Present Status and Future Prospects of Simulation Models for Predicting the Microstructure of Cold-rolled Steel Sheets

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Annealing treatment of cold-rolled steel sheets are mostly subjected to the continuous annealing line. In such process recovery, recrystallization, reverse transformation and dissolution or formation of precipitates occur during heating and homogenizing treatments, and the transformation and precipitation in the steel during the subsequent cooling and aging treatments. The required microstructure and mechanical properties are obtained by controlling these metallurgical phenomena appropriately. Because a simulation model that predicts the microstructure is a useful tool for the minute control of these phenomena, many models have been developed. In this review, the present status of these models is surveyed and their problems and future tasks are discussed.

KEY WORDS: bake-hardening recovery; cold-rolled steel sheet; continuous annealing; simulation model; recrystallization; phase transformation; reverse transformation; precipitation.

1. Introduction

Minute control of the microstructure of cold-rolled steel sheets is necessary to develop advanced high-strength and high-performance steel sheets, and many models for predicting the microstructure have been developed to support such a minute control of the microstructure. The need for such models has been increasing. A number of models for predicting the microstructural evolution through all steps in the process have been developed for hot rolling. On the other hand, models for predicting the microstructural evolution of cold-rolled steel sheets through all steps during the annealing process are underdeveloped. The microstructural models of cold-rolled steel sheets have been mostly developed for predicting individual metallurgical phenomena in the annealing process, such as recrystallization, transformation, reverse transformation, and precipitation. Figure 1 shows the metallurgical phenomena that occur under the individual processes in the continuous annealing line, and Fig. 2 shows a schematic drawing representing its typical thermal history. During the heating and homogenizing processes, recovery, recrystallization, dissolution, and formation of precipitates occur, and if the steel is heated higher than the Ac1 temperature, a reverse transformation occurs. In the cooling and aging processes, transformation and precipitation occur.

Because the prediction of the recrystallization temperature, ferrite grain size, and recrystallization texture is useful for the production of interstitial-free (IF) steel sheets, the corresponding prediction models have been developed. For the production of bake-hardenable (BH) steel sheets, the dissolution and precipitation of carbides have been modeled to predict the amount of carbon in solution after annealing, which has a close relationship with the BH value. To suppress the reappearance of yield-point elongation in lowcarbon steel sheets, the amount of solute carbon in solution is reduced by optimizing the over-aging treatments, and models that predict the cementite precipitation are useful for designing the optimum over-aging condition. In the cases of high-strength steel sheets, the microstructural evolution owing to transformation and reverse transformation has been modeled.

In this paper, the present status of models applied to predict the microstructural evolution of cold-rolled steel sheets in continuous annealing processes is reviewed, and the

<table>
<thead>
<tr>
<th>Phenomena</th>
<th>Heating</th>
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<th>Cooling</th>
<th>Aging</th>
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<td>Precipitation</td>
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Fig. 1. Metallurgical phenomena that occur sequentially in the continuous annealing process.
remaining problems to be investigated are discussed.

2. Prediction Models for Recovery and Recrystallization

Recovery models are usually described by the evolution of the dislocation density as a function of the existing dislocation density and temperature. One of the simplest models is given by Eq. (1). Using this type of a model, Guerenu et al. studied the recovery behavior of cold-rolled steel sheets during annealing processes in detail and determined the fitting parameters of the equation.

$$\frac{d\rho}{dt} = -a\rho^{n} \exp\left(\frac{-Q}{RT}\right)$$  \hspace{1cm} (1)

Regarding a recent development of the recovery model, the model developed by Zurob et al., which was based on the model of Verdier et al., has attracted our attention. Even though the Zurob model was developed to predict the recovery behavior under hot working, it is also applicable for the recovery behavior in the annealing of cold-rolled steel sheets. The model took into consideration the retardation of recovery owing to the pinning effect of precipitates on the movement of dislocations. Because the dislocations pinned by precipitates are hard to move and annihilate, the annihilation rate of dislocation density decreases. The annihilation rate of dislocation density is influenced by not only the existence of precipitates but also the interaction of solid solution elements with vacancies and dislocations. In addition, clusters or dipoles of alloying elements formed during annealing also retard the recovery. A model considering all of these retarding effects is expected to be developed in the future.

