



Influence of cooling rate upon weld metal microstructural evolution behaviors of EH36 shipbuilding steel

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Abstract

Microstructural evolution features have been systematically investigated for the weld metal of EH36 shipbuilding steel under an in situ confocal scanning laser microscope. The influence of cooling rate on microstructural changes during the transformation from austenite to ferrite has been clarified. It is found that ferrite side plates form preceding to acicular ferrites, although the starting temperature of respective component decreases as the cooling rate is raised. In particular, the growth rate of acicular ferrite is measured to increase significantly, rising from 30.4 $\mu\text{m/s}$ at a cooling rate of 3 K/s to 109.0 $\mu\text{m/s}$ at 15 K/s, driven primarily by an ever-increasing degree of undercooling. These findings highlight the critical role of cooling rate in dictating the sequence and growth rate of microstructural transformations, which is crucial for optimizing welding processes to obtain desired microstructures while avoiding the formation of deleterious components.

Keywords Cooling rate · Shipbuilding steel · Weld metal · Microstructural evolution

1 Introduction

Low-carbon low-alloy EH36 shipbuilding steel, acclaimed for exceptional strength and excellent weldability, finds widespread application in submerged arc welding (SAW) procedures [1, 2]. Microstructural features play a vital role in dictating the performance of the weld metal (WM). Among these microstructural attributes, acicular ferrite (AF) stands out as an active phase component with interlocking characteristics capable of refining the microstructure. It is a highly desirable constituent for enhancing mechanical properties. Extensive efforts have been dedicated to elucidating the factors that actuate the formation of AF and microstructural evolution behaviors, encompassing investigations with regard to inclusion characteristics,

chemical composition, and notably, cooling rate experienced during the phase transformation process [3–5].

In terms of chemical compositions and inclusion characteristics, the concept of inclusion-induced AF formation within appropriately sized austenite grains has gained gravitational recognition [4, 6–8]. Ilman et al. [9] discovered that the addition of Ti in C–Mn steel WM accelerated phase transformation kinetics and modified the ultimate microstructure, leading to the replacement of Widmanstätten with AF. Cooling rate has been widely acknowledged as a paramount variable in determining the volume fraction and morphology of different microstructures. Guo et al. [5] investigated phase transformation and microstructural evolution under different cooling conditions. They noted that in inclusion-engineered steels, a higher cooling rate promoted the formation of AF, which led to an increased degree of interlocking of the AF phases. Variations in cooling rate exert a pronounced impact on the growth kinetics of AFs, which, in turn, govern the final microstructures of WMs and consequently affect the resulting mechanical properties, albeit the underlying mechanisms remain a subject of ongoing debate [10, 11].

Numerous studies have provided evidence for the influence of cooling rate on various aspects of microstructural evolution, but the focus has predominantly

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Table 1 Chemical compositions of investigated WM (wt.%)

C	Mn	Si	Al	Ti	O
0.071	1.090	0.072	0.012	0.041	0.017

been on utilizing thermal expansion experiments and traditional metallographic observations [3, 12]. Recently, high-temperature confocal scanning laser microscopy (CSLM) has emerged as a valuable tool, overcoming technical limitations of traditional inspection methods and enabling real-time detection of high-temperature phase transformation [13–15]. It could serve as a satisfactory technique in determining ferrite types and respective formation locations, as well as evaluating transformation kinetics coupled with quantified in situ measurements. As an illustration, Zou et al. [16] reported that at a cooling rate of 92 K/min, the preferred nucleation sites of ferrite laths shifted from grain boundaries to inclusion surfaces. Additionally, a critical cooling rate was identified, at which the effectiveness of AF formation reached parity with ferrite side plate (FSP). Xie et al. [17] demonstrated that at a cooling rate of 530 K/min, the growth rates of bainite laths ranged from 70.6 to 122.4 $\mu\text{m/s}$, whereas corresponding growth rates of ferrite laths significantly decreased to 0.6 to 0.7 $\mu\text{m/s}$ when the cooling rate was lowered to 35 K/min.

