Experimental and Numerical Prediction of Austenite Grain Size Distribution in Round-oval Shape Rolling

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To determine austenite grain size (AGS) distribution during round-oval shape rolling of mild carbon steel (0.2 wt% C), three-dimensional non-isothermal finite element analysis in couple with an AGS evolution model available in the literature was carried out in the present investigation. A hot rolling test was conducted at the laboratory to verify the numerical data obtained from the analysis. For more accurate prediction of the temperature history during the process, interface heat transfer coefficient between the billet and rolls was determined by comparing the numerical temperature history with the measured data at three locations in the billet. Metallographic investigation of the rolled specimen after quenching showed that recrystallization behavior could be classified into two regions: interior zone with refined grain dominated by metadynamic recrystallization, and exterior zone with less refined grain due to static recrystallization. It was found that the numerically predicted AGS distribution was in good agreement with the experimentally measured one. In addition, the conventional additivity rule generally used for the prediction of recrystallization behavior for non-isothermal rolling conditions was modified in order to bear a better comparison with the experimentally measured AGS values.

KEY WORDS: austenite grain size; shape rolling; non-isothermal finite element method; interface heat transfer coefficient; additivity rule.

1. Introduction

In hot metal forming processes, geometric accuracy and better mechanical property are two major objectives to achieve the quality of final products. In rolling the first requirement has been gradually met due to development of versatile control technology in couple with the enormous processing experiences from shop floor. To satisfy the second requirement, however, basic understanding of the metallurgical evolution occurring inside the deforming workpiece under the complex thermal history during the process must be established.

Therefore, many studies had been carried out in metallurgical modeling after the pioneering work by Sellars and Whitman. Most of the applications of mathematical models have been concentrated on hot plate or strip rolling for the last two or three decades, but recently the interest in steel industry is about to change toward shape rolling, especially in bar rolling. Thus, the finite element (FE) analysis in couple with the microstructure model was used to predict the local microstructure evolution in bar rolling. Recently Yanagimoto summarized a wide range of studies on the FE based analyses for the prediction of microstructure distribution in metal forming.

In an effort to make better prediction of microstructural change, Karhausen and Kopp proposed an incremental model for dynamic recrystallization which can predict the microstructural change and flow stress at any given instant during the forming process. Based on this incremental approach, Yanagimoto et al. proposed a new incremental formulation in which the dislocation density was used as a representative variable based on the microstructure evolution model suggested by Yada and Senuma. Later, this incremental modeling approach was expanded to the whole bar rolling process including phase transformation and the characteristics of microstructure evolution under different rolling conditions were analyzed.

Recently, Kwon et al. applied the isothermal FE analysis in combination with austenite grain size (AGS) evolution model proposed by Hodgson and Gibbs to the four pass oval-round rolling process. They investigated the effect of roll gap change and rolling speed on the AGS distribution, but the accuracy of the predicted local grain size was not examined through rolling experiments. Jimenez et al., however, showed the accuracy of the FE based modeling of microstructural change and verified their results using the bar stretching test, whose deformation mode was assumed to be similar to bar rolling. In this comparative work, a quite good agreement was obtained in spite of isothermal assumption in the FE analysis. This could be attributable to the relatively high speed ($v \approx 100$) so that local die chilling may not be significant at this forming speed. In result, the effect of temperature history on microstructure evolution might be small.

For low working speed level, however, the temperature history can produce a considerable effect on the microstruc-
tural evolution. In such a case, the selection of interface heat transfer coefficient (HTC) between the workpiece and rolls plays a crucial role in determining the thermal response during rolling. Thus, accurate prediction of HTC is of importance. Im and Burte investigated the heat transfer behavior in hot upsetting and estimated HTC values. Recently, an integrated model for hot strip rolling was proposed by Zou and in this work a delicate thermal modeling was conducted for the AGS prediction in seven stands rolling schedule. Most of these investigations were on strip or slab rolling.

In this study the local austenite grain size distribution for low speed bar rolling process of plain carbon steel (0.2 wt% C) was investigated where the thermal variation due to the contact between the workpiece and roll grooves was considered. A single oval pass rolling experiment was conducted using a laboratory mill and local AGS distribution measured after quenching the specimen. To analyze the recrystallization behavior quantitatively during the rolling and subsequent cooling, the three-dimensional non-isothermal FE analysis in couple with the AGS evolution model proposed by Hodgson and Gibbs has been made.

