

Lath martensite plasticity enabled by apparent sliding of substructure boundaries

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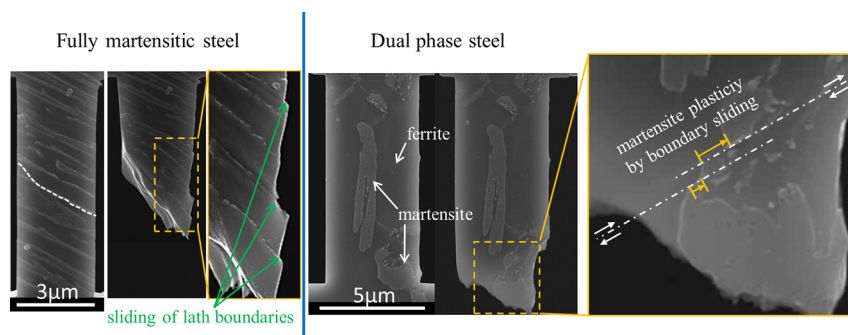
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HIGHLIGHTS

- Apparent sub-structure boundary sliding in lath martensite occurs even for sub-optimal loading orientations.
- The required shear stress to initiate apparent boundary sliding is lower than for intra-lath crystallographic slip.
- Apparent boundary sliding seems also a relevant martensite plasticity mechanism in engineering steels.

GRAPHICAL ABSTRACT



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ABSTRACT

Lath martensite is widely present in advanced high strength steels as the key strengthening phase. Unexpectedly high ductility of lath martensite has been reported in the literature in both single-phase and multi-phase steels, however, without systematic identification of the underlying plasticity mechanisms. In this study, first, well-defined micro-tensile tests are carried out on fully martensitic steel with a clean large substructure and a variety of substructure boundary orientations with respect to the loading direction. Two deformation mechanisms of lath martensite were identified, namely, intra-lath crystallographic slip and apparent substructure boundary sliding, that compete with each other to carry the overall plasticity. The condition under which these two mechanisms are active has been clarified. It is found that boundary sliding is more easily activated than intra-lath crystallographic slip. In addition, for dual phase steel, as an example of multi-phase steels, the probability for sliding of lath martensite boundaries was estimated by boundary orientation characterization and micro-tensile tests. The results suggest that the apparent boundary sliding is also important for lath-martensite-containing multi-phase steels, which would explain prior reports in the literature of unexpectedly high local strains in the martensite regions.

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Introduction

Among all types of ferrous martensite, lath martensite can be found in a large variety of commercial steels such as dual phase (DP) steel, transformation induced plasticity (TRIP) steel, maraging steel and quenching

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and partitioning steel where it serves as a strengthening phase. Its strengthening mechanisms include dislocation hardening [1,42], solid solution hardening [22], precipitation strengthening [22,36] and substructure boundary strengthening [8,14,30,32,3,35,36,40,49] which has the most important contribution to the strength of lath martensite. On the other hand, similar to most strong materials, lath martensite is generally known to be brittle [4,16,18,22,28,31]. Beside the tetragonal distortion caused by the super saturated interstitial carbon atoms, one origin of its brittleness has been attributed to sub-structure boundary (region) related mechanisms, either due to segregation of phosphorus [4,5] or due to tempering embrittlement [3,16,18,22,28].

On the contrary, apparent ductility of lath martensite has been reported in the literature. Ghadbeigi et al. [13] showed evidence of remarkably high strains in martensite within DP steel up to 120%. Similar observations were reported by [19,20] with local martensite strain exceeding 70%. Moreover, the fracture surfaces of DP steel show dimples [6,7,20,43,44], which is a signature of ductile fracture, also in the martensite islands [6,17,44,48]. Finally, in single-phase lath martensite steels, the overall fracture strain can still be as large as 13% with dimples at the fracture surfaces [37,22], and even 20% in sandwiched specimens [29,34]. Therefore, in both single-phase and multi-phase steels, lath martensite may reveal (much) higher ductility than what is usually expected for a brittle material.

In order to investigate the ductility of lath martensite, we performed uniaxial tensile tests on micro-specimens of lath martensite cut out of a single packet/block [8,9]. Specimens with a specific orientation of the martensite substructure boundaries were selected, yielding the maximum resolved shear stress on the boundary planes, i.e. boundaries perpendicular to the specimen front surface and with an angle of $\sim 45^\circ$ with respect to the loading direction. These micro-tensile tests revealed a novel apparent boundary sliding mechanism in lath martensite, as shown for instance in Fig. 1, which is reproduced from Ref. [9]. The sliding clearly occurs at certain lath boundaries, as indicated by the white arrows at the steps. From this and other tests, it was established that this sliding mechanism can be active at all types of tested boundaries, lath, sub-block, and block boundaries, and that it is in competition with intra-lath crystallographic slip [9]. However, as stated above, these micro-tests on martensite boundaries were only performed at

the most favorable orientation for boundary sliding, hence, whether or not boundary sliding is important under sub-optimal orientations or in bulk (multi-phase) steels remains an open question.

Therefore, the goal of this work is to study the condition under which the apparent boundary sliding mechanism is activated and to investigate the relevance of this mechanism in engineering (multi-phase) steels containing lath martensite. In particular, the following questions will be addressed. (1) Does this apparent boundary sliding also occur for random boundary orientations? (2) Under which conditions is this mechanism active? (3) In a multi-phase steel, is the sliding promoted or demoted by the morphology and orientation of the substructure boundaries in the lath martensite islands? And (4), how important is this sliding mechanism as a plastic deformation mechanism in lath-martensite-based (multi-phase) steels?

In an effort to provide conclusive answers to these questions, the research focuses on two lath-martensite-containing engineering steels with the same overall chemical composition, a fully martensitic steel and a DP steel, which are at the extremes of the spectrum with respect to (i) strength vs. ductility, (ii) martensite volume fraction, and (iii) external loading constraint acting on the martensite domains in the microstructure. The fully martensitic steel has a cleaner substructure arrangement enabling clear microstructural, crystallographic and micro-mechanical analysis [4,8,9,30,29,34,40]. The DP steel is a well-known example of lath-martensite-based multi-phase steels, for which rich data is available in the literature on all aspects, see e.g. Refs. [6,7,13,14,19,20,43,44,46].

The investigation starts with a more extensive study of the sliding mechanism using micro-tensile tests of specimens with unfavorably oriented boundaries, including boundaries that are not perpendicular to the front surface and/or at angles other than $\sim 45^\circ$ to the loading direction. Subsequently, the activation threshold for the sliding mechanism is analyzed. Next, the substructure boundaries of lath martensite in the multi-phase steel are analyzed to assess the activation potential of the sliding mechanism. And, finally, micro-tensile tests of DP steel specimens with embedded lath martensite islands are conducted to determine the activity of the sliding mechanism in a multi-phase steel, after which the relevance of this mechanism for bulk metals is discussed.

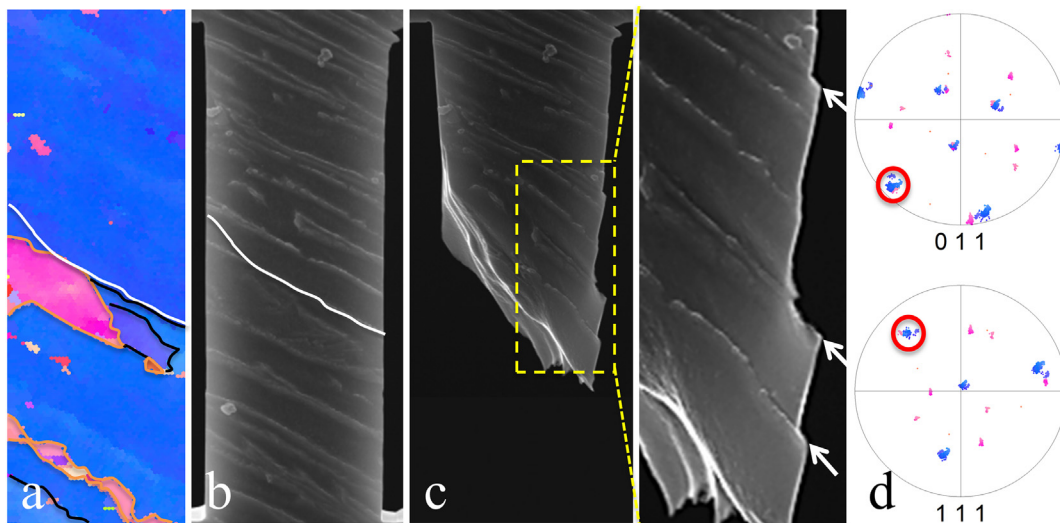


Fig. 1. The gauge part of a $9\ \mu\text{m}$ -long lath martensite micro-tensile specimen, reproduced from [9]. This specimen is produced from the same piece of bulk materials as all the micro-tensile specimens of the current work. (a) Inverse pole figure map (IPF) on which the substructure boundaries are marked with colors: Orange and black lines mark the block and sub-block boundaries, while the white line marks the fracture surface. Block boundaries and sub-block boundaries are differentiated by the different misorientation angles across them, as identified from the pole figures in (d). (b,c) Scanning electron microscopy (SEM) secondary electron (SE) images of the gauge section before and after fracture, together with a zoom of the area in the yellow dashed frame. The white arrows indicate the positions where apparent boundary sliding occurred. (d) {011} and {111} pole figures of the specimen, in which the slip system with the highest Schmid factor is marked by red circles.

Material and methodology

Two materials are used in this study, a DP steel (0.092C-1.68Mn-0.24Si-0.57Cr) with ~25% lath martensite volume fraction and a fully martensitic steel obtained by heating the DP steel for 2 h at 1000 °C, followed by water quenching. The heat treatment transforms the DP steel into a fully martensitic steel with clean and large substructures which makes it straightforward to analyze the deformation behavior. Fig. 2(a,b) shows the microstructure of two materials.