A newly proposed recrystallization model that will be introduced later in this review assumes that recrystallization occurs from the abnormal growth of selected subgrains. Therefore, the prediction of subgrain formation is indispensable for predicting the recrystallization behavior in the annealing process. It is difficult for the aforementioned recovery models to predict any morphological changes in the dislocation structure, such as the formation of subgrains. A relationship between the completion of subgrain formation and some threshold value of the decrease in the dislocation density calculated by the aforementioned models needs to be established. Furthermore, the development of a model for predicting both the decrease in dislocation density and the formation of subgrains simultaneously is the subject of future studies.

Next, the models for predicting recrystallization behavior during annealing are reviewed. Recrystallization behavior is often described using a Johnson–Mehl–Avrami–Kolmogorov (JMAK)-type equation, expressed by Eq. (2).

$$X = 1 - \exp\left(-bt^n\right)$$  \hspace{1cm} (2)

Here, $X$ is the fraction of recrystallization, and $t$ is the time. The parameter $b$ is the function of temperature, reduction in cold rolling and initial grain size, and $n$ is a constant that varies from 1 to 4 depending on the growth behavior of the recrystallizing grains. Equation (3) is obtained by differentiating Eq. (2). Using Eq. (3), the fraction of recrystallization can be calculated under an arbitrary thermal history.

$$\frac{dX}{dt} = nb^{1/n}\left[-\ln(1-X)\right]^{(n-1)/n} (1-X)$$  \hspace{1cm} (3)

Recent research on the prediction of the recrystallization behavior of cold-rolled extra-low-carbon steel sheets using a JMAK-type equation is introduced in the following. Figure 3 shows the recrystallization behavior of steel samples cold-rolled to 80% and subsequently heated with various heating rates. The parameters in Eq. (3) were determined by fitting the experimental data obtained by both 1°C/s and 10°C/s with the calculation data. The above equation fitted by the observed recrystallization behavior under lower heating rates is hard to predict accurately the recrystallization behavior at higher heating rates of 300°C/s and 1000°C/s. The results of calculation predicted a slower recrystallization rate than that observed by experiment. The influence of the driving force for recrystallization and the mobility of the boundary between the deformed matrix and the recrystallizing grains are discussed to clarify the reason for this discrepancy. The driving force for recrystallization was estimated from the hardness of the samples before the start of recrystallization. The difference in the driving force of the samples heated with 10°C/s from those heated with 1000°C/s was small, and could not explain the marked difference between the results of experiment and calculation. The calculations of the sample heated with 1000°C/s could be fitted to the experimental data if the apparent thermal activation energy, $Q$, was changed by 4%. A possible explanation of the decrease in the apparent thermal activation energy with a high heating rate is given by the solute drag effect caused by alloying elements and were difficult to segregate at the
grain boundary under higher heating rates because the time was too short to segregate during annealing. Even when using the simple recrystallization model described by the JMAK equation, the influence of various factors on the recrystallization behavior can be discussed and considered.

Next, a sophisticated recrystallization model that was developed recently is introduced. Considering various experimental findings, Humphreys established a theory that recrystallization occurred through the abnormal growth of selected subgrains and proposed a model based on this theory. The growth rate of an abnormally growing subgrain with a radius $R$ is described by Eq. (4). Here, $A$ is a constant, $M$ is the mobility of the boundary between the abnormally growing subgrain and the surrounding normally growing subgrains, and $G$ is the driving force for recrystallization.

The mobility and interface energies are changed with the high-energy boundaries. If precipitates exist, then the driving force is reduced by the pinning force of the precipitates, as formulated in Eq. (4). The term $\gamma$ is the energy of the interface between the abnormally growing subgrains and surrounding normally growing subgrains, $\gamma'$ is the energy of the interface between two normally growing subgrains, $F_\gamma$ is the fraction precipitated, and $d$ is the mean diameter of the precipitates. The growth rate of the normally growing subgrains with radius $\bar{R}$ is described by Eq. (5).

$$\frac{d\bar{R}}{dt} = AMG = AM\left(\frac{\gamma}{\bar{R}} - \frac{\gamma'}{R} - \frac{3F_\gamma}{d}\right)$$ \hspace{1cm} (4)

$$\frac{d\bar{R}}{dt} = A\bar{M}\gamma'\left(\frac{1}{4R} - \frac{3F_\gamma}{d}\right)$$ \hspace{1cm} (5)