For an accurate analysis of thermal response during the rolling and cooling, heat transfer coefficients were numerically determined based on the temperature history obtained from experiments. A modified additivity rule was also introduced in applying the AGS evolution model for a better AGS prediction. Finally, the predicted AGS distribution was compared well with the experimentally measured one.

2. Experimental

A series of tests was conducted on a single stand two-high laboratory mill at the POSCO Technical Research Laboratory, which was driven by a 75 kW constant torque DC motor. Ductile casting iron rolls of the maximum diameter of 310 mm and the roll barrel length of 320 mm were used and the rolling speed were set to be 34 rpm. The rolling experiments were conducted for an oval pass only. Plain carbon steel was used as material of the specimen and its chemical compositions are summarized in Table 1.

2.1. Experimental Procedure

Figure 1 illustrates three main experimental devices and their arrangements: (a) reheating furnace (R/F), (b) 2-high pilot rolling mill, and (c) water bath for quenching after rolling. Specimen was soaked at about 50°C above the desired rolling temperature for 20 min to ensure a uniform temperature distribution all over the specimen. Then, the specimen was transported into the rolling mill and rolled with the oval groove as shown in the upper right zone of Fig. 1. The rolled hot specimen was quenched right after the given delay time on the supporting steel plate at the rear end of the rolling mill to freeze the austenite grain boundary. Quenching was carried out by immersing the specimen into the water bath and stirring with a gripper.

2.2. Temperature Measurement

For accurate measurement of the temperature, it is necessary to obtain the full local thermal histories of the specimen from the reheating stage to the quenching stage. Since the high stiffness of the thick thermocouple makes it difficult to stir the rolled specimen during the quenching stage, it was necessary to separate the thermocouples from the specimen before quenching.

Thus, the temperature measurements were carried out in two groups (group I and group II) as summarized in Table 2. The group I (I-1, I-2, I-3, and I-4) focused on investigating the overall temperature distribution at the specimen and determining the heat transfer coefficients which will be used in the FE analysis. Among four experiments in the group I, three data sets (I-1, I-3, and I-4) were obtained without significant noise. In the three data, two (I-1 and I-4) showed the abnormal thermal responses and the third experiment (I-3) was adopted in this study. These failures were attributed to the complexity such as initial misalignment of the specimen and abrupt breakage of thermocouples during rolling. In the group II, all the three data were used in the AGS measurement.

Experiments in the group II (II-1, II-2, and II-3) were carried out for the measurement of the final AGS distribution of the specimen which was rolled at 1,000°C after the

| Table 1. Chemical composition of S20C steel used for experiments. |
|-----------------------------|-----------------------------|-----------------------------|
| C  | Mn  | Si  | P (max) | S (max) |
| Wt. % |      |     |         |         |
| 0.20 | 0.15 | 0.25 | 0.030   | 0.035   |

Fig. 1. Experimental set-up and its arrangement with the groove profile (unit: μm).
specified delay time. In the second group thermal history was just monitored in the data acquisition system and when the given temperature (1000°C) was reached the single thermocouple at the center of the workpiece was separated from the specimen. Then, the specimen was pushed into the roll gap and quenched.

In the group I, three thermocouples were embedded in 100 mm deep holes drilled from the tail of the specimen with the diameter of 28 mm and the total length of 250 mm as shown in Fig. 2, to investigate thermal history of the specimen during rolling and interpass time. In the group II, a single thermocouple was used at the center of the specimen. During the whole experiments K-type thermocouples with the diameter of 4.6 mm were used because the sensing device should be robust in measuring the thermal history under the current heavy deformation. For smooth biting of the specimen between the roll gap, the round specimen was tapered axi-symmetrically at the front with the end diameter of 14 mm. For each test, the data acquisition system (YOKOKAWA® mobile view recorder-MV100) was used to capture the thermal data through channels (CH) of 1, 2, and 3 with minimum sampling time of 0.125 s. The data acquisition system shows the on-line thermal history graphically on the screen and also the temperature-time history can be saved in a Zip® disk.

