

Effect of carbon content on boundary deformation behaviour in martensitic steels

Karel Blanken^{a,1} , Sien Liu^{a,*} , Jiaqiang Dang^a, Fabien Briffod^b, Wujun Yin^a ,
Shoichi Nambu^{a,**} 

^a Department of Materials Engineering, Graduate School of Engineering, The University of Tokyo, 7-3-1, Hongo, Bunkyo-ku, Tokyo, 113-8656, Japan

^b Research Center for Structural Materials, National Institute for Materials Science, 1-2-1 Sengen, Tsukuba, Ibaraki, 305-0047, Japan

ARTICLE INFO

Keywords:

Martensite
Plastic deformation
Boundary
Fracture behaviour

ABSTRACT

This study aims to clarify the different boundary deformation mechanisms between medium-carbon lath martensite and high-carbon martensitic steels. In medium-carbon martensite, boundary deformation can trigger the activation of secondary deformation bands in surrounding blocks, leading to local strain partitioning. Compared with normal intra-block slips along in-lath-plane slip systems, these boundary-induced secondary deformation bands are activated on planes aligned with the deformed boundaries. In contrast, boundary deformation in high-carbon martensite is strictly confined along the boundaries, forming intense single-peak strain concentration band. However, when the high-carbon martensite is tempered at 500 °C, the boundary-induced secondary deformation bands are observed again. This indicates that the activation of boundary-induced secondary deformation bands depends primarily on the local critical resolved shear stress (CRSS) around the boundaries, instead of the carbon-induced microstructure variation. When primary boundary deformation activated, nearby microstructure with high CRSS values remain hardly deformed, leading to strain concentration only at these boundaries. With low CRSS, secondary deformation bands can be activated near the primary boundaries, leading to strain partitioning effect and improved overall ductility.

1. Introduction

Plastic deformation in low-carbon lath martensite highly depends on its hierarchical structure, consisting of laths, blocks, packets and prior austenite grains (PAGs). According to the slip activation sites, the plasticity can be divided into two categories: slips within blocks and boundary deformation [1,2].

In lath martensite, intra-block slips are preferentially activated along specific crystallographic systems, which are $\{110\}<111>$ and $\{211\}<111>$ in bcc lattice structure. Moreover, the majority of intra-block slips are activated along in-lath-plane slip systems, owing to their relatively lower critical resolved shear stress (CRSS) [3,4]. So far, studies showed that Schmid effect can effectively evaluate the intra-block slipping behaviour [5,6]. On the other hand, the boundary deformation behaviour remains less understood, because it can affect local plasticity through different ways. For instance, low-angle boundaries can hinder or tangle with local dislocation gliding [7,8], block boundaries mainly

impede dislocation movement and exhibit a Hall-Petch like effect [9,10], while packet boundaries and prior austenite grain boundaries (PAGBs) are the potential sites for boundary sliding [11].

Much attention has been paid to low-carbon lath martensite, whereas studies on high-carbon martensite remain limited. High-carbon martensite consists of both lath-based blocks with smaller size and twin-based coarse blocks (i.e. plate martensite blocks, butterfly martensite blocks, and coarse blocks (CBs) along PAGBs) [12,13]. Compared with low-carbon lath martensite, the lath-based blocks consist of laths with only single Kurdjumov-Sachs (K-S) variant [14]. Under deformation, high-carbon martensite exhibits improved strength but poor ductility. So far, there are different explanations for the origin of its brittleness. Some studies proposed that the poor deformability in twin-based blocks leads to premature brittleness of high-carbon martensite, due to both twin-dislocation mutual impediment effect [15] and carbide precipitation on twin boundaries [16]. However, recent studies indicated that the brittleness of high-carbon martensite is

* Corresponding author. 7-3-1 Hongo, Bunkyo-ku, Tokyo, 113-8656, Japan.

** Corresponding author.

E-mail addresses: sien_liu@metall.t.u-tokyo.ac.jp (S. Liu), nambu@metall.t.u-tokyo.ac.jp (S. Nambu).

¹ These authors contributed equally to this work.

more likely to originate from boundary fracture [17–20]. This suggests that the low- and high-carbon martensite may have different boundary deformation behaviours, where surrounding blocks play a significant role.

Furthermore, the relationship between carbon content and boundary strength still remains contradictory. Theoretically, atomic-scale calculations on binding energies showed that carbon is a cohesion enhancer in the Fe grain boundary [21–23]. While in practice, martensite with elevated carbon content is prone to the intergranular brittle fracture [24], particularly along packet and prior austenite grain boundaries (PAGBs) [17]. Such contradiction also infers that high carbon content may introduce unclarified effects on the boundary deformation in high-carbon martensite.

This study aims to elucidate the plastic deformation, particularly the boundary deformation mechanisms, in martensite with different carbon additions. A combination of electron backscatter diffraction (EBSD) and high-resolution digital image correlation (HR-DIC) analysis was employed to evaluate the microstructure-deformation relationship at the sub-micron scale.

2. Method

In this study, a 0.4 wt% medium-carbon steel and a 0.6 wt% high-carbon steel were employed. Table 1 shows their chemical compositions. The raw steels were rolled into 1-mm thick plates. Then the steels were austenitized at 950 °C for 10 min and water quenched to room temperature, followed by a 1-h tempering at 200 °C. The two 200 °C-tempered specimens are named 04C and 06C specimens. Additionally, the 06C martensite was also subjected to 500 °C tempering for 10 min. This high-temperature tempered 06C specimen is referred to as the 06C-HT specimen. Subsequently, dog-bone tensile specimens with gauge size of $20 \times 3 \times 1 \text{ mm}^3$ were machined from the plates. The gauge length is parallel to the longitudinal axis of the rolling direction.

For the observation purposes, the specimens were mechanically ground by 320, 1200 and 2000 grit silicon carbide papers and polished by colloidal silica suspension (0.04 µm particle size). EBSD analysis (accelerating voltage: 15 kV, probe current: 8 nA, JSM-7001F, JEOL) was conducted with a step size of 100 nm. Uniaxial tensile tests were carried out at a uniform strain rate of $1 \times 10^{-3} \text{ s}^{-1}$ (AGX-V2, Shimadzu).

To perform the HR-DIC analysis, a uniform 5-nm Au layer was coated on the observing surface. The uniform coating layer was then remodelled in a vapor environment into nano-island particles of about 30 nm in size. A detailed methodology description was introduced in this literature [25]. Subsequently, backscattered electron images (BSE, $\times 7000$ magnification, accelerating voltage: 15 kV, probe current: 8 nA, JSM-7900F, JEOL) were taken before and after the deformation. For each sample, a region of interest (ROI) of $150 \times 150 \text{ µm}$ was scanned by a mosaic of 11×15 images (1280×960 pixels). NCORR toolbox in MATLAB [26] was employed to calculate the in-plane strain, and MTEX was used to correlate the initial microstructure with the deformation plasticity [27].

Finally, in order to clarify the various plastic deformation behaviours in martensite, the following expressions are adopted in this study: plasticity within lath-based blocks are referred to as slips; sliding at block boundaries is referred to as deformation band; deformation at PAGBs and packet boundaries is referred to as boundary deformation. The boundary deformation includes both boundary sliding and plastic deformation in local blocks.

Table 1

Chemical compositions of the medium-carbon 04C and high-carbon 06C steels (wt.%).

