1. Introduction

One direction in the design of steel sheets used for automobile bodies is to reduce the emission of carbon dioxide by improving fuel efficiency while also improving the safety of people who drive or ride the automobiles. To this end, it is necessary to reduce the thickness of components and control deformation during a crash. Different types of steel are used for different components depending on the required properties, such as deformability and strength. Dual-phase steels are mainly used for components that are required to have deformability while fully martensitic steels are used for components that are required to have high strength above 1 470 MPa.1) Such high-strength steels used in automobiles inevitably deform in accidents, and it is thus important to understand the deformation mechanism of fully martensitic steels that have limited deformability compared with dual-phase steels.

Generally, ductility (i.e., total elongation) decreases with increasing tensile strength. The schematic of the relationship is known as a banana curve.2) In the case of fully martensitic steels, however, there is no strong correlation between the total elongation and the tensile strength (i.e., the total elongation is nearly constant)3) in contrast to the case for interstitial free steels, whose total elongation reduces approximately 20% with a change in the tensile strength of 200 MPa.

Martensitic steels are usually used after tempering, the temperature of which depends on the use. Characteristic deformation properties, such as the independency of the total elongation with respect to the tempering temperature, indicate that the deformation mechanism for fully martensitic steels is not the same as that for ferritic single-phase steels. Many studies have investigated the deformation mechanism of martensitic steels focusing on the microstructure or the hieratical structure of martensite.4–9) Badinier et al.7) reported that the macroscopic yield strength is affected by variations in the local flow stress due to differences in
the amount of cementite in different laths resulting from differences in the increase in dislocation density during quenching. Shibata et al. 8) conducted bending tests with micro-cantilevers, showing that blocks play as obstacles for dislocation gliding.

Relating the hierarchical microstructures of martensite to macroscopic mechanical properties, it is necessary to understand the changes of mechanical properties with tempering temperature. The present study focuses on the change in the uniform elongation with the tempering temperature. The mechanism behind the decrease in the uniform elongation with increasing tempering temperature is discussed considering inhomogeneous plastic deformation after macroscopic yielding and the experimentally measured yield stress, tensile strength, uniform elongation, local elongation, distributions of equivalent plastic strain and nano-hardness. Additionally, the decrease in the uniform elongation with increasing tempering temperature is analysed with a finite element method (FEM) calculation.

2. Experimental Procedure

Table 1 gives the chemical composition of the material used in the present experiments. A 50-kg ingot was made in the laboratory with a vacuum furnace. The ingot was heated at 1 473 K for 600 s and hot rolled to a thickness of 20 mm and then air cooled. Four samples were then produced from the rolled ingot. In producing each sample, a plate was cut using an electro-discharge machine (Mitsubishi MV1200R). The initial strain rate was set at 0.005 s⁻¹. The nano-hardness was measured using a nano-indentor (Elionix, ENT-1100a) with a Berkovich indent. FEM calculations were performed (ANSYS, ver. 2020R2) to investigate the effect of inhomogeneous deformation on the macroscopic stress-strain (s-s) curve.

Precision markers were drawn on the specimen surface to measure local plastic strain after yielding. The width of a square cell was 70 nm and the grid interval was 500 nm. Details of the drawing method have been given elsewhere. 10) Changes in the positions of the apexes of the square markers were observed adopting scanning electron microscopy (SEM) after terminating tensile tests. The Green–Lagrange strain is given as:

\[
\varepsilon_{eq} = \sqrt{\frac{1}{3} (\varepsilon_i^2 + \varepsilon_i \varepsilon_j + \varepsilon_j^2)} \quad (i = x, y, z),
\]

where \(\varepsilon_i\) and \(\varepsilon_j\) are displacement in x direction and y direction after deformation, respectively. \(\varepsilon_i\) and \(\varepsilon_{eq}\) are normal plastic strain, the plastic shear strain and the equivalent plastic strain in the square.

3. Results and Discussion

3.1. Mechanical Testing

Figures 2(a)–2(d) shows SEM images for TP100, TP200, TP300 and TP400 while Figs. 2(e), 2(f), 2(g) and 2(h) shows enlarged images of (a), (b), (c) and (d), respectively, revealing that the microstructures are nearly independent of the tempering temperature. No cementite was found at 100 °C. This result can be explained in that the precipitation hardening of cementite is appreciable in TP200, and the proof stress then decreased with the decreasing dislocation density during tempering. Shibata 8) conducted bending tests with micro-cantilevers, showing that blocks play as obstacles for dislocation gliding.

Table 1. Chemical composition of the material used in this study.

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>N</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.20</td>
<td>0.05</td>
<td>2.0</td>
<td>0.005</td>
<td>0.0005</td>
<td>0.03</td>
<td>0.003</td>
<td>bal.</td>
</tr>
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</table>

mass%
tempering. The tensile strength decreased with the tempering temperature.

Figures 5(a) and 5(c) respectively show maps of the equivalent plastic strain after the plastic deformation of TP200 and TP300, where the tensile direction is horizontal. The observed areas are nearly the center of the tensile specimen. The average value of the strain in a measurement area of approximately $100 \times 100 \, \mu m^2$ was 0.07 in (a) and 0.06 in (c). The average strains are close to the uniform elongation limit macroscopically obtained from s-s curves shown in Fig. 4(b). It is noted that the distribution of the strain accumulation in TP200 was inclined approximately $45^\circ$ to the tensile direction, while the distribution of strain in TP300 was relatively uniform compared with that in TP200. Figure 6 presents histograms of the equivalent plastic strain taken from the areas shown in Figs. 5(a) and 5(c). The cumulative relative frequency of TP300 exceeds 0.9 at an equivalent plastic strain of 0.075 while that of TP200 exceeds 0.9 at the equivalent strain of 0.125. In addition, the cumulative relative frequency of TP200 increases more gently above a value of 0.9 than that of TP300. These results indicate that the distribution of the relative frequency over 0.9 for TP200 is larger than that for TP300. That is to say, the tendency of the histograms quantitatively shows that the distribution of the equivalent plastic strain in TP200 (Fig. 5(a)) is more inhomogeneous than that in TP300 (Fig. 5(c)).

Figures 5(b) and 5(d) show orientation maps before plastic deformation for the same areas shown in Figs. 5(a) and 5(c), respectively. Figure 5(a) indicates that plastic strain accumulates near some block boundaries in TP200; however, this does not mean that all block boundaries can be sites for strain accumulation. Some of the strain accumulations inclined $45^\circ$ to the tensile direction penetrate the block boundaries. In TP300, as shown in Fig. 5(c), the plastic strain accumulates vertically in the figure; i.e., the plastic strain accumulates near block or prior austenite grain boundaries that are perpendicular to the tensile direction. This is in good agreement with past reports that the minimum unit for the strain to concentrate is blocks.\textsuperscript{4,9,12} We do not discuss further the strain accumulation site because it is not the main focus of this paper. Instead, it is again stressed here that the distribution of the equivalent plastic strain in TP200 is more inhomogeneous than that in TP300, the detail of which is discussed later.

Figure 4(b) indicates that the total elongation is nearly independent of the tempering temperature although that...
Fig. 4. (a) Nominal stress–strain curves for TP100, TP200, TP300 and TP400. (b) Dependence of the tensile strength, 0.2% proof stress, total elongation, uniform elongation and local elongation on the tempering temperature. (Online version in color.)

at 100°C is slightly lower than those at other tempering temperatures. The uniform elongation decreases with the tempering temperature while the local elongation increases with the tempering temperature. These results indicate that the nearly constant trend of the total elongation with the tempering temperature is due to the balance between the decrease in the uniform elongation and the increase in the local elongation with the tempering temperature. It is noted that the uniform elongation decreases with tempering temperature in the low-temperature tempering of low-carbon martensitic steels used in this study. We focus on the decrease in the uniform elongation with tempering temperature hereafter, which is a characteristic of martensitic steels. True stress–true strain curves and curves of the work-hardening rate, $d\sigma/d\varepsilon$, are next obtained to investigate the deformation behaviour immediately after macroscopic yielding in s-s curves.