2.3. Measured Thermal Response of the Experiment (I-3)

The successfully measured temperature data for three points of the specimen (CH 1, 2, and 3) in the experiment (I-3) was shown in Fig. 3. These data was used in determination of the heat transfer coefficients. The thermal response could be classified into three intervals: Stage I (before rolling), Stage II (rolling and stabilizing), and Stage III (steady cooling).

Stage I: Even when the specimen was soaked in the furnace, there was some temperature deviations of 7.8°C between position A (1090.7°C) and position C (1082.9°C). It was partly attributed to the fact that the door of reheating furnace was slightly opened during heating due to the thermocouples embedded in the specimens. After the specimen was taken out from the furnace, considerable temperature decrease was recorded at the three positions during transfer from the furnace to the rolling set-up. When the hot specimen was just about to be rolled, the temperature difference reached 21.7°C between the position A and C (1063.3°C at position A and 1041.6°C at position C).

Stage II: During rolling, there was a sharp temperature hike at the center position due to heat generation by plastic deformation. The temperature increased up to 1078.0°C (ΔT=14.7°C) and then experienced steep decrease due to heat loss toward the cold roll contact zone. Contrary to the thermal response at position A, position C suffered sharp temperature decrease due to the heat loss to the roll contact zone, followed by temperature recovery from the energy redistribution in the specimen. At point B, thermal response followed the intermediate history between those of point A and point C.

Stage III: The temperature redistribution in the specimen finished almost in 6 s after rolling (around 35 s) and then the specimen showed the steady cooling rate of 3.28°C/s, resulting in the final temperature of 935.0°C. The cooling rate was determined by calculating the temperature gradient between two points at 35 and 60 s.
2.4. AGS Measurements

To investigate the microstructure in austenite state, quenched rolled specimen was polished and etched with 2% nital solution to capture the microstructure in the cross section normal to the rolling direction. The experimental grain size was determined using the planimetric method, as described in ASTM standard E112.\(^{12}\) Figure 4 shows a typical microstructure distribution at three points of the deformed specimen: at the center, below the roll contact zone, and at the right side of the cross section. At the center, smaller grain was measured (31.5 μm), while at the two boundary points, less refined grains were measured, respectively (39.5 μm at position (1) and 43.4 μm at position (3)).

Near the boundary zone, martensite structures were visible with the original grain boundary frozen from the austenite state. At the center region blurred structure was found at the edges between the grains. The blurred microstructure might be bainite structure because the cooling rate at the center region was intermediate between the high cooling rate usually linked to martensite formation and low cooling rate required for pearlite formation.

3. Numerical

3.1. Integrated AGS Prediction Modeling

For numerical prediction of the AGS evolution during rolling and subsequent cooling, three-dimensional finite element method was used in couple with the microstructural evolution model proposed by Hodgson and Gibbs\(^7\) as mentioned before. The current FE analysis was carried out using CAMProll\(^{13,14}\) which was developed based on the rigid-thermo-viscoplastic approach and coupled transient temperature analysis. Since the detailed information is available in reference, the finite element formulation is omitted in this paper.

Figure 5 illustrates numerical procedure for the AGS prediction in non-isothermal rolling conditions adopted in this study. The AGS at each pass was predicted by using the deformation and thermal history obtained from the FE analysis at each point with the initial austenite grain size. The effective strain and time-averaged mean effective strain rate computed from the deformation analysis were given to input values for the AGS prediction model. Then the numerically determined temperature right after rolling and subsequent temperature history for the interpass time were also used in the AGS evolution model by applying the conventional additivity rule. Based on these inputs the grain size at each position and its residual strain was calculated. The results obtained from the integrated AGS prediction were used for the AGS prediction of the next pass.

Note that the AGS prediction equations proposed by Hodgson and Gibbs\(^7\) were derived under isothermal conditions. Therefore, to apply the isothermal kinetic data to non-isothermal conditions, conventional additivity rule was applied in this study and its concept was illustrated in Fig. 6. For the kinetic equations of recrystallization and grain growth, the thermal history was discretized into a series of isothermal steps. The equations were then applied to each of these segments assuming the additivity for recrystallization, while for grain growth the grain size at the end of each step was then used as the initial grain size in the following time step.