	C	Si	Mn	Cr	Mo	P	Fe
04C	0.40	0.27	0.79	1.13	0.15	<0.02	Bal.
06C	0.59	0.27	0.79	1.12	0.15	<0.02	Bal.

3. Results

3.1. Microstructure and tensile response

Fig. 1(a–f) compare the initial microstructure of the two 200 °C-tempered specimens, 04C and 06C. Grain boundaries with misorientation angles higher than 10° were depicted in the IPF maps. Although the two specimens have similar PAG and packet sizes, the 06C specimen includes much smaller block size [6,28]. In contrast, many blocks in 04C specimen exhibited an elongated columnar morphology.

It is notable that the 06C high-carbon martensite also includes some coarse blocks (CBs) along PAGBs (Fig. 1(e and f)), whereas the 04C medium-carbon martensite exhibited full-lath-martensite substructure with uniform size distribution (Fig. 1(c)). Previous studies attempted to clarify the transformation origins of these CBs along PAGBs, whereas no lath structure was observed inside the grains [29]. Existing hypotheses proposed that the coarse blocks are variants of butterfly martensite, where blocks preferentially grow along the PAGBs rather than developing the second symmetric wing as in conventional butterfly martensite [30].

Fig. 1(g) compares the stress-strain curves of the 04C and 06C specimens. The 04C specimen exhibited about 3.8% of uniform elongation strain, whereas the 06C specimen underwent a brittle fracture soon after yielding. Fig. 1(h, i) show the SEM observations on the fracture surfaces of the 04C and 06C specimens. The 04C specimen showed a combination of cleavage and dimple features, whereas the 06C specimen showed an intergranular fracture feature at the crack initiation spot. This suggests that the brittle fracture in high-carbon martensite is attributed to its boundary deformation behaviour, whereas the boundary deformation in low- and medium-carbon martensite does not lead to a premature brittle fracture [28].

3.2. Overall plastic deformation behaviour

Fig. 2 shows the deformation behaviour of the two specimens, including their (a, h) initial IPF maps, (b, i) habit-plane (HP) Schmid factor (SF) maps, (c, j) packet maps, (e, l) plastic strain maps, (f, m) grains' average strain value maps and (g, n) corresponding strain distributions. The HP SF is defined as the SF values of the slip systems along habit planes. Previous studies suggested that in-lath-plane SF, especially the HP SF, can effectively evaluate the intra-block slip activation [5,12].

During tensile loading, strain localization bands evenly appeared within the ROIs in both specimens (Fig. 2(e3, l3)). In the 04C specimen, these bands were more evenly distributed with both intense and shallow features, while the 06C specimen exhibited predominantly intense bands [31]. Correlated to the initial microstructure, the long and intense strain localization bands were mainly aligned with PAGBs and packet boundaries, while the short and shallow bands primarily located within blocks. This indicates that both intra-block slips and boundary sliding were activated in the 04C specimen, whereas boundary sliding dominated the plastic deformation in the 06C specimen.

The grains' average strain maps (Fig. 2(f–m)) further highlighted the different plastic strain distributions between the two specimens. In the 04C specimen (Fig. 2(f)), high-strain grains formed near PAGBs and packet boundaries, but the average strain values of many blocks increased due to sustained intra-block slips. This is also reflected in the progressive shift of the strain distribution as shown in Fig. 2(g). In contrast, in the 06C specimen (Fig. 2(m)), the grains adjacent to PAGBs and packet boundaries exhibited intensified high strain values, while the activation of intra-block slips was extremely limited. Consequently, the plastic deformation in high-carbon martensite strongly concentrated along the boundaries, which may contribute to premature intergranular fracture.

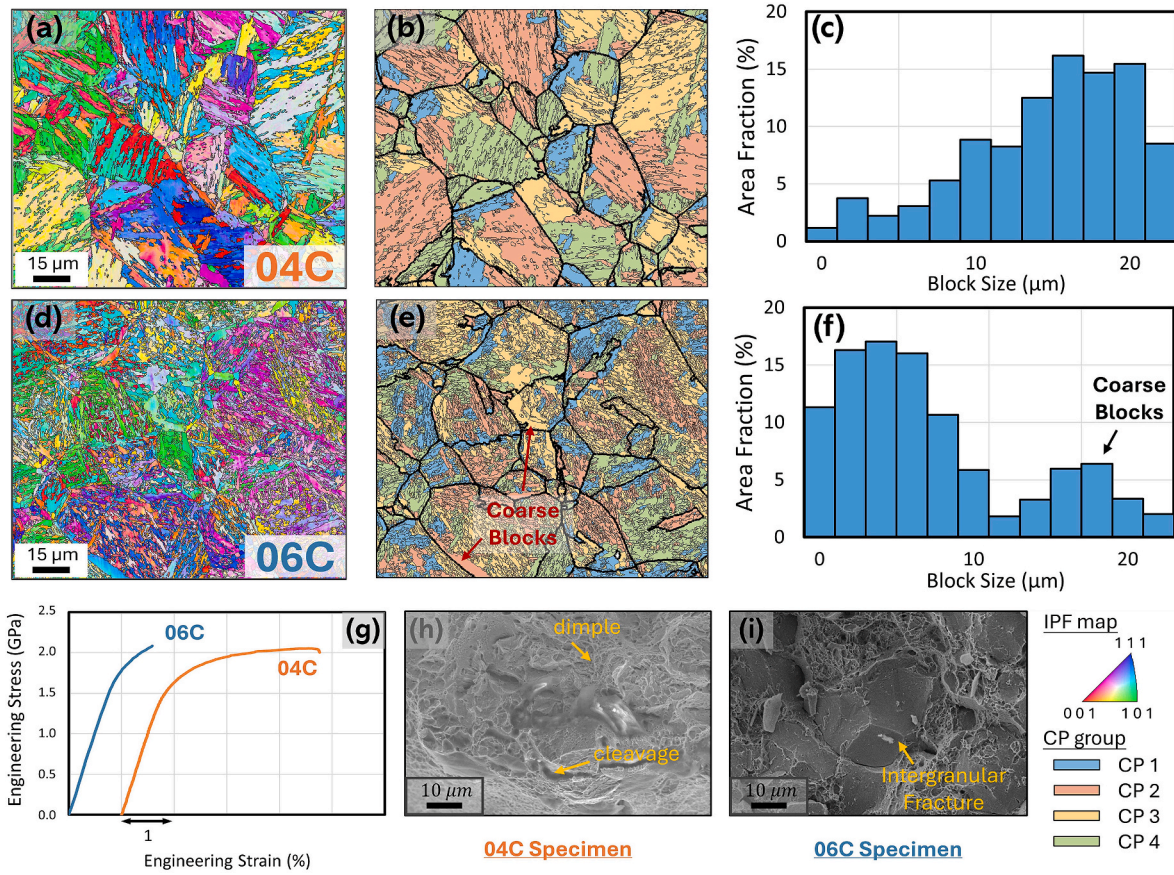


Fig. 1. The IPF maps of the (a) 04C and (d) 06C specimens before tensile deformation. GBs with misorientation angles higher than 10° are labelled. (b) and (e) show corresponding packet group maps, and (c) and (f) show the block size distributions of the two specimens. (g) stress-strain curves of the two specimens. (h) and (i) show the SEM observation results on the fracture surface of the 04C and 06C specimen, respectively.

