Nucleation Analysis of Variant Transformed from Austenite with Σ_3 Boundary in High-Strength Low-Alloy Steel

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Abstract

We elucidate here the effects of annealing twins on phase transformation products based on electron back-scatter diffraction analysis and corresponding microstructure visualization and quantification methods. Martensite and bainite were obtained by rapid continuous cooling and isothermal processing at different temperatures, respectively, which were designed to study variants formed in austenite with Σ_3 boundary. The isothermal transformation near martensite start (M_s) temperature was most conducive in obtaining the highest content of twin-related V_1/V_2 variant pair. Based on the classical nucleation theory of martensite and bainite, respectively, the role of austenite Σ_3 boundary in martensite and bainite transformation is illustrated.

Keywords Steel · Austenite Σ_3 boundary · Variant · Phase transformation · Martensite · Bainite

1 Introduction

Hall–Petch relationship describes the refinement of grain size to achieve strength and toughness in metals [1–3]. For martensitic or bainitic steels, the key to improving strength and toughness is not only to refine the prior austenite grains, but also the internal units such as packets, blocks and lath [4]. The packet is composed of a plurality of blocks arranged in parallel, and the block can be composed of two sub-blocks with misorientation of $\sim 10^\circ$. Laths having the same orientation arranged in parallel constitute a sub-block or block, which can also be called a variant. There are many studies on strengthening and toughening of steels, which are key aspects in improving the performance of steel. The studies on nano-bainitic steels [5–7] show that refining lath

contributes to improvement of strength and toughness. Morito et al. [8] and Wang et al. [9] have studied the block size to define the role in governing strength and toughness of martensite/bainite. Some studies [10, 11] noted that increasing the carbon content of steel can effectively reduce the size of the packet, while some studies [12–14] observed that reducing the packet can improve toughness.

The matrix microstructure of many advanced high strength steels generally consists of martensite or bainite, which has a near Kurdjumov–Sachs (K–S) orientation relationship (OR) [15] with the parent phase-austenite. According to K–S OR, 24 variants are defined and adopted to describe microstructure, and corresponding characterization by electron back-scatter diffraction (EBSD) provides more accurate description than the morphology. Both Takayama et al. [16] and our previous studies [17] showed that martensite has a higher content of V_1/V_4 variant pairs, while bainite has a higher content of twin-related V_1/V_2 variant pairs. The rotation angle/axis of V_1/V_4 and V_1/V_2 in the standard K–S OR were $10.5^\circ/[0-1-1]$ and $60.0^\circ/[11-1]$, respectively. Martensite has higher strength and hardness, which is mainly related to its higher content of V_1/V_4 variant pairs. In contrast, bainite has better strength and toughness combination because it has a higher content of V_1/V_2 variant pairs, resulting in a higher density of block boundaries. Therefore, the mechanical properties of steel are influenced by these differently oriented grain boundaries, that is, variant selection determines the density of different types of

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boundaries, which is determined by phase transformation conditions.

From the perspective of crystallography, there are many studies [9–11, 16] on the effects of phase transformation conditions on products, including cooling rate, transformation temperature, alloying elements, etc. In addition, the austenite state has significant influence on the microstructure, but related studies mainly focused on the size of prior austenite grains [18–22]. Studies [23, 24] have shown that the austenite grain boundary is the nucleation site for bainite transformation, and the variant selection becomes stronger with the increase of transformation temperature and decrease of carbon content. There are also studies [25, 26] on variant selection of austenite with Σ_3 boundaries, and it was found that coinciding variants can be produced after bainite transformation, which maintains near K–S OR with both of the two twin-related austenite. According to Abbasi et al. [25], the content of coinciding variants gradually decreases with increasing distance from the Σ_3 grain boundary. Since annealing twins are extremely common during austenitizing process of steel, thus, to study the effect of austenite state on the transformation products, it is necessary to investigate the variant nucleation from austenite with Σ_3 boundaries and the corresponding relationship with the mechanical properties. In addition, the role of austenite Σ_3 boundary in martensite transformation has not been revealed.

In this study, the variant selection of austenite Σ_3 boundary in martensitic and bainite transformation is studied. According to the traditional theory of martensite transformation, the transformation of a thermal martensite is independent of temperature and related to cooling rate. When the cooling rate is not sufficient to trigger martensite transformation, bainite transformation occurs. Therefore, the effects of austenite Σ_3 boundaries in martensite and bainite transformation are studied using rapid continuous cooling transformation and isothermal transformation, respectively.