Therefore, the objective of the current work is to, by utilizing in situ CSLM observations, systematically shed light on the relationship between cooling rate and microstructural evolution behaviors in the WM of EH36 shipbuilding steel. This endeavor seeks to contribute to a refreshed understanding of the formation and growth characteristics of salient microstructure constituents and is anticipated to furnish a reliable approach for obtaining desired WM microstructures.

2 Experimental

Fluxes for SAW were prepared using the fusion method. Corresponding proportions of CaF_2 (> 98.5 mass%) and TiO_2 (> 99.0 mass%) reagent powders were weighed and thoroughly mixed. Subsequently, these raw materials were placed into a graphite crucible and heated to 1823 K in an induction furnace for 1 h to ensure homogeneity. The molten fluxes were then quenched in a water bath. Afterward, solidified flux particles underwent a baking process at 973 K to remove residual moisture and graphite. Following this, the fluxes were crushed and sieved into 350 to 1397 μm before being used for welding. EH36 shipbuilding steel with a thickness of 24 mm was targeted as the base metal, and the double-wire SAW technique was carried out to obtain single-pass bead-on-plate WMs. The welding

parameters were set as follows: a welding speed of 500 mm/min, a forward welding wire with a current of 850 A and voltage of 32 V, and a backward welding wire with a current of 625 A and voltage of 36 V. The method for detecting WM chemical compositions are based on proven procedures [18]. Corresponding chemical compositions are presented in Table 1.

Cylindrical CSLM specimens with the diameter of 3 mm and thickness of 1.5 mm were obtained from WM cross-sections and subjected to standard mechanical grinding and polishing. Thereafter, specimens were placed in high purity alumina crucibles and mounted in the CSLM equipment (VL2000DXSVF18SP, Yonekura Manufacturing Co., Ltd., Japan). Specimens were heated to 473 K at a rate of 50 K/min, followed by a rate of 300 K/min to 1673 K, and held for 120 s. Subsequently, specimens were subjected to continuous cooling to 573 K with different cooling rates of 3, 5, and 15 K/s, as these fall within the typical range of 2–15 K/s commonly chosen [5, 19, 20]. In order to capture detailed information regarding ferrite nucleation and growth, continuous images were collected and recorded at 15 frames per second by a charge-couple device. Following in situ CSLM observations, specimens were mechanically ground, polished and etched with a solution of 4 vol.% Nital to reveal and confirm microstructural features by using an optical microscope (OM, GX51, OLYMPUS, Japan). The relative volume fractions of various typical microstructural phases were quantitatively determined by the standard manual point counting method. At a magnification of 1000 $\times$, analyses were conducted on a minimum of 20 randomly visual fields for each specimen using OM [7, 21].

3 Results and discussion

3.1 In situ observation of ferrite formation

Figure 1 depicts representative CSLM images demonstrating microstructural evolution of WMs under varying cooling rates. Due to the formation of surface relief induced by the displacive transformation from austenite to ferrite, distinct morphologies and microstructural features of FSPs, AFs and bainitic ferrite (BFs) can be clearly visualized from interference contrast enabled by CSLM snapshots [22, 23]. FSPs typically initiate from austenite grain boundaries or polygonal ferrites and grow directly