3.3. Intra-block deformation in 04C martensite

Fig. 3 shows the intra-block plastic deformation mechanism within a PAG of the 04C lath martensite. Fig. 3(a) shows the HP SF map calculated from the initial EBSD, the HP SF and corresponding slip system traces are labelled in the figure. Fig. 3(b) shows the blocks' average strain map at plastic tensile strain of 1.71%. It can be seen that the blocks with high plastic strains (red and orange colours) highly consist with those with high HP SF values (Fig. 3(a)). This agrees with the previous conclusions that HP SF determines the selection of intra-block slip activation [1,12]. Moreover, it is noteworthy that not all blocks with high HP SF values are activated under tensile deformation. This is because the boundary effect and dislocation density also affect the CRSS of the intra-block slip [32].

Fig. 3(c, d) show the HR-DIC strain map at 0.36% and 1.71%. In the labelled blocks (by red arrows in Fig. 3(d)), slip bands were mainly observed near the block boundaries, whereas the inner region of the blocks remains undeformed. With increased tensile deformation, these blocks intensified with higher strains, whereas new slip activation in low-SF blocks is rarely preferred. In this PAG, the intra-block slips are the main plastic deformation mechanisms, whereas boundary deformation along PAGB or packet boundary is not observed.

Moreover, the boundary sliding behaviour in the 04C and 06C specimens also differs in terms of their strain distribution mechanisms. To further demonstrate the difference, representative regions were selected as shown in Figs. 4 and 5.

3.4. Boundary deformation in 04C martensite

Fig. 4 presents (a) the HP SF map, (b) grain's average strain map at 1.71%, (c-f) high-resolution strain maps and (g, h) strain profiles across the boundary as shown in Fig. 4(c). In this PAG, a pronounced strain localization band developed along the inclined packet boundary, accompanied by numerous short and shallow bands inside the surrounding blocks. Notably, many of these grains exhibited relatively low HP SF values, indicating that intra-block slips were not strongly favoured. Nevertheless, the strain map at 1.71% (Fig. 4(e)) exhibited that both the packet boundaries and surrounding blocks were activated, forming a complex and river-like deformation system.

Selecting two strain profiles as shown in Fig. 4(c)–4(g, h) show their strain distribution across the packet boundary. In addition to the major boundary sliding at the packet boundary, many other strain value peaks appeared due to intra-block slips inside the surrounding blocks. Such additional intra-block plasticity can redistribute the strain concentration at the boundaries and reduce the risk of premature intergranular fracture.

Moreover, the slip system selection of the boundary-induced secondary deformation differs from the normal intra-block slips along in-lath-plane slip systems (Fig. 3) [33]. Fig. 4(e, f) show two magnified regions near the packet boundary. In Fig. 4(e), the maximum SF plane is aligned with the packet boundary and has a high SF value of 0.487, whereas the HP has a low SF value of 0.211. During deformation, adjacent intra-block slips were activated along the out-of-lath-plane maximum SF slip system [25]. In Fig. 4(f), the HP is aligned with the packet boundary and has an elevated SF value of 0.401, whereas the maximum SF value is 0.497. Despite this, most intra-block slips were

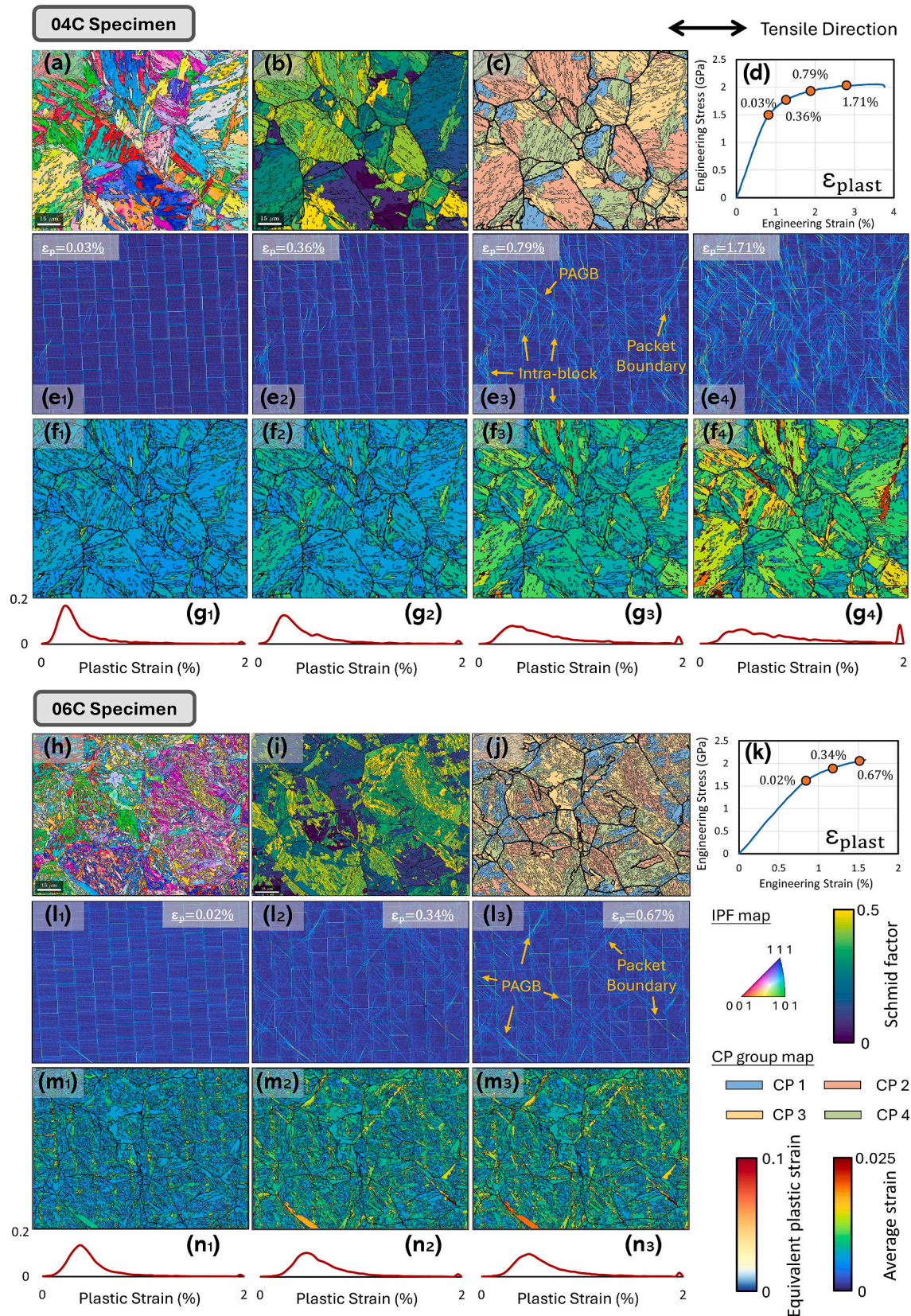


Fig. 2. Initial microstructure and strain maps by HR-DIC of the ROIs in the (a-g) 04C and (h-n) 06C specimens. Initial microstructure includes (a, h) IPF maps, the corresponding (b, i) habit-plane Schmid factor maps and the (c, j) packet maps. (d, k) show the stress-strain curve of the tensile test. (e, l) show the strain maps of the ROIs at various tensile strains. (f, m) show the grains' average strain maps, where the average strain was calculated by overall equivalent plastic strain values versus the grain size in μm . (g, n) show the distribution of grains' average strain values.

