Microstructure evolution of roll core during the preparation of composite roll by electroslag remelting cladding technology

Abstract: In the present study, the comprehensive analyses based on the numerical simulation, in situ observation, and metallographic detection were carried out for the roll core of GCr15/45 carbon steel composite roll manufactured by the new electroslag remelting cladding (ESRC) method. During the ESRC process, the temperature distributions at the different radial and longitudinal positions of the roll core have great changes due to the different degrees of heat conduction from the slag bath, as a result, various microstructure and properties were obtained at the different positions of the roll core. The results illustrated that the ESRC process tended to be stable as the composite height reached a certain value and the high-temperature austenitization process mainly occurred in the radial regions where $R > 60$ mm, whereas no significant changes occurred at the areas where $R \leq 60$ mm. The coarse grains and few Widmanstatten structures with proeutectoid ferrite were generated in the roll core areas near the bimetallic interface, while the fine grains with more proeutectoid ferrite were obtained in the roll core area away from the interface. Therefore, a higher tensile strength and a better plasticity were obtained for the specimens away from the interface.

Keywords: composite roll, electroslag remelting cladding, in situ observation, microstructure evolution, austenitization

1 Introduction

A composite roll is composed of two different materials, which mainly includes three parts of the roll core, composite layer, and bimetallic bonding interface. According to the operating condition of the roll, high strength and toughness of the roll core are needed to bear the heavy loading and impact forces that occur during operation [1]. The commonly used materials of roll core are gray cast iron, ductile cast iron, cast steel, forged steel, and so on. However, with an increase in rolling size, rolling speed, and rolling reduction, the higher performance of the roll core is required to ensure a safe and smooth rolling process. Under this circumstance, the composite roll with cast steel and forged steel as core materials has been widely used due to its higher strength and toughness [2–4].

The preparation of bimetallic composite roll is always a liquid–solid bonding process. Except for the centrifugal casting method [5], whether it is the continuous pouring process for cladding method [6], electroslag surfacing with liquid metal method [7], or the electroslag remelting cladding (ESRC) method [8], the liquid metal of outer layer is bonded to the solid roll core through the direct contact and interaction with each other. During this process, the roll core is always heated to a very high temperature to promote the element diffusion and form metallurgical bonding interface.
between the bimetals. However, due to the different degrees of heat conduction from the molten liquid metal or slag bath, various temperature distributions at different radial and longitudinal positions of the roll core are acquired, which have a significant influence on the phase transition and microstructure evolution of the roll core at different positions.

The previous studies on the composite roll were mainly focused on the solidification segregation, microstructure, properties of outer layer [9–11] and element diffusion, interface microstructure, and reaction of bimetallic bonding interface [12–14]. There is a lack of systematic studies on the contrastive analyses of roll core before and after the composite process or the microstructure evolution at the different positions of roll core. The ESRC method developed by Jiang et al. [15] is an innovative method for manufacturing a composite roll, which can effectively improve the solidification segregation, crystal structure of the outer layer, and realize the control of interfacial bonding temperature based on the electroslag remelting (ESR) and current supplying mold [16] technology. Cao et al. [17] have carried out a study on the changes of graphite morphology, distribution, size, and matrix structure of the roll core (ductile cast iron) before and after the ESRC process. It illustrates that the graphite phase and matrix structure also change obviously due to the high temperature austenitization process that occurred in the roll core during the ESRC process. However, the differences at the different radial and longitudinal positions of the roll core have not investigated in detail, and the microstructure evolution in the roll core of cast or forged steel is less well understood.

In the present study, a comprehensive analysis based on the numerical simulation, in situ observation, and metallographic detection was carried out for the roll core of the GCr15/45 carbon steel composite roll manufactured by the ESRC method. Its microstructure and tensile strength were analyzed.