2 Experimental

A 0.04C-0.21Si-1.35Mn-2.36(Cr + Ni + Mo)-0.35Cu-0.002B (wt%) steel was studied, which was obtained in the form of 12 mm thick hot-rolled plates. Cylindrical samples of 10 mm in length and 4 mm in diameter were machined and martensite start and final transformation temperature (M_s and M_f) were measured by DIL 805A/D dilatometer. The sample was heated to 1050 °C at a rate of 10 °C/s, and after 15 min of incubation, it was cooled to room temperature at a rate of 100 °C/s. As shown in Fig. 1a, the experimental M_s and M_f temperatures were 452 and 368 °C, respectively. The heat treatment before cooling was consistent with the measurement of M_s and M_f temperature. After isothermal treatment of 400, 450 and 500 °C for 10 min, the samples were cooled to room temperature at a cooling rate of 20 °C/s, as shown in Fig. 1b. Samples for microstructure characterization were mounted and mechanically polished using standard metallographic procedure. The samples were etched with 4% nital solution for microstructure observations, which were carried using TESCAN MIRA3 field emission scanning electron microscopy (FE-SEM). The samples were electropolished using a solution containing 85% alcohol, 10% perchloric acid and 5% glycerol to obtain crystallography information by Oxford-EBSD system, respectively. Since the observed microstructure are all transformed, and it is difficult to distinguish the austenite Σ_3 boundaries from the morphology, there were difficulties in studying the phase transformation products at austenite Σ_3 boundaries. The method of austenite reconstruction program developed by the author's research group was adopted and austenite Σ_3 boundaries were determined in 2D EBSD figure by corresponding experimental and theoretical pole figure.

Fig. 1 Transition temperatures during rapid continuous cooling (100 °C/s) **a**, schematic diagram showing the heat treatment process **b**

3 Results

3.1 Microstructure and properties

Figure 2 shows SEM and EBSD results of isothermally processed samples. As shown in Fig. 2a–c, the microstructure of quenched sample and isothermal transformation at 400 and 450 °C exhibited similar lath characteristics. At 450 °C isothermal process, fine martensite–austenite (M–A) constituents are obtained, which is not apparent in 400 °C isothermal sample. The microstructure obtained at 500 °C showed bainite with coarser M–A constituents (Fig. 2d). Studies [27, 28] have shown that promoting complete bainite transformation is also beneficial to refine bainitic ferrite and M–A constituents, which is also consistent with our experimental results. The M–A constituents nucleate during cooling after the isothermal process, because isothermal bainite transformation is incomplete, which is confirmed in Fig. 3. The change in temperature–length curve during the final cooling process is not straight, but there is a deviation, indicating that a phase transformation occurred during the final cooling process. The difference between martensite and bainite is sometimes difficult to distinguish, but the EBSD grain boundary distribution map showed differences, as shown in Fig. 2e–h. Quenched sample showed high density of intricate high-angle grain boundaries (HAGBs) and low-angle grain boundaries (LAGBs), which is consistent with the previous studies [16, 17]. When the isothermal temperature

was increased, many long and straight HAGBs appeared and were arranged in parallel. When the isothermal temperature was 450 °C, the content of long and straight HAGBs was the highest. When the isothermal temperature reached 500 °C, the density of HAGBs and LAGBs was decreased, and the size of block increased.

In the case of 400 °C isothermal process, the transformation consisted to two stages: continuous cooling transformation between M_s temperature and 400 °C and isothermal transformation at 400 °C, as shown in Fig. 3a. When the isothermal temperature was increased to 450 °C, a small portion of continuous cooling transformation occurred between M_s temperature and 450 °C, followed by an isothermal transformation at 450 °C, as shown in Fig. 3b. Figure 3c shows that the 500 °C isothermal process included only isothermal transformation. It is confirmed that athermal martensite transform during cooling process and stop at isothermal temperature [29, 30]. Thus, the microstructure of the quenched sample was martensite, while the isothermally transformed structure at 450 °C was a typical lath bainite. According to reports on the boundary characteristics of martensite and bainite in the literature [16, 17], bainite has higher density of HAGBs than martensite, which mostly originate from twin-related variant pairs. Combining the characteristics of grain boundaries, it may be noted that in 400 °C isothermal process, isothermal transformation produced bainite with high density of parallel-arranged HAGBs, while continuous cooling transformation at a cooling rate of 100 °C/s produced

Fig. 2 SEM micrographs and band contrast (BC) maps depicting boundary distribution: **a, e** quenching, **b, f** 400 °C, **c, g** 450 °C, **d, h** 500 °C (white line: $15^\circ > \theta > 5^\circ$, black line: $45^\circ > \theta > 15^\circ$, yellow line: $\theta > 45^\circ$)

Fig. 3 Dilatation versus temperature during the whole transformation process of isothermal process: **a** 400 °C, **b** 450 °C, **c** 500 °C

martensite with lower density of HAGBs and higher density of LAGBs.