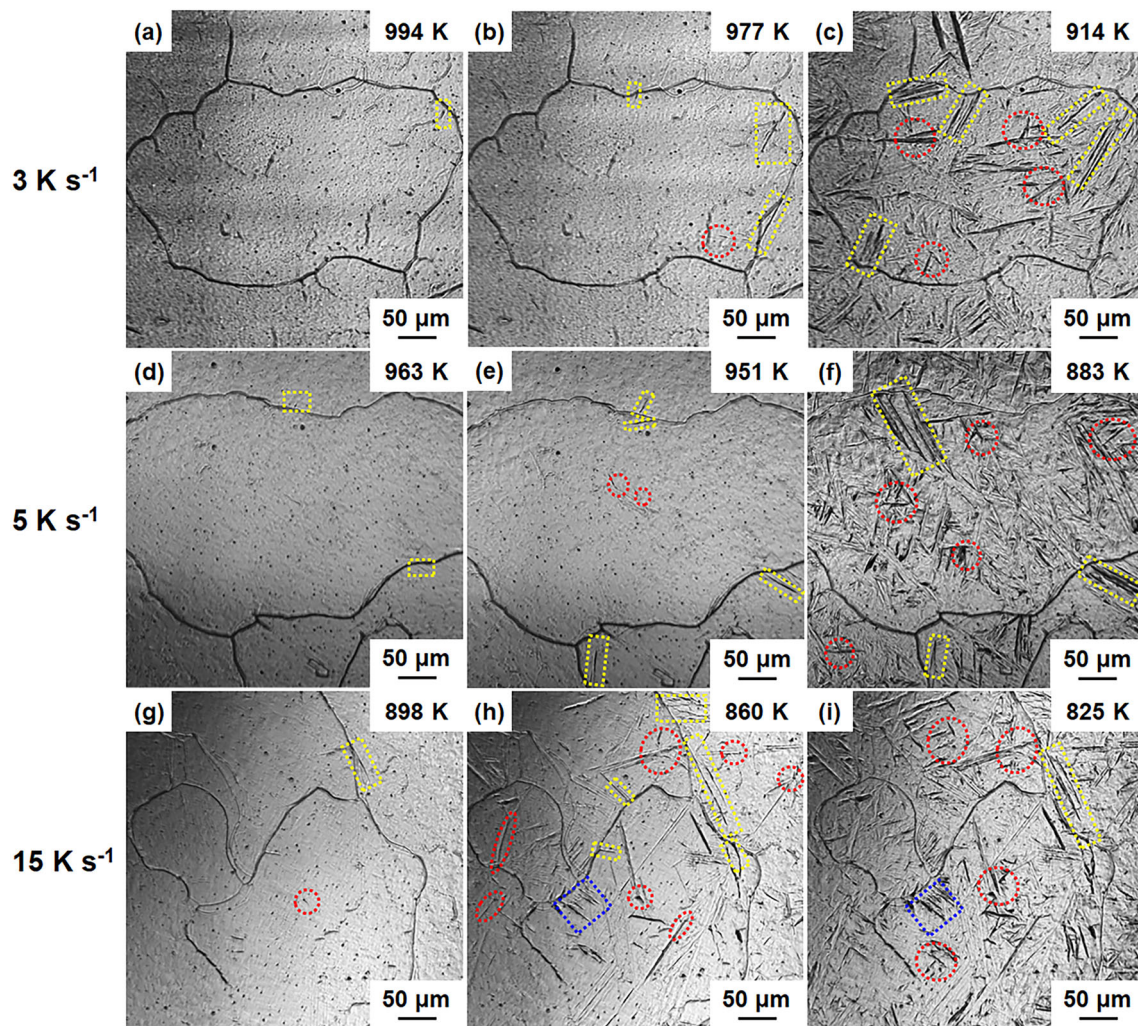


Fig. 1 In situ observation of ferrite formation and growth at different cooling rates. **a–c** 3 K s^{-1} ; **d–f** 5 K s^{-1} ; **g–i** 15 K s^{-1} . Yellow rectangle, red circle, and blue rectangle represent ferrite side plate, acicular ferrite, and bainitic ferrite, respectively

protruding into the interior [24]. Compared with FSPs, AFs nucleate on inclusion surfaces in the austenite grains and are characterized by fine grain sizes, high dislocation density, large-angle grain boundaries, and randomly chaotic arrangement structures. Furthermore, BFs generally nucleate at austenite grain boundaries and grow parallel in the grain interior at lower temperatures compared with FSPs and AFs. Moreover, the inclination between these BF laths is small, and the dislocation density of BF is high [25]. As depicted in Fig. 1a–c, when the cooling rate is 3 K/s , an FSP lath, circled by the yellow dashed rectangle, is observed on the grain boundary at about 994 K . When the temperature drops to 977 K , AFs form on inclusion surfaces and grow in multiple directions, as indicated by red dashed circles. At the cooling rate of 5 K/s , FSPs preferentially form along grain boundaries and grow toward the interior of grains at 963 K . Subsequently, when the cooling temperature drops to 951 K , AFs are formed.