The mobility and interface energies are changed with the misorientation angle $\theta$ between the neighboring subgrains and are expressed by Eqs. (6) and (7), respectively. An interfacial energy of 0.5 J/m$^2$ was used for $\theta \geq 15^\circ$, which is often used for an incoherent boundary. The term $M_m$ is the mobility of high-angle boundaries.

$$\gamma = 0.5\theta \left(\frac{1}{15} - \ln \frac{\theta}{15}\right)$$ \hspace{1cm} (6)

$$\bar{M} = M_m \left[1 - \exp\left(-5\frac{\theta}{15}\right)^4\right]$$ \hspace{1cm} (7)

Equation (8) describes the changing rate of the ratio of the size of an abnormally growing grain and the average size of a normally growing grain. Humphreys considered the condition for the start of recrystallization in detail by substituting Eqs. (4) and (5) into Eq. (8).

$$\frac{d}{dt}\left(\frac{R}{\bar{R}}\right) = \frac{1}{\bar{R}}\left(\frac{dR}{dt} - \frac{d\bar{R}}{dt}\right) > 0$$ \hspace{1cm} (8)

Suwa et al. combined the Phase-Field method with the model of Humphreys and examined the changes in the morphology of the microstructure during the progress of recrystallization. Suwa also reviewed recrystallization models for predicting microstructural evolution during recrystallization by considering the local alignment of the subgrain structure using Monte Carlo, Vertex, and Phase-Field simulations.

Nishitani et al. extended the model of Humphreys by considering a microscale textural analysis to predict the recrystallization behavior of cold-rolled steel sheets. Hashimoto et al. measured the orientations of the recovered microstructure of cold-rolled extra-low-carbon steel sheet using scanning electron microscopy–electron backscatter patterns (SEM–EBSP) and classified the deformed grains into several types based on their orientation and degree of scattering. Nishitani et al. applied the modified model of Humphreys to three representative types of deformed subgrain structures, and developed a model for predicting the recrystallization behavior of cold-rolled extra-low-carbon steel sheets, succeeding in a reasonable prediction. Because this model could estimate each effect of pinning and solute dragging on the recrystallization, Nishitani et al. studied the influence of the addition of 0.05% Ti in extra-low-carbon steel sheets on the retardation of recrystallization. Their analyses using the model developed indicated that the retardation of recrystallization owing to the addition of Ti can mainly be attributed to the retardation of the subgrain formation under the recovery process; both the pinning effect by the precipitates, and the solute dragging effect by Ti in solution were relatively small on the progress of recrystallization.
rotation angle between the orientations of the recrystallized and deformed grains; and the other axis was normal to the maximum active slip plane.

Hayakawa et al. proposed a sophisticated model for predicting recrystallization textures using Monte Carlo simulations. The deformed microstructure was represented using 400 deformed grains. Each of the deformed grains was divided into smaller elements that were regarded as subgrains. The orientations and stored energy of the deformed grains, which were determined using X-ray and neutron diffraction analysis, were assigned to these subgrains. Subgrains close to the grain boundaries were assumed to rotate to fulfill the continuity of the orientation of the neighboring grains. The degree of crystal rotation between two grains with different stored strain energies was assumed to be inversely proportional to their stored energy. Nucleation was assumed to occur at the grain boundaries, where a large orientation gradient and high level of stored energy existed. The nucleus then grew with a rate expressed by the product of the mobility and the driving force. The mobility was assigned as a function of the misorientation angle between the nucleus and deformed grain. This model was applied to predict the recrystallization texture of IF steel, and good agreement between the experimental and calculated results was obtained. However, some of the recrystallization behaviors predicted by the simulations did not coincide with experimental results. For example, the calculation results suggested that more than 90% of the nuclei were nucleated at grain boundaries, while observations revealed that a large number of nuclei were formed inside the {111} deformed grain boundaries. Caballero et al. proposed the following phenomenological equation for the nucleation rate of austenite:

\[
N = f_0 \exp \left( \frac{-Q_N}{kT} \right) \tag{9}
\]