2 Materials and methods

A detailed description is given for the manufacture of the composite roll by the ESRC method [14] in Figure 1; the hot-rolled 45 carbon steel of 0.45C–0.21Si–0.68Mn–0.07Cr (wt%) was used as the roll core and 20 rod-like GCr15 electrodes with a uniform distribution are used. For a particular material, the heating temperature and cooling rate are the main factors affecting its microstructure evolution. As the actual temperature distribution is always difficult to measure, a 2D electromagnetic-flow-thermal coupling simulation for the ESRC process has been done by Fluent software due to its axial symmetry [15]. Similar to the ESR process [18], a quasi-steady-state is assumed when the composite layer reaches a certain height. According to the steady withdrawal speed, the temperature varying with coordinate in the longitudinal direction of the roll core can be transformed into temperature varying with time convincingly. Based on the relationship between temperature and time, an in situ observation of the austenite grain growth and acicular ferrite precipitation behaviors with the temperature variations for the roll core specimens was carried out by the high-temperature laser scanning confocal microscope to explain the changes in roll core during the ESRC process. For the in situ analysis, the cylindrical specimens with 5 mm in diameter and 5 mm in length were taken from the near-surface region of roll core in the radial direction. The specimen was mounted into alumina crucibles, and a thermocouple with a precision of 0.1°C was positioned under the crucible to measure the temperature in the furnace. Before the experiment, the furnace chamber was pumped to a vacuum state of $6 \times 10^{-3}$ Pa and the protective gas (argon) was injected to prevent the surface oxidation of the specimen. Then, the specimen was heated according to the stated function of temperature varying with time. The allowable maximum heating rate was 5°C/s.

In addition, a series of $10 \times 10 \times 8$ mm specimens were taken from the different positions of roll core before and after the ESRC process to analyze the microstructure evolution. A comparative analysis was made for the grain size and proeutectoid ferrite content (analyzed by using Image-Pro Plus 6.0 software through lots of fields for each specimen) of the different specimens. The tensile properties of specimens taken from different positions of roll core were tested by a universal tester with a tensile speed of 0.5 mm min$^{-1}$. The morphology of
the fractured surface was inspected by a scanning electron microscope.

3 Results and discussion

3.1 Temperature distribution at different radial regions of roll core

During the ESRC process, the Joule heat generated in the slag bath is the heat source of the whole composite system. The roll core was heated rapidly by liquid slag bath from initial room temperature to high temperature, and then, it participated in the cladding process with the molten metal of the outer layer. As the model parameters and simulation settings have been described in detail in the previous report [15], the simulation result of temperature distribution in the composite system is given directly in the present study as shown in Figure 2.

As shown in Figure 2b, for a research point with a certain $R$-value, if $z = 0$ mm, $t = 0$ s is set, when it passes through the different areas due to withdrawal operation and reaches the position of $z = 500$ mm, the time $t = 3,529$ s is obtained with the withdrawal speed of 8.5 mm min$^{-1}$ in the experimental test. Here, $R = 110$ mm represents the position where it is 110 mm from the roll core center and 10 mm from the roll core surface as the roll core radius is 120 mm.

Based on the simulation results in Figure 2b, the temperature varying with coordinate in the longitudinal direction as well as the temperature varying with time at different radial positions of the roll core is extracted as shown in Figure 3. It is used to analyze the affected depth and degree of roll core caused by the high-temperature austenitization. The $A_1/A_3$ temperature of 45 carbon steel calculated by Thermo-Calc software according to its composition is marked in Figure 3. It shows that the heating temperature of the roll core in the radial direction decreases gradually with an increase in

![Figure 2](image-url)  
**Figure 2:** The (a) geometric model and (b) temperature distribution of the composite billet.

![Figure 3](image-url)  
**Figure 3:** Temperature varying with coordinate and time at different radial positions of roll core.
distance from its surface. In other words, the larger the \( R \)-value, the higher the austenitizing temperature which is higher than the \( A_3 \) temperature.

In this work, as the heating process of roll core is continuous, the formation of austenite conforms to the basic law that the larger the heating rate, the shorter the incubation period, the higher the temperature at which the austenite transformation starts and finishes, the shorter the time required for the complete transformation [19]. Therefore, the actual austenitizing temperature of the region with a large \( R \)-value is higher than that with a small \( R \)-value, and all these temperature values are higher than the \( A_3 \) temperature (1,044 K) obtained by the equilibrium calculation. Since the 45 carbon steel (roll core) belongs to hypoeutectoid steel, if the heating temperature only exceeds the \( Ac1 \) temperature, only the pearlite can be transformed into austenite and the proeutectoid ferrite will remain, and if the heating temperature exceeds the \( Ac3 \) temperature and holds it for a sufficient time, we can hope to achieve uniform single-phase austenite. Therefore, the area where austenitization actually occurs of roll core is smaller than that above the \( A_3 \) temperature line in Figure 3; in other words, it is smaller than the area where the \( R \)-value is larger than 60 mm. From Figure 3, it can also be seen that the region closer to the roll core surface (the larger \( R \)-value zone) has not only a higher austenitizing temperature but also a longer austenitizing time, as a result, the austenite grains coarsening phenomenon only occurs at the radial regions where \( R > 60 \) mm and no significant changes will occur at the radial regions where \( R \leq 60 \) mm.

