Model for Predicting the Microstructural Evolution of Extralow Carbon Steels

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The microstructural evolution of extralow carbon steels in hot working processes has been studied. Considering the experimental results of the microstructural evolution of austenite and the transformation behavior, a metallurgical model for predicting the microstructural evolution of extralow carbon steels in hot working processes was developed. The model was applied to hot rolling processes to evaluate the quantitative effect of various operation conditions on the grain refinement of ferrite and the calculated results were found to correlate relatively well to experimental ones.

KEY WORDS: extralow carbon steel; metallurgical model; grain refinement; microstructure; transformation; recrystallization.

1. Introduction

Since the refinement of ferrite grains improves both the strength and toughness simultaneously, grain refinement has attracted attention in the development of modern high strength steels and technology achieving the refinement of ferrite grains less than 1 \( \mu \text{m} \) in the industry has been developed.\(^1\)

On the other hand, the aim of grain refinement of hot rolled mild steels is not to improve the strength and toughness but rather to get superior deep drawability of cold rolled sheets. This latter quality is known to improve if the ferrite grain size before cold rolling, namely the grain size of hot bands, is refined. The reason and mechanism were discussed and some hypotheses were given.\(^2\)–\(^4\)

To automotive steel sheets for panels requiring high press formability, extralow carbon steels are applied because the decreased carbon content in the steels improves formability, although it is also apt to coarsen the ferrite grain in hot bands. With this in mind, several processing measures were considered for refining the ferrite grain size and their quantitative effect was discussed.\(^5\) The measures are ① the increase in reduction, especially in the reduction at final deformation, ② the increase in the cooling rate, ③ the reduction of the interval between the final deformation and the onset of rapid cooling and ④ the proper selection of the finishing temperature. To determine the optimum operating condition through combining these grain refining measures, a model for predicting the microstructure is useful.

For low carbon steels, a model for predicting the microstructure has been already developed and published.\(^6\) It is, however, difficult to apply the developed model to extralow carbon steels because certain metallurgical phenomena of extralow carbon steels differ quite significantly from those of low carbon steels. For example, in low carbon steels, the transformation from \( \gamma \) to \( \alpha \) is controlled by carbon diffusion in austenite while the transformation of extralow carbon steels is mainly controlled by the interface mobility.

In this paper, we will introduce a newly developed model for predicting the microstructural evolution of extralow carbon steels with these differences in mind.

2. Experimental Conditions

2.1. Material

As is well known, the grain boundary of austenite of low carbon steels can be visualized by a sophisticated etching technique if the steels are quenched and the martensitic microstructure is formed. In a case of conventional extralow carbon steels, the visualization of the austenitic microstructure using the same technique is difficult because the martensitic microstructure cannot be obtained by quenching. In this investigation, 2.5% Mn is added to extralow carbon steels to obtain the martensitic microstructure by quenching. Table 1 shows the chemical compositions of steels used in the experiment. Steel A is a conventional extralow carbon steel, Steel B is an extralow carbon steel with 2.5% Mn, Steels C and D are low carbon steels with different amounts of C and Steel E is used to clarify the influence

| Steels | C   | Mn  | Si  | P   | S   | Al  | N   |
|--------|-----|-----|-----|-----|-----|-----|-----|
| A      | 0.0013 | 0.18 | 0.01 | 0.005 | 0.006 | 0.006 | 0.0021 |
| B      | 0.0020 | 2.45 | 0.01 | 0.016 | 0.004 | 0.020 | 0.0017 |
| C      | 0.05 | 2.51 | 0.01 | 0.016 | 0.004 | 0.025 | 0.0017 |
| D      | 0.1  | 2.51 | 0.01 | 0.015 | 0.004 | 0.023 | 0.0029 |
| E      | 0.1  | 1.51 | 0.01 | 0.015 | 0.003 | 0.023 | 0.0024 |
of the Mn content on the evolution of the austenitic microstructure. Steel A is used both in the rolling experiment, to verify the model developed, and in the cooling experiment, to study the transformation behavior, while the other steels are used in hot working experiments to study the evolution of the austenitic microstructure. They were melted in a vacuum furnace and cast slabs 110 mm in thickness were reheated at 1200°C for 1 h and hot rolled into a 15 mm-thick plate at a finishing temperature of around 950°C. Specimens of 7 mm in diameter and 12 mm in height were machined from the plates.

2.2. Hot Working Experiment

The hot working experiment was carried out on the hot deformation simulator. Figure 1 shows the procedure of the experiment. The deformation temperature, strain and strain rate were varied from 900 to 1000°C, from 0.25 to 2.3 and from 1 to 50 s⁻¹, respectively.

2.3. Cooling Experiment

The cooling experiment was also carried out on the hot deformation simulator. The dilatation of the specimen during cooling was measured by laser to estimate the fraction transformed. Specimens were heated at 1000°C for 10 min and subsequently cooled at constant rates of 3°C/s, 10°C/s and 30°C/s respectively.