where \(f_0\) is a function that depends on the initial microstructure and heating rate, \(Q_N\) is the activation energy of nucleation, and \(\Delta T\) denotes the degree of overheating. The function \(f_0\) can be a very complicated function because Huang et al. revealed that the heating rate dependence of \(f_0\) was significantly affected by the initial microstructure. Huang et al. found that the interface is between not only pearlite and ferrite, but also ferrite grain boundaries as nucleation sites of austenite if the sample was heated with high heating rate. They also reported an unusual phenomenon whereby a sample heated with high heating rate showed a higher fraction of transformed austenite than a sample heated with lower heating rate if the holding time at the reaustenitization temperature was short. The nucleation phenomenon observed under high heating rates becomes complex if the formation of austenite and the recrystallization of ferrite occur simultaneously.

In almost all models, the growth of austenite is assumed to be controlled by the diffusion of carbon in austenite, which is supplied from the dissolution of cementite. As shown in Fig. 4, the cementite–austenite interface and the austenite–ferrite interface move simultaneously during reaustenitization. The growth rates of these two interfaces were calculated by solving the diffusion equations for carbon fulfilling the flux balance of carbon. The influence of alloying elements is considered to determine the carbon concentration at the interfaces assuming various equilibrium conditions, such as paraequilibrium (PE) and negligible partition local equilibrium (NPLE).

Taking into account the effect of the pressure caused by the curvature of the interface, Jacot et al. calculated the growth rate of austenite under NPLE conditions. They observed the evolution of the microstructure of steel with a ferrite + pearlite microstructure and classified the austenite transformation behavior into three stages: (1) very rapid growth of austenite into pearlite leading to complete dissolution of pearlite, (2) slower growth of austenite into ferrite that was primarily controlled by carbon diffusion in the austenite, and (3) very slow growth of austenite controlled by the diffusion of substitutional alloying elements in austenite. The rate of the pearlite-to-austenite transformation in Stage 1 increased with the decrease in the lamellar spacing of the pearlite. Speich et al. also used cold-rolled samples, and revealed that the formation of austenite was accelerated by cold rolling. They inferred that the cause of this acceleration was caused by an increase in the number of ferrite grain boundaries owing to the recrystallization and deformation of pearlite, both of which increase the number of nucleation sites for austenite.

Reaustenitization proceeds with the nucleation and growth of austenite. Austenite primarily nucleates at the interface between cementite and ferrite or pearlite and ferrite.
applied a finite element method (FEM) to calculate and predict the two-dimensional distribution of the concentration of carbon atoms in austenite and the morphological changes of pearlite occurring during its dissolution. Besides this, some JMAX-type models have also been developed to predict the reaustenitization behavior for simple handling.\textsuperscript{34,35}

Some experiments with high heating rates have shown that remnants of cementsites were often observed after the completion of reaustenitization, even though the cementsites dissolved rapidly. It has also been observed in a ferrite–cementite microstructure that some cementites had dissolved into the ferrite matrix before acting as a nucleation site for austenite. A model taking these observations into consideration is needed to be developed.

4. Model for Predicting Microstructures of Advanced High-strength Steel Sheets

Dual phase and transformation-induced plasticity (TRIP)-type steel sheets are representative advanced high-strength steel sheets. Both steel sheets are annealed in an intercritical temperature range. Many studies have revealed that alloying elements are partially partitioned during the intercritical annealing.\textsuperscript{28,39,40} In the four-year research project from 2006 to 2009 called “Mathematical Models for Predicting Microstructures and Mechanical Properties of Steels” organized by the ISIJ, the partitioning behavior of Mn and Si during annealing at intercritical temperature and its influence on the transformation behavior was studied, and some of the results of this study are introduced here.\textsuperscript{55}