Adopting the micro-tensile test methodology described in Ref. [10], multiple-parallel micro-tensile specimens of the two materials are tested. The specimens are fabricated using focused ion beam (FIB) milling at the tip of a wedge that is prepared from the bulk material. Since the main focus is on the block and sub-block boundaries, single-packet specimens have been fabricated from large packets. The milling parameters and procedures have been carefully chosen to minimize the influence of Ga^+ ions on the specimens. The gauge part of the fully martensitic specimens is 9 μm in length with a cross section of $\sim 3 \times 2 \mu\text{m}^2$. Fig. 2(c,d) show examples of micro-tensile specimens of DP steel and fully martensitic respectively. Each specimen is characterized using scanning electron microscopy (SEM) and electron backscattered diffraction (EBSD) from both the front and backside of the specimen (note, that only the front side images are shown in this paper). The tensile tests are conducted with a micro-force tensile stage with accurate specimen alignment (<0.1 mrad angular alignment and near-perfect co-linearity) and precise force (force range of 0.07 μN to 250 mN) and displacement (<6 nm reproducibility) resolution [2].

Results and discussion

In total 15 fully martensitic steel micro-specimens with clean substructures have been tested and grouped by the types of deformation exposed. The specimens that showed fracture surfaces along the boundary traces are presented in Fig. 3, whereas the specimens that fracture in other directions are shown in Fig. 4. For each specimen, from left to right, the following data is shown: (i) a backscattered electron (BSE) SEM image of the untested specimen, which shows the boundary traces due to the electron channeling contrast; (ii) and (iii) an SE and BSE image of the fractured specimens respectively; (iv) an inverse pole figure (IPF) map from EBSD measurement of the untested specimen, with substructure boundaries marked in the same colors as used in Fig. 1 and the fracture surface marked also in white; (v) the $\{110\}$ and $\{111\}$ pole figures of the specimen gauge section.

The substructure boundaries are parallel to the habit plane of phase transformation, i.e. one of the $\{110\}$ planes of lath martensite (bcc) or one of the $\{111\}$ planes of the parent austenite (fcc) [24]. Therefore, the $\{110\}$ pole figures of lath martensite can be used to determine the substructure boundary orientations. Since for the lath martensite used in our study, it was demonstrated that the $\{112\}\langle 111 \rangle$ or $\{123\}\langle 111 \rangle$ slip systems are not activated [8], here only the $\{110\}\langle 111 \rangle$ slip systems are considered for intra-lath crystallographic slip. Out of the twelve $\{110\}\langle 111 \rangle$ slip systems, there are always two slip systems parallel to the boundary planes, which will be referred to as the in-plane bcc slip systems [29]. The in-plane system with the higher Schmid factor is marked in red in the $\{110\}$ and $\{111\}$ pole figures in Fig. 3 and Fig. 4. In addition, the other ten slip systems are distributed over the five $\{110\}$ planes that are not parallel to the substructure boundaries, which will

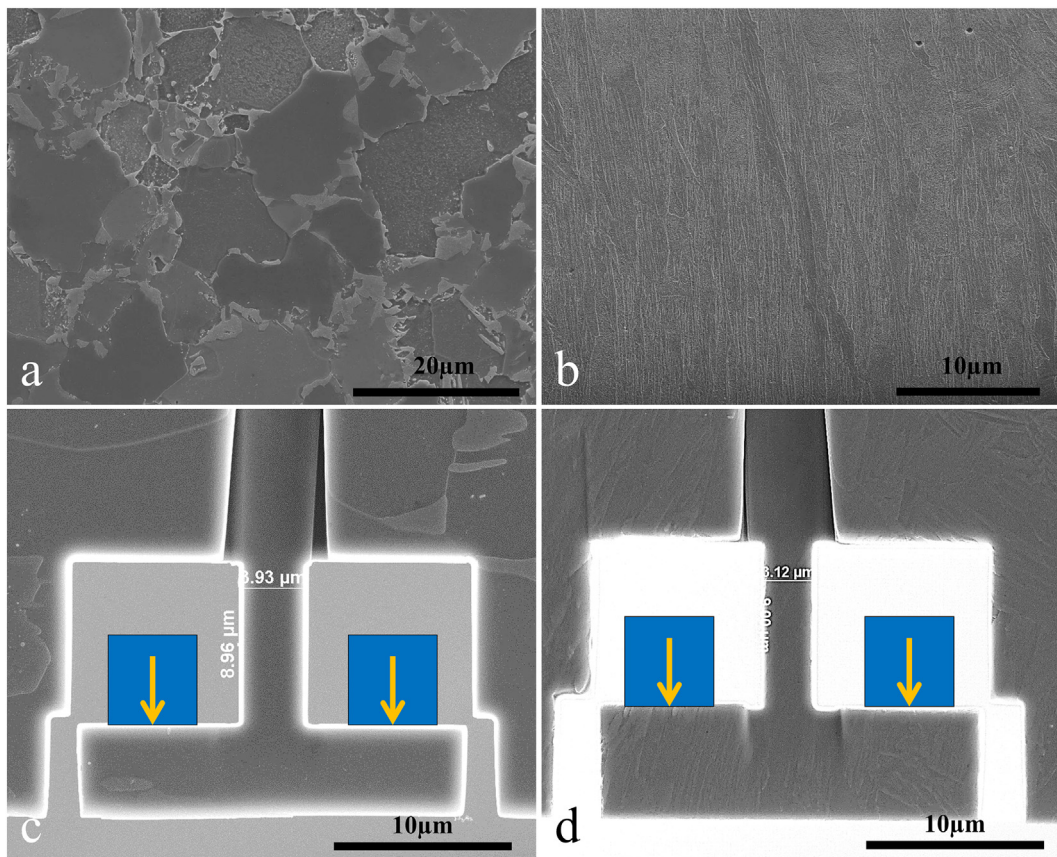


Fig. 2. (a) SEM SE image of the microstructure of DP steel tested in this study. (b) SEM SE image of the microstructure of fully martensitic steel. A region with parallel laths is selected, which is large enough for the fabrication of a micro-tensile specimen. (c,d) Examples of micro-tensile specimens extracted from DP steel and fully martensitic steel, respectively. The two blue blocks represent the pins of the loading gripper which pulls the specimen cross-bar downwards (in the images) to apply a uni-axial loading to the specimen $\sim 9 \mu\text{m}$ -long gauge section, as detailed in Ref. [2,10].

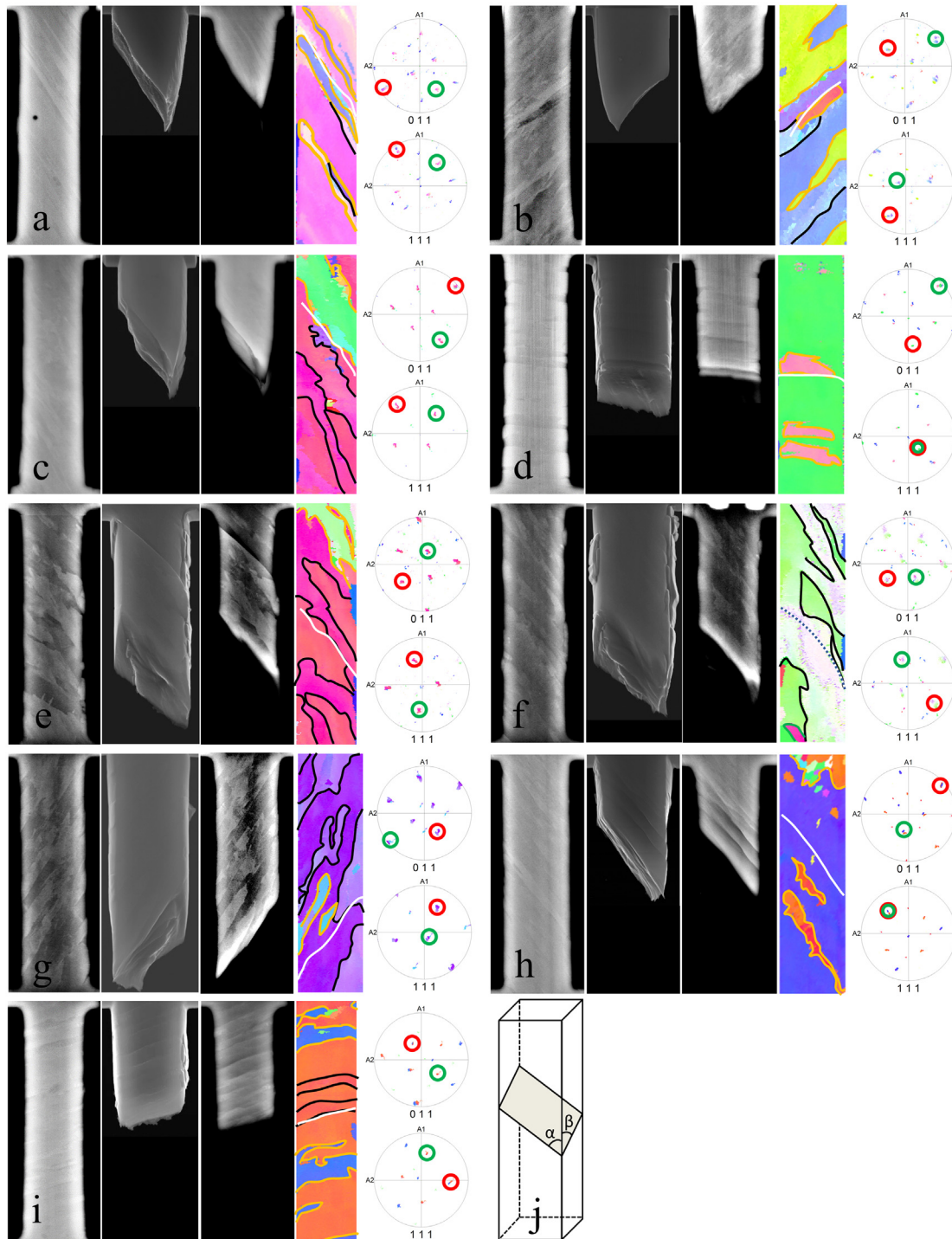


Fig. 3. The nine specimens (Specimen S3 a-S 3i) that fractured along the substructure boundaries, which is attributed to apparent boundary sliding (see text). The initial length of the specimens is 9 μm . For each specimen, the following data is shown, from left to right: a backscattered electron (BSE) image of the original specimen gauge section; an SE and BSE image of the fractured gauge section, respectively; an IPF map of the gauge section, with the substructure boundaries marked in different colors: block boundaries in orange, sub-block boundaries in black, lath boundaries left unmarked, and fracture surfaces marked in white; and the $\{110\}$ and $\{111\}$ pole figures of the gauge section. The red and green circles indicate, respectively, the in-plane and out-of-plane slip system with the highest Schmid factor. Note that the white fracture line is copied from the deformed specimen onto the IPF map of the undeformed specimen (only in specimen S3f a blue-white dashed line is used to contrast the fracture line with the EBSD map). (j) Illustration of the angle definitions, used in Table 1, of the lath martensite boundary planes. The shaded plane represents the substructure boundaries between laths in a packet of lath martensite, where α is the angle between the trace of the boundary plane on the specimen front surface and the loading direction, which is the specimen length direction, and β is the angle between the boundary trace on side surface of the specimen and the load axis. Note that, although the SE and BSE images are recorded simultaneously, their magnification may appear different, because too few primary electrons are back-scattered into the BSE detector from the slightly tapered specimen sides and rounded-off edges to observe them, in contrast to SE mode where even the very steep side surface is visible.

be referred to as the out-of-plane bcc slip systems [29]. The out-of-plane bcc system with the highest Schmid factor is marked with green circles in the $\{110\}$ and $\{111\}$ pole figures. These here-named ‘maximum

Schmid factors of the out-of-plane bcc slip systems’ and the ‘Schmid factors of the in-plane bcc slip systems’ of martensite ($2\times$) are presented in Table 1 and Table 2.