3.2 Microstructure evolution at different radial regions of roll core

To analyze the microstructure evolution, a series of specimens are taken from the roll core before and after the ESRC process as shown in Figure 4. The specimens at the radial position of \( R = 110 \) mm, \( R = 60 \) mm, and \( R = 0 \) mm are marked as edge, middle, and center, respectively. Figure 4b shows a partial longitudinal section of the GCr15/45 carbon steel composite billet produced by the ESRC method.

Figure 5 shows the microstructure of the above specimens; when compared with the microstructure of specimens taken from the initial roll core, an obvious austenite grain coarsening phenomenon occurs at the edge area (\( R = 110 \) mm) while no significant changes occur at the middle and center area (\( R = 60 \) mm and \( R = 0 \) mm), respectively. This is in good agreement with the analysis results based on the temperature simulation in Figure 3. Figure 6 shows the proeutectoid ferrite content of the above specimens, which decreases basically with an increase in composite height whether it is at the edge, middle, or center. It is worth noting that the grain size and proeutectoid ferrite content of the specimens at bottom zone are different from the analysis results of those at middle and top zones; in these zones, no element diffusion from the composite layer to roll core occurs, a rapid heat transfer is hindered due to the solidified slag shell, and a large amount of proeutectoid ferrite precipitated after the austenitization with the rapid heating and slow cooling. As a result, a uniform microstructure and more proeutectoid ferrite occur at the bottom zone, which are similar to the results of specimens taken from the initial roll core. Through the comprehensive analyses of the results shown in Figures 5 and 6, the grain size and the content and distribution characteristics of proeutectoid ferrite in the specimens at the middle and top zones are very similar, which shows good consistency. It proves that the ESRC process gradually enters the stable stage after the entrapped slag defect at the bimetallic interface disappears.

Temperature and carbon content are the main factors affecting the proeutectoid ferrite content in the roll core. Four specimens in the top zone of the longitudinal section were taken to measure the carbon content through the carbon-sulfur analyzer (CS600). The \( R \) values of the specimens are 15, 75, 105, and 115 mm, respectively, and the theoretical content of proeutectoid ferrite was calculated with the lever law of equation (1) based on the carbon content. Both the results of theoretical calculation and experimental measurement are shown in Figure 7.

\[
\omega_a = \frac{0.77 - x}{0.77 - 0.0218} \times 100\%
\]

where \( \omega_a \) is the theoretical content of the proeutectoid ferrite and \( x \) is the carbon content.

From Figure 7, it can be seen that the carbon content increases gradually from the roll core surface (large \( R \)-value) to the center position (small \( R \)-value), and a decreasing trend in the theoretical content of proeutectoid ferrite is observed. The experimental results have a consistent variation tendency with the calculated results, and the latter is always larger than the former. To determine the macrosegregation of roll core, the segregation coefficient \( k \) was employed, which is defined as [20]

\[
k = \frac{C_L}{C_0} \times 100\%
\]
where $C_0$ is the actual content of the element at a given point in the ingot and $C_i$ is the mean content of the element in the whole ingot.

According to equation (2), the segregation coefficients of carbon element at four positions ($R = 15, 75, 105,$ and $115$ mm) are $1.10, 1.08, 1.01,$ and $0.93,$ respectively. It shows a consistent variation tendency with the research on a casting billet of $0.60\%$ C that the carbon segregation coefficient on the centerline of its cross-section is $0.890 \leq k_c \leq 1.103$ [21]. After the rolling or forging process, the macrosegregation in the casting billet can be reduced to some extent, but it cannot be avoided. Generally, all the solidification processes of molten steels are nonequilibrium crystallization processes, due to the insufficient diffusion of alloy elements in the solid phase and liquid phase, not only the microscopic crystal segregation but also the macroscopic regional segregation occurs. In addition, the actual precipitation content of the high-melting point phase (e.g., the proeutectoid ferrite) is always lower than that of the theoretical analysis due to the repaid cooling rate, which is the reason for the differences between experimental and calculated results of the proeutectoid ferrite shown in Figure 7. The segregation behavior of carbon element is called as positive segregation, that is, more content is obtained in the final solidified core areas.