As the temperature gradually decreases, new AFs continue to emerge and existing AFs grow further, as illustrated in Fig. 1d–f. Different from the cooling rates of 3 and 5 K/s , for the specimen with 15 K/s cooling rate, the formation of FSP and AF occurs almost simultaneously, as shown in Fig. 1g. The final microstructure mainly consists of AFs and FSPs but also with a small amount of BF, as represented by the blue rectangle in Fig. 1i.

An interesting observation is that FSPs tend to form more readily than AFs when subjected to lower cooling rates. This preference could potentially be attributed to the presence of different nucleation sites, which might incur variations in energy barriers [15, 26]. It is well-established that the formation of both FSPs and AFs demands a substantial driving force to overcome associated energy barriers [17, 27]. Thewlis et al. [3] conducted calculations pertaining to activation energy barriers involved in the nucleation of ferrite at various heterogeneous sites. It was

shown that austenite grain boundaries carried notably lower energy barriers compared to intragranular inclusions, thereby facilitating more favorable conditions for the formation of FSPs [27, 28]. As a result, the initial stages of phase transformation are expected to take place at austenite grain boundaries [17, 20]. However, as the transformation progresses, participation of inclusions becomes increasingly possible, and larger inclusions, by aiding in reducing energy barriers, demonstrate enhanced efficacy in initiating the formation of AF formation as contrast to smaller inclusions [26, 29]. Ti-rich inclusions, serving as inert matrices characterized by high surface energy, possess the capacity to facilitate the formation of AF by offering favorable nucleation sites attributed to the low lattice misfit between the TiN layer and ferrite, concomitant with the high-energy strain field [30, 31]. Yuan et al. [18] revealed that the distribution of Ti-containing inclusions in the current investigated WM predominantly fell within the range of 0.3 to 2 μm , and these sub-micron inclusions could be widely considered highly effective toward the formation of AFs [32–34]. Furthermore, Mu et al. [35] reported that, with a constant inclusion size, the energy barrier for heterogeneous nucleation decreased as the temperature was lowered.

3.2 AF growth behaviors at different cooling rates

In order to obtain the growth rates of AFs during the phase transformation, certain AFs were randomly selected for measurements. Figure 2a shows the growth length of each AF versus the corresponding different growth time, and the real-time AF length was measured using the ImageJ software. Due to the hindrance from other ferrites or grain boundaries, the growth rate of each individual AF is significantly different for a given cooling rate [30]. Furthermore, the growth rate of AF indicates a notable increase as the cooling rate is raised. At the cooling rate of 3 K/s, the average growth rate of AF falls within the range of 12.9–30.4 $\mu\text{m/s}$. When the cooling rate is elevated to 5 K/s, the corresponding average growth rate increases to a range of 16.5–58.2 $\mu\text{m/s}$. With a higher cooling rate of 15 K/s, the average growth rate is further elevated, ranging from 35.9–109.0 $\mu\text{m/s}$. It is clear that as the cooling rate increases from 3 to 5 and 15 K/s, the difference between the maximum and minimum values of AF growth rate shows an upward trend, escalating from 17.5 to 41.7 and 73.1 $\mu\text{m/s}$, respectively. In addition, the actual AF growth rate (G_{AF}) is claimed as an important parameter to evaluate AF growth kinetics and can be represented by the maximum value of the average growth rate [36]. Therefore, at cooling rates of 3, 5 and 15 K/s, corresponding G_{AF} values for three specimens are 30.4, 58.2 and 109.0 $\mu\text{m/s}$, respectively. The