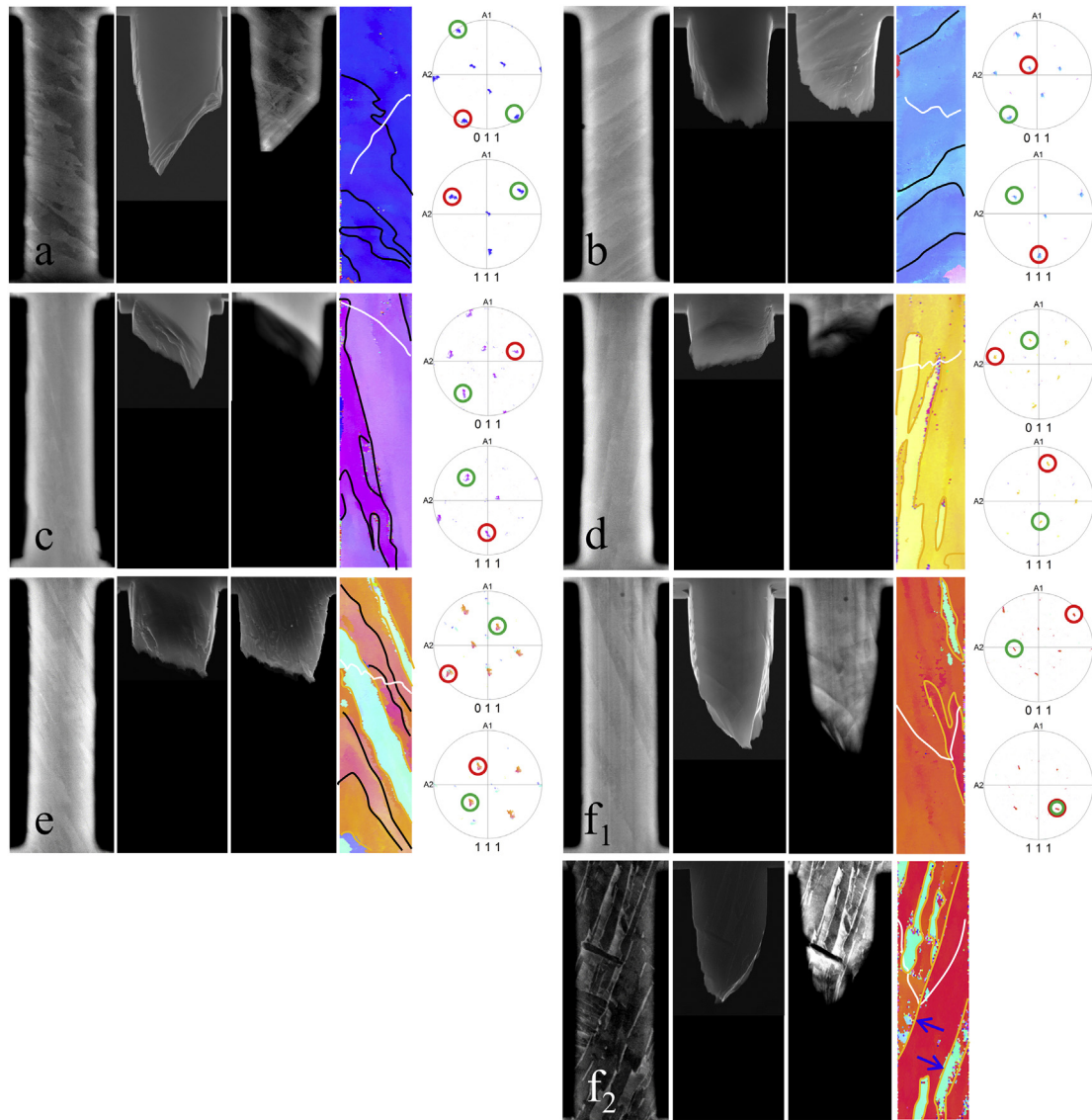


Fig. 4. The six 9 μm -long specimens (Specimen S4a – S4f) that show a fracture surface that cuts through the substructure boundaries. For each specimen, the following data is shown, from left to right: a BSE image of the original specimen gauge section; an SE and BSE image of the fractured gauge section respectively; an IPF map of the gauge section, with substructure boundaries marked in different colors: block boundaries in orange, sub-block boundaries in black, lath boundaries left unmarked, and fracture surfaces marked in white; and the $\{110\}$ and $\{111\}$ pole figures of the gauge section. The red and green circles indicate, respectively, the in-plane and out-of-plane slip system with the highest Schmid factor. Note that the white fracture line is copied from the deformed specimen onto the IPF map of the undeformed specimen. The backside of specimen (f_1) is given in (f_2). The fracture surface and substructure boundaries are marked the same as (a-f). The blue arrows indicate two boundaries that connect to the specimen base and specimen cross-bar, thereby constraining these boundaries for sliding.

For none of the specimens, shown in Fig. 3, in-plane slip traces inside the lath were observed on the top surface or side surfaces, therefore, it is

fair to conclude that the dominant plastic deformation that contributes to final failure does not occur inside the laths but along the substructure

Table 1

Characteristics of the micro-specimen (S1) in Fig. 1 and the nine specimens (S3a to S3i) of Fig. 3 with, from top to bottom: the fracture stress; the maximum Schmid factors (SF) of the out-of-plane systems; The SF of the two in-plane slip systems of martensite, with the maximum one in bold; the SF of the three in-plane slip systems of the possibly existed inter-lath retained austenite films, with the maximum one in bold; the upper limit of SF for boundary sliding; the angle between the boundary trace on front surface of the specimen and the load axis (α); and the angle between the boundary trace on side surface of the specimen and the load axis (β) (for the definition of α and β , see Fig. 3(j)).

Specimen No.	S1	S3a	S3b	S3c	S3d	S3e	S3f	S3g	S3h	S3i
Fracture stress (MPa)	760	858	871	875	842	779	799	859	791	944
Max. SF of out-of-plane bcc slip systems (—)	0.45	0.50	0.48	0.48	0.34	0.42	0.44	0.48	0.43	0.42
SF of the two in-plane bcc slip systems (—)	0.50	0.46	0.45	0.46	0.41	0.45	0.45	0.49	0.43	0.46
(Max. SF in bold)	0.13	0.17	0.26	0.30	0.10	0.02	0.33	0.15	0.08	0.02
SF of the three in-plane fcc slip system (—)	0.50	0.46	0.45	0.46	0.41	0.45	0.45	0.49	0.43	0.46
(Max. SF in bold)	0.28	0.27	0.12	0.37	0.04	0.41	0.39	0.26	0.01	0.35
	0.21	0.19	0.33	0.09	0.35	0.01	0.06	0.23	0.42	0.11
Upper limit of SF for boundary sliding	≤ 0.50	≤ 0.46	≤ 0.46	≤ 0.49	≤ 0.48	≤ 0.47	≤ 0.48	≤ 0.49	≤ 0.49	≤ 0.48
α (°)	46	34	−46	39	−83	42	62	−54	41	−61
β (°)	−77	−81	51	87	−54	−59	−61	−53	85	43

Table 2

Characteristics of the six specimens (S4a to S4f) of Fig. 4 with, from top to bottom: the fracture stress; the maximum Schmid factors (SF) of the out-of-plane systems; The SF of the two in-plane slip systems of martensite, with the maximum one in bold; the SF of the three in-plane slip systems of the possibly existed inter-lath retained austenite films, with the maximum one in bold; the upper limit of SF for boundary sliding; the angle between the boundary trace on front surface of the specimen and the load axis (α); and the angle between the boundary trace on side surface of the specimen and the load axis (β) (for the definition of α and β , see Fig. 3(j)).

Specimen No.	S4a	S4b	S4c	S4d	S4e	S4f
Fracture stress (MPa)	873	874	857	1053	1013	954
Max. SF of out-of-plane bcc slip systems (—)	0.45	0.40	0.47	0.46	0.48	0.45
SF of the two in-plane bcc slip systems (—)	0.38	0.23	0.22	0.12	0.42	0.46
(Max. SF in bold)	0.03	0.10	0.15	0.09	0.34	0.34
SF of the three in-plane fcc slip system (—)	0.38	0.23	0.22	0.12	0.42	0.46
(Max. SF in bold)	0.29	0.14	0.04	0.02	0.02	0.41
	0.10	0.10	0.18	0.10	0.39	0.05
Upper limit of SF for boundary sliding	≤ 0.40	≤ 0.23	≤ 0.24	≤ 0.13	≤ 0.47	≤ 0.50
α (°)	64	−42	17	−7	35	−46
β (°)	−87	21	58	77	−80	−84

boundaries. The question remains: what is the cause of failure exactly at or along the substructure boundaries? Two possible mechanisms can explain this apparent boundary sliding phenomenon, depending on the presence of ultra-thin inter-lath retained austenite films between the laths: (i) crystallographic slip inside the inter-lath retained austenite films, which functions as a 'greasy' plane to mediated the relative sliding of the martensite laths [25]. Note that, while continuous uninterrupted inter-lath retained austenite films have been reported in the literature (e.g. [38]), the austenite films may also be nonuniform and interrupted [33], in which case they would still be the weakest boundary sections for plastic deformation to start, triggering subsequent plasticity in the intermediate austenite-free boundary sections. (For the blocky retained austenite, which has a random orientation relationship with respect to the neighboring martensite, this mechanism may not work because of the in-plane slip systems of martensite do not align with any of the austenite slip systems.) And (ii) sliding of the phase or grain boundary itself, induced by motion of, respectively, bcc-fcc or bcc-bcc boundary dislocations, respectively, in the presence or absence of inter-lath retained austenite [26,27], or a combination of both mechanisms.