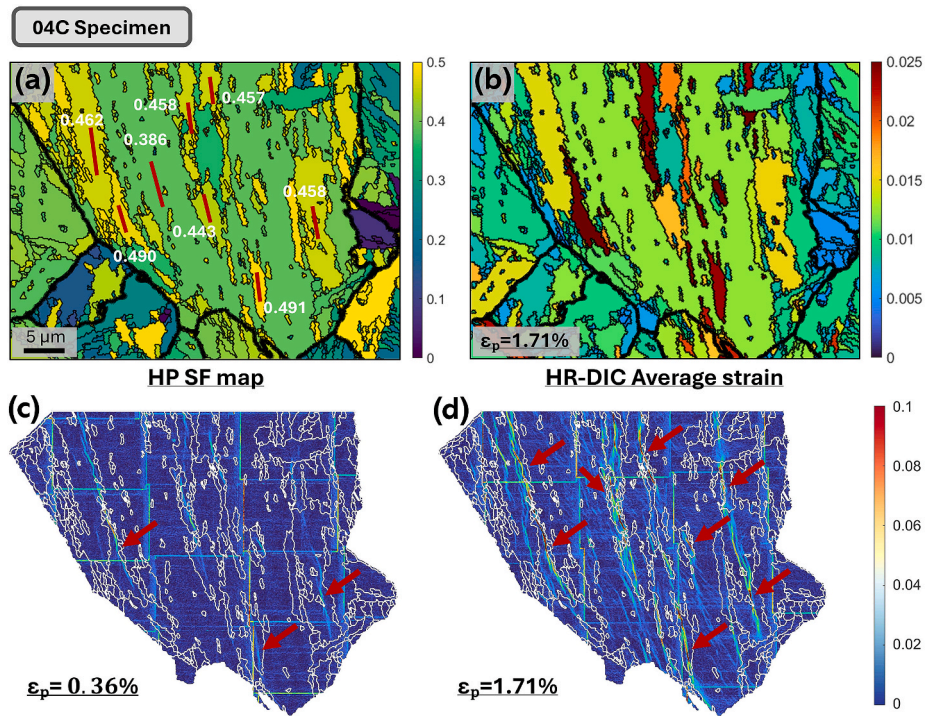


Fig. 3. The intra-block deformation mechanism of a selected PAG in the 04C specimen. (a) the habit-plane Schmid factor map of the PAG. (b) average strain map at plastic tensile strain of 1.71%. (c, d) HR-DIC equivalent strain map of the PAG at tensile plastic strains of 0.36% and 1.71%.

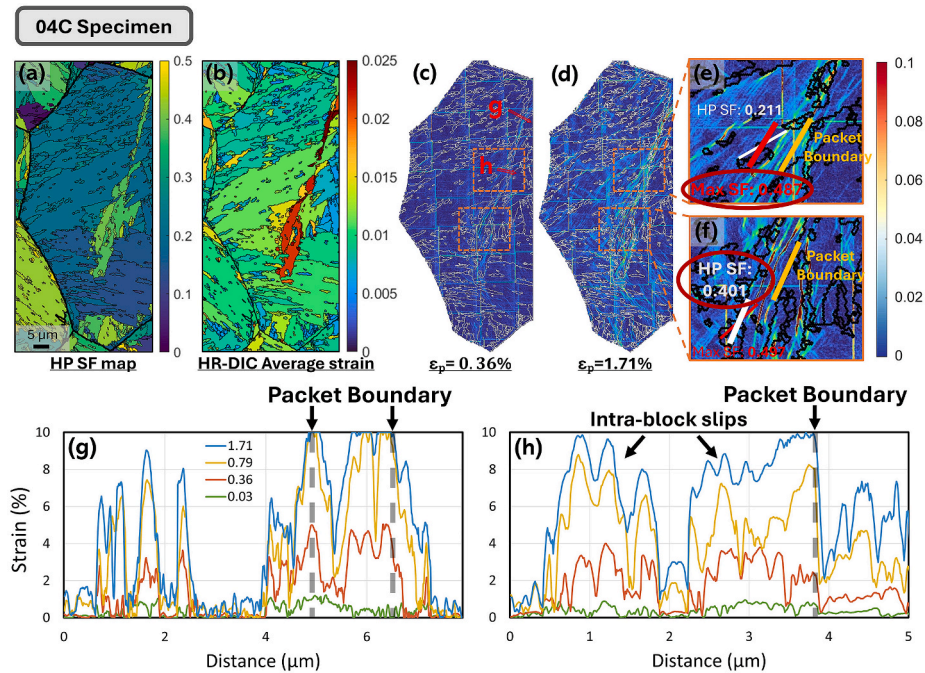


Fig. 4. The deformation mechanism of a selected PAG in the 04C specimen. (a) the habit-plane Schmid factor map of the PAG. (b) average strain map at plastic tensile strain of 1.71%. (c, d) strain maps of the PAG at plastic tensile strains of 0.36% and 1.71%, respectively. (e, f) the zoomed in regions of the 1.71% strain map as labelled in figure (d). (g, h) the strain value profiles across the packet boundary as labelled in figure (c).

developed along the habit-plane slip systems instead of others with even higher SF values. Therefore, the boundary-induced secondary deformation bands in the surrounding blocks were activated aligned with the activated boundary. In this case, either habit-plane slips with lower CRSS value or out-of-lath-plane slips with higher SF values could be activated, provided they have good alignment with the sliding

boundary. Such boundary-induced deformation systems were mainly observed in low- and medium-carbon martensite steels [28].