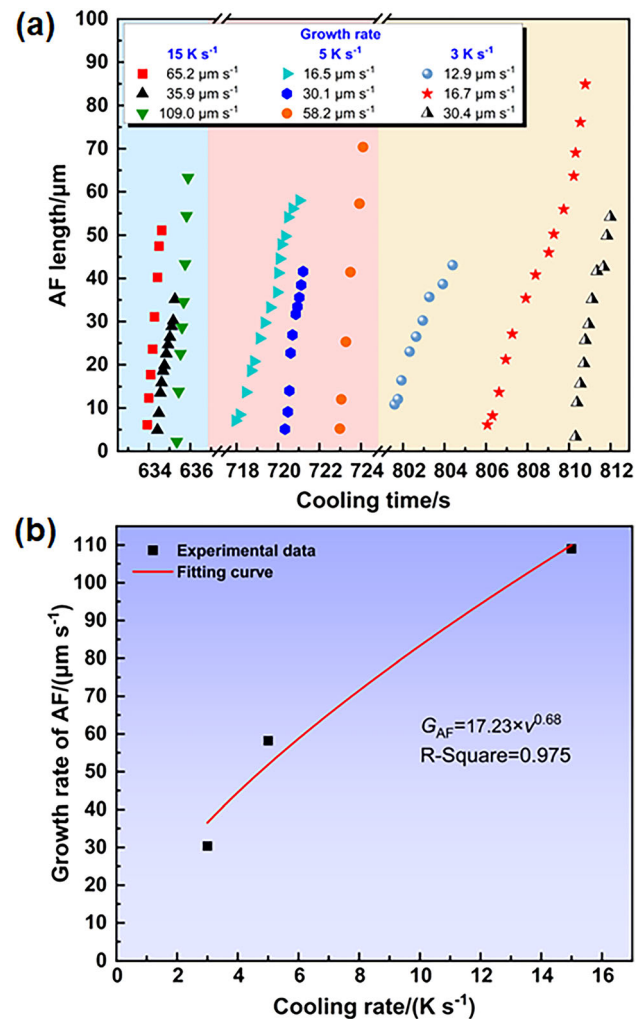


Fig. 2 Relationship between AF growth length against cooling time at different cooling rates (a) and variation of AF actual growth rate with cooling rate (b)

relationship between G_{AF} and cooling rate is illustrated by the red curve in Fig. 2b, as can be represented by Eq. (1):

$$G_{\text{AF}} = 17.23 \times v^{0.68} \quad (1)$$

where v is the cooling rate, K/s. It is consistent with the phenomenon observed in previous studies showing a parallel increase in G_{AF} with increasing cooling rate [30].

It is widely recognized that the thermodynamic driving force of the phase transformation is directly influenced by the degree of undercooling, ΔT , which is required to overcome the energy barrier to initiate the transformation [37]. ΔT can be expressed using the following equation:

$$\Delta T = T_e - T_s \quad (2)$$

where T_e represents the equilibrium temperature for austenite to ferrite phase transformation; and T_s denotes the observed starting temperature for the appearance of the initial ferrite [17, 38]. In the current work, T_e was

calculated to be approximately 1118 K using FactSage 8.1 software, consequently, $\Delta T = 1118 \text{ K} - T_s$. When the cooling rate increases, T_s decreases. Since the value of T_c is fixed, the degree of ΔT will increase. Such similar results have been reported by Wang et al. [19]. Therefore, a significant rise in AF growth rate with the increase in cooling rate can be attributed to the increase in ΔT .