The presence of retained austenite at the boundaries has been found by transmission electron microscopy (TEM) studies of lath martensite, in the form of blocks but also as ultra-thin (~10 nm thick) inter-lath films between the laths [1,21,33,38]. Recently, it was shown using crystal plasticity simulations that the presence of such ultra-thin austenite films might explain sliding of the boundaries through mechanism (i) because the activation of austenite (fcc) slip systems is much easier than those of martensite (bcc), since three austenite slip systems are always aligned with the habit plane (and thus with the boundaries) [25]. To this end, multiple attempts were conducted to reveal the presence of the inter-lath retained austenite films in our samples using high-resolution TEM. Unfortunately, for the two steels investigated here, a positive identification of the austenite films could not be made, because of their very small thickness, consistent with the comments in [39] that thin inter-lath retained austenite films are difficult to identify. This leaves open the possibility for both boundary plasticity mechanisms ((i) and (ii)). It is therefore necessary to calculate Schmid factors of the slip systems for both mechanisms, along with that of intra-lath crystallographic slip, as all three can carry the plasticity of lath martensite.

For crystallographic slip inside the inter-lath retained austenite, the orientations of the three austenite slip systems which lie in the boundary plane between the martensite laths can be obtained from the K—S orientation relationship ($\{111\}_{\text{fcc}} // \{011\}_{\text{bcc}}$, $\{10-1\}_{\text{fcc}} //$

$\langle 11-1 \rangle_{\text{bcc}}$) as follow. First, the orientations of the 12 $\{110\} \langle 1-11 \rangle$ bcc slip systems can be directly measured by EBSD technique. Out of these 12 systems, one is parallel to an in-plane $\{111\} \langle 1-10 \rangle$ fcc system and this parallel system is shared by more than one variant from the same packet, which can be identified from the overlaying projections (of multiple colors) in the $\{110\}$ and $\{111\}$ pole figures. The other two in-plane systems of austenite can be obtained by rotation of this shared $[110]$ fcc (or $[111]$ bcc) direction around the normal of the boundary plane for 60 and 120° respectively. These so-called 'Schmid factors of the in-plane fcc slip systems' of austenite ($3\times$) are given in Table 1 and Table 2.

For the case of sliding of the phase or grain boundary itself (mechanism (ii)), calculation of the relevant Schmid factors is unfeasible, as it remains unknown whether this would be caused by bcc-fcc or bcc-bcc boundary sliding, whereas the precise boundary configuration with its specific boundary dislocations arrangement should be known to enable a correct Schmid factor calculation. Nevertheless, assuming that the in-plane slip direction is unknown, then the upper limit of the Schmid factor for boundary sliding is equal to the highest Schmid factor in the slip plane. This Schmid factor is called here the 'upper limit of the Schmid factor for boundary sliding' and its values are added in Tables 1 and 2.

Interestingly, for the specimens of Fig. 3 in which apparent boundary sliding is observed, the highest 'Schmid factors of the in-plane fcc slip systems' are very close to the 'Upper limit of SF for boundary sliding'. Therefore, it will not be possible to identify from these tests which of the two apparent boundary sliding mechanism is active, or both. However, the fact that the SF values are so close for the specimens of Fig. 3 does enable an investigation of the competition between intra-lath crystallographic slip and apparent boundary sliding, without knowing the latter's underlying cause. In the following, to enhance readability, only the 'maximum Schmid factors of the out-of-plane bcc slip systems' and the maximum 'Schmid factor of the in-plane bcc slip systems' are compared, realizing that the resulting observation and conclusions regarding the activation of apparent boundary sliding does not rule out the second mechanism for apparent boundary sliding, i.e. sliding of phase or grain boundaries.

Most of the specimens in Fig. 3 (all except S3a, S3c and S3h) have their boundary planes clearly not perpendicular to the specimen front surface (β much different from $\pm 90^\circ$), as evident by the fact that the projection points marked by the red circles in the $\{110\}$ pole figure do not touch the periphery of the projection circle. In addition, the angles of the boundary surface traces, α , deviate considerably from $\pm 45^\circ$ for most specimens. This is the main difference between the samples tested in this study compared to the ideal specimens tested in our previous study [9], which all had $\alpha \approx \pm 45^\circ$ and $\beta \approx \pm 90^\circ$, as shown e.g. by the red marker on the pole figure periphery in Fig. 1.

From Table 1 it can be seen that specimen S3 a, S3 b, and S3 c all contain an out-of-plane slip system with a higher Schmid factor than the two in-plane slip systems. Therefore, in case of crystallographic slip, this out-of-plane slip system should be activated first. However, for none of these three specimens, out-of-plane slip activity could be observed. In addition, no in-plane slip traces are visible inside the laths, which would be parallel to the boundary surface traces, nor are there other indications of in-plane slip, e.g., on the side of the specimens. In contrast, the fracture surfaces of these specimens are along one of the substructure boundaries, in all three cases along a block boundary, as indicated by the white line in the IPF map (The exact fracture position is determined with the help of the contrast of substructures obtained in the backscattered electron image and the electron backscatter diffraction inverse pole figure map.). Therefore, there must be a preferential deformation mechanism that occurs at the boundaries of lath martensite substructures. This is the apparent sliding mechanism as was already identified in the previous study (see, e.g., Fig. 1) [9]. Considering that, for S3 a and S3 b, the maximum out-of-plane SF is higher than the maximum in-plane SF, these new measurements indicate that this sliding

mechanism is easier activated than intra-lath crystallographic slip. Moreover, specimens S3a – S3c all have a non-optimal boundary orientation in the sense that $\alpha \approx \pm 45^\circ$ and $\beta \approx \pm 90^\circ$ are not fulfilled simultaneously, nevertheless, the deformation is still predominantly mitigated by boundary sliding.

For specimen S3d, S3e and S3f, it is less of a surprise that apparent boundary sliding is the main deformation mechanism, since the max SF for the in-plane systems is higher than the maximum SF for out-of-planes. Still, these specimens are interesting because of their surface orientation is further from $\alpha \approx \pm 45^\circ$ and $\beta \approx \pm 90^\circ$ compared to specimens S3a – S3c while still boundary sliding is activated.

For specimen S3g and S3h the out-of-plane slip systems have Schmid factors that are approximately equal to that of the in-plane slip systems. Given the high density of initial dislocations and dislocation sources in lath martensite, it is logical to assume that, when plasticity were to be carried solely by intra-lath crystallographic slip, both in-plane and out-of-plane systems should be activated roughly equally to accommodate the plastic deformation, probably resulting in a high density of slip planes that are distributed rather homogeneously or concentrated slip steps within the laths. However, for specimens S3g all deformation is concentrated at the fracture surface, while specimen S3h does show a few distinct slips or sliding traces but these are all nicely aligned with the lath boundaries and parallel to the fracture surface (see the parallel boundary traces in the BSE image of the undeformed specimens). A similar story holds for specimen S3i, i.e. it shows a few distinct slips or sliding traces parallel to the lath boundaries and the fracture surface is along a sub-block boundary. Therefore, the sliding mechanism also seems to be the dominant plastic mechanism for specimen S3g, S3h and S3i.

Finally, as argued above, intra-lath crystallographic slip would probably result in a rather homogeneous, high-density distribution of surface traces and steps that can't be discriminated in SEM images, therefore, the effect of intra-lath crystallographic slip would only be visible from a gradual, overall deformation of the specimen shape. Indeed, specimens S3a, S3b, S3c and S3g seem to show an indication of (a small degree of) bending, which is attributed to crystallographic slip activity within laths. This agrees with the conclusion from Ref. [9] that apparent boundary sliding and intra-lath crystallographic slip compete with each other to carry the overall specimen plasticity.

A result of the competition between apparent boundary sliding and intra-lath crystallographic slip is that there are also cases where boundary sliding is not activated. Fig. 4 shows six specimens with substructure boundaries tilted to the loading direction with different angles. The boundary types are again marked by the same colors as in Fig. 3. Specimens S4a – S4c are single-block specimens while the others are multi-block specimens. The in-plane slip system and the out-of-plane system with the highest Schmid factors are again indicated with red and green circles respectively. The Schmid factors of the samples are given in Table 2.

Contrary to the specimens in Fig. 3, the specimens in Fig. 4 all fractured by the out-of-plane slip systems (green systems in the pole figure), which cut through the substructure boundaries (including the many lath boundaries, which are not indicated in the IPF maps), as indicated by the white lines in the IPF maps. This is reflected in the fact that the averaged fracture stress of all specimen in Fig. 4 (937 MPa) is much higher than that in Fig. 3 and Fig. 1 (838 MPa), see Tables 1 and 2. This is mainly caused by the inhibition of the easier boundary sliding mechanism in the specimens of Fig. 4 due to the unfavorable substructure boundary orientations. All six specimens of Fig. 4 except specimen S4f have a much higher maximum SF of the out-of-plane slip systems compared to the maximum in-plane SF (Table 2). Among the three single-block specimens (S4a – S4c), it is not surprising that the sliding mechanism is not activated in S4b and S4c since the resolved shear stress in the boundary planes is significantly lower than that of out-of-plane crystallographic slip. Moreover, the fracture surface is straight for specimens S4a and S4c, indicating that the dislocations are able to pass through

the sub-block and lath boundaries. This observation agrees with the micro-tests on other single-block specimens in Ref. [8].