3.2. Deformation Analysis

The accuracy of the current FE analysis for the isothermal case had been verified using the laboratory scale rolling tests for two pass round-oval-round rolling sequence.\(^8\) In this study to check the accuracy of the non-isothermal FE analysis in predicting the workpiece geometry, the measured workpiece geometry was compared with the FE analysis result and it was illustrated in Fig. 7. A quarter FE model was used in the current deformation analysis considering the computational efficiency. It is clear that the deformed geometry based on the FE analysis is in good agreement with the experimentally measured surface profile.

Table 3 shows the typical geometry of the deformed workpiece and related FE modeling method for flow stress, friction, and mesh information used in this study.
3.3. Heat Transfer Analysis

The finite element thermal analysis was not carried out for the roll according to the result of the previous study.\textsuperscript{14} The roll temperature was assumed to be 60\degree C considering the moderate temperature of the roll surface during rolling experiment. In taking into account the heat generation from plastic work, it was assumed that 90\% of the plastic work was transformed to heat generation. In the modeling of heat generation from the frictional work, it was assumed that half of the frictional work was transferred into the workpiece using the following heat flux term:

\begin{equation}
q = (0.5) (f \cdot |v_r|) \text{.}
\end{equation}

Here, \( f \) represents the frictional force and \( v_r \) indicates the relative velocity between the workpiece and rolls.

Table 3. Predicted geometry and FE analysis conditions.

| Flow stress modeling | Friction modeling | Mesh information | Predicted workpiece geometry |
|----------------------|------------------|-----------------|----------------------------|
| C = 0.2 \( \nu \) \( \alpha \) | \( m_r = 0.6 \) | 920 | 46 in section, 20 in the rolling direction |
| Shida's model \textsuperscript{15} | Constant shear friction model | |

(1) Assumption for the Initial Temperature Distribution in the Experiment (1-3)

In the heat transfer analysis, the initial temperature distribution of the specimen must be known. In the current study, however, only three thermocouples were used and it is necessary to assume the initial temperature distribution, especially in the outer region of the round specimen (between 9.0 mm and 14.0 mm in radius). This is because the temperatures were measured at the inner region of the specimen (A: 0.0 mm, B: 4.5 mm, and C: 9.0 mm). Therefore it was assumed that the temperature distribution has a linear relation with the radius of the workpiece and the slope of the temperature was determined with the temperatures at the two positions A and C. The linear assumption is somewhat reasonable because the steel usually has such high conductivity that the specimen can reach the steady cooling state in a few seconds. This trend was observed in the measured thermal response in the group I in Fig. 3. The specimen with very steep internal temperature gradients showed a quick temperature recovery and the temperature reached the steady cooling state in 6 s according to the thermal responses at the stages II and III in Fig. 3.

(2) Determination of the Heat Transfer Coefficients

For an accurate prediction of temperature distribution of the workpiece, the optimal mesh density for a cross section of the workpiece to characterize the thermal history was determined at first. Three mesh layouts (for each case the boundary layer of elements was divided into two layers in the thickness direction) were tested as shown in Fig. 8 and the second mesh layout with 46 elements was found to be acceptable from the perspectives of the computational effi-
In this calculation, the rolling speed was set to be zero because the workpiece was not moving on the steel plate during interpass time. Then, radiation term was determined by finding the proper emissivity which produced the feasible solutions, as represented by the dashed line in this figure.

(b) Interface HTC during Rolling

To obtain the interface HTC value between the workpiece and rolls, it is necessary to compare the numerically predicted thermal history with experimental results at the roll gap position. As mentioned in the previous section, however, the minimum sampling time of the current data acquisition system was 0.125 s. It was almost the same as the total deformation time of the workpiece. To solve this problem, thermal responses just after rolling were predicted with three different interface HTC values (3000, 30000, and 300 000 W/m$^2$K) in the FE analyses and compared with the measured temperature data in Fig. 11. From the comparison, it was found in this figure that the thermal history using the interface HTC of 30 000 W/m$^2$K agreed well with experimental measurements.