3.3 Effect of cooling rate on microstructural evolution

According to the observations from CSLM images, starting temperatures of FSP and AF formation, along with the finishing temperature of ferrite phase transformation at different cooling rates, were extracted and interconnected to construct the schematic continuous cooling transformation diagram for the WM, as depicted in Fig. 3. In the CCT diagram, a clear tendency is evident, showing that values of both T_s and T_f decrease as the cooling rate is increased. It shows that $T_{s,\text{FSP}}$ is consistently higher than $T_{s,\text{AF}}$ for each cooling rate, which is anticipated due to the higher energy barrier for ferrite nucleation on inclusion surfaces as compared to grain boundaries [27, 28]. It is also found that the difference between $T_{s,\text{FSP}}$ and $T_{s,\text{AF}}$ is reduced from 17 to 4 K as the cooling rate is increased from 3 to 15 K/s. However, for the specimen with a cooling rate of 15 K/s, $T_{s,\text{FSP}}$ and $T_{s,\text{AF}}$ are similar in the range of 898 to 902 K. Furthermore, difference between T_f and T_s remains relatively stable at different cooling rates, approximately 80 K. As mentioned above, ΔT serves as one of necessary conditions for driving phase transformation. At reduced cooling rates, a decrease in degree of ΔT could lead to destabilization of undercooled austenites, significantly

facilitating phase transformation from austenite to ferrite [30]. Conversely, a larger cooling rate results in a higher degree of ΔT , which enhances the stability of austenite, leading to a delay in the phase transformation, thus decreasing the T_s of the two target phases, $T_{s,\text{FSP}}$ and $T_{s,\text{AF}}$ [5, 30].

Figure 4 illustrates the microstructural characteristics, along with the relative fractions of typical constituent phases, of specimens subjected to different cooling rates. As presented in Fig. 4a, b, for specimens cooled at 3 and 5 K/s, the microstructure mainly consists of AF, FSP, GBF (grain boundary ferrite), and PF (polygonal ferrite). Analogous microstructural features are observed for the specimen with a cooling rate of 15 K/s. Apart from identified AF, FSP, GBP, and PF, a small amount of BF also forms within the austenite grain, as shown in Fig. 4c, and such morphological disparities of these microstructural phases are reported by Zhang et al. [22]. Generally, BFs grow into crystallographic packets composed of a colony of parallel ferrite laths with similar orientations, most of which are low-angle boundaries [39]. Figure 4d shows relative volume fractions and distributions of dominant microstructural constituents. It can be seen that the relative phase fraction changes as the cooling rate increases from 3 to 5 and 15 K/s. It can be seen that the fractions of GBF, FSP, and PF decrease, whereas that of AF increases from 49.9% to 70.4%. Such findings were well verified by Di et al. [32] through statistical analysis and microstructural observations. A faster cooling rate will shorten the duration for phase transformations within the high-temperature region, consequently limiting the growth of FSP and GBF. On the other hand, due to increased ΔT , the driving force for AF formation will increase, enabling more austenite being transformed into AF [35, 40, 41]. Therefore, it is reasonably expected that at faster cooling rates, the volume fraction of AF will be higher as compared to that under lower cooling rate.

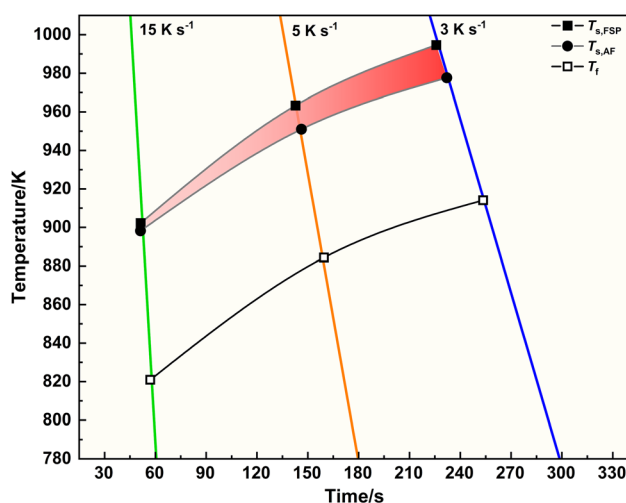


Fig. 3 Schematic continuous cooling transformation (CCT) diagram for WM. $T_{s,\text{FSP}}$ —Starting temperature for FSP formation; $T_{s,\text{AF}}$ —starting temperature for AF formation; T_f —finishing temperature for ferrite phase transformation

4 Conclusions

1. During the transformation from austenite to ferrite, the evolution sequence is characterized by the initial appearance of ferrite side plates, followed by the formation of acicular ferrites. As the cooling rate increases from 3 to 15 K/s, starting temperatures of these two microstructural constituents decrease from 994 to 902 K for ferrite side plate and 977 to 898 K for acicular ferrite, respectively.
2. The growth rate of acicular ferrite significantly increases from 30.4 to 109.0 $\mu\text{m/s}$ as the cooling rate increases from 3 to 15 K/s, which is predominantly attributed to the enhanced degree of undercooling.