The microstructure of 400 °C isothermal process is between the quenched sample and isothermally transformed microstructure at 450 °C. Although 450 and 500 °C isothermal transformation products are both bainite transformations, they showed different boundary characteristics because of two reasons: (a) the former has a lower phase transformation temperature and therefore has a greater driving force; (b) the former undergoes a small degree of martensite transformation, and then accelerates the bainite transformation. According to previous studies [31–34], prior martensite can provide more uniformly distributed nucleation sites for subsequent bainite transformation, and introduce dislocations in untransformed austenite region, further promoting the transformation.

The hardness decreased with the increase of isothermal temperature, and is presented in Table 1. The quenched martensite had a relatively higher hardness. As the isothermal temperature was increased, the content of lath bainite was increased and the hardness decreased. When the temperature was above M_s temperature, the transformation product tends to be granular bainite, and hardness was further decreased.

Table 1 Hardness of isothermal transformation

Sample	Quenching	Isothermal treatment at 400 °C	Isothermal treatment at 450 °C	Isothermal treatment at 500 °C
Hardness (HV)	325 (± 16)	305 (± 7)	285 (± 3)	273 (± 11)

Table 2 Average orientation relationships (ORs)

Sample	OR
Quenching	119.9°, 9.0°, 196.6°
400 °C	120.6°, 9.2°, 196.0°
450 °C	122.1°, 9.2°, 194.6°
500 °C	117.4°, 8.6°, 198.7°

The OR of variant microstructure respect to austenite was calculated by MATLAB regression program, and the average OR of the four samples is presented in Table 2. Based on these ORs, the variant pair content was calculated, as shown in Fig. 4. All the four samples showed that block boundaries are dominant. With the increase of isothermal

Fig. 4 Length fraction of inter-variant boundaries

temperature, the percentage of twin-related V_1/V_2 variant pair increased significantly and achieved the highest value of 58% at 450 °C. When the temperature was higher than the M_s temperature, the percentage of V_1/V_2 variant pair was significantly reduced. These characteristics showed that long and straight HAGBs are mainly from V_1/V_2 variant pairs. The above results provide the basis and conditions for us to study the transformation from austenite with Σ_3 boundary to martensite, lath bainite and granular bainite, respectively.

3.2 Identification of austenite Σ_3 boundary: martensite

Austenite forms a significant content of twin structure at high temperatures, and higher the austenitizing temperature or the larger the austenite grain size, the higher is the proportion [35, 36]. At room temperature, it is difficult to distinguish whether the austenite before phase transformation has

a twin boundary from the morphology of martensite/bainite. By using EBSD characterization results in combination with the corresponding pole figures, it can be determined whether the austenite before the phase transformation had a twin structure. There are two cases of variants formed near the austenite Σ_3 boundary: one is to maintain a near K–S OR with both the parent phase and neighboring austenite, and the other is only with the parent phase, which can be referred as having coinciding variants and unique variants, respectively.

If the austenite Σ_3 boundary can be identified, it will help us understand the influence of Σ_3 boundary on the variant microstructure. Austenite has four equivalent $\{111\}$ planes, austenite γ_1 and γ_2 are twinned with each other, a certain close-packed plane $\{111\}$ of γ_1 and γ_2 , respectively, are parallel, and has an angular axis relationship of $60^\circ/[111]$. Therefore, if a $\{110\}$ close-packed plane of variant which is transformed from the parallel $\{111\}$ plane of twin austenite, the variant maintains near K–S OR with both γ_1 and γ_2 . But if the $\{110\}$ close-packed plane of the variant is not parallel to the $\{111\}$ plane, the variant maintains near K–S OR only with γ_1 or γ_2 . In the two-dimensional EBSD characterization results, the variants in the same packet are approximately parallel to each other [37], as long as the variant is parallel to the coinciding variant, it remains near K–S OR with γ_1 and γ_2 , regardless of whether it is near the Σ_3 boundary. According to Morito et al. [15, 38, 39] and Cayron et al. [37], variants within a packet contain three habit planes: $\{575\}$, $\{755\}$ and $\{557\}$, and the lines of the different habit planes in the two-dimensional image exhibit small differences in angle. Therefore, the coinciding variants are approximately parallel, and the ones that are not parallel with coinciding variants are only in a near K–S OR with one of the austenite.