3. Experimental Results

Figure 2 is an Avrami plot diagram showing the change of the fraction recrystallized over time following deformation at various holding temperatures. The inclination of the Avrami plots is around 2, which coincides with that in the case of low carbon steels.7,8) Figure 3 shows a comparison of the experimental results of the time for 50% recrystallization with that calculated via the model developed for low carbon steels (strain: 0.45, strain rate: 10 s⁻¹, deformation temperature: 900–1000°C).

Figure 4 shows the transformation behavior of Steel A, heated at 1000°C for 10 min and cooled at various cooling rates. The austenite grain size of 160 μm was determined with a polished specimen by thermal etching. The dotted curves indicate the fraction transformed calculated by the model described in the following section.
4. Model

4.1. Concept of the Integrated Model

Figure 5 shows a conceptual scheme of the integrated model for predicting the evolution of microstructures. Because this model is thought to apply to hot rolling processes, the three relevant processes involved are considered to be: the slab reheating process, the hot rolling process and the cooling process. During the first process, austenite grains grow, while in the second, the austenitic microstructure is refined due to recrystallization and in the cooling process, the $\gamma \rightarrow \alpha$ transformation occurs. The modeling of these metallurgical phenomena will be described in the following section.

4.2. Model for Predicting Grain Size in Slab Reheating

It has been confirmed that the austenite microstructure is hardly influenced by the initial grain size in the slab reheating process just prior to transformation, due to the heavy total reduction in the hot rolling process. Therefore, a simplified model is applied for predicting the grain size after slab reheating, which is used for low carbon steels. More sophisticated models based on the grain growth theory of Hillert have been developed. If the total reduction is relatively small, it is recommended to use these models.

4.3. Model for Predicting the Evolution of Austenitic Microstructure in the Hot Rolling Process

The austenitic microstructure in the hot rolling process changes due to the occurrence of dynamic and static recrystallization and grain growth. The model describes these metallurgical phenomena and predicts the resultant microstructure.

The previous investigation revealed that carbon content of between 0.05% and 0.8% hardly affects the evolution of the austenite microstructure and this study confirms that C content of less than 0.05% has similarly negligible influence on the evolution of austenitic microstructure, which means that the model developed for low carbon steel can also be applied to extra-low carbon steels. The equations used in the model are given in Table 2.

4.4. Model for Predicting Transformation Behavior

The transformation mechanism of extra-low carbon steels differs from that of low carbon steels. The latter is controlled by carbon diffusion in austenite while the former is controlled at a very early stage by carbon diffusion and then interface mobility. This means the transformation kinetics of extra-low carbon steels are controlled by interface mobility if the partition of carbon is no longer thermodynamically required.

In this model, the kinetics of transformation controlled by carbon diffusion in austenite are described using the Triveli equation to calculate the velocity $v$ of the $\gamma \rightarrow \alpha$ boundary. Triveli succeeded in developing a more accurate model based on strict mathematical treatment of the Zener–Hillert equation. The equations used are given in the following:

$$v_c = \frac{27D_T}{256\pi r_c} \left( \frac{\Omega_0}{1 - 2\Omega_0/\pi - \Omega_0^2/\pi^2} \right)^{\frac{1}{3}}$$

$$\Omega_0 = \frac{C_{\gamma} - C_{\alpha}}{C_{\gamma} - C_{\alpha}}$$

$$r_c = \frac{\sigma V}{RT(C_{\alpha} - C_{\gamma})}$$

Here, $D_T$ is diffusion coefficient of C in austenite, $C_{\gamma}$, and $C_{\alpha}$ are the equilibrium carbon content in austenite at the $\gamma \rightarrow \alpha$ boundary and is assumed to be 0.2 J/m$^2$. $V_{\alpha}$ is the mol volume of ferrite.

The transition from the carbon diffusion controlled process to the interface mobility controlled process occurs
if the condition $C_a=C_\gamma$ is fulfilled. Using Eq. (1), $v_c$ for $C_a=C_\gamma$ becomes a very large value and can exceed the interface velocity $v_k$ calculated assuming the interface mobility controlled process. In the practical calculation, $v_c=v_k$ is used if $v_c>v_k$.