The experiments were carried out using samples from six steels with various C, Mn, and Si contents. Table 1 shows the measured and calculated results of the constituents of a microstructure with 0.046% C, 0.02% Si, and 1.54% Mn, as an example. The samples were heated with 10°C/s to the temperatures at which the fractions transformed to austenite were 100%, 75%, 50%, and 25%. The samples were kept for 300 s at the selected temperature and cooled with various cooling rates. Only the volume fraction of ferrite is given in the experimental results, because the other constituents in the microstructure were hard to distinguish. The calculations were carried out using a transformation model developed by Suchiro et al.\textsuperscript{50} The outline of the model is given in Table 2. The content of carbon in austenite used in the calculations was the equilibrium value at the heating temperature, and the Mn and Si contents were measured in a sample cooled with 140°C/s using an electron probe microanalyzer (EPMA). The calculated results fairly agreed with the experimental results, which means the transformation model for predicting the transformation behavior from full austenite can also predict the transformation behavior of samples that were heated to the intercritical temperatures with a reasonable accuracy, if the amount of C, Mn, and Si partitioned are applied. The partitioning behavior of some alloying elements, such as Mn and Si, during the annealing at the intercritical temperature has been investigated in detail. In the early stage of reaustenitization, the partitioning of any substitutional alloying element does not occur, and reaustenitization proceeds under NPLE conditions. In the later stages of reaustenitization, the diffusion of alloying elements through the \( \gamma \rightarrow \alpha \) interface occurs, and the growth rate of the austenite is controlled by the diffusion of the substitutional alloying elements.\textsuperscript{42} Although alloying elements often present a sharp concentration in the austenite at the vicinity of the \( \gamma \rightarrow \alpha \) boundaries after soaking at the intercritical temperature, the \( \gamma \rightarrow \alpha \) transformation was calculated using an average concentration of the alloying element partitioned, and the results shown in Table 1 were obtained using this assumption. The agreement between the results of calculation and experiment is good, but to improve the calculation accuracy, the concentration gradient at the vicinity of the austenite–ferrite interface should be taken into consideration in the future.

In the case of an initial austenite fraction of 75%, the model of Suchiro et al. assumed that the nucleation of ferrite did not occur. This means that a site saturation was assumed, and the existing ferrite grew into austenite. However, the experimental results revealed a grain refinement of the ferrite, which indicated that some ferrite grains had nucleated during cooling. Holding the sample at the intercritical temperature promotes the segregation of alloying elements at the ferrite–austenite boundary, which suggests that the solute drag effect on the transformation needs to be taken into account. Some improvement in the model by considering these phenomena should be carried out in the future. Because of space limitations, other models for predicting the microstructure of dual phase steel sheets are indicated only in the References.\textsuperscript{57–59}

Regarding the cold-rolled TRIP steel sheets, a number of models for predicting the microstructure have been developed, and most of these assume that the initial microstruc-

<table>
<thead>
<tr>
<th>Heating temp. [°C]</th>
<th>Cooling rate [°C/s]</th>
<th>Mn in ( \gamma ) [wt%]</th>
<th>Si in ( \gamma ) [wt%]</th>
<th>Fraction of ( \alpha ) [%]</th>
<th>XF [%]</th>
<th>XP [%]</th>
<th>XB [%]</th>
<th>XM [%]</th>
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The microstructure, namely the fraction of austenite and ferrite, is given by equilibrium calculations at the intercritical temperature. The bainite transformation plays an important role in controlling the mechanical properties of TRIP steels. Minote et al. have developed a model for predicting the microstructure of intercritically annealed low-carbon TRIP steels using two bainite transformation models based on a diffusional and a displacive transformation mechanism. Their model simulations revealed that the bainite transformation kinetics above 350°C followed the model based on the diffusional transformation mechanism, while the bainite transformation kinetics below 350°C followed the model based on the displacive transformation mechanism.

An accurate prediction of the fraction of retained austenite and its carbon content is required to predict the mechanical properties of TRIP steel sheets. For an accurate prediction of these quantities, an improvement of the bainite transformation model was carried out by Azuma et al. Referring to the studies of Bahdeshia et al., Azuma et al. considered four phenomena simultaneously, as shown in Fig. 5, namely, a diffusionless transformation from austenite to bainite, carbon diffusion from oversaturated bainite to the remaining austenite, and cementite precipitates in both the bainite and the retained austenite. From this treatment, typical characteristics of the bainite transformation, for example, the temporary cessation of the bainite transformation and the