Based on the parallel relationship of variants of the same CP group and the non-parallel relationship of different CP groups, we can determine the position of twin boundaries of austenite that is transformed into martensite. Figure 5 shows

Fig. 5 Inverse pole figure (IPF) showing coinciding variants **a**, band contrast map **b**, corresponding experimental and theoretical pole figure **c** of Quenching sample (white line: $15^\circ > \theta > 5^\circ$, black line: $\theta > 15^\circ$)

the variant microstructure and corresponding pole figures of martensite transformed from twin austenite, and the blue broken line in Fig. 5b is the position of austenite Σ_3 boundary. The variants that maintain the near K–S OR with the two-part austenite are marked in red, and the rest are band contrast (BC) maps, as shown in Fig. 5b. The corresponding experimental and theoretical pole figure is shown in Fig. 5c. The orientation of γ_1 and γ_2 in Fig. 5 are $(230.0^\circ, 39.2^\circ, 7.3^\circ)$ and $(132.0^\circ, 24.8^\circ, 59.5^\circ)$, respectively, which are calculated by regression program. According to the orientation of austenite and actual OR, all the coinciding and unique variants can be determined. The variants in the red coincident regions are all identified as the same variants, but are different variants for γ_1 and γ_2 . The coinciding variants belong to a same packet (CP1), and the high content of LAGBs come from variant pairs of V_3/V_6 for γ_1 or V_2/V_5 for γ_2 (Fig. 5a), which are both equivalent variant pair of V_1/V_4 .

3.3 Identification of austenite Σ_3 boundary: bainite

Similar to identifying coinciding variants of martensite, the red regions in Fig. 6b represent the coinciding variants formed in the austenite with Σ_3 grain boundary of 450 °C isothermal transformation, and Fig. 6c is the corresponding pole figure. In Fig. 6a, the broken lines parallel to the length of the variant can simply be considered to represent the habit plane. The variants (V-I and V-II) have the same orientation, but they have two different habit planes, so it can be determined that they are coinciding variants. V-I is V_{23} in γ_1 and V_{22} in γ_2 , and V-II is V_{24} in γ_1 and V_{21} in γ_2 . Accordingly, the habit planes of γ_1 and γ_2 at this point are $\{57-5\}$ and $\{75-5\}$, respectively.

It is worth noting that the austenite Σ_3 boundary also has a blocking effect on other packet variants at 450 °C isothermal transformation, as shown in Fig. 6a, b. According to the hindering variants of other three packets, the position of Σ_3 boundary of the parent austenite can be determined, as

shown by the blue broken lines in Fig. 6b. Therefore, austenite Σ_3 boundary plays an important role in refining the size of the packet, and accordingly, the length of the block can be shortened, and the block boundary density can be increased.

For 500 °C isothermal transformation, the temperature is large enough to produce bainite with granular characteristic, and it is also easy to identify the austenite Σ_3 boundary, as shown in Fig. 7a–c. It is worth noting that the coinciding variants nucleating at the Σ_3 boundary tend to run through the entire prior austenite grain. The variant selection for transformation at higher temperature is likely to be enhanced by the formation of a large number of coinciding variants, which is discussed in detail.

4 Discussion

4.1 Analysis of martensite nucleation

As shown in Fig. 8, the product of 400 °C isothermal process had characteristics of quenching and 450 °C isothermal transformation. The boundary characteristics of area “A” indicated by two dashed square boxes in Fig. 8 show that the variants are arranged in parallel. The other part-area “B” shows a larger block with a large number of LAGBs. The OR of area “A” and “B” is $(122.2^\circ, 9.4^\circ, 194.4^\circ)$ and $(120.4^\circ, 8.9^\circ, 195.9^\circ)$, respectively, which is close to quenched and 450 °C isothermal transformation samples. Therefore, the main structures of area “A” and “B” are bainite and martensite, and the transformation products at 400 °C isothermal process are mixture of bainite and martensite. The phase transformation process is divided into two parts: (1) continuous cooling transformation from M_s temperature to 400 °C; (2) isothermal transformation at 400 °C. According to the widely used Koistinen–Marburger (K–M) model [40] and modified model [41] of martensite