The multi-block specimens S4d – S4f all show zigzag fracture surfaces. Further analysis reveals that the turning points of the fracture surfaces correspond to the positions of the block boundaries in general, which agrees with observations that the dislocations cannot move through block boundaries [8,40]. The zig-zag shape of the fracture surface is built up by the small sections of fracture surface in the individual blocks each having a different trace angles with respect to the loading direction, in line with the observations in [8]. For specimen S4d and S4e, due to the low Schmid factors for the in-plane slip systems, the out-of-plane slip systems are activated, even though the dislocations need to pass through the block boundaries, which are stronger barriers to dislocation motion than the sub-block boundaries [8]. Specimen S4f has high Schmid factor for the in-plane slip systems, it is therefore expected to be deformed by boundary sliding at the first glance. However, since this is the only specimen with the backside (shown in Fig. 4(f₂)) much different from the front side, the mechanical behavior should be analyzed with the backside also taken into consideration. The boundaries in the middle part of the specimen are indicated by two blue arrows, both of which has one end runs into either the specimen base or specimen cross bar. These two boundaries are therefore constrained and not free. This could be the reason that boundary sliding did not occur even with high boundary sliding Schmid factor. In conclusion, for both single-block and multi-block cases shown in Fig. 4, sliding is not possible due to the low resolved stress on the boundary planes, forcing crystallographic slip to carry the plastic deformation. The different mechanisms of the specimens in Fig. 3 and Fig. 4 agree with the recent work of Kwak et al., which clearly showed the strength difference of lath martensite specimens with different substructure boundary orientations with respect to the loading direction, which they named “anisotropy of strength” [23].

The key factor that determines the main deformation mechanism is the difference in Schmid factor between the in-plane systems (which indicates the possibility for the apparent sliding) and the out-of-plane systems (Tables 1 and 2). That is, neither the angle between the boundary plane and the specimen front surface (β) alone nor the trace angle (α) alone determines the dominant deformation mechanisms, namely crystallographic slip or apparent sliding of the boundaries. For instance, while specimen S3a, S3c, S3h and S4a each have their block boundaries perpendicular to the front surface ($\beta \approx \pm 90^\circ$), the former three specimens activated boundary sliding while the latter did not. In addition, the boundary type is also important because it determines the threshold stress at which dislocations can across a certain boundary. A lath boundary is the weakest in terms of blocking dislocation movement, whereas the sub-block boundaries are stronger and block boundaries even stronger [8]. This explains the clear difference in strength between specimens S4a – S4c and S4d – S4e. For boundaries strongly impeding crystallographic slip, the sliding mechanism might still be more favorable even with less favorable orientations, as observed in specimen S3a and S3b.

To obtain a clearer view on threshold for apparent boundary sliding activation and its variability, all experimental data is gathered in Fig. 5, which plots upper limit of SF for apparent boundary sliding (a) F and max. fcc in-plane SF (b) (which is equal to Max. bcc in-plane SF) against max. out-of-plane SF. The specimens in Fig. 3 and Fig. 4 are marked as circles, while the specimens tested in the previous works are marked as crosses [8,9]. The first group of data points, marked in red, represent specimens that show substructure boundary sliding as the main deformation mechanism. Since the specimens from ([9]), marked with red crosses, were selected to have the optimum 3D inclination for boundary sliding, they typically have higher Max. fcc in-plane SF than the ones in Fig. 3. This is reflected by the fact that the red crosses are generally higher than the red circles in the scatter plotting. Similarly, the second data point group (the blue data points) correspond to those specimens whose plastic deformation was mostly carried by crystallographic slip. The specimens from the previous work [8], marked with blue crosses,

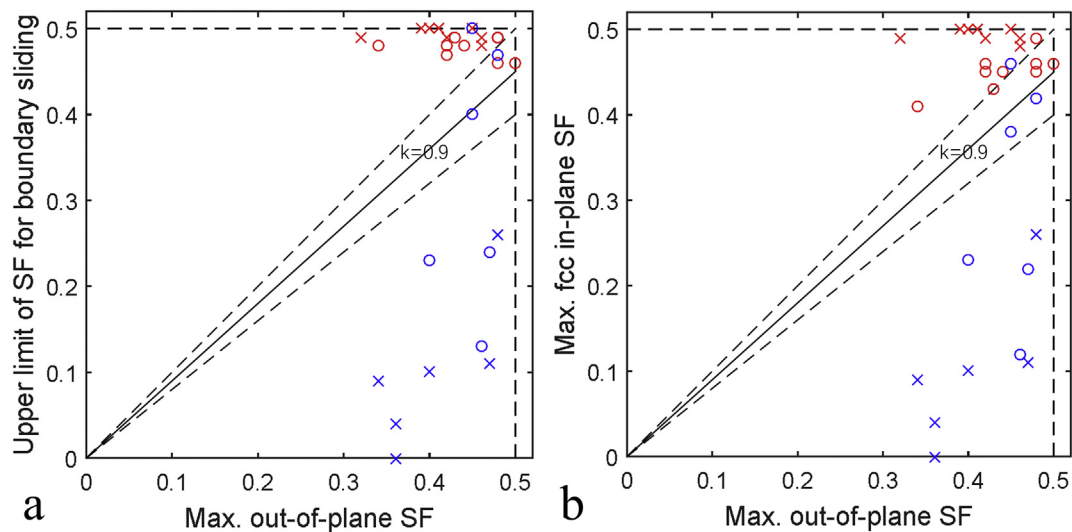


Fig. 5. Upper limit of SF for apparent boundary sliding (a) and Max. fcc in-plane SF (which is equal to Max. bcc in-plane SF) (b) plotted against Max. out-of-plane SF for specimens in Fig. 3 and Fig. 4 (marked as circles) and the specimens from previous research (marked as crosses) [8,9]. The red and blue data points represent the specimens which show, respectively, boundary sliding and intra-lath crystallographic slip as the main deformation mechanism. 'k' represents the slope of the lines. The dashed lines represent the tolerance of k (± 0.10).

generally have lower Schmid factors for boundary sliding due to the more optimal boundary configurations for crystallographic slip. It is clear that the red and blue data points lie in different regions of the graph. An attempt has been made to separate the region of boundary sliding from the region of intra-lath crystallographic slip, resulting in the solid line with $k = 0.9$ in both Fig. 5(a) and (b) with a tolerance of ± 0.10 (dashed lines). This means the apparent boundary sliding is slightly easier and no distinct difference between the fundamental mechanisms for the apparent boundary sliding ((Fig. 5(a)) or retained austenite deformation (Fig. 5 (b))) is observed. In addition, the higher critical resolved shear stress for crystallographic slip leads to even higher chance for the apparent boundary sliding. The tolerance is estimated by considering the following factors: the uncertainty in the morphology of the boundary (straight or curved), the uncertainty in the nature of the boundary (e.g. possible presence of second phase at the boundary, as discussed below), and the inaccuracy from the measurement.

Reconsidering Fig. 1 and Fig. 3 in terms of boundary types, the sliding mechanism is observed to be activated at all kinds of substructure boundaries, e.g. block boundary in specimen S3a – S3d, sub-block boundaries observed in specimen S3e – S3g, S3i and lath boundary in specimen S1 shown in Fig. 1 and specimen S3h. Interestingly, a preference in terms of a boundary type that is more easily activated cannot be identified, even though in most cases the resolved shear stress on most boundaries is approximately equal due to the (near) parallelism of boundaries from the same packet. Consequently, the boundaries seem to have equal probability to activate sliding. Yet, in general, only one of the parallel boundaries exhibits sliding. Hence, as mentioned above, there seems to exist an intrinsic variability in the stress level to activate boundary sliding, which must be related to the intrinsic structure of the boundaries at smaller scales. Therefore, the boundaries are studied in detail with TEM and transmission Kikuchi diffraction next.

Interesting observations of the boundaries were made that might explain the variability in boundary sliding activation. Fig. 6 shows an IPF map from transmission Kikuchi diffraction measurement on top of a TEM dark field image and a bright field image. It can be seen that various boundaries of different substructure levels are present in the region covered by the TEM dark field images. The block boundaries are marked in orange as done in Fig. 3 and Fig. 4. The sub-block boundaries are marked in green (instead of black) for enhanced visibility. The unmarked boundaries are lath boundaries, which are the most abundant.

The white precipitates are carbides, which were found at all types of boundaries. Their presence is attributed to the high M_s temperature of this material, so that auto-tempering can occur in the quenching process [11]. The carbides are present only at some of the boundaries and their presence does not depend on the type of boundary. Moreover, their density changes strongly even along a single boundary. Therefore, whatever the precise mechanism for boundary sliding may be, the presence of the carbide precipitates in the boundaries will certainly hamper and probably obstruct the apparent boundary sliding mechanism. Therefore, the variability of the carbide density between different boundaries and along the boundary length appears to be one of the underlying reasons for the large variability in boundary sliding activation and the accompanying differences in stress levels.

So far, all presented results of boundary sliding (in Fig. 1 and Fig. 3 and Ref. [9]) were obtained for fully martensitic steel. As stated in the introduction, lath martensite is the primary strengthening component of multi-phase steels. The next aim is therefore to investigate whether the apparent sliding mechanism is still activated in industrially relevant lath-martensite-containing multi-phase steels. To this end, first, the microstructural morphology of the lath martensite islands in such steels is investigated. Fig. 7 shows an EBSD measurement of a martensite band in DP steel. The IPF map is plotted together with the image quality (IQ) map, in which the martensite domains appear darker due to the higher dislocation density. The first important observation is that many boundary planes cross over the full width of the martensite band. Most of these boundaries are tilted to the length direction of martensite band. Note that the martensite bands in rolled steel sheet are always parallel to the sheet surface [45] and are therefore typically loaded under tension or compression, also when the sheet is loaded under bending. There are almost no boundaries aligned with the length direction of the map, which is likely caused by the orientation-dependent stress built-up in the austenite-to-martensite transformation. In particular, the necking regions of martensite domains have boundaries that cover the complete cross section and are tilted with respect to the band direction, i.e. a configuration that is favorable for boundary sliding. Therefore, based on the particular microstructural morphology of lath martensite islands, one would expect that the apparent boundary sliding may well be an important plasticity mechanism in lath martensite in DP steel, and possibly other multi-phase steels.