### Table 2. Transformation model used for calculations whose results are given in Table 1.

<table>
<thead>
<tr>
<th>Transformation</th>
<th>Basic equation of transformation rate</th>
<th>Factor corresponding to nucleation rate and growth rate</th>
<th>Coefficient</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ferrite</td>
<td>( I = T^{-1/2}D \times \exp \left( \frac{-k_1}{RT} \right) )</td>
<td>( k_1 = 17476 ) ( k_2 = 8.933 \times 10^{-12} \exp \left( \frac{21100}{T} \right) )</td>
<td>( G^* = \frac{1}{2\pi} \frac{D}{C_\gamma C_\alpha} )</td>
</tr>
<tr>
<td></td>
<td>site saturation</td>
<td>( S = 6/\gamma )</td>
<td>( k_3 = 0.957 \times 10^9 (J/mol^3) )</td>
</tr>
<tr>
<td>Pearlite</td>
<td>( I = T^{-1/2}D \times \exp \left( \frac{-k_1}{RT} \right) )</td>
<td>( k_4 = 2.27 \times 10^9 (J/mol^3) )</td>
<td>( S = 6/\gamma )</td>
</tr>
<tr>
<td></td>
<td>site saturation</td>
<td>( G^* = \frac{1}{2\pi} \frac{D}{C_\gamma C_\alpha} )</td>
<td>( k_2 = 2.01 \times 10^3 )</td>
</tr>
<tr>
<td>Bainite</td>
<td>( G^* = \frac{1}{2\pi} \frac{D}{C_\gamma C_\alpha} )</td>
<td>( k_2 = 6.816 \times 10^{-4} \exp \left( \frac{3431.5}{T} \right) )</td>
<td></td>
</tr>
</tbody>
</table>

(Note): \( \gamma \): austenite grain size (cm), \( D \): diffusion coefficient of carbon in \( \gamma \) (cm²/s) \( C_\gamma \): carbon content in \( \gamma \) \( C_\alpha \): carbon content in \( \alpha \) \( C_{\gamma \alpha} \): carbon content in \( \gamma \) at \( \gamma / \alpha \) boundary \( C_{\gamma \cem} \): carbon content in \( \gamma \) at \( \gamma / \cem \) boundary \( \Delta T \): undercooling below \( A_\epsilon \) \( G^* \): Zener-Hillert equation (the value was calculated with the method by Kaufman et al.) \( r \): radius of curvature of advancing phase \( \gamma \): austenite, \( \alpha \): ferrite, \( \cem \): cementite

![Fig. 5. Schematic depiction of bainitic transformation.](image)
5. Model for Predicting the Precipitation of Cementite in Over-aging Processes

The amount of carbon in solution of almost all mild steel sheets is reduced below a certain threshold level through a precipitation treatment in the over-aging process to suppress not only the deterioration of formability by strain aging after temper rolling, but also the occurrence of the stretcher strain during press forming that fatally deteriorates the surface quality of automotive outer panels. Koyama et al. developed a model for predicting the evolution of carbon in solution in continuous annealing processes that was used to design a compact continuous annealing process. Their model used the following assumptions.

1. The microstructure consists of ferrite with a known grain size.
2. The initial content of carbon in solution is the equilibrium content at the homogenizing temperature.
3. Grain boundaries are unlimited sink for carbon.
4. Ferrite grains are spherical shape.

Under these assumptions, the change in carbon concentration in ferrite was calculated by solving diffusion equations in spherical coordinates. As a boundary condition of the diffusion equation describing the carbon diffusion to the ferrite grain boundaries, the carbon concentration at the grain boundaries was assumed to be zero, because the boundaries were considered to be an unlimited sink for carbon. The amount of carbon in solution decreased from the flow into the ferrite grain boundaries and from cementite precipitation inside the ferrite grains. The cementite precipitation model was formulated based on classical nucleation and growth theory. Although observations strongly indicate that MnS is a preferential nucleation site of cementite, Koyama et al. did not develop a model for predicting the size and number of MnS precipitates; instead, they set an arbitrary value to a fitting parameter for the number of nucleation sites for cementite.

Residual carbon in solution is the difference between the initial content of carbon in solution and the sum of the carbon that has flowed into the ferrite grain boundaries and the carbon precipitated as cementite inside the ferrite grains. Ushioda et al. discussed efficient ways to reduce the carbon in solution from various overaging patterns, and proposed a compact continuous annealing line based on their results. Figure 6 shows a schematic diagram of the thermal history of a compact continuous annealing line. Through rapid cooling to a low temperature, a high nucleation rate of cementite can be realized from the high supersaturation of carbon in solution. The temperature is subsequently raised to promote the nucleation and growth of cementites. Finally, the temperature is decreased gradually to reduce the equilibrium amount of carbon in solution. In this way, low-carbon steel sheets with low solute carbon can be produced in a remarkably short continuous annealing line.