For more accurate determination of the interface HTC, however, an objective function was defined as Eq. (3).

$$f = \sum_{i=1}^{N} \left\{ \left( T_{Ai} - T_{Ai} \right)^2 + \left( T_{Ci} - T_{Ci} \right)^2 \right\} \quad \text{(3)}$$
where $\tilde{T}_A$ and $\tilde{T}_C$ are the experimentally measured values at the two different points $A$ and $C$, respectively. $T_A$ and $T_C$ are numerically obtained values at the same two points.

The final interface HTC value was determined by minimizing the objective function. According to this approach, the interface HTC value was finally determined to be 24,000 W/m²K as shown in Fig. 12. If this value was used for the temperature calculation, the only difference was 4.3°C. Thus, the result was not plotted in Fig. 11.

Heat transfer coefficient determined in this way was later used for the FE analysis for the rolling conditions in the group II to predict the AGS. Note the center temperature of the specimen was 1063°C just before rolling (at 29 s in Fig. 3) when interface HTC during rolling was determined, while the AGS measuring experiment (group II) was conducted when the center temperature was 1000°C. Although there is some discrepancy in the initial temperature, the heat transfer coefficients currently determined were not much affected according to the initial temperature difference (63°C). Each HTC and thermal properties used in the current FE analysis are summarized in Table 4.

![Fig. 12. Determination of the interface HTC during rolling by minimizing the error between the measured and predicted temperatures at two points of the specimen.](image1)

**Table 4.** Thermal properties and heat transfer coefficients used.

|           | value     | unit  |
|-----------|-----------|-------|
| Interface HTC | 24000     | W/m²K |
| Convection HTC | 2.33      | W/m²K |
| Emissivity | 0.505     |       |
| Stefan-Boltzmann const. | 5.6705x10⁻⁸ | W m⁻² K⁻⁴ |
| Thermal conductivity | 29        | W/mK  |
| Specific heat | 0.651     | kJ/kgK |
| Density     | 7833      | kg/m³ |
| Roll temperature | 60        | °C     |

II-2, and II-3), as previously mentioned, only the center temperature was measured as 1000°C and the delay time (interpass time, IPT) was set as 2, 5, 20 s before quenching, respectively. This was intended to investigate how the AGS evolution occurs depending on time.

In the AGS prediction, the approach described earlier was followed. As previously mentioned, only the center temperature was measured as 1000°C just before rolling in the group II experiments. Thus, the initial temperature distribution must be assumed and in this study the temperature distribution was assumed to have the same linear temperature gradient as the gradient assumed in the experiment (I-3), while the center temperature was fixed at 1000°C.

The measured grain size distribution (expressed with solid symbols) can be mainly divided into two groups: the inner part of the specimen (positions 2, 3, 4, 5, 9, 10) with the grain size of around 30–35 µm and the outer region (positions 1, 6, 7, 8) with coarser grain between 40–50 µm. Overall tendency of the AGS distribution predicted by the FE based approach was in good agreement with the measured distributions. At the center position (no. 3), measured values and predicted values at each delay time were almost the same, while at positions 2 and 9 some discrepancies were found.

4.2. Numerical Results: Strain, Strain Rate, and Temperature Distribution

For better understanding of the above AGS analyses, local distributions of three major variables influencing the recrystallization and grain growth kinetics were illustrated in Fig. 14 under current rolling conditions, the equivalent strain had its maximum value at the center (0.60 and minimum value was predicted at the right side (0.30, almost half the maximum value) and strain value exceeded 0.40 at most of the section. Equivalent strain rate (more exactly, time-averaged mean effective strain rate) followed similar contours and these values were predicted as almost 10 times the strain values. Figure 14(c) shows the temperature distribution at the exit of roll gap. The temperature at the roll contact zone was 710°C but in most of other regions it was around 1000°C with its maximum value of 1018°C at the center (temperature increase of 18°C due to heat generation by plastic work). Near the free surface area (at the right...
temperature decrease due to convection and radiation heat loss was observed. Figure 14(d) shows the uniform temperature distribution 5 s after rolling, demonstrating temperature redistribution in the specimen, which instantly removes the steep temperature gradient at the upper surface of the specimen.