Based on earlier conclusions, the boundaries which are expected to slide during the deformation are the ones for which the ratio of the

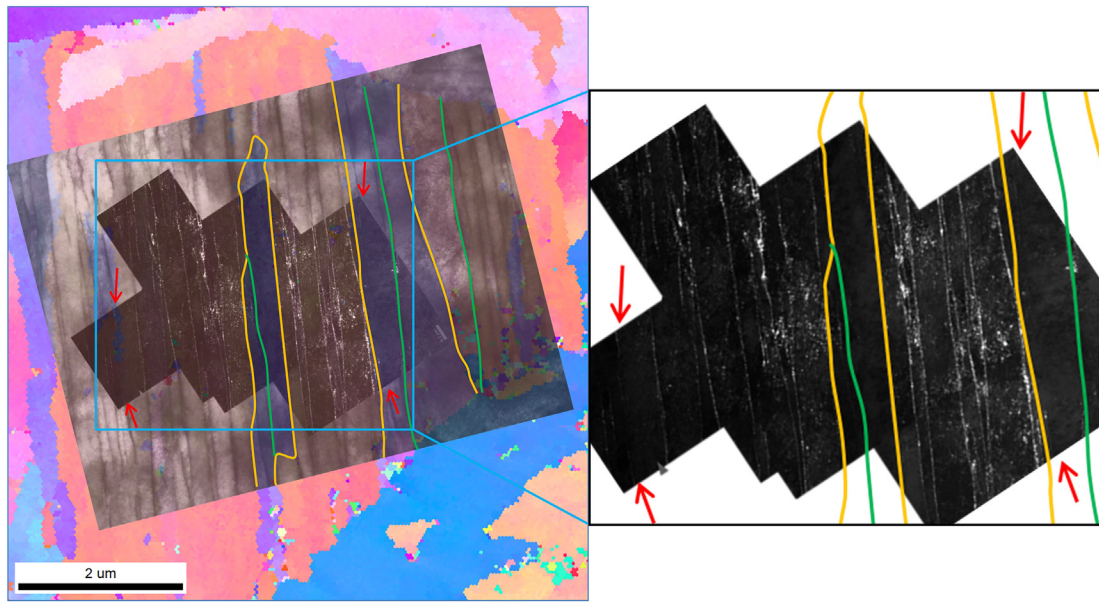


Fig. 6. An IPF map from TKD measurement overlaid with a TEM dark field image, and a bright field image taken by the transmission mode in a SEM with a zoom-in of the blue frame on the right side. The white particles are carbides in the dark field image. The block boundaries are marked in orange and the sub-block boundaries are marked in green (instead of black in Fig. 3 and 4) for enhanced visibility. The red arrows indicate the positions of two boundaries that do not contain carbides at all.

maximum in-plane SF to the maximum out-of-plane SF in the two adjacent laths exceeds (0.9 ± 0.1) . For these martensite measurements on bulk DP steel (Fig. 7), a positive identification of the boundary

orientation (α and β) can only be made for (i) boundaries that lie inside a packet, because in that case the boundary plane orientation equals that of the shared habit plane of the two adjacent laths, and (ii) high

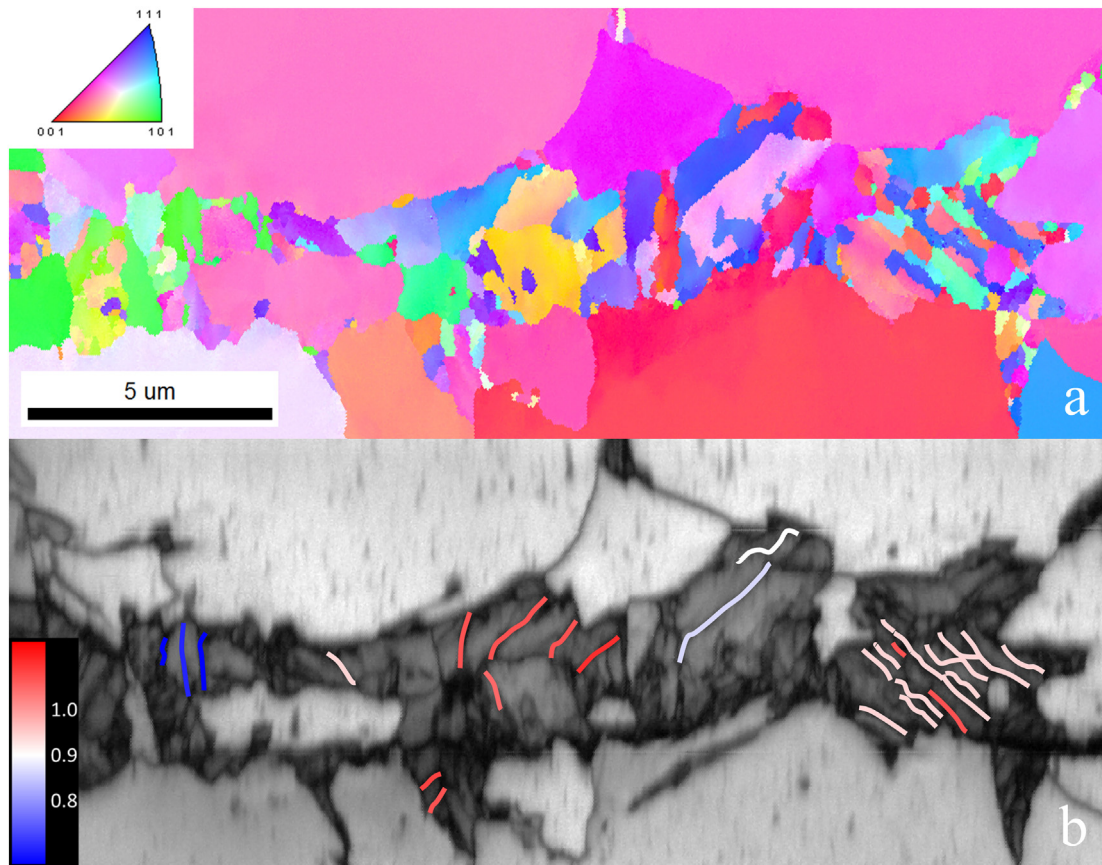


Fig. 7. EBSD scan of lath martensite in DP steel. (a) IPF map of the measurement and (b) the corresponding image quality map in which the boundaries inside identified packets are marked with colors, which reflect their probability of sliding under the assumption that the local stress state is equal to a globally applied uniaxial tension in the horizontal direction. The colour coding is based on the conclusion from Fig. 5 that the threshold for boundary sliding is a ratio of 'maximum in-plane SF' to 'maximum out-of-plane SF in the two adjacent laths' of (0.9 ± 0.1) , i.e. red means a high probability for boundary sliding and vice versa.

quality of the EBSD measurement in both adjacent laths. For these identified boundary orientations, under the assumption that the local stress state is equal to a globally applied uniaxial tension in the horizontal direction, the ratio of the maximum in-plane SF to the maximum out-of-plane SF in the two adjacent laths has been determined from the EBSD data of Fig. 7. This ratio is indicated in Fig. 7(b) with a colour coding, where a white colour corresponds to an equal probability for boundary sliding and out-of-plane slip in one of the adjacent laths, while red and blue mark a preference for, respectively apparent boundary sliding and intra-lath crystallographic slip. Therefore, all calculated boundaries except the dark blue ones, i.e. the majority of boundaries, are expected to slide during the above-assumed loading state, which supports the hypothesis that apparent martensite boundary sliding is important in DP steel and possibly also other multi-phase steels. This is consistent with the results on localized plasticity in martensite banding observed in [15].

Although the loading constraint on a martensite island in a micro-specimen cut from DP steel is quite different in comparison to the loading constraint inside the bulk DP material, uniaxial micro-tensile tests on DP steel micro-specimens have been carried out in an attempt to characterize the activation of apparent boundary sliding in DP steel. In total, eight specimens were tested, of which seven specimens show the expected behavior: an easy deformation percolation path in the ferrite is formed over the entire specimen cross-section, along which all deformation continued to concentrate until fracture (not shown). This yields no insight in the potential role of apparent boundary sliding, because the loading constraint on the martensite islands in a micro-specimen is much lower than in bulk material. In the bulk, before localization in ferrite grains can initiate, the martensite islands need to deform to accommodate the overall deformation [15,45], and martensite plasticity is known to be important. It is, therefore, informative to investigate whether the martensite plasticity in the remaining DP micro-specimen deformed by apparent boundary sliding or by intra-lath crystallographic slip.

In Fig. 8 this DP micro-specimen is shown, for which the fracture propagated around the bottom martensite domain. Interestingly, the top part of the bottom martensite island is cleanly sheared off along two parallel planes, which are marked by the white dash-dotted lines in the zoom of Fig. 8(c), resulting in two distinct steps in the left edge of the martensite domain, denoted by the orange arrows in Fig. 8(c). This indicates that shearing in the martensite domain can be activated almost as easily as plasticity in ferrite, suggesting the activity of an easy sliding mechanism across the complete domain of the martensite island, which is supported by the sharp, clean nature of the sheared parts. This clean shearing across a martensite domain bears the signature of apparent boundary sliding, similar to the many examples of apparent boundary sliding in fully martensitic steel in Fig. 1 and Fig. 3 and Ref. [9]. Because the presence of substructure boundaries over the full cross section of the martensite domains in DP steel was already demonstrated in Fig. 7 and because no clear evidence of regular martensite plasticity was observed in any of the eight DP micro-tests, all available evidence seems to be in support of the occurrence of apparent boundary sliding in DP steel.

In general, considering that martensite plasticity is known to occur frequently in bulk lath-martensite containing multi-phase steels [13,15], apparent boundary sliding may be the main mechanism responsible for the observed plasticity. This mechanism, when it occurs, can introduce large localized (shear) strains within the martensite domains. For example, Ghadbeigi et al. showed that the localized (shear) strain inside martensite can reach up to 120% [13] while more evidences of (high) localized martensite ductility has been found in DP steels [6,7,19,20,43,44]. In addition to DP steels, lath martensite in other multi-phase steels have shown indications of apparent boundary sliding as well, as indicated in [41,47] in TRIP steel. Therefore, the sliding mechanism may be an important mechanism explaining ductility of lath martensite at local regions.