6. Model for Predicting the Carbon in Solution in Extra-low-carbon BH Steel Sheets

BH steel sheets have good formability in press forming and a high yield strength after the bake-hardening treatment. They are used to produce automotive outer panels. As a good press formability is required for the steel sheets of outer panels, most BH steels are produced from Ti and/or Nb bearing extra-low-carbon steels. There are two ways to produce extra-low-carbon BH steel sheets. The first one is the steels that contained more carbon than the stoichiometric balance with Ti and/or Nb are used, and a relatively low annealing temperature is employed. The second one is the steels that contained a lower amount of carbon than the stoichiometric balance with Ti and/or Nb are used, and a relatively high annealing temperature is employed. The former steels always have carbon in solution, independent of the annealing temperature. The existence of carbon in solution during cold rolling adversely influences the formation of a recrystallization texture, which is good for deep drawability. In the latter steels, carbon in solution is only obtained if the annealing temperature is higher than the solubility temperature of TiC and/or NbC. Their deep drawability is superior to the former case because the amount of carbon in solution is very low during the cold rolling caused by the high-temperature hot coiling. In the latter steels, the degree of BH is strongly influenced by the annealing history. A model for predicting the amount of carbon in solution has been developed to estimate the degree of BH. The procedure for the calculation of this model is as follows.

1. The average size and number of precipitates, such as...
1. TiC and/or NbC, before annealing are measured and used as the initial calculating conditions.

2. The dissolution of the precipitates into ferrite during annealing is calculated by solving the diffusion equation under local equilibrium conditions during heating and homogenizing.

3. Assuming that any nucleation of new precipitates does not occur, and only existing precipitates grow during cooling, the change in carbon in solution is calculated using the diffusion equation used in Item 2.

4. Parallel to the calculation of Item 3, the precipitation of cementite is calculated using the model developed by Koyama et al. [68]

**Figure 7** shows the influence of the annealing temperature on samples heated with 10°C/s, maintained for a period of 30 s at the annealing temperature, and then cooled with 90°C/s for the amount of BH steel formed. Good agreement was obtained between the experimental and calculated results.

7. Nitriding Model for a Continuous Annealing Process

Recent research [72] indicates that the formation of a surface layer hardened by nitriding provides highly dent-resistant steel sheets. **Figure 8** shows the hardness distribution in such a surface layer of samples with various Ti contents nitrided at 750°C for 40 min in an NH3 atmosphere. The hardness and thickness of the nitried surface layer are influenced by the amount of nitride-forming alloying elements and the nitriding conditions. To elucidate the optimum nitriding conditions, a model for predicting the increase in hardness of the surface layer after nitriding in a continuous annealing process has been developed. In this model, a diffusion equation for nitrogen is solved using the value of nitrogen at the surface as a boundary condition, and the distribution of nitrogen in the surface layer is determined. Simultaneously, the precipitation of TiN is calculated in the surface layer based on the nucleation and growth mechanism using the variation in nitrogen concentration with the distance from the surface.

**Figure 9** shows the calculation results using this model, which indicate that the nitrogen diffused from the surface rapidly forms TiN, and the further diffusion of nitrogen is suppressed. The calculated distribution of Ti coincides with the abrupt change in the hardness shown in Fig. 8.

8. Concluding Remarks

Many models for predicting the evolution of the micro-structure of cold-rolled steel sheets in continuous annealing processes have been developed. The target for the future is to develop a model for predicting the evolution of micro-structure through the entire process, from hot rolling to cold rolling to continuous annealing. It is important to predict the distribution and size of cementites or pearlites in hot bands, as these become the nucleation sites during reaustenitization. It is also necessary to predict quantitatively the influ-
ence of cold rolling conditions on the recrystallization and reaustenitization behaviors. Another important subject to be modeled is the partitioning behavior of alloying elements during intercritical annealing.

As mentioned, there are many challenging tasks to develop the models for predicting the evolution of microstructure and the mechanical properties of cold-rolled steel sheets. The author hopes that this review will give the reader an overview of the present status and future prospects of models for predicting the microstructure and mechanical properties of cold-rolled steel sheets.

REFERENCES