3.5. Boundary deformation in 06C martensite

In the 06C specimen, Fig. 2(13) shows that most localized

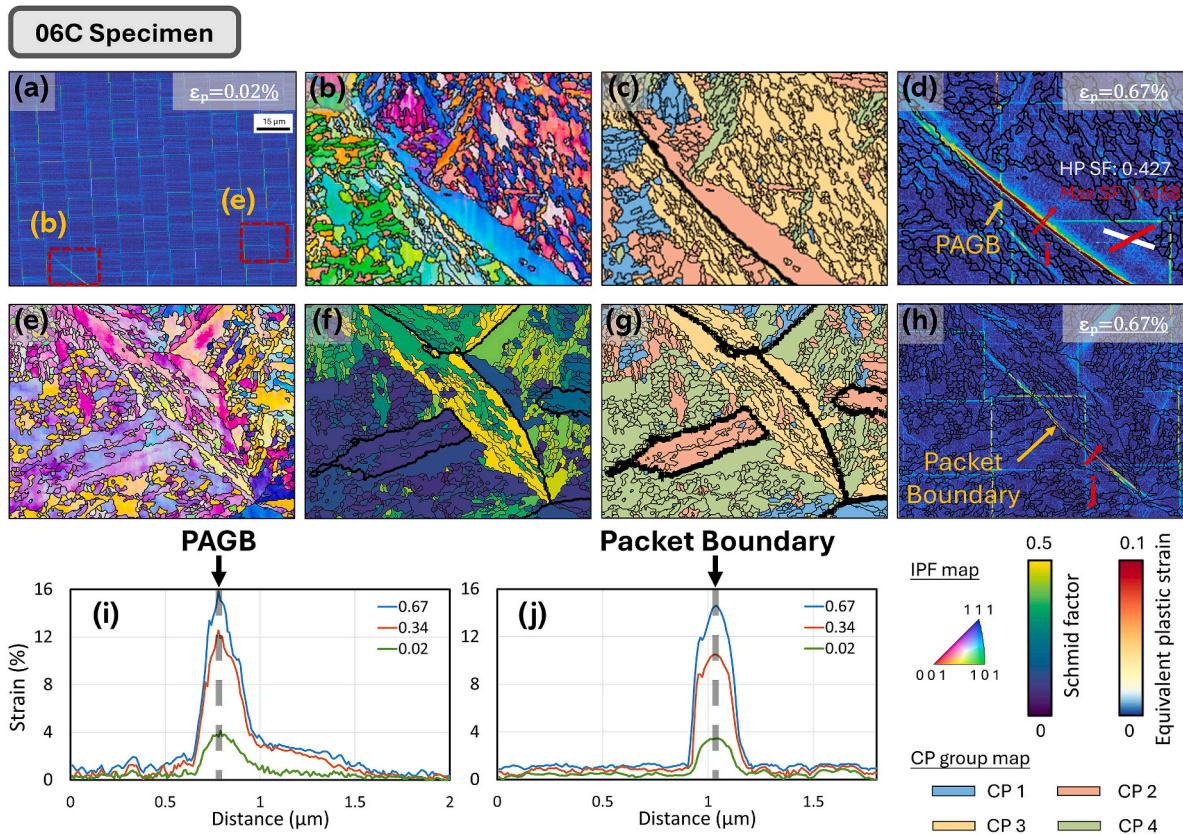


Fig. 5. The deformation mechanism of the 06C specimen. (a) the strain map of the ROI just after yielding. Selecting two regions as examples, region (b) shows the boundary sliding along a PAGB with its (b) IPF map, (c) packet map and (d) strain map at plastic tensile strain of 0.67%, and region (e) shows boundary sliding along a packet boundary with its (e) IPF map, (f) habit-plane Schmid factor map, (g) packet map and (h) strain map at plastic tensile strain of 0.67%. (i, j) show the strain value profiles across the two boundary sliding bands.

deformation bands correspond to the packet boundaries and PAGBs, whereas the intra-block deformation bands are substantially reduced compared with that in the 04C specimen. A possible explanation is that the refined block size in 06C hinders the slip activation inside the blocks [20]. Consequently, several shallow bands were observed along block boundaries, instead of inside the blocks. Therefore, the present section will focus on the boundary deformation behaviour in 06C, which plays a more important role during the tensile deformation.

The boundary deformation behaviour in the 06C specimen is depicted in Fig. 5. Fig. 5(a) shows the strain map of the ROI just after yielding, where several straight and sharp strain localization bands inclined at about 45° with respect to the loading direction. Two regions (b) and (e) were selected as examples of boundary sliding along a PAGB and a packet boundary, respectively.

In region (b) (Fig. 5(b–d)), an intense and sharp band formed strictly along the PAGB. It is notable that although the neighbouring grain exhibits a relatively large size and maximum SF value of 0.457, no slip was activated within the surrounding grains. It is inferred that the higher carbon content concentrated at the boundaries and varied the boundary sliding behaviour. Similar intense and sharp band was also observed in the region (e) (Fig. 5(e–h)) where the strain localization band remained restricted to the boundary and did not extend into surrounding blocks.

Fig. 5(i, j) show the strain value profiles across the two bands. Two severe strain localization bands were developed along the boundaries with thickness of approximately $0.2\text{--}0.3\text{ }\mu\text{m}$. As tensile strain increased, the local strain values intensified and evolved into distinct peaks, while the surrounding blocks remained undeformed. Such single-peak boundary sliding mechanism leads to pronounced strain concentration at PAGBs and packet boundaries, which may contribute to the brittle intergranular fracture in high-carbon martensite steels [20].

Additionally, the 06C specimen also exhibited smaller block sizes and a higher density of LAGBs, which could suppress the activation of intra-block slips near the boundaries [7]. However, the coarse blocks adjacent to PAGBs also showed no slip activation (Fig. 5(b)), indicating that the block size alone cannot explain the boundary strain concentration in high-carbon martensite.

So far, there are two possible explanations for the different boundary deformation behaviours between 04C and 06C martensite, especially the activation of boundary-induced secondary deformation bands: the effect of microstructure variations (block size and CBs), and the effect of CRSS on local slip activation. To clarify their contributions, the following discussion was performed.

4. Discussion

Increasing carbon content in martensite can induce both microstructure variation [13,14] and increase of CRSS for intra-block slip activation [34–37]. The aforementioned results indicated that low-carbon martensite can induce secondary deformation bands surrounding boundary deformation, whereas high-carbon martensite exhibited single-peak strain concentration at deformed boundaries. It is necessary to clarify the fundamental origin of the difference, either because of the carbon-induced microstructure variation or the CRSS difference. Hence, new 06C specimens were tempered at 500°C for 10 min (named 06C-HT specimen) to decrease the CRSS values [38–40], and to investigate its effect on boundary deformation behaviour.

4.1. Plastic deformation of high-temperature tempered 06C martensite

Fig. 6(a) shows the stress-strain curves of the 06C martensite

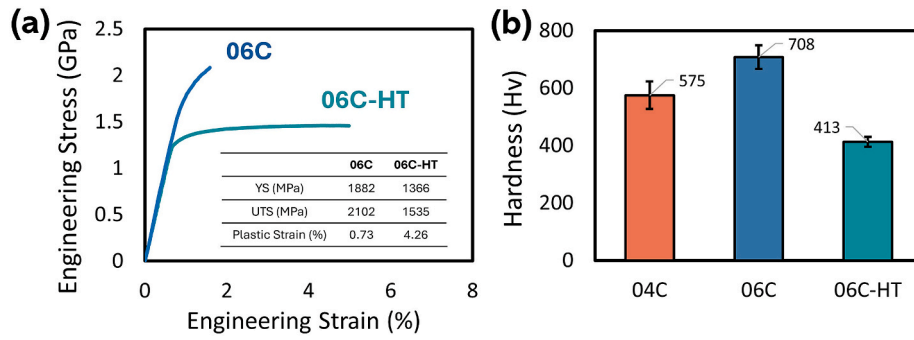


Fig. 6. (a) The engineering stress-strain curves of the 06C martensite specimens after low temperature tempering and high-temperature tempering (06C-HT). YS and UTS stand for yield strength and ultimate tensile strength, respectively. (b) The hardness of the 04C martensite, 06C martensite and 06C-HT martensite specimens. The tempering condition for both 04C and 06C specimen is 200 ° C for 1 h, and that for the 06C-HT specimen is 500 ° C for 10 min.

specimens under the two tempering conditions. Compared with the 06C specimen, the 500 ° C-tempered 06C-HT specimen exhibited significantly decreased yield strength and ultimate tensile strength, but the uniform elongation has improved to about 5%. Notably, the 06C specimen exhibits pronounced strain hardening after yielding, which was not observed for the 06C-HT specimen [41]. Fig. 6(b) compares the hardness values of the three martensite specimens. The 06C-HT specimen shows hardness of 413 Hv, which is lower than the 04C and 06C specimen. This indicates that the plastic deformation activating stress for the 06C-HT specimen is lower than 04C and 06C specimens [42,43].