### 3.3 Microstructure of roll core near the bimetallic interface

In the former part, the temperature variation and microstructure evolution of roll core at different radial positions were analyzed. Here, the microstructure of the roll core near the bimetallic interface was investigated. Figure 8 shows the temperature distribution of the roll core surface obtained from Figure 2b. Different areas of the roll core surface are marked as ①, ②, ③, and ④ according to its different temperature characteristics which are corresponding to that in Figure 2b and have been described in detail in the previous report [15]. It illustrates that a more severe austenitization will happen here than that occurs for the specimens in Figure 4. A specimen including the interface of $35 \times 40$ mm was taken from the top zone of composite billet in Figure 4b, and the sampling schematic of metallographic and tensile specimens is shown in Figure 9a. From Figure 9a, a uniform bimetallic interface (marked as a yellow line) between the bimetals is obtained, and a metallographic specimen of $10 \times 10$ mm marked as S1 is taken for the microstructure observation.

For the roll core (hypoeutectoid steel), the microstructure examination in Figure 10 reveals the presence of plenty of perlite and a small quantity of proeutectoid ferrite precipitated along the prior austenite grain boundaries. The microstructure of the white reticular area is proeutectoid ferrite and that of the gray area is pearlite. Near the bimetallic interface, the proeutectoid ferrite precipitates in a network along the grain boundary, which is thin and uniform. However, when it is away from the interface, most of the proeutectoid ferrite precipitates along the grain boundary and a small amount of that precipitates within the grain, which is relatively thick. It can also be seen that the grain size and proeutectoid ferrite content shown in Figure 10b and c after the ESRC experiment have an obvious increase when compared with that in Figure 10a.

![Figure 4: Diagram of specimens taken from the roll core (a) before and (b) after the ESRC process.](image-url)
before the experiment. This phenomenon gradually weakens in the roll core side with an increase in distance from the bimetallic interface. For the hypoeutectoid steel, the prior-austenite grain size is strongly increasing with an increase in the austenitization temperature [22], and a higher isothermal transition temperature will lead to the much thicker and more proeutectoid ferrite precipitated along the prior-austenite grain boundary [23]. In this work, except for the temperature, an obvious diffusion of C and Cr elements occurred in the bimetallic interface due to the composition difference between GCr15 and 45 carbon steel is also the main factor affecting the precipitation of proeutectoid ferrite. As is known to all, C is the element of expanding austenite phase zone and Cr is opposite; in addition, a rapid cooling rate is not beneficial to the precipitation of high-melting point phase.

Figure 10b and c shows that partial Widmanstatten structure is generated in the heat-affected zone (HAZ) of roll core. Through the above analyses, a high-temperature austenitization occurs in roll core especially its surface areas and the austenite grains become coarsely

Figure 5: Microstructure of the specimens taken from the roll core (j–l) before and (a–i) after the ESRC process.
enough as shown in Figure 10b. On this condition, with a faster or even though a normal cooling rate, part of the proeutectoid ferrite grows into the grain interior from the austenite grain boundary along with a certain crystal face and precipitates as an acicular morphology \[24\].

The remaining austenite between the acicular ferrite is finally converted to lamellar pearlite, and this kind of complex structure is called as Widmanstatten structure.

3.4 In situ analysis of grain growth and cooling process of roll core

To analyze the grain growth of austenite grain and precipitation of acicular ferrite, an in situ observation was carried out. At HAZ locations where the temperature just exceeds the solidus temperature of the matrix, liquation initiates at the grain boundaries to form a continuous film of grain-boundary liquid \[25\]. As the solidus temperature of 45 carbon steel is 1,686 K (1,413°C), the highest temperature of 1,400°C was selected in the in situ analysis to observe the crystal growth behavior clearly. Both the marks of A and B in Figure 8 represent the temperature of 1,673 K (1,400°C); about 7 min (420 s) between A and B positions will be taken according to its coordinates and the withdrawal speed. Based on the values shown in Figure 8, the stated function of temperature varying with time for the in situ analysis was obtained as shown in Figure 11. The specimen was heated from room temperature to 1,400°C over a period of about 604 s and held at this temperature for a further 420 s.

"Missing" grain boundaries are common at the early stage of the structure evolution but rapidly become rare as the process continued \[26\], so a partial process of heat preservation at 1,400°C is adopted to analyze the growth behavior of austenite grain as shown in Figure 12. It should be noted that the scale label in Figures 12 and 13 is 100 µm as the \(\times 5\) represents the magnification of the lens.