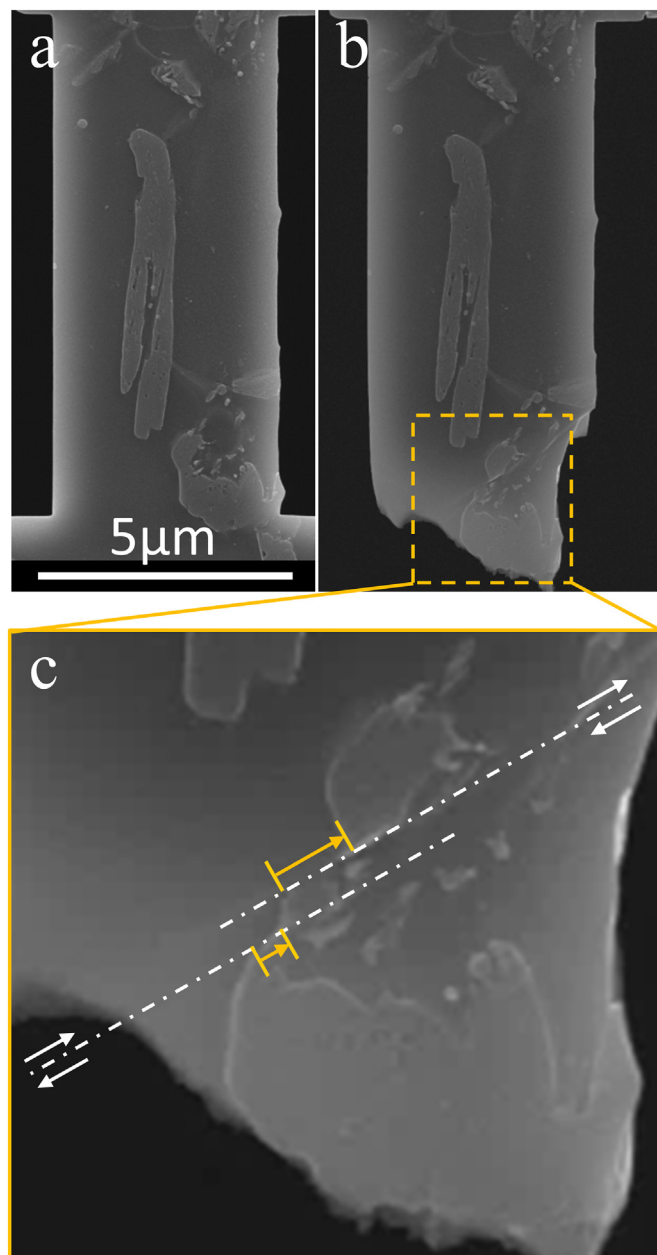


Fig. 8. SEM SE images of the gauge part of a DP micro-tensile specimen (a) before and (b) after fracture. (c) Zoom-in of the orange frame, which highlights the bottom martensite island that is sheared off along two planes, as indicated by the white dash-dotted lines; the orange arrows indicate the two distinct steps in the left edge of the island.

Also for bulk fully martensitic steel, indications of apparent boundary sliding on the sample surface can be found in the literature [29,34]. Nambu et al. reported many examples for which 'at strain levels less than the transition strain, slip bands were found to develop parallel to the lath plane only within martensite blocks that have a large Schmid factor for the slip systems where the Burgers vector is parallel to the lath growth direction.' [34]. When analyzing their results, the 'slip bands parallel to the lath plane' appear to be traces of apparent boundary sliding, exactly because the traces were perfectly aligned with the boundaries. Note that in the current study and in Ref. [9] the exact positions of the steps/traces were identified from higher magnification SEM images while also including side view SEM images. Nambu et al. also pointed out that the slip traces only occur in 'martensite blocks that have a large Schmid factor for the slip systems' [34], namely the ones with high in-plane SF. As argued above, the SF for apparent boundary

sliding is always equal to the maximum in-plane SF, which supports the reasoning that the observed in-plane slip is probably boundary sliding. It is worth mention that in bulk samples (both fully martensitic steels and multi-phase steels that contains lath martensite), the constraints in the interior of the specimens on the boundaries function as a barrier for the apparent gliding mechanism to take place. But still, the boundary regions can be stress concentrators, because deformation is more likely to occur along them, causing the local anisotropy of lath martensite bulk materials. This is critical for the plastic deformation of lath martensite, and even for its final failure. It is expected that this apparent boundary sliding mechanism does not depend significantly on the grain size, but does depend strongly on the morphology of the martensite. For example, with higher carbon content [32], for which the shape of the laths is not straight anymore, one would expect that the probability that boundary sliding is activated is lower. In addition, the influence of the evolving loading conditions (i.e. the increase of lateral force after the first plastic deformation starts) on the micro-sample behavior should not be neglected, as shown in a parallel study on micro-tensile tests of fully ferrite specimens [12]. Note, however, that the focus of the current analysis was limited to which plastic mechanism activates first in the lath martensite, therefore, the influence of the evolution of the loading conditions in the elastic or early-plastic regime is of minor importance. Finally, at high strain levels, Nambu et al. observed crossing of the substructure boundaries [29,34]. This suggests that in bulk fully martensitic steel, the sliding mechanism is important for the deformation at the early stages, however, at higher strain levels, intra-lath crystallographic slip will also be activated (across the substructure boundaries). Therefore, apparent boundary sliding appears to be not only important for micro-tensile specimens but also in lath-martensite-containing bulk materials.

Conclusions

In this work, micro-tensile specimens of various lath-martensite boundary configurations have been tested using a nano-force tensile stage. Although the verification of the fundamental mechanism of the apparent lath boundary sliding mechanism requires further experimental investigation, the following conclusions can be drawn:

1. Many cases of apparent boundary sliding have been observed for sub-optimal boundary orientations, i.e. it is not necessary that the boundaries are perpendicular to the specimen front surface and have a tilting angle of $\sim 45^\circ$ with respect to the loading direction.
2. The ratios between the maximum in-plane Schmid factor of fcc austenite and maximum out-of-plane slip factor determine the dominant deformation mechanism, namely the apparent sliding mechanism or intra-lath crystallographic slip. The threshold of this ratio is (0.9 ± 0.10) , showing that the apparent boundary sliding is more easily activated.
3. For micro-tests on dual phase steel, as an example of multi-phase steels, plastic deformation in martensite was observed in only one specimen due to the low external loading constraint. Nevertheless, this martensite island exhibited two clean parallel shearing planes over its full cross-section, which shows all signatures of boundary sliding.
4. For bulk samples, although the lath martensite substructure boundaries are constrained compared to the micro-specimens, the apparent boundary sliding mechanism can create many local stress concentrators, triggering subsequent plastic deformations and even leading to local failures.

Credit author statement

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R. Petrov: Investigation, Writing – Original draft.

M.G.D. Geers: Conceptualization, Methodology, Writing – original draft, Supervision, Funding acquisition.

J.P.M. Hoefnagels: Conceptualization, Methodology, Investigation, Validation, Writing – original draft, Writing – Review & Editing, Visualization, Supervision, Project administration, Funding acquisition.

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References

- [1] A.L.T. Azevedo, E.G. da Silva, Mossbauer study of retained austenite in a low C low alloy steel, *Scr. Metall.* 12 (2) (1978) 113–117, [https://doi.org/10.1016/0036-9748\(78\)90146-1](https://doi.org/10.1016/0036-9748(78)90146-1).
- [2] L.J. Bergers, J.P.M. Hoefnagels, M.G.D. Geers, On-wafer time-dependent high reproducibility nano-force tensile testing, *J. Phys. D. Appl. Phys.* 47 (48) (2014), 495306, <https://doi.org/10.1088/0022-3727/47/49/495306>.
- [3] H.K.D.H. Bhadeshia, R.W.K. Honeycombe, *Steels, Microstructure and Properties*, Butterworth-Heinemann, Oxford, UK, 2006.
- [4] P. Bowen, C.A. Hipsley, J.F. Knott, Effects of segregation on brittle fracture and fatigue crack growth in coarse-grained, martensitic A533B pressure vessel steel, *Acta Metall.* 32 (5) (1984) 637–647, [https://doi.org/10.1016/0001-6160\(84\)90137-8](https://doi.org/10.1016/0001-6160(84)90137-8).
- [5] C.L. Briant, S.K. Banerji, Tempered martensite embrittlement in phosphorus doped steels, *Metall. Mater. Trans. A* 10 (11) (1979) 1729–1737, <https://doi.org/10.1007/BF02811708>.
- [6] X.L. Cai, J. Feng, W.S. Owen, The dependence of some tensile and fatigue properties of a dual-phase steel on its microstructure, *Metall. Trans. A* 16 (8) (1985) 1405–1415, <https://doi.org/10.1007/BF02658673>.
- [7] M. Calcagnotto, D. Ponge, D. Raabe, Effect of grain refinement to 1 μm on strength and toughness of dual-phase steels, *Mater. Sci. Eng. A* 527 (29–30) (2010) 7832–7840, <https://doi.org/10.1016/j.msea.2010.08.062>.
- [8] C. Du, J.P.M. Hoefnagels, R. Vaes, M.G.D. Geers, Block and sub-block boundary strengthening in lath martensite, *Scr. Mater.* 116 (2016) 117–121, <https://doi.org/10.1016/j.scriptamat.2016.01.043>.
- [9] C. Du, J.P.M. Hoefnagels, R. Vaes, M.G.D. Geers, Plasticity of lath martensite by sliding of substructure boundaries, *Scr. Mater.* 120 (2016) 37–40, <https://doi.org/10.1016/j.scriptamat.2016.04.006>.
- [10] C. Du, J.P.M. Hoefnagels, M.G.D. Geers, A uni-axial nano-displacement micro-tensile test of individual constituents from bulk material, *Exp. Mech.* 57 (8) (2017) 1249–1263, <https://doi.org/10.1007/s11340-017-0299-6>.
- [11] C. Du, J.P.M. Hoefnagels, S. Kölling, M.G.D. Geers, J. Sietsma, R. Petrov, V. Bliznuk, P.M. Koenraad, D. Schryvers, B. Amin-Ahmadi, Martensite crystallography and chemistry in dual phase and fully martensitic steels, *Mater. Charact.* 139 (2018) 411–420, <https://doi.org/10.1016/j.matchar.2018.03.011>.
- [12] C. Du, F. Maresca, J.P.M. Hoefnagels, M.G.D. Geers, Ferrite slip system activation investigated by uniaxial micro-tensile tests and simulations, *Acta Mater.* 146 (2018) 314–327, <https://doi.org/10.1016/j.actamat.2017.12.054>.
- [13] H. Ghadbeigi, C. Pinna, S. Celotto, J.R. Yates, Local plastic strain evolution in a high strength dual-phase steel, *Mater. Sci. Eng. A* 527 (18–19) (2010) 5026–5032, <https://doi.org/10.1016/j.msea.2010.04.052>.
- [14] H. Ghassemi-Armaki, R. Maass, S.P. Bhat, S. Sriram, J.R. Greer, K.S. Kumar, Deformation response of ferrite and martensite in a dual-phase steel, *Acta Mater.* 62 (2014) 197–211, <https://doi.org/10.1016/j.actamat.2013.10.001>.
- [15] J.P.M. Hoefnagels, C.C. Tasan, F. Maresca, F.J. Peters, V.G. Kouznetsova, Retardation of plastic instability via damage-enabled microstrain delocalization, *J. Mater. Sci.* 50 (21) (2015) 6882–6897, <https://doi.org/10.1007/s10853-015-9164-0>.
- [16] R.M. Horn, R.O. Ritchie, Mechanisms of tempered martensite embrittlement in low alloy steels, *Metall. Trans. A* 9 (8) (1978) 1039–1053, <https://doi.org/10.1007/BF02652208>.
- [17] D. Hull, *Fractography*, Cambridge university press, Cambridge, 1999.
- [18] T. Inoue, S. Matsuda, Y. Okamura, K. Aoki, The fracture of a low carbon tempered martensite, *Trans. JIM* 11 (1) (1970) 36–43, <https://doi.org/10.2320/matertrans1960.11.36>.
- [19] J. Kang, Y. Ososkov, J.D. Embury, D.S. Wilkinson, Digital image correlation studies for microscopic strain distribution and damage in dual phase steels, *Scr. Mater.* 56 (11) (2007) 999–1002, <https://doi.org/10.1016/j.scriptamat.2007.01.031>.
- [20] J. Kadkhodapour, A. Butz, S. Ziaei Rad, Mechanisms of void formation during tensile testing in a commercial, dual-phase steel, *Acta Mater.* 59 (7) (2011) 2575–2588, <https://doi.org/10.1016/j.actamat.2010.12.039>.
- [21] N. Kim, G. Thomas, Effects of morphology on the mechanical behavior of a dual phase Fe/2Si/0.1C steel, *Metall. Mater. Trans. A* 12 (3) (1981) 483–489, <https://doi.org/10.1007/BF02648546>.
- [22] G. Krauss, Martensite in steel: strength and structure, *Mater. Sci. Eng. A* 273–275 (1999) 40–57, [https://doi.org/10.1016/S0921-5093\(99\)00288-9](https://doi.org/10.1016/S0921-5093(99)00288-9).
- [23] K. Kwak, T. Mayama, Y. Mine, K. Takashima, Anisotropy of strength and plasticity in lath martensite steel, *Mater. Sci. Eng. A* 674 (2016) 104–116, <https://doi.org/10.1016/j.msea.2016.07.047>.