Fig. 6 IPF figure showing coinciding variants **a**, band contrast map **b** and corresponding experimental and theoretical pole figure **c** of 450 °C isothermal transformation sample (white line: $15^\circ > \theta > 5^\circ$, black line: $\theta > 15^\circ$)

Fig. 7 IPF figure showing coinciding variants **a**, band contrast map **b** and corresponding experimental and theoretical pole figure **c** of 500 °C isothermal transformation sample (white line: $15^\circ > \theta > 5^\circ$, black line: $\theta > 15^\circ$)

Fig. 8 Band contrast map **a** showing coinciding variants and grain boundary distribution and corresponding experimental **b** and theoretical pole figure **c** of 400 °C isothermal process (white line: $15^\circ > \theta > 5^\circ$, black line: $\theta > 15^\circ$)

transformation, the volume fraction of martensite V_M can be described by Eqs. (1) and (2), respectively:

$$V_M = 1 - \exp[-\alpha_1 (M_s - T)]. \quad (1)$$

$$V_M = 1 - \exp[-\alpha_2 (M_s - T)^\beta]. \quad (2)$$

where T is the temperature in Kelvin, α_1 is a constant with value of 0.011, α_2 and β represent transformation rate parameter and exponent, respectively. Regardless of the traditional K–M or modified model, when the temperature is constant,

the value of $\frac{\partial V_M}{\partial T}$ is zero and the martensitic transformation stops. Martensite is formed during continuous cooling transformation, and bainite is formed during isothermal transformation. Since the nucleation mechanism of martensite transformation is different from that of bainite, martensite transformation can be nucleated with stacking faults and dislocations in addition to the grain boundaries. Classical heterogeneous nucleation theory [42–44] considers that during the cooling process, a martensite embryo is nucleated, which is a region with a diameter of several tens of nanometers, surrounded by dislocations. The growth of martensite embryo requires continuous cooling and corresponding higher driving force. For a thermal martensite, the activation energy required for its growth is substantially zero, and thus the growth rate is independent of temperature. However, the amount of martensite transformation is related to temperature. Unlike bainite nucleation theory at boundaries, it is the martensite nucleation that mainly revolves around dislocations, and the resulting variants show great differences with bainite. The austenite Σ_3 boundary does not serve as a preferential nucleation site for martensite transformation. If all the four packets have the identical opportunity for formation, the austenite Σ_3 boundary not only hinders the growth of most of the variants, but also favors the formation of coinciding variants, which is same as the bainite transformation. Therefore, the austenite Σ_3 boundary plays a role in refining the packet and decreasing the block size for martensitic transformation.

4.2 Bainite nucleation at austenite Σ_3 boundary

Abbasi et al. [25] found that austenite Σ_3 boundary can act as a nucleation site and enhances variant selection in the form of coinciding variants, which minimizes the ferrite–neighboring austenite interfacial energy. They proposed a model for nucleation at Σ_3 grain boundary, that is, a variant- V_i that maintains a K–S OR with the parent austenite is first grown on the side of the Σ_3 boundary, and then another one- V_j is formed on the other side. To reduce the driving force required for phase transformation as much as possible, V_i and V_j both hold K–S OR with the parent phase austenite. From the results of isothermal transformation, it seems that the austenite Σ_3 boundary becomes the nucleation site, but it must also be explained why some of the austenite Σ_3 boundary becomes a barrier to some unique variants, such as that shown in Figs. 5 and 6.

Nucleation of martensite occurs at dislocations and does not depend on grain boundaries. Therefore, the austenite Σ_3 grain boundaries mainly play a role in blocking the growth of packets and refine the packets. Thus, it is necessary to examine the characteristics of grain boundary nucleation for bainite transformation. From the results of 500 °C isothermal

transformation, the coinciding variants penetrate the entire austenite Σ_3 boundary, while in the austenite grains at 450 °C isothermal transformation, part of the austenite Σ_3 boundary is occupied by coinciding variants, and the other part still plays a role in hindering the growth of unique variants. This indicates that with the decrease of driving force for bainite transformation, the austenite Σ_3 boundary reduces the free energy required for phase transformation, which will cause the variant to preferentially appear at austenite Σ_3 boundary, and strengthen the variant selection.