For the kinetics of transformation controlled by interface mobility, the velocity of the $\gamma/\alpha$ boundary is expressed by Eq. (4).

$$v_c=MA\Delta G_v$$

(4)

Here, $M$ is the mobility expressed by a form $M=M_0\exp(-Q_0/T)$ and $\Delta G_v$ is the difference between the Gibbs free energy of austenite and that of ferrite of the same chemical composition. As $M$, the equation proposed by Hillert$^{18}$ can be applied with a value of $M_0$ corrected from 0.035 to 0.00525 to fit the experimental results and the following equation is given:

$$M=0.00525 \exp(-17700/T)$$

(5)

The transformation behavior of extralow carbon steels was observed on a microscopy equipped with a hot stage and it was recognized that the nucleation occurred only at a very early stage. It is, therefore, reasonably assumed that the transformation proceeds under site saturation conditions, while the progress of transformation can be then expressed by Eq. (6).

$$\frac{dx}{dt} = k_2(1-x)$$

(6)

$k_2$ is determined by fitting with experimental results and given:

$$k_2=1.6748 \cdot 10^8 \exp(21100/T)$$

(7)

### 4.5. Model for Predicting Ferrite Grain Size

In a previous study,$^{17}$ the ferrite grain size of low carbon steels was successfully predicted by a model considering the transformation temperature $T_{0.05}$ (temperature at which 5% transformation occurs) and effective austenite grain size. Comparison of the calculation results with experimental equivalents confirmed the fact that the model is applicable to extralow carbon steels with a minor change in the coefficient value. The model is expressed by Eqs. (8) and (9):

$$d_\alpha = 4 \cdot 10^9 d_{\gamma \text{eff}}^{1.75} \exp\left(-\frac{21430}{T_{0.05}}\right)^{1/3}$$

(8)

$$d_{\gamma \text{eff}} = d_{\gamma} / (1+10^{-11}P^{1.54})$$

(9)

### 5. Discussion

The coefficients of the equations used in the model are fitted through comparing with experimental results on a hot deformation simulator. To verify the applicability of the model to hot rolling processes, the published data of hot rolling experiments carried out on a laboratory 2 stand mill whose schematic depiction is given in Fig. 6 are compared with the data calculated by the developed model. The detailed rolling condition and the experimental results were given in Ref. 5.

**Figure 7** shows the influence of the onset time for cooling on the ferrite grain size. The conditions are a final reduction of 25%, a finishing temperature of 930°C and a cooling rate of 70°C/s. The circles and square indicate the experimental and calculated results, respectively. The calculated results correlated relatively well with experimental ones. An analysis of the calculation result indicates that the grain refinement through the shorter onset time for cooling is achieved by the presence of higher dislocation density which decreases with time due to recovery and recrystallization.

**Figure 8** shows the influence of the final reduction on the ferrite grain size. In this case, the onset time for cooling, the finishing temperature and the cooling rate are 0.6 s, 930°C and 70°C/s, respectively, and relatively good correlation between the calculated and experimental results is obtained. An analysis of the calculation result indicates that a saturated tendency of the grain refinement by increasing the...
strain of the final reduction comes from the less increase in dislocation density introduced by deformation in the higher strain region.

Figure 9 shows the influence of the cooling rate on the ferrite grain size. The conditions are a final reduction of 25%, a finishing temperature of 930°C and an onset time for cooling of 1.6 s. Although some fluctuation of experimental data is recognized, the overall dependence of cooling rate on the ferrite grain size is relatively small. The calculated results correlated well with experimental figures. The calculation result indicates that the dislocation density at 1.6 s after the final rolling is very low because the recrystallization is completed, and the dislocation density just before onset of transformation correspondently low. Equation (8) shows not only that the ferrite grain size is influenced by the retained dislocation density but also that the 5% transformation temperature affects it. To discuss the slight influence of the cooling rate on the ferrite grain size, the 5% transformation temperature was determined by a dilatation measurement during transformation. Figure 10 is an experimental result showing a little decrease in the 5% transformation temperature from 883 to 858°C by increasing the cooling rate from 30 to 120°C. A marked decrease in the transformation temperature, it means a significant ferrite grain refinement of extralow carbon steels by increasing the cooling rate cannot be achieved since their transformation proceeds extremely rapidly.

The comparisons of the calculated and experimental results in Figs. 7 to 9 indicate that the model developed is applicable to hot rolling processes to predict the ferrite grain size of extralow carbon steel sheets. The prediction model is also proved as useful means to discuss the effect of rolling and cooling conditions on the grain refinement quantitatively.

6. Conclusion

The recrystallization and transformation behavior of extralow carbon steels has been studied quantitatively during laboratorial hot working experiments. Considering the experimental results, a metallurgical model was used to predict the microstructural evolution of extralow carbon steel sheets in hot rolling processes was developed and the following results were obtained:

(1) The influence of the amount of carbon in the steel on the recrystallization behavior of austenite is relatively small so that the model previously developed for low carbon steels can be applied to predict the microstructural evolution of extralow carbon steels.

(2) The transformation kinetics was modeled as a process controlled by carbon diffusion at first and subsequently by interface mobility.

(3) If nucleation occurs, the created phase boundary moves very rapidly and the transformation is completed, mainly due to the growth of the grain. With this fact in mind, site saturation is assumed in the transformation model.

(4) The model developed was applied to predict the ferrite grain size of hot bands produced by a laboratory mill. The calculated result and experimental observation correlate with each other with tolerable accuracy.

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