1192. https://doi.org/10.1016/j.scriptamat.2010.08.029.

D. Xie, S. Li, M. Li, Z. Wang, P. Gumbsch, J. Sun, E. Ma, J. Li, Z. Shan, Hydrogenated vacancies lock dislocations in aluminium, Nat. Commun. 7 (2016) 1–7. https://doi.org/10.1038/ncomms13341.

J. Song, W.A. Curtin, Mechanisms of hydrogen-enhanced localized plasticity: An atomistic study using $\alpha$-Fe as a model system, Acta Mater. 68 (2014) 61–69. https://doi.org/10.1016/j.actamat.2014.01.008.

Y. Zhu, Z. Li, M. Huang, H. Fan, Study on interactions of an edge dislocation with vacancy-H complex by atomistic modelling, Int. J. Plast. 92 (2017) 31–44. https://doi.org/10.1016/j.ijplas.2017.03.003.

I. Gutierrez-Urrutia, S. Zaefferer, D. Raabe, The effect of grain size and grain orientation on deformation twinning in a Fe-22wt.% Mn-0.6wt.% C TWIP steel, Mater. Sci. Eng. A. 527 (2010) 3552–3560. https://doi.org/10.1016/j.msea.2010.02.041.

P. Yu, Y. Cui, G. zhen Zhu, Y. Shen, M. Wen, The key role played by dislocation core radius and energy in hydrogen interaction with dislocations, Acta Mater. 185 (2020) 518–527. https://doi.org/10.1016/j.actamat.2019.12.033.

I.M. Robertson, M.L. Martin, J.A. Fenske, Influence of hydrogen on the behavior of dislocations, Woodhead Publishing Limited, 2012. https://doi.org/10.1533/9780857095374.1.166.
[58] H. Li, H. Zong, S. Li, S. Jin, Y. Chen, M.J. Cabral, B. Chen, Q. Huang, Y. Chen, Y. Ren, K. Yu, S. Han, X. Ding, G. Sha, J. Lian, X. Liao, E. Ma, J. Sun, Uniting tensile ductility with ultrahigh strength via composition undulation, Nature. 604 (2022) 273–279. https://doi.org/10.1038/s41586-022-04459-w.

[59] A.M. Brass, J. Chêne, Hydrogen uptake in 316L stainless steel: Consequences on the tensile properties, Corros. Sci. 48 (2006) 3222–3242. https://doi.org/10.1016/j.corsci.2005.11.004.

[60] J.D. Hermida, A. Roviglione, Stacking fault energy decrease in austenitic stainless steels induced by hydrogen pairs formation, Scr. Mater. 39 (1998) 1145–1149. https://doi.org/10.1016/S1359-6462(98)00285-1.

[61] D.P. Abraham, C.J. Altstetter, The effect of hydrogen on the yield and flow stress of an austenitic stainless steel, Metall. Mater. Trans. A. 26 (1995) 2849–2858. https://doi.org/10.1007/BF02669643.

[62] V.G. Gavriljuk, V.N. Shivanyuk, J. Foct, Mechanisms of Hydrogen Embrittlement of Austenitic Stainless Steels, J. Mech. Behav. Mater. 16 (2011) 163–176. https://doi.org/10.1515/jmbm.2005.16.3.163.