However, because of coherent relationship between the crystal structure on both sides of the Σ_3 boundary, the interface energy of the crystal structure is very low [45, 46], so it cannot directly become a nucleation site. The single austenite Σ_3 boundary is not the most preferential nucleation site, and the conditions of preferential nucleation in the phase transformation need to be investigated. The nucleation of coinciding variant may be related to the junction of austenite Σ_3 boundary and the prior austenite grain boundary.

The classical heterogeneous nucleation theory suggests that nucleation at the grain boundary is related to the contact angle [47, 48]. The smaller the contact angle, the better is the nucleation. The driving force for nucleation can be described by Eqs. (3)–(5):

$$\Delta G = \Delta G^* \cdot f(\varphi). \quad (3)$$

$$\Delta G^* = \frac{4}{3} \pi r^{*3} \Delta G_l^V + 4 \pi r^{*3} \gamma. \quad (4)$$

$$f(\varphi) = \frac{1}{4} (2 + \cos \varphi) (1 - \cos \varphi)^2. \quad (5)$$

where ΔG is the energy required to form crystal embryo, φ is contact angle, ΔG^* is the critical driving force for forming core, ΔG_l^V is the driving force for per unit volume, r^* is critical radius of the core, and γ is the interface energy of old and new phase. When $0 \leq \theta \leq \pi$, $\frac{df(\varphi)}{d\varphi} \geq 0$ and $\frac{\partial \Delta G}{\partial \varphi} \geq 0$. As φ increases gradually, $f(\varphi)$ increases. The nucleation of the double interface, triangle edge and quadruple corner have dimensions of 1, 2 and 3, respectively. According to the difference in driving force for nucleation, the order of nucleation at grain boundaries is: quadruple corner > triangle edge > double interface. The most favorable nucleation sites are generally quadruple corners, which are three-line junctions in a two-dimensional image, as shown in Fig. 9a. Due to $\varphi_1 < \varphi_2 < \varphi_3$, the driving force required for the phase transformation is $\Delta G_{\varphi_1} < \Delta G_{\varphi_2} < \Delta G_{\varphi_3}$. It is worth noting that in the variants transformed from twin austenite shown in Fig. 9b, there are two nucleation sites: I and II. At the I nucleation point, θ_1 and θ_2 are close to 90°, respectively, and θ_3 and θ_4 at II nucleation point are close to 45° and 135°, respectively. The junction of Σ_3 boundary and the prior

Fig. 9 Schematic diagram **a**, experimental observation **b** of grain boundary nucleation for bainite transformation and schematic diagram of nucleation at the austenite Σ_3 boundary for 450 °C **c** and 500 °C **d** isothermal transformation

austenite grain boundary is not a complete quadruple corner, because the other side of the prior austenite grain boundary is too large ($\sim 180^\circ$). The austenite grain in the studied twin portion has half of the angular structure, which we call a semi-quadruple corner. At the junction, the γ_1 and γ_2

portions are quadruple corners, and γ_3 is a general triangle edge. In this situation, we can call it semi-quadruple corner. Figure 9c shows a schematic diagram of nucleation at austenite Σ_3 boundary. The first stage is nucleation at the junction of Σ_3 boundary and prior austenite grain boundary, or

nucleation at other sites. The difference is that the former forms coinciding variants, and the latter can form unique or coinciding variants. Subsequent phase transformation is based on the lowest nucleation barrier, as shown in Stages 2 and 3 of Fig. 9c. As the isothermal temperature increases, the fcc to bcc phase transformation driving force decreases, such that the reduction of the required driving force is more favorable, and Σ_3 grain boundary becomes the preferential nucleation sites, and the common variants along the Σ_3 boundary runs through the entire prior austenite grain.

5 Conclusions

The role of austenite Σ_3 boundary on variant nucleation in martensite and bainite transformation was studied, and the effect of transformation temperature on variant selection of bainite was also revealed. The results obtained are summarized as follows:

1. The crystallographic method was used to quantitatively calculate and visualize the isothermal transformation microstructure, indicating that the isothermal transformation slightly was lower than M_s temperature and is conducive to produce the highest content of V_1/V_2 variant pair and block boundaries.
2. The formed packet may be blocked by austenite Σ_3 boundary, which refines the packet size and decreases the block size. This effect also provides strengthened and toughened martensite.
3. The preferred nucleation site for bainite transformation was grain boundaries, especially the junction of Σ_3 boundary and the prior austenite grain boundary. The austenite Σ_3 boundary can play the role of strengthