Effects of Solute Carbon on the Work Hardening Behavior of Lath Martensite in Low-Carbon Steel

Taku NIINO,1) Junya INOUE,1,2)* Mayumi OJIMA,1) Shoichi NAMBU1) and Toshihiko KOSEKI1)

1) Department of Materials Engineering, The University of Tokyo, 7-3-1 Hongo, Bukyo, Tokyo, 113-8656 Japan.
2) Research Center for Advanced Science and Technology, The University of Tokyo, 4-6-1 Komaba, Meguro, Tokyo, 153-8904 Japan.

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The work hardening behavior and the change in the dislocation density of lath martensite at strain levels of less than 15% under uniaxial tensile loading were investigated. It was clarified that the work hardening rate and the multiplication of dislocation become more prominent as the solute carbon content increases. The change in the mobile dislocation density during deformation was evaluated by studying dynamic strain aging behavior, and it was found that the annihilation of mobile dislocations becomes slower at a higher carbon content. The findings were further examined by a modified Kocks-Mecking-Estrin model proposed in order to explicitly clarify the changes in the mobile and sessile dislocation densities during deformation. From the model-based analysis, it is also suggested that the solute carbon retards the formation of dislocation cells by reducing the mobility of dislocations. These findings were also corresponded well with the observation of the dislocation structure using a transmission electron microscope.

KEY WORDS: martensite; low carbon steel; work hardening; mobile dislocation.

1. Introduction

Even though lath martensite in low carbon steel has received considerable attention for high-strength applications, its low ductility significantly limits its applicability. Thus, a number of studies1–12) have been conducted to clarify the deformation behavior of lath martensite. The unique characteristics of lath martensite, such as an extremely high initial dislocation density and supersaturated solute carbon, have been suggested to be responsible for its deformation behaviors. For instance, Nakashima et al. demonstrated that the extremely high dislocation density of lath martensite in ultra low-carbon Ni steel is the source of its low elastic limit, and further suggested that it also induces the rapid annihilation of mobile dislocations with opposite signs at a very early stage of deformation, hence resulting in the formation of a dislocation cell structure below 10% strain.1) Leslie and Sober investigated the mechanical response of lath martensite with various carbon contents under a very small tensile strain, and it was found that the annihilation of mobile dislocations becomes slower at a higher carbon content.2) The applied strain was limited to around the yield point because lath martensite easily fractures with a small plastic strain under tension. Accordingly, in the present literature on martensitic steels, the local deformation behavior as well as the strain-hardening behavior of lath martensite at a strain level of larger than 10% has not yet been completely elucidated. To improve this situation, we have employed a multilayered steel composite, in which lath martensite can be uniformly elongated under uniaxial tension to a strain of over 50% by selecting an appropriate layer thickness and mechanical properties of each component.13,14) This enhanced elongation enabled us to conduct detailed studies to clarify the unknown deformation behavior of lath martensite at larger strain levels,15–19) and a number of findings regarding the local deformation behavior were obtained. In this study, special emphasis is placed on understanding the strain-hardening behavior of lath martensite to clarify the effects of the solute carbon on the dislocation behavior at a strain level of around 10%. The dynamic strain aging effect at a low temperature was investigated to clarify the change in the mobile dislocation density, and a modified Kocks-Mecking-Estrin model was developed to clarify the effect of the solute carbon on the rate of dislocation cell structure formation.

2. Experimental

2.1. Materials

In this study, four lath martensitic steels with different solute carbon contents were selected, and a nanocrystalline...
ferrite (NCF) film and an interstitial free steel (IF steel) were employed as reference materials to obtain the typical mechanical behavior of the ferrite. To achieve uniform elongation of over 10%, each lath martensitic steel was embedded in a multilayered structure. As a ductile component of the multilayered structure, type 316L austenitic stainless steel (SS316L) was selected. The chemical compositions of the steels prepared are shown in Table 1.

A martensitic steel layer was sandwiched by two SS316L layers, and very thin Ni films were inserted between the martensitic steel layer and the SS316L layers to minimize carbon diffusion during the manufacturing process. The thickness of the multilayered steel composite was reduced to 1.0 mm by hot-rolling and subsequent cold-rolling. The resulting thickness of the martensitic steel layer was controlled approximately 150 μm. Tensile test specimens were machined from the multilayered steel composite, the geometry of which is shown in Fig. 1. After machining, solution heat treatment was applied to the multilayered specimens at 1 373 K for 120 s, which was followed by water-quenching. The specimens were further heat-treated at a temperature of 1 273 K for 300 s followed by water-quenching to obtain a full lath martensitic structure in the martensitic steel layers. To obtain equiaxial ferrite grains, the IF steel was air-cooled after austenitizing treatment at 1 273 K for 300 s. Tensile specimens were also fabricated with the same geometry as the multilayered specimens from the IF steel.

NCF films with grain sizes from 100 nm to 500 nm were fabricated on IF steel substrates. On a mirror polished IF steel surface, a 1 μm-thick pure silver layer was fabricated by sputter deposition, and an ultra low-carbon steel film was further formed on the silver layer by electron beam vapor deposition. The sample was heat-treated at 773 to 923 K for 1 h to obtain the desired grain size.

**2.2. Estimation of Total Dislocation Density**

The dislocation densities of the martensite at various strain levels were estimated by X-ray diffraction.\(^{11,20,21}\) Prior to the X-ray diffraction analysis, prestrain was applied to the specimen at a strain rate of 1.0×10^{-3} s^{-1}, and one side of the SS316L layers was completely removed from the multilayered specimen to expose the lath martensite layer to the X-ray beam. The total dislocation density was estimated by the following procedure.

First, the microscopic lattice strain \( \varepsilon \) was estimated from the equations\(^{20}\)

\[
\beta \frac{\cos \theta}{\lambda} = 2\varepsilon \frac{\sin \theta}{\lambda}, \quad \text{........................... (1)}
\]

\[
\beta^2 = \beta_1^2 - \beta_0^2, \quad \text{........................... (2)}
\]

where \( \beta_0 \) and \( \beta_1 \) are the half widths of the (211)\(_6\) peak of the well-annealed IF and martensitic steels, respectively. Then, the dislocation density was estimated using the Williamson-Smallman equation:\(^{21}\)

\[
\rho = 14.4 \frac{\varepsilon^2}{b^2}, \quad \text{........................... (3)}
\]

where \( b \) is the magnitude of the Burgers vector, and a value of 0.247 nm was used in this present study. It has been reported that the conventional method, so-called Williamson-Hall (WH) method, tends to overestimate the dislocation density of martensitic steels, and the modified Warren-Averbach analysis has been suggested as the alternative method for the more precise estimation of the dislocation density.\(^{22}\) However, in this present study the conventional method was applied, in order to maintain the consistency with the previous study of the evolution of dislocation density.\(^{11(1)}\)

**2.3. Evaluation of Mobile Dislocation Density**

The mobile dislocation density in the lath martensite was evaluated by investigating the dynamic strain aging effect. Dynamic strain aging occurs when the solute carbon atoms have sufficient mobility to keep up with moving dislocations and form a solute atmosphere in the vicinity of the dislocation cores. However, as the applied strain rate increases and the solute carbon atoms start to escape from the solute atmosphere, the dynamic strain aging effect is gradually lost. Accordingly, dynamic strain aging emerges as negative strain rate dependence of the flow stress, and a critical strain rate, \( \dot{\varepsilon}_{\text{min}} \), exists where the solute carbon atoms can no longer keep up with the moving dislocations. In the present study, \( \dot{\varepsilon}_{\text{min}} \) was estimated by finding the strain rate giving the minimum flow stress.

The velocity of a solute carbon atom around a dislocation core is given as

\[
\nu = \frac{AD \sin \alpha}{kT} \frac{1}{r^2}, \quad \text{........................... (4)}
\]

where \( A \) is a constant, \( D \) is the diffusion coefficient, \( k \) is the Boltzmann constant, and \( T \) is the absolute temperature. When the solute atmosphere around a dislocation is unstable

\[
\alpha = \frac{2\pi}{\beta}
\]

where \( \beta \) is the Burgers vector.

**Table 1.** Chemical compositions of the steels prepared in the present study in mass%.

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
<th>Cr</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.001% C</td>
<td>0.001</td>
<td>0.23</td>
<td>0.24</td>
<td>0.005</td>
<td>0.0004</td>
<td>16.0</td>
<td>–</td>
<td>bal.</td>
</tr>
<tr>
<td>0.06% C</td>
<td>0.06</td>
<td>0.50</td>
<td>0.40</td>
<td>0.001</td>
<td>0.001</td>
<td>14.3</td>
<td>0.41</td>
<td>bal.</td>
</tr>
<tr>
<td>0.15% C</td>
<td>0.15</td>
<td>0.21</td>
<td>0.20</td>
<td>0.001</td>
<td>0.001</td>
<td>13.2</td>
<td>0.001</td>
<td>bal.</td>
</tr>
<tr>
<td>0.32% C</td>
<td>0.32</td>
<td>0.20</td>
<td>0.20</td>
<td>0.001</td>
<td>0.001</td>
<td>10.3</td>
<td>0.001</td>
<td>bal.</td>
</tr>
<tr>
<td>SS316L</td>
<td>0.02</td>
<td>0.63</td>
<td>0.84</td>
<td>0.026</td>
<td>0.001</td>
<td>12.09</td>
<td>17.76</td>
<td>bal.</td>
</tr>
<tr>
<td>IF</td>
<td>0.001</td>
<td>0.002</td>
<td>0.14</td>
<td>0.001</td>
<td>0.001</td>
<td>–</td>
<td>–</td>
<td>bal.</td>
</tr>
</tbody>
</table>

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where $D$, $k$, and $T$ are the diffusion coefficient, the Boltzmann constant, and the temperature, respectively. $r$ and $\alpha$ are the distance and direction of the carbon atom from the dislocation core, respectively. $A$ is a material constant and takes a value of $3 \times 10^{-28}$ Nm$^{-2}$ for solute carbon atoms in iron. From the Orowan equation, $\dot{\varepsilon} = \rho_m b v$, where $\rho_m$ is the mobile dislocation density. $\dot{\varepsilon}_{\text{min}}$, can be derived as

$$\dot{\varepsilon}_{\text{min}} = \rho_m b A D \frac{1}{kT R} r_0^2,$$  

(5)

where $r_0$ is the radius of a dislocation core.

Since the dynamic strain aging effect is known to become more prominent at lower temperatures, the uniaxial tensile tests were performed at 208 K in the present study. At this temperature, the diffusion coefficient in the vicinity of a dislocation core has been reported to be approximately $10^{-19} - 10^{-18}$ m$^2$/s$^{-1}$. Finally, by taking $r_0 = 10^{-9}$ m, and assuming the mobile dislocation density during deformation to be in the order of $10^{14}$ m$^{-2}$, the critical strain rate is estimated to be around $10^{-5} - 10^{-2}$ s$^{-1}$. Accordingly, the uniaxial loading tests were performed at strain rates within this range in the present study. The detailed experimental procedure is as follows.

First, prestrains of 0%, 5%, and 10% were applied to the multilayered steel specimens at 298 K with a strain rate of $10^{-3}$ s$^{-1}$. Then, the specimens were unloaded and fully statically strain-aged at 298 K to have the same amount of solute carbon in the atmosphere around a dislocation core for all the specimens. The typical true stress-true strain curves of the multilayered steel with 0.32 mass% carbon martensite for the three different aging times are shown in Fig. 2. Since no clear differences exist in the stress-strain relationship after $9.0 \times 10^4$ s, full static aging is assumed to have been completed after $9.0 \times 10^4$ s in the present study.

For the prestrained and fully aged specimens, uniaxial tensile tests were conducted at 208 K and strain rates of $10^{-5} - 10^{-2}$ s$^{-1}$.

### 2.4. Modified Kocks-Mecking-Estrin Model

It is well known that the multiplication or annihilation of dislocations in mild steel during deformation is well described by the Kocks-Mecking-Estrin (KME) model:23)

$$\frac{d\rho}{de} = k_1 \sqrt{\rho} + k_2 \frac{1}{bd} - k_3 \rho,$$  

(6)

where $\rho$ and $d$ are the total dislocation density and average grain size, respectively, and $k_1$, $k_2$, and $k_3$ are material constants. The first and second terms on the right-hand side of the equation represent the multiplication of dislocations due to the interaction between dislocations and that between dislocations and grain boundaries, respectively. The third term represents the annihilation of the dislocations due to dynamic recovery. Since the total dislocation density in Eq. (6) is the sum of the mobile and sessile dislocation densities, the rate of the formation of the dislocation cell structure and the corresponding decrease in the mobile dislocation density in lath martensitic steels are not explicitly given. Therefore, the KME model is slightly modified as to separately evaluate the changes in the mobile and sessile dislocation densities during deformation. Note that the unstable rapid decrease in the mobile dislocation density,1) observed for a small strain, is not considered in the present model. The key ideas of the model are explained as follows:

In the KME model, there is no distinction between total and mobile dislocation densities, but in the reality the multiplication and annihilation of dislocations are induced by the glide of mobile dislocations. Accordingly, the rates of multiplication and annihilation are considered to be proportional to the frequency for the mobile dislocations to encounter with obstacles, such as grain boundaries.

From the Orowan equation, the average distance traveled by each mobile dislocation is estimated to be $l = \frac{de}{\rho_m b}$ for an infinitesimally small applied strain, $de$. Since the average spacing, $s$, between mobile dislocations is proportional to $\rho_m^{1/2}$, the rate of multiplication associated with the interaction between mobile dislocations is estimated to be $\rho_m \frac{l}{s} \approx \frac{\sqrt{\rho_m}}{b} de$. Likewise, the rate of multiplication induced by the interaction between grain boundaries and mobile dislocations is estimated to be $\rho_m \frac{l}{d} \approx \frac{1}{bd} de$, where $d$ is the mean free path of dislocation glide, which corresponds to the lath width in the case of lath martensite. On the other hand, the average diameter of the dislocation cell wall, $R$, is considered to be inversely proportional to the sessile dislocation density, $\rho_s$, as schematically shown in Fig. 3. Thus, the rate of annihilation at the dislocation cell walls is estimated to be $\rho_s \frac{l}{R} \approx \frac{\rho_s}{bd} de$. Therefore, the total change in the dislocation density is given as

$$\frac{d\rho}{de} = k_1 \sqrt{\rho} + k_2 \frac{1}{bd} - k_3 \rho,$$  

(7)

Since the rate of transition from mobile dislocations to sessile dislocations during deformation is considered to be proportional to both the mobility of dislocations and the interaction between mobile dislocations,26) the rate is described by the following equation:

$$\frac{d\rho}{de} = F(Xc)\rho_m,$$  

(8)

where $F(Xc)$ is a function proportional to the dislocation mobility determined from the experimentally obtained evolution of the total dislocation density, and $Xc$ is the mass fraction of solute carbon.
As described above, the material constants, \( k_1, k_2, \) and \( k_3, \) in the modified KME model are unlikely to be affected by the amount of solute carbon and are assumed to be constant in the present study. Note that the effects of solute carbon on both the rate of multiplication and annihilation of dislocations are taken into account through the third term in Eq. (7). These material constants and the function \( F(X_c) \) for a very low carbon content were estimated by fitting the modified KME model to the experimentally obtained changes in the total dislocation density of the ultra low-carbon steels with various grain sizes (NCF and IF steel). After that, the changes in the dislocation density of the lath martensite with various carbon contents were fitted to the model to derive \( F(X_c) \), using the same material constants obtained from the ultra low-carbon steels.

2.5. Formation of Dislocation Cell Structure

Transmission electron microscopy (TEM) was performed to qualitatively clarify the evolution of the dislocation structure in lath martensite. A martensite layer with a thickness of 80 \( \mu \)m was machined from 0–10% prestrained multilayered specimens by removing SS316L layers. Small disks (3 mm diameter) were cut out from the martensite layer, from which thin-film samples were fabricated by standard electrolytic double-jet polishing.

3. Results and Discussions

3.1. Initial Microstructure of Each Specimen

The size of the microstructure and the estimated initial dislocation density of each specimen are summarized in Table 2. It was confirmed that the block width is much finer than the thickness of the martensite layer. The TEM observation revealed that the lath width of the martensite used in the present study was approximately 300 nm and almost independent of the solute carbon content.

3.2. Work-hardening Behavior of Martensite with Different Solute Carbon Contents

The true stress-true strain curves of martensite estimated by the equistrain rule of mixtures\(^{14}\) and the change in the total dislocation density for various strain levels are shown in Fig. 4. As demonstrated in our previous study,\(^{14}\) the estimated curves correspond well with those of monolithic counterparts up to the fracture strain. It was clarified that an increase in the solute carbon content leads to considerable work-hardening and the multiplication of dislocations.

From the experimental results and those reported elsewhere,\(^{11,27}\) the solid-solution strengthening by the solute carbon and dislocation strengthening were separately estimated. Here, we assume that the total strengthening is the sum of each strengthening component, and the result is as follows:

\[
\sigma [\text{MPa}] = 0.07 + 1.3 \times 10^8 \sqrt{\rho} + 0.86 \sqrt{X_c}, \quad \ldots \ldots \ (9)
\]

where \( X_c \) is the mass percentage of solute carbon. The other strengthening components, such as Hall-Petch strengthening, are assumed constant considering the fact that the lath width is almost independent of the solute carbon content.

<table>
<thead>
<tr>
<th>Block width (( \mu )m)</th>
<th>0.001%C</th>
<th>0.06%C</th>
<th>0.15%C</th>
<th>0.32%C</th>
</tr>
</thead>
<tbody>
<tr>
<td>Initial dislocation density (10(^{15}) m(^{-2}))</td>
<td>3.8</td>
<td>5.2</td>
<td>6.8</td>
<td>7.0</td>
</tr>
</tbody>
</table>

Table 2. Block widths and initial dislocation densities of the martensitic steels.

![Fig. 3. Schematic illustration of dislocation cell structure.](image)

![Fig. 4. True stress-strain curves (a) and the change in dislocation density with applied strain (b) of martensitic steels.](image)
This result corresponds well with those obtained in previous studies. However, there is a small discrepancy in the coefficient for solute carbon strengthening. The reason for this is that in the previous studies the effect of dislocation strengthening was neglected, even though a higher carbon content induces a higher initial dislocation density in lath martensite. The relationship between true stress and dislocation density after subtracting the strengthening effect of solute carbon using Eq. (9) is shown in Fig. 5. The Bailey-Hirsch relationship is clearly maintained, and it is clear that the dislocation density has a predominant effect on the work-hardening behavior of lath martensite irrespective of its carbon content.

3.3. Change in Mobile Dislocation Density

3.3.1. Estimation from Dynamic Strain Aging Effect

Typical stress-strain curves of a multilayered steel specimen obtained at 208 K after applying 10% prestrain are shown in Fig. 6. There are regions where the stress gradually decreases after yielding at each strain rate. In these regions, part of the solute carbon atoms in the atmosphere formed around the dislocation core by static strain aging treatment start to gradually leave the dislocation core, and finally a state is achieved where only some of the carbon atoms can follow the moving dislocations. In the final state, the strain-hardening rate recovers to an almost constant as shown by the stress-strain curves in Fig. 6. Thereby, the flow stress at 208 K and a strain rate of $10^{-2}$ s$^{-1}$ was estimated at the point where the line representing linear elasticity and the extrapolated strain-stress curve intersect as indicated by the arrow in Fig. 6, and the total difference in the flow stress, $\Delta\sigma$, was estimated as shown in Fig. 6. Since $\Delta\sigma$ is the sum of the changes in flow stress in both the martensitic and austenitic layers, the temperature and strain rate dependences of the flow stress in the austenitic steel layers should be subtracted from $\Delta\sigma$ to obtain the value of $\Delta\sigma$ for the martensitic steel layer only. Furthermore, to make the dynamic strain aging effect clearer, the temperature and strain rate dependences of the dislocation motion in the ferrite were obtained using the IF steel. Finally, the amount of dynamic strain aging, $\Delta\sigma_{DSA}$, was estimated as

$$\Delta\sigma_{DSA} = \frac{1}{f} \left[ \Delta\sigma - (1 - f) \Delta\sigma' \right] - \Delta\sigma'' \quad \text{(10)}$$

where $\Delta\sigma'$ and $\Delta\sigma''$ are the temperature and the strain rate dependences in the monolithic austenitic steel and IF steel, respectively, and $f$ is the volume fraction of the martensitic steel layer. $\Delta\sigma'$ and $\Delta\sigma''$ were evaluated using the same method as the multilayered steel to obtain the total difference in the flow stress $\Delta\sigma$. The dynamic strain aging evaluated using Eq. (10) for all the strain rates are summarized in Fig. 7. The solid curved lines are guides to the eye, and the vertical lines represent the critical strain rate, $\dot{\varepsilon}_{\text{min}}$, estimated from the result obtained from the modified KME model.
model, which will be discussed later.

From Fig. 7, \( \varepsilon_{\text{min}} \) was estimated to be around \( 10^{-4} - 10^{-2} \) s\(^{-1} \) in each martensitic steel, which corresponds to mobile dislocation densities around \( 10^{13} - 10^{15} \) \( \text{m}^{-2} \) according to Eq. (5). From the change in \( \varepsilon_{\text{min}} \), it was found that the mobile dislocation density in martensite with solute carbon contents of both 0.06% and 0.15% monotonically decreases with increased strain. On the other hand, the mobile dislocation density for a carbon content of 0.32% appears to be almost constant up to a strain of 10%. This difference in the change in the mobile dislocation density suggests that the annihilation of mobile dislocations at dislocation cell walls is more pronounced at a strain level from 5% to 10% in martensite with a lower carbon content.

3.3.2. Modified KME Model

The parameters \( k_1, k_2, k_3 \) and \( F(X_c) \) were estimated by fitting all the experimental evolutions of the dislocation density of the NCF and IF steels to the modified KME model. The changes in the total dislocation density in these steels during deformation and the best-fit curves are shown in Fig. 8(a). Note that all the curves were fitted using the same parameters and the validity of the modified KME model is clearly demonstrated. Using the estimated parameters, it is also demonstrated that the total dislocation densities of the as-quenched and tempered martensites of 0.001%C steel can also be predicted accurately as shown in Fig. 8(b). From these results, it was concluded that the modified KME model can be used to estimate the change in the total dislocation density in ultra low-carbon steels by using fixed parameters regardless of their microstructure. The same material constants, \( k_1, k_2, \) and \( k_3 \), were also employed in the prediction of the total dislocation density in the low carbon martensites.

For the lath martensite with various solute carbon contents, the change in the total dislocation density was fitted by modifying \( F(X_c) \). The experimental values and predictions obtained using the modified KME model are shown in Fig. 9, and the changes in the mobile dislocation density estimated by the model are shown in Fig. 10. Figure 10 clearly indicates that the rate of transition from mobile dislocations to sessile dislocations increases with decreasing solute carbon content. Using the predicted mobile dislocation densities, the values of \( \varepsilon_{\text{min}} \) were estimated and are shown by vertical lines in Fig. 7. Even though the prediction by the modified KME model overestimates the critical strain a little bit presumably due to the application of the WH method, the results correspond well with the experimental findings. The parameters estimated from the experimentally obtained dislocation densities are summarized in Table 3.
3.4. Formation Behavior of Dislocation Cells
The effect of solute carbon on the formation behavior of dislocation cells at true strain of 0%, 5%, and 10% was investigated by TEM observation. Typical lath structures were observed in all samples before deformation, and dislocation cells started to form with increased strain. In particular, in the martensite with solute carbon contents of 0.06% and 0.15%, dislocation cells were clearly observed at a true strain of 5%, and they were widely distributed at a true strain of 10%, as shown in Fig. 11(a). Meanwhile, at a solute carbon content of 0.32%, dislocation cells appeared at a true strain of 10%, but there were some areas where the clear formation of dislocation cell was not observed as shown in Fig. 11(b). From these results, it is inferred that the formation of the dislocation cell structure starts at a relatively small strain in martensite with a low solute carbon content. These findings show good agreement with the change in the mobile dislocation density estimated from both the dynamic strain aging experiment and the modified KME model. In addition, this tendency is in good agreement with previous reports that the formation of dislocation cells was much more clearly observed in ferrite with lower carbon content after severe cold-rolling.29,30)

3.5. Dislocation Mobility and the Modified KME Model
The relationship between \( F(X_c) \) used in the predictions and the solute carbon content, \( X_c \), is presented in Fig. 12 as a double logarithmic plot. It was inferred from Fig. 12 that these values are proportional. The slope indicates that \( F(X_c) \) is inversely proportional to the solute carbon contents. Since it is considered that mobile dislocations have to glide against a friction force from solute carbon atoms, the dislocation mobility is considered proportional to average spacing between the solute carbon atoms in lath martensite. Thus, \( F(X_c) \), which is assumed to be proportional to the dislocation mobility as mentioned above, is inversely proportional to the solute carbon content, \( X_c \).

Table 3. Parameters and constants used in the modified KME model.

<table>
<thead>
<tr>
<th>Phase</th>
<th>( X_c )</th>
<th>( k_1 )</th>
<th>( k_2 )</th>
<th>( k_3 )</th>
<th>( F(X_c) )</th>
<th>( \rho_0^{++} ) (m(^{-2} ))</th>
<th>( d ) (( \mu )m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ferrite</td>
<td>0.001</td>
<td>0.102</td>
<td>1.51</td>
<td>5.4</td>
<td>0.00952</td>
<td>( 10^{13} )</td>
<td>0.1</td>
</tr>
<tr>
<td>F</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>( 10^{13} )</td>
<td>0.5</td>
</tr>
<tr>
<td>F</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>( 10^{12} )</td>
<td>50</td>
</tr>
<tr>
<td>Martensite</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>( 2.2 \times 10^{15} )</td>
<td>0.3*</td>
</tr>
<tr>
<td>Temper M</td>
<td>0.06</td>
<td></td>
<td></td>
<td></td>
<td>0.00305</td>
<td>( 6.8 \times 10^{14} )</td>
<td></td>
</tr>
<tr>
<td>M</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.00220</td>
<td>( 3.6 \times 10^{15} )</td>
<td></td>
</tr>
<tr>
<td>M</td>
<td>0.32</td>
<td></td>
<td></td>
<td></td>
<td>0.00130</td>
<td>( 4.2 \times 10^{15} )</td>
<td></td>
</tr>
</tbody>
</table>

* lath width (estimated by TEM observation).
| indicates that the value is same as above.

Fig. 11. Dislocation cell structures after 10% elongation in (a) 0.06%C and (b) 0.32%C steels.

Fig. 12. Relationship between \( X_c \) and \( F(X_c) \) used in the modified KME model.

4. Conclusion
(1) The work-hardening rate and the multiplication of dislocations during deformation are much more notable in
lath martensite with a higher solute carbon content.

(2) From our evaluation of the change in the mobile dislocation density by the analysis of dynamic strain aging behavior, the enhanced work-hardening rate can be explained by the decreased annihilation rate of the mobile dislocations.

(3) On the basis of the Kocks-Mecking-Estrin model and some theories about the formation of the dislocation cell structure, a modified KME model, which can separately predict the changes in the mobile and sessile dislocation densities in lath martensite during deformation with high accuracy, was proposed.

(4) Both the result obtained from the dynamic strain aging study and the accuracy of our model indicate that the solute carbon enhances the increase rate of the dislocation density, i.e. the work hardening, by retarding the formation rate of dislocation cells, which is confirmed also by the TEM observations.

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