Fig. 7 shows the plastic deformation behaviour of the 06C-HT specimen. The IPF map of initial microstructure after the 500 ° C tempering is shown in Fig. 7(a), where the block size is similar to that in the 06C specimen [14]. Fig. 7(b, c) show the overall strain map and grains' average strain map at plastic tensile strain of 1.76%. Generally, most deformation bands were still activated near the boundaries, especially near the PAGBs [20]. Moreover, the boundary-induced secondary deformation bands were again observed in this high-carbon martensite.

Taking the red-labelled region in Fig. 7(a) as an example, Fig. 7(d–f) show the local boundary deformation behaviour. Adjacent to the primary deformation along the PAGB, numerous secondary deformation

bands are observed with similar slip directions aligned with the primary boundary. The secondary deformation can dissipate strain concentration at the PAGB, enhanced resistance to premature crack initiation at the boundaries [36,44]. Moreover, Fig. 7(f) shows the locations of the secondary deformation bands, where most bands were activated along block boundaries [45]. It differs from the 04C martensite where the secondary deformation is activated inside the surrounding blocks. This is because the higher C content reduces the average block size, where activation of intra-block slips is hindered by the block boundaries [14].

4.2. Activation mechanism of boundary-induced secondary deformation

The activation of boundary-induced secondary deformation bands is attributed to local CRSS near the primary deformed boundary. In contrast, the role of carbon-induced microstructure variation is less important. When primary boundary deformation is activated, mainly along PAGBs or packet boundaries [17,46], the activation of surrounding secondary deformation is also mechanically preferential [47]. If there are potential low-CRSS sites near the boundary, the activation of these slips can be thereby triggered [10,48], forming surrounding secondary deformation bands. This explains why the slip directions are in

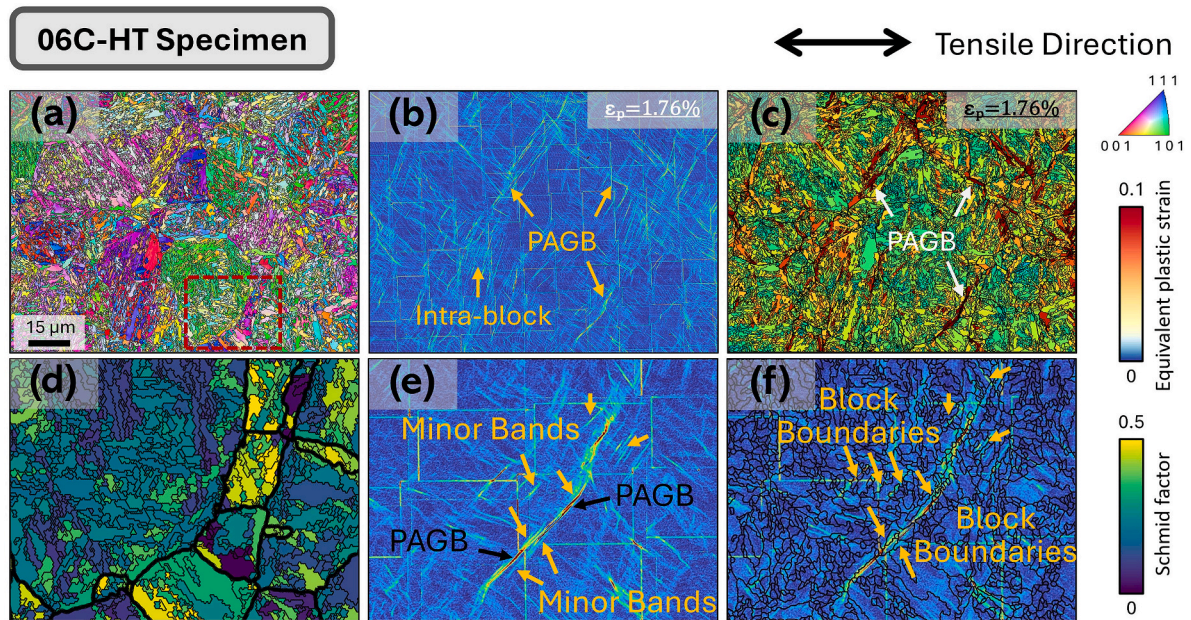


Fig. 7. The deformation mechanism of the 06C-HT specimen, which was tempered at 500 ° C for 10 min. (a) IPF map of the initial microstructure. (b) Strain map by HR-DIC analysis at plastic tensile strain of 1.76%. (c) Grains' average strain value map at plastic tensile strain of 1.76%. Selecting the region highlighted in (a) as an example, (d–f) show the local deformation mechanism along a PAGB, including the (d) habit-plane Schmid factor map, (e) strain map at plastic tensile strain of 1.76%, and (f) a superposition of high-angle grain boundaries and the strain map.

similar orientations with the primary boundaries [1,12]. These potential sites can be intra-block slip systems in 04C or block boundaries in 06C-HT specimens. However, the 06C high-carbon martensite includes much higher CRSS where no preferential deformation site can be activated near the primary boundaries [34,35]. As a result, strain continuously accumulates along the primary boundaries and forms sharp and intense single-peak slip bands [49]. Such strain concentration at boundaries is likely to induce premature crack initiation, which corresponds to the brittle intergranular fracture in high-carbon martensite (Fig. 1(i)) [17,50].

4.3. Deformation of coarse blocks in high-carbon martensite

High-carbon martensite is characterized by small lath-based blocks and twin-based coarse blocks [14]. Compared with the lath blocks, the CBs exhibit lower hardness values and less crystal defects [29,51]. Fig. 8 shows the plastic deformation behaviour of the CBs in the 06C and 06C-HT specimens. In the present 0.6 wt%-carbon martensite, mainly butterfly martensite blocks and single CBs along PAGBs were observed. The DIC results show that plastic slip bands were observed only in a few blocks, while most CBs remain largely undeformed. There is no obvious improvement in the deformability of the CBs by the 500 °C tempering process. Following deformation behaviours of CBs can be concluded.

1. The deformation of CBs follows Schmid law, where slip activation was mainly observed in CBs with high HP SF values.
2. Compared with concentrated slip bands within lath-based blocks (Fig. 3(d)), the deformed CBs exhibit a branch of parallel slip bands with lower strain values (Fig. 8(d)).
3. Different types of CBs may exhibit different deformation mechanisms.

To investigate the Schmid effect in CBs, all 24 slip systems were analysed for the four CBs in Fig. 8. In the CB-1 (Fig. 8(a)), the activated slip system is along the HP with SF value of 0.434, instead of along the highest-SF slip system with value of 0.498 [12]. It is notable that the HP is parallel to the longitudinal direction in this block. In contrast, other CBs (2, 3 and 4) in Fig. 8 have low HP SF values. Despite that there are other high-SF slip systems, these CBs were hardly deformed.

Moreover, different types of CBs include distinct intra-block substructure, i.e., butterfly martensite has dislocations and {112} twin pile

ups along longitudinal directions, while thin-plate martensite consists of {112} twins between smooth and planar interfaces [52–54]. Such substructure leads to a complex deformation system which affects slip activation, dislocation mobility and twin-boundary migration [55]. In this study, plastic deformation of CBs was mainly observed in butterfly martensite blocks, especially those wings with high SF values. In contrast, the single CBs along PAGBs were hardly deformed throughout the tensile test. Further efforts should be made on plastic deformation behaviours of CBs with various types of substructures.

4.4. CRSS variation and effects on plastic deformation

In this study, the CRSS value is employed to measure the stress required for activating plastic deformation. Among the 04C, 06C, and 06C-HT specimens, the CRSS values are expected to vary due to differences in carbon content and tempering condition [38]. Increasing carbon content enhances solid-solution strengthening and increases the interaction between carbon atoms and dislocations, thereby hindering dislocation glides and resulting in a higher CRSS [2]. Consequently, slip activation becomes more difficult in the 06C specimen than in the 04C specimen [37]. Additionally, high-temperature tempering treatment promotes dislocation recovery. In specific, carbon diffusion and carbide precipitation reduce the carbon concentration remaining in the martensitic matrix [38]. Therefore, the CRSS of the 06C-HT specimen is expected to decrease compared with that of the as-quenched 06C specimen [56]. Moreover, the evaluation of CRSS and its effects on the plastic deformation was investigated by EBSD-based CPFEM simulations. Further information is provided in the supplementary material.

Notably, the inhomogeneous CRSS distribution plays an important role in the martensitic steels. Regions possessing a lower local CRSS are more readily activated and tend to accommodate plastic strain, whereas regions with a higher CRSS remain elastically constrained until higher applied stresses are reached [36]. In the 06C specimens, as the CRSS increases inside the martensite blocks, plastic deformation is increasingly accommodated by boundaries. Such preferential deformation at the boundaries forms severe strain concentration and behaves as potential fracture sites [57]. Here, the intra-block CRSS determines the activation of secondary deformation bands near the deformed boundaries. The presence of secondary deformation bands can disperse the strain distribution and contribute to the boundary enhancement (i.e. 04C and 06C-HT specimens).

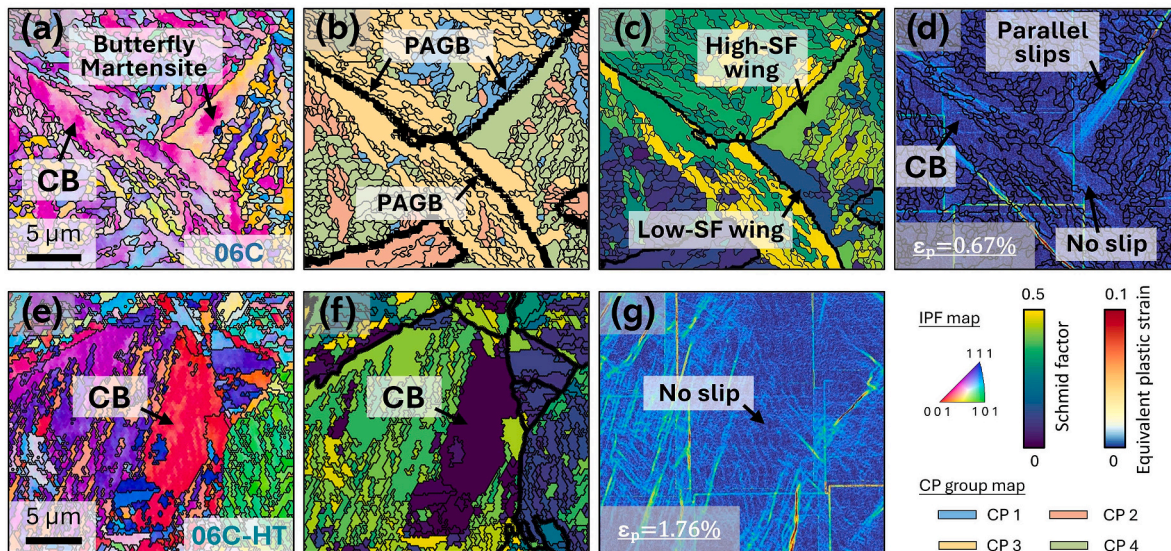


Fig. 8. Deformation mechanisms of the coarse blocks (CBs) in the 0.6 wt%-C martensitic steels under 200 °C tempering (06C) and 500 °C tempering (06C-HT). (a) IPF map of the 06C martensite; (b) reconstructed map; (c) habit plane Schmid factor (HP SF) value map; (d) strain map at 0.67% plastic strain by HR-DIC. (e) IPF map of the 06C-HT martensite; (f) HP SF value map; (g) strain map at 1.76% plastic strain by HR-DIC. Four CBs were labelled by 1-4 as shown in (a) and (e).

Notably, deformation of martensitic microstructure is influenced by various factors, including dislocation density, boundary features and carbide precipitation. Each of these factors affects the deformation behaviour through distinct underlying mechanisms. In this study, their influence on plastic deformation is primarily reflected in the local CRSS value. Plastic deformation occurs only when the resolved shear stress exceeds the local CRSS [36]. Therefore, spatial variations in CRSS governs the activation of secondary deformation bands, strain partitioning, and the degree of deformation localization observed in the present study.

4.5. A comparison on two models of boundary deformation

Fig. 9 shows the statistical results on the activated deformation types in the three specimens. In the 04C specimen, both boundary deformation and intra-block slips were activated to a comparable extent; in contrast, plastic deformation in high-carbon martensite, both the 06C and 06C-HT specimens, concentrates at the PAGBs and packet boundaries. This is because the smaller block size in high-carbon martensite suppresses the activation of intra-block slip, leading to a grain boundary strengthening [14]. This can be observed in the 06C-HT specimen, where the preferential slip activation site is the block boundaries instead of the intra-block sites (Fig. 7(f)) [45].

Moreover, various boundary deformation behaviours were observed in the three specimens, as depicted in Fig. 10. Generally, single deformation bands along boundaries were observed in the 06C martensite, and secondary deformation bands aligned with primary boundary deformation were observed in the 04C and 06C-HT specimens. This indicates that the CRSS dominates the activation of secondary deformation bands, instead of the microstructure difference due to the carbon content difference.

In high-carbon martensite, plastic strain concentrates at PAGBs and packet boundaries, whereas other parts hardly deform. As a result, strain concentration at the deformed boundaries is likely to induce local premature crack initiation, leading to decreased overall tensile strain [20, 39]. In contrast, the boundary deformation in 04C and 06C-HT martensite can trigger surrounding secondary deformation bands, leading to strain partitioning around the deformed boundaries. The presence of these boundary-induced secondary deformation bands can effectively suppress the crack initiation at these boundaries and improve the overall ductility [25]. It is notable that the secondary bands in low-carbon martensite are active within surrounding blocks, whereas secondary deformation bands are observed along block boundaries in high-carbon martensite. However, the activation of secondary deformation in turn requires relatively low CRSS sites near the boundaries [37]. This explains the strength-ductility trade-off of martensitic steels in terms of their deformation mechanisms [58]. The results also suggest

that enhancing the boundary deformability in high-carbon martensite can improve the strength-ductility balance simultaneously [20,40,59].

5. Summary

In summary, the present study investigated the boundary deformation behaviours in martensite with 0.4 wt% (04C) and 0.6 wt% (06C) carbon content. In the 04C martensite, plastic deformation exhibited a combination of intra-block slips and boundary deformation. **In contrast, plasticity in the 06C martensite is dominated by boundary deformation along PAGBs and packet boundaries, whereas intra-block and block-boundary deformation bands accounted for only a minor fraction among the overall deformation features.** Further details about their boundary deformation behaviours can be concluded as follows.

1. The boundary deformation in medium-C martensite (04C) can trigger the activation of secondary intra-block deformation within surrounding blocks. Such boundary-induced secondary deformation bands are activated along high-SF planes aligned with the deformed boundaries, exhibiting local strain partitioning effect.
2. In low-temperature tempered high-C martensite (06C), the boundary deformation exhibits sharp and intense single-peak slip bands strictly along the boundaries. Secondary deformation bands are hardly observed surrounding the primary boundaries. This leads to severe strain concentrations along boundaries.
3. In high-temperature tempered 06C high-C martensite (06C-HT), secondary deformation bands are again observed surrounding the primary boundaries. Compared with the 04C martensite, these boundary-induced secondary deformation bands are activated along block boundaries, instead of within the blocks.

The activation of the boundary-induced secondary deformation is determined by the local CRSS of potential deformation sites near the primarily deformed boundary. Although high C content reduces the average size of martensitic block, it could not suppress the activation of secondary deformation bands in the 06C-HT martensite. Without the secondary deformation bands, high-C martensite exhibits strain concentration at PAGBs and packet boundaries, which may contribute to premature intergranular fracture. In contrast, the secondary deformation along primary boundary deformation leads to local strain partitioning and hence the ductilization of martensitic steels.

CRediT authorship contribution statement

Karel Blanken: Conceptualization, Data curation, Formal analysis, Investigation, Methodology, Software. **Sien Liu:** Conceptualization,

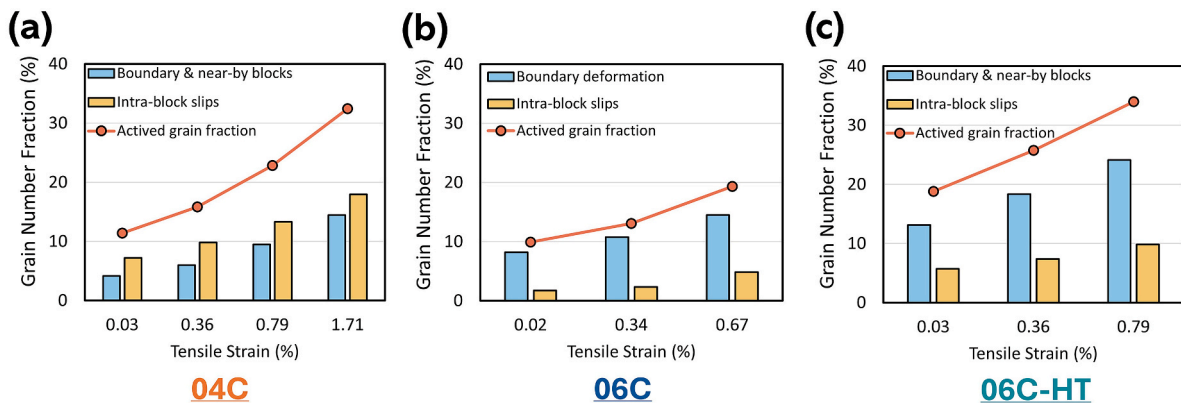


Fig. 9. Statistical results on the deformed grain fraction at various tensile strains for the (a) 04C, (b) 06C and (c) 06C-HT specimens. The boundary deformation here mainly refers to the deformed PAGBs, packet boundaries, and surrounding secondary deformation bands.

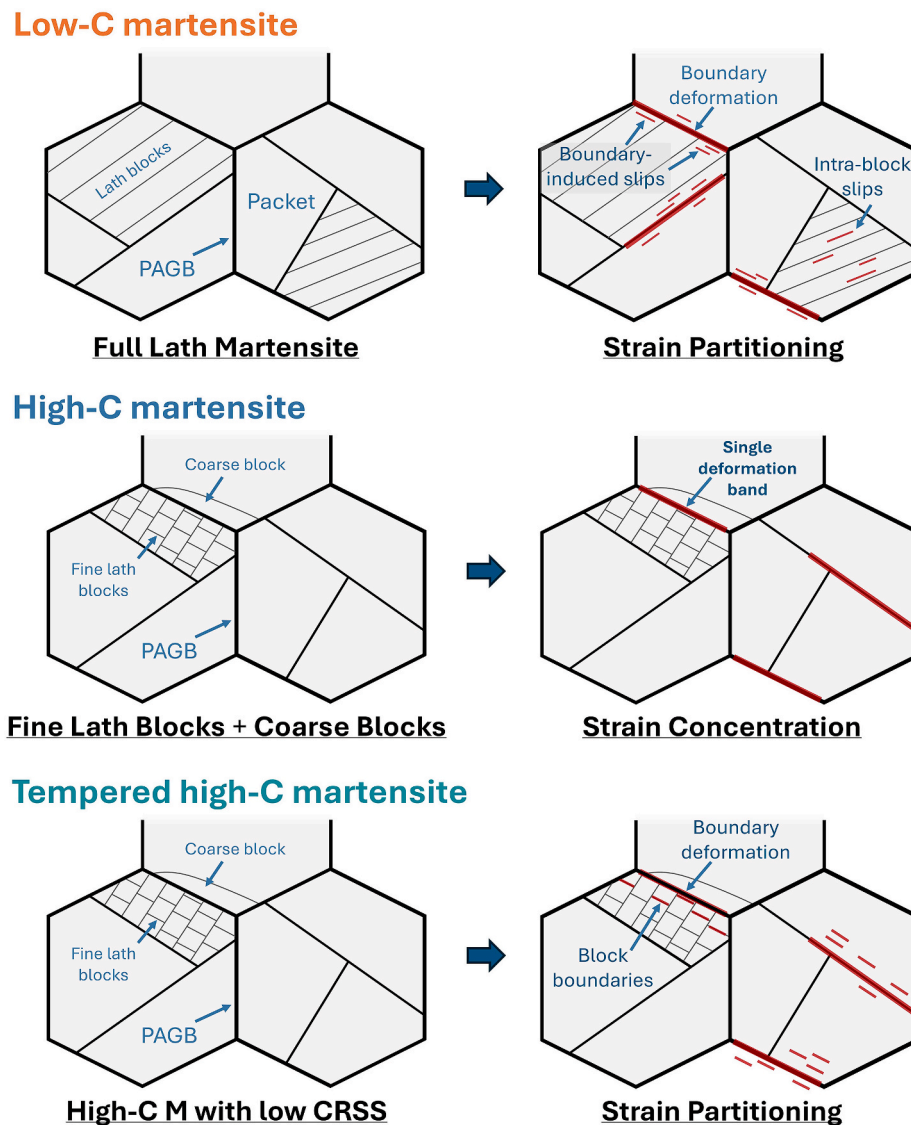


Fig. 10. Schematics of the boundary deformation mechanisms of three different types of martensitic steels, including (a) low- and medium-carbon martensite, (b) high-carbon martensite with low-temperature tempering, (c) high-carbon martensite with high-temperature tempering.

Data curation, Formal analysis, Investigation, Methodology, Software, Supervision, Writing – original draft. **Jiaqiang Dang:** Data curation, Investigation, Methodology. **Fabien Briffod:** Investigation, Software, Writing – review & editing. **Wujun Yin:** Data curation, Methodology, Software. **Shoichi Nambu:** Conceptualization, Funding acquisition, Project administration, Resources, Supervision, Writing – review & editing.

Declaration of competing interest

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

Appendix A. Supplementary data

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.msea.2026.150709>.

Data availability

Data will be made available on request.

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