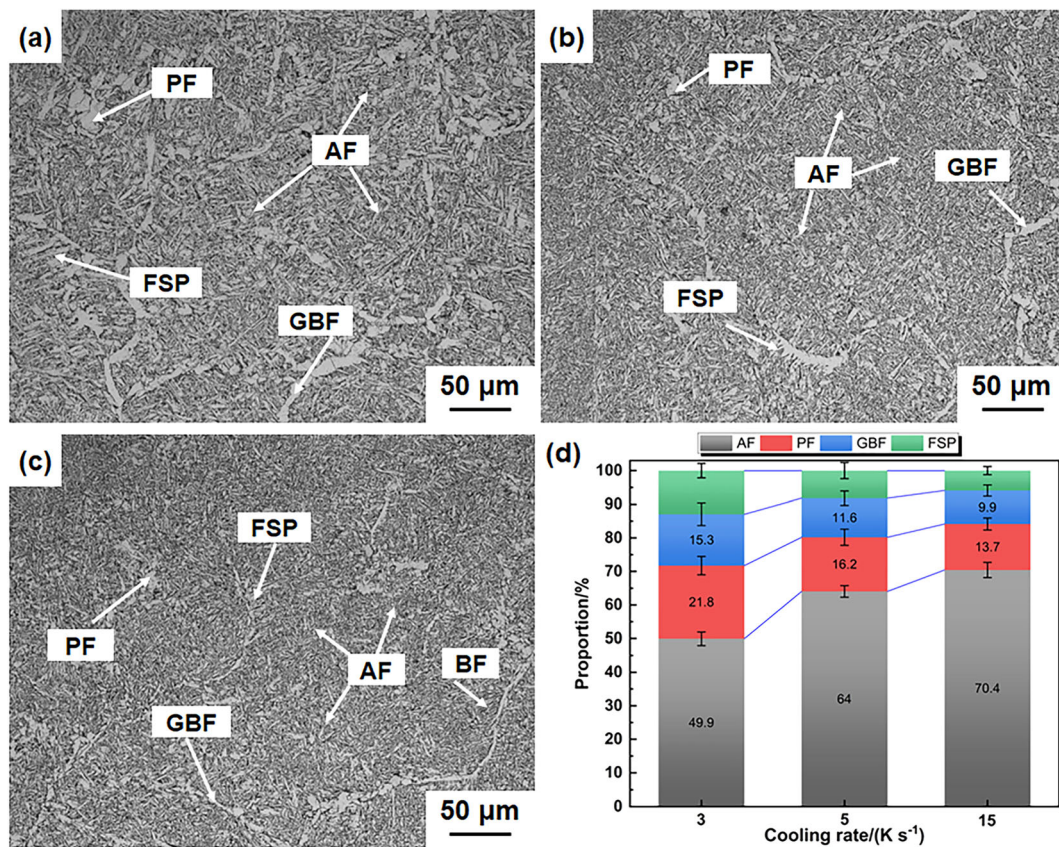


Fig. 4 Typical microstructural features of samples at different cooling rates. **a** 3 K s⁻¹; **b** 5 K s⁻¹; **c** 15 K s⁻¹; **d** relative fraction of typical constituent phases

With the increase in cooling rate, the volume fraction of acicular ferrite increases, which is concurrent with the decrease in ferrite side plate, grain boundary ferrite, and polygonal ferrite.

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Declarations

Conflict of interest Cong Wang is an editorial board member for *Journal of Iron and Steel Research International* and was not involved in the editorial review or the decision to publish this article. The authors have no competing interests to declare that are relevant to the content of this article.

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