4.3. Additivity Rule Revised: Transition from SRXN to MDRXN

As previously mentioned, it was shown that the roll-workpiece interface experienced steep temperature decrease and then quick temperature recovery in 5 s. Thus, application of the additivity rule in discrete isothermal intervals might lead to sudden change of initial static recrystallization into metadynamic recrystallization, resulting in more grain refinement numerically. This phenomenon can be explained as follows.

At the roll contact zone the amount of deformation was relatively large ($\varepsilon \geq 0.45$) but due to the relatively low temperature ($710^\circ C$) the Zener–Hollomon parameter ($Z = \dot{\varepsilon} \exp(Q/RT)$) increased, resulting in the increase of critical strain ($\varepsilon_c = 5.6 \times 10^{-4} d^{0.3} Z^{0.17}$) and then static recrystallization ($\varepsilon < \varepsilon_c$) occurred. As the temperature increased, however, the additivity rule could predict that the recrystallization mechanism was replaced by the metadynamic behavior ($\varepsilon > \varepsilon_c$), which led into far more grain refinement. But the metadynamic recrystallization could not develop without dynamic recrystallization occurring during deformation, because metadynamic recrystallization developed only by continued growth of the nuclei formed by dynamic recrystallization during straining. Thus, the transition from SRXN to MDRXN could not occur during the interpass time in spite of the steep temperature increase.

In this context, it was constrained in applying the additivity rule that the statically recrystallized zone could not develop the metadynamic behavior and the static phenomenon continued during the subsequent interpass time. Before this restriction, there was a noticeable error between the measured and predicted AGS at positions 7 and 8. The symbols with cross lines in Fig. 13 indicate the predicted AGS when the metadynamic recrystallization was likely even after the static recrystallization. In this case, too much grain refinement was predicted at these two positions. However, much improved prediction was obtained after introducing the constraint on the transition of recrystallization behaviors in numerical calculation.

4.4. Predicted AGS Evolution along Interpass Time

To examine the recrystallization kinetics, AGS evolution...
for the 10 measurement points between 0 and 20 s after rolling is shown in Fig. 15. Inner positions (2, 3, 4, 5, 9, 10) were governed by the metadynamic recrystallization, resulting in relatively fine grain structure. Contrary to this, the boundary region (positions 1, 6, 7, 8) had relatively large grain size due to static recrystallization: (a) positions 1, 7, 8 showed full recrystallization and subsequent grain growth and (b) position 6 showed partial recrystallization.

MDRXN at the inner part was partly attributed to the large amount of deformation and full MDRXN (FMD) finished in 0.2–0.3 s. Among the boundary zone showing slow grain refinement due to SRXN, at positions 1, 8, and 7, recrystallization completed at 7.1, 5.9, and 5.3 s, respectively. Then, grain growth followed, resulting in the final grain size of 45.7, 45.2, and 45.2 μm for each position. But the position 6 predicted the final grain size as 37.9 μm due to partial recrystallization, which was attributed to the small amount of deformation (strain value of 0.31).

5. Conclusions

In this study, a single round-oval pass rolling experiment was conducted for mild carbon (0.2 wt% C) steel in the laboratory rolling mill. The local AGS distribution and temperature history across the cross section were measured and investigated. For deeper understanding of the recrystallization behavior under the given rolling condition, nonisothermal three dimensional FE analysis combined with the AGS evolution model available in the literature was used to predict the local grain size evolution during rolling and subsequent cooling. These experimental and numerical studies lead to the following conclusions:

(1) Grain refinement was dominant at the center of the specimen due to metadynamic recrystallization and grain refinement was relatively low at the boundary region because of static recrystallization. With initial microstructure of about 110 μm, AGS at the center was predicted to be around 30–35 μm, while at the outer zone grain size of 40–50 μm was measured. The grain size was reduced by 60–70% in a single pass compared to the initial grain size.

(2) Interface heat transfer coefficient used in the nonisothermal FE analysis was determined to be 24000 W/m²K. The convection heat transfer coefficients and radiation emissivity were also determined based on the thermal responses obtained from the rolling experiments. By using such heat transfer coefficients, the numerically predicted thermal response was in good agreement with the measured one during rolling and interpass time.

(3) In applying the additivity rule for the highly transient thermal condition during rolling and subsequent cooling, the transition from static recrystallization to metadynamic recrystallization must be constrained to prevent overestimation of grain refinement in numerical modeling.

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