Figure 7 shows the true stress–true strain curves and $d\sigma/d\varepsilon$ for TP100, TP200, TP300 and TP400. $d\sigma/d\varepsilon$ initially decreases abruptly from the value of Young’s modulus when the plastic deformation begins macroscopically and the slope of $d\sigma/d\varepsilon$ then becomes gradual as the tensile elongation increases. Arrows in the figure indicate changes in the slope of $d\sigma/d\varepsilon$; and the corresponding strains are given in Table 2. The area in which the strain is lower than that at the slope change is called stage I while that in which the strain is higher than that at the slope change called stage II.\(^{13}\)

$\sigma/d\varepsilon$ for TP400 decreases nearly vertically in stage I when the plastic deformation starts macroscopically and more gradually in stage II, which is similar to the case for a ferritic steel. Meanwhile, $\sigma/d\varepsilon$ for TP100, TP 200 and TP300 remains higher until a higher strain in stage I; i.e., a higher value of $\sigma/d\varepsilon$ is maintained for a lower tempering temperature. Although the transition temperatures of TP100, TP200 and TP300 are not obvious compared with that of TP400, the transition of the trend of $d\sigma/d\varepsilon$ is between the flow stress of 3 500 and 4 000 GPa. Because the decrease rates in stage II is not so sensitive to the tempering temperature, the strain corresponding to the end of the local deformation at $d\sigma/d\varepsilon = \sigma$, known as the Considère condition, increases as the tempering temperature decreases. It is therefore concluded that the change in the magnitude of

\(^{1}\) It is noted that the definitions of stages I and II are different from those for work-hardening in single crystals.
uniform elongation with the tempering temperature in fully martensitic steels is due to the change in the magnitude of the strain in stage I with the tempering temperature. This finding can be specific to fully martensitic steels which have small magnitudes of uniform elongation, compared with ferritic steels.

It has been reported from neutron diffraction analysis that martensitic steels undergo both elastic and plastic deformation immediately after macroscopic yielding.\textsuperscript{14} Ungár et al.\textsuperscript{15} and Harjo et al.\textsuperscript{16} made in-situ tensile observations based on neutron diffraction and found that an as-quenched 1 100-GPa class of lath martensite with 0.2% carbon exhibited load transfer between soft packets and hard packets immediately after macroscopic yielding. That is to say, there was a distribution of areas in which plastic deformation began and elastic deformation continued even after the onset of macroscopic yielding. The load transfer is often observed in dual-phase steels, such as those containing phases or microstructures with different yield stresses.\textsuperscript{17–20} It is noted that the fully martensitic steels used in this study have a load transfer even though there is a single microstructure owing to the inhomogeneity of the microstructure of martensite. This suggests that the change in the load transfer with the tempering temperature affects the magnitude of stage I shown in Fig. 7. In TP100 having the lowest tempering temperature, the volume fraction maintaining elastic deformation locally is the largest while the volume fraction having plastic deformation locally is the smallest among the specimens. The volume fraction having plastic deformation locally increases with the tempering temperature. To reveal the relationship between the length of stage I and the volume that maintains elastic deformation even after the onset of macroscopic plastic deformation, s-s curves were obtained using the FEM and considering the variation of the local yield stress.

### 3.2. FEM Modelling Considering the Inhomogeneity of Local Yield Stress

Figure 8(a) shows the model used in the FEM. The $x \times y \times z$ dimensions of the model are $1 \times 50 \times 50$. One of the apexes of the model is at the origin. The interior of the model has a honeycomb structure as shown in the figure. There are approximately 114 000 nodes and 20 000 elements. The nodes at interfaces between the honeycombs are sheared. The honeycomb structure has a uniform cross section in the $x$ direction. To investigate the inhomogeneity of the local yield stress due to the microstructure of martensite, different mechanical properties, such as hard, medium and soft properties, as given in Table 3, are assigned to the hon,
Assigned Young’s modulus and Poisson’s ratio are 200 GPa and 0.3, respectively. A linear approximation is adopted in the stress – strain curve after yielding. Three models are used. A model comprising hard honeycombs and soft honeycombs with a ratio of 75:25 is denoted H75S25, a model comprising medium honeycombs and soft honeycombs with a ratio of 25:75 is denoted M25S75, and a model comprising only soft honeycombs is denoted S100. Blue honeycombs in Fig. 8(b) have the mechanical property of a larger volume fraction in the models of H75S25 and M25S75. H75S25 and S100 correspond to TP100 and TP400, respectively. The bottom surface is fixed as indicated in Fig. 8(b), and the top surface is moved in the y direction the macroscopic tensile strain of 0.02.

Figure 9 presents macroscopic s-s curves obtained from the force acting on and the displacement of the top surface of the model presented in Fig. 8. The dashed lines in the figure represent $\sigma/\varepsilon$, and an enlarged curve near yield points is shown at the top right of the figure. The s-s curve for S100 is plotted in the figure; however, it is not clearly seen as it nearly overlaps the line for M25S75, as shown in the enlarged image at the top right. In S100, plastic deformation starts at all honeycombs together when the stress reaches the yield stress of 1 100 MPa for soft honeycombs. $\sigma/\varepsilon$ suddenly decreases from a Young’s modulus of 200 GPa and the Considère condition is fulfilled immediately after the macroscopic yielding. It is noted that $\sigma/\varepsilon$ suddenly decreases if all the regions begin plastic deformation at the same time in S100. There is then a transition from stage I to stage II immediately after the macroscopic yielding of S100.

Figure 10(a) presents the equivalent plastic strain in M25S75 at a macroscopic tensile strain of 0.006. Soft honeycombs begin plastic deformation when the local stress reaches 1 100 MPa while medium honeycombs maintain elastic deformation at a macroscopic tensile strain of 0.006. Then, the transition strain of stage I and stage II is approximately 0.006 in the model calculated with the FEM. Medium honeycombs maintain elastic deformation, where the stress does not reach the yield stress of 1 400 MPa for medium honeycombs even at the onset of stage II. Figure 9 shows that $\sigma/\varepsilon$ abruptly decreases in M25S75, however, the transition of stage I and stage II shifted to higher strain than that in S100.

Figure 10(b) presents the equivalent plastic strain in H75S25 at a macroscopic tensile strain of 0.006. As seen in M25S75, soft honeycombs begin plastic deformation when the local stress reaches 1 100 MPa upon macroscopic yield-

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Table 3. Mechanical properties of each unit used in FEM calculations.

<table>
<thead>
<tr>
<th>Unit</th>
<th>Yield stress (MPa)</th>
<th>Work-hardening rate (MPa)</th>
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<tbody>
<tr>
<td>Hard</td>
<td>1 700</td>
<td>6 000</td>
</tr>
<tr>
<td>Medium</td>
<td>1 400</td>
<td>6 000</td>
</tr>
<tr>
<td>Soft</td>
<td>1 100</td>
<td>6 000</td>
</tr>
</tbody>
</table>
ing while hard honeycombs maintain elastic deformation. In the s-s curve of H75S25 in Fig. 9, the model macroscopically yields and shows work-hardening. In addition, the s-s curve of H75S25 obtained using the FEM is round around the yield point and has higher apparent work-hardening rate as seen for the experimental curve of TP100. The FEM calculation indicates that $\frac{d\sigma}{d\varepsilon}$ gradually decreases after macroscopic yielding and the transition from stage I to stage II is vague if the volume fraction which maintain elastic deformation even after macroscopic yielding is larger. The distributions of the equivalent strain in Figs. 10(a) and 10(b) roughly correspond to those in Figs. 5(a) and 5(c), respectively.

It is possible to consider that stage I is the case that the specimen/model has both areas exhibiting plastic deformation and areas maintaining elastic deformation while stage II is the case that the specimen deforms fully plastically. The volume fraction showing the plastic or elastic deformation should change with the tempering temperature. That is to say, the duration of stage I in Fig. 7 should depend on the inhomogeneity of the local yield stress in the specimen due to the microstructure of martensite. Summarising the above, at a lower tempering temperature, the variation of the local yield stress reflects the inhomogeneity of the microstructure of martensite, where the volume fraction that maintains elastic deformation locally is larger at the macroscopic yield point. This allows $\frac{d\sigma}{d\varepsilon}$ to remain higher after yielding. Meanwhile, in TP400, the volume fraction maintaining elastic deformation locally is small enough for $\frac{d\sigma}{d\varepsilon}$ to abruptly reduce after the yield point. The drop of $\frac{d\sigma}{d\varepsilon}$ should reduce the uniform elongation as shown in Figs. 7 and 9 in such a martensitic steel with high strength. The fact that the local elastic deformation in stage I influences the magnitude of the uniform elongation is specific to martensitic steels whose uniform plastic strain is no more than a few percent.

It is reasonable to consider that elastic deformation remains even after the onset of macroscopic plastic deformation in martensite, which is supported by the results of previous works that showed that s-s curves immediately after yielding can be reproduced by assuming the distribution of flow stress. The variation of the local yield stress is next discussed.

3.3. Nanoindentation for the Estimation of the Variation of the Local Yield Stress

The comparison of s-s curves experimentally obtained and those obtained from FEM analysis suggest two points. (1) In the specimens with lower tempering temperatures, there should be a distribution of the values of local yield stress due to the microstructure. The range of the distribution should be wide, and the frequency of high yield stress should be larger. (2) The range of the distribution for the local yield stress should narrow with an increase in the tempering temperature, and the magnitude of local yield stress should decrease. To investigate the distribution of the variation of local yield stress, the variation of the magnitude of nanohardness was measured with the large number of measured points were performed, eliminating the measurement error. 676 points were measured every 2 $\mu$m interval with an applied load of 0.5 mN.

Figure 11 presents box plots of the nanohardness of TP100, TP200, TP300 and TP400, showing that the variation in hardness decreases with tempering temperature and the variation and average values for TP300 and TP400 are nearly the same. Although the nanohardness and yield stress do not exactly correspond, Fig. 11 suggests that there is a combination of areas for higher yield stress and lower yield stress, which supports the pervious discussion that some regions maintain elastic deformation locally even after macroscopic yielding. The circle equivalent diameters of indentation marks in TP100, TP200, TP300, and TP400 are 260 $\pm$ 160, 270 $\pm$ 30, 370 $\pm$ 50, 390 $\pm$ 120 nm, respectively. The sizes of the indentation marks are smaller than that of block size. Additionally, the results support a previous finding that martensitic steels have a lower elastic limit.

Although there are variations in nanohardness lath to lath, it is unlikely that the variations affect the macroscopic yield stress. Nanoindentation tests were performed to reveal the size scale relating to the macroscopic mechanical property. Figures 12(a) and 12(b) respectively shows an SEM image of the area indented and a heat map of nanohardness obtained for TP200. Note that the indents shown in this figure are not included in the data for Fig. 11. The magnitude of nanohardness roughly varies among three levels, the areas of which are denoted A, B and C as shown Fig. 12(b). Although the magnitude of nanohardness varies lath to lath, it is noted that the magnitude varies much more at a length scale over 10 $\mu$m. This indicates that a variation in nanohardness beyond at least the length scale of blocks affects the macroscopic yielding.

The assumption that the lath-to-lath variation in nanohardness does not contribute to the macroscopic mechanical properties is supported by the following. Ohmura et al. investigated the correlation between microhardness and nanohardness using single-crystal body-centred-cubic metals and lath martensite. They concluded that the block size controls the hardness of lath martensite because the nanoindent size was a few times larger than the lath width but smaller than the block and the coefficient in Eq. (6) is lower than that in Eq. (5). Shibata et al. mechanically tested Fe-23Ni fully martensitic steels using a micro-cantilever and found that the yield stress of a specimen with block boundaries is higher than that of a specimen without block

![Figure 11](https://example.com/fig11.png)
boundaries. It is thus concluded that the variation in the local yield stress among blocks or prior austenite grains controls the local elongation in the uniform elongation.

4. Conclusions

The mechanism behind the change in the uniform elongation with tempering temperature was investigated in fully lath martensitic steels adopting tensile tests, strain mapping, the FEM, and nano indenting. The following findings were made from the results of the study.

(1) The uniform elongation decreased while local elongation increased with tempering temperature, resulting in nearly constant total elongation irrespective of the tempering temperature.

(2) Equivalent plastic strain maps and a nanohardness map indicated the distributions of variations should have a length scale larger than at least the block size.

(3) Finite element analysis indicated that the distribution of the local yield stress changes macroscopic stress-strain curves immediately after macroscopic yielding. The decrease in the uniform elongation with the tempering temperature can be explained by the decrease in the volume fraction that maintains elastic deformation immediately after the macroscopic yielding.

(4) A large number of nanoindents indicated that the distribution of the magnitude of nanohardness became narrower with the tempering temperature. Additionally, the absolute value of the hardness decreased with the tempering temperature. This suggests that the variation of local yield stress narrow with tempering temperature.

(5) There is the variation in the magnitude of the local yield stress lath to lath; however, the variation in the local yield stress that controls the uniform elongation should have a length scale larger than the block size.

(6) The effect of the inhomogeneity of local yield stress on the uniform elongation is specific to martensitic steels that show the uniform elongation no more than a few percent.

(7) Because the magnitude of yield stress depends on strain rate, it is expected that, in the case of low temperature tempering, the higher global strain rate should induce larger uniform elongation for the same type of martensitic steels as those used in this study.

Acknowledgement

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