From Figure 12, it can be seen that the growth of austenite grain mainly occurs through the following three ways \[27\]. The first one is the migration and expansion of austenite grain boundaries, such as the growth of grains marked as \(\odot\) and \(\odot\). The second one is that the intermediate grain segmented and absorbed by the peripheral grains, for example, the grain marked as \(\odot\) is absorbed by the grains of \(\odot\), \(\odot\), and \(\odot\), the same phenomena occur with the grains of \(\odot\) and \(\odot\) as shown in Figure 11. The third one is always that multiple small
grains coalesce into a single large grain [28] which is not detected due to the limited field of view. Hu et al. have carried out a dynamic observation of austenite grain growth behavior of advanced bainite steel, which illustrates that the austenite grain coarsening mainly occurs at temperatures higher than 1,100°C, and no significant phenomenon occurs when that is lower than 1,100°C [27]. The results fully illustrate that the temperature has a very important influence on the grain size. In Figure 3, the closer to the roll core surface, the higher the temperature and the longer the heating time, so the larger the grain size obtained in the edge area of roll core, especially that near the bimetallic interface as shown in Figure 10.

Dynamic analysis for the cooling process from 1,400°C to room temperature is also carried out based on the temperature varying with time in Figure 11, and the precipitation behavior of acicular ferrite (proeutectoid ferrite) during this process is shown in Figure 13.

Figure 13a illustrates that the acicular ferrite lath is nucleated on the grain boundary with a decrease in temperature in the earlier stage. With a further decrease in temperature, there is an increase in the number of acicular ferrite laths or plates (Figure 13b and c), and the
Acicular ferrite grows into the grain interior from the austenite grain boundary along with a certain crystal face. The remaining austenite between the adjacent acicular ferrite will be finally converted to lamellar pearlite, and this kind of complex structure is called as the Widmanstatten structure. In this work, the precipitation process of acicular ferrite mainly occurs in the temperature range from 646°C to 539°C in Figure 13. Generally, the coarse austenite grain and a rapid cooling rate are the main factors affecting the formation of Widmanstatten structure [19].

### 3.5 Property of roll core near the bimetallic interface

The tensile specimens with different microstructure marked as S2 and S3 were taken from the roll core side near the interface as shown in Figure 9a and the size diagram is shown in Figure 9b. The actual distance from the central axis of tensile samples S2 and S3 to the bimetallic interface are 2 and 12.5 mm, respectively. Figure 14 shows that the average tensile strength of S2 and S3 specimens are 622 and 665 MPa, respectively. The stress varying with strain for one of the S2 and S3 specimens, respectively, is given in Figure 14, a proportional stress–strain zone is first formed, and then an increasing extent of plastic deformation occurs before its

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**Figure 12:** In situ observation of grain growth in the heat preservation process at 1,400°C of (a) 773.04 s, (b) 893.04 s and (c) 1013.04 s.

**Figure 13:** In situ observation of acicular ferrite precipitation during the cooling process at (a) 646.4°C, (b) 583.0°C and (c) 539.2°C.

**Figure 14:** Dependence of stress varying with strain for the S2 and S3 specimens.
fracture especially for the S3 specimens. In conclusion, tensile strength and plasticity of S3 specimens are higher and better than that of the S2 specimens.

Figure 15 shows the macrostructure and microstructure of the tensile fracture for the S2 and S3 specimens. It illustrates that both the specimens near or away from the interface present the mixed features of both ductile and brittle fractures with the partial cleavage fracture morphology (river patterns) and lots of dimples. According to the flow direction of the river, the crack propagation direction can be determined and the crack source can be found. For both specimens, the cracks were developed from the near edge area and extended to the core.

As is known to all, at room temperature, the finer the grain, the higher the strength and the better the plasticity and toughness. The S3 specimens that are much further from the bimetallic interface than S2 specimens have the finer grains and no Widmanstatten structure. In addition, generally, the Widmanstatten structure is an overheating defect structure. Its existence will lead to a serious separation of the matrix, which not only decreases the material’s mechanical properties of toughness and plasticity but also increases the brittle transition temperature that can promote a brittle fracture of the material.

4 Conclusions

(1) Temperature distribution and element diffusion have great effects on the microstructure at different positions of roll core. It illustrates that the ESRC process gradually enters the stable stage after the entrapped slag defect at the bimetallic interface disappears and the high-temperature austenitization process mainly occurs at the radial regions where \( R > 60 \) mm and no significant changes occur at the areas where \( R \leq 60 \) mm.

(2) During the ESRC process, the closer to the roll core surface, the higher the temperature and the longer the heating time, so the larger the grain size obtained in the roll core side, especially the regions near the bimetallic interface. As a result, the tensile strength and plasticity of the roll core specimens near the bimetallic interface are lower than that away from the interface.

(3) The austenite grain growth and acicular ferrite precipitation behaviors with the temperature variations can be observed effectively through the in situ analysis. The growth of austenite grain mainly occurs through the three ways, and the precipitation process of acicular ferrite mainly occurs in the temperature range from 646°C to 539°C.

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