- [24] T. Maki, Microstructure and mechanical behaviour of ferrous martensite, *Mater. Sci. Forum* 56–58 (1990) 157–168, <https://doi.org/10.4028/www.scientific.net/MSF.56-58.157>.
- [25] F. Maresca, V.G. Kouznetsova, M.G.D. Geers, On the role of interlath retained austenite in the deformation of lath martensite, *Model. Simul. Mater. Sci. Eng.* 22 (2014), 045011, <https://doi.org/10.1088/0965-0393/22/4/045011>.
- [26] F. Maresca, W.A. Curtin, The austenite/lath martensite interface in steels: structure, athermal motion, and in-situ transformation strain revealed by simulation and theory, *Acta Mater.* 134 (2017) 302–323, <https://doi.org/10.1016/j.actamat.2017.05.044>.
- [27] F. Maresca, V.G. Kouznetsova, M.G.D. Geers, W.A. Curtin, Contribution of austenite-martensite transformation to deformability of advanced high strength steels: from atomistic mechanisms to microstructural response, *Acta Mater.* 156 (2018) 463–478, <https://doi.org/10.1016/j.actamat.2018.06.028>.
- [28] J.P. Materkowski, G. Krauss, Tempered martensite embrittlement in SAE 4340 steel, *Metall. Trans. A* 10 (11) (1979) 1643–1651, <https://doi.org/10.1007/BF02811697>.
- [29] M. Michiuchi, S. Nambu, Y. Ishimoto, J. Inoue, T. Koseki, Relationship between local deformation behavior and crystallographic features of as-quenched lath martensite during uniaxial tensile deformation, *Acta Mater.* 57 (18) (2009) 5283–5291, <https://doi.org/10.1016/j.actamat.2009.06.021>.
- [30] Y. Mine, K. Hirashita, H. Takashima, M. Matsuda, K. Takashima, Micro-tension behaviour of lath martensite structures of carbon steel, *Mater. Sci. Eng. A* 560 (2010) 535–544, <https://doi.org/10.1016/j.msea.2012.09.099>.
- [31] J.W. Morris, On the ductile-brittle transition in lath martensitic steel, *ISIJ Int.* 51 (10) (2011) 1569–1575, <https://doi.org/10.2355/isijinternational.51.1569>.
- [32] S. Morito, H. Yoshida, T. Maki, X. Huang, Effect of block size on the strength of lath martensite in low carbon steels, *Mater. Sci. Eng. A* 438–440 (2006) 237–240, <https://doi.org/10.1016/j.msea.2005.12.048>.
- [33] S. Morito, K. Oh-Shi, Carbon enrichment in retained austenite films in low carbon lath martensite steel, *ISIJ Inter.* 51 (7) (2011) 1200–1202, <https://doi.org/10.2355/isijinternational.51.1200>.
- [34] S. Nambu, M. Michiuchi, Y. Ishimoto, K. Asakura, J. Inoue, T. Koseki, Transition in deformation behavior of martensitic steel during large deformation under uniaxial tensile loading, *Scr. Mater.* 60 (4) (2009) 221–224, <https://doi.org/10.1016/j.scriptamat.2008.10.007>.
- [35] J. Li, T. Ohmura, K. Tsuzaki, Evaluation of grain boundary effect on strength of Fe–C low alloy martensitic steels by nanoindentation technique, *Mater. Trans.* 46 (6) (2005) 1301–1305, <https://doi.org/10.2320/matertrans.46.1301>.
- [36] T. Ohmura, T. Hara, K. Tsuzaki, Evaluation of temper softening behavior of Fe–C binary martensitic steels by nanoindentation, *Scr. Mater.* 49 (12) (2003) 1157–1162, <https://doi.org/10.1016/j.scriptamat.2003.08.025>.
- [37] M. Saegeltz, G. Krauss, Deformation, fracture, and mechanical properties of low-temperature-tempered Martensite in SAE 43xx steels, *Metall. Mater. Trans. A* 27 (1997) 377–387, <https://doi.org/10.1007/s11661-997-0139-x>.
- [38] B.P.J. Sandvik, C.M. Wayman, Crystallography and substructure of lath martensite formed in carbon steels, *Metallography* 16 (2) (1983) 199–227, [https://doi.org/10.1016/0026-0800\(83\)90005-8](https://doi.org/10.1016/0026-0800(83)90005-8).
- [39] K. Wakasa, C.M. Wayman, Crystallography and morphology of ferrous lath martensite, *Metallography* 14 (1) (1981) 49–60, [https://doi.org/10.1016/0026-0800\(81\)90019-7](https://doi.org/10.1016/0026-0800(81)90019-7).
- [40] A. Shibata, T. Nagoshi, M. Sone, S. Morito, Y. Higo, Micromechanical characterization of deformation behavior in ferrous lath martensite, *J. Alloy. Compd.* 577 (1) (2013) S555–S558, <https://doi.org/10.1016/j.jallcom.2012.01.066>.
- [41] X. Sun, A. Soulati, K.S. Choi, O. Guzman, W. Chen, Effects of sample geometry and loading rate on tensile ductility of TRIP800 steel, *Mater. Sci. Eng. A* 541 (2012) 1–7, <https://doi.org/10.1016/j.msea.2011.12.115>.
- [42] S. Takaki, K. Ngo-huynh, N. Nakada, T. Tsuchiyama, Strengthening mechanism in ultra low carbon martensitic steel, *ISIJ Int.* 52 (4) (2012) 710–716, <https://doi.org/10.2355/isijinternational.52.710>.
- [43] C.C. Tasan, J.P.M. Hoefnagels, C.H.L.J. ten Horn, M.G.D. Geers, Experimental analysis of strain path dependent ductile damage mechanics and forming limits, *Mech. Mater.* 41 (11) (2009) 1264–1276, <https://doi.org/10.1016/j.mechmat.2009.08.003>.
- [44] C.C. Tasan, J.P.M. Hoefnagels, M.G.D. Geers, A critical assessment of indentation-based ductile damage quantification, *Acta Mater.* 57 (17) (2009) 4957–4966, <https://doi.org/10.1016/j.actamat.2009.06.057>.
- [45] C.C. Tasan, J.P.M. Hoefnagels, M.G.D. Geers, Microstructural banding effects clarified through micrographic digital image correlation, *Scr. Mater.* 62 (11) (2010) 835–838, <https://doi.org/10.1016/j.scriptamat.2010.02.014>.
- [46] C.C. Tasan, J.P.M. Hoefnagels, M.G.D. Geers, Identification of the continuum damage parameter: an experimental challenge in modeling damage evolution, *Acta Mater.* 60 (8) (2012) 3581–3589, <https://doi.org/10.1016/j.actamat.2012.03.017>.
- [47] C. Wang, H. Ding, M. Cai, B. Rolfe, Characterization of microstructures and tensile properties of TRIP-aided steels with different matrix microstructure, *Mater. Sci. Eng. A* 610 (2014) 65–75, <https://doi.org/10.1016/j.msea.2014.05.029>.
- [48] R. Wouters, L. Froyen, Scanning electron microscope fractography in failure analysis of steels, *Mater. Charact.* 36 (4–5) (1996) 357–364, [https://doi.org/10.1016/S1044-5803\(96\)00070-8](https://doi.org/10.1016/S1044-5803(96)00070-8).
- [49] C. Zhang, Q. Wang, J. Ren, R. Li, M. Wang, F. Zhang, K. Sun, Effect of martensitic morphology on mechanical properties of an as-quenched and tempered 25CrMo48V steel, *Mater. Sci. Eng. A* 534 (2012) 339–346, <https://doi.org/10.1016/j.msea.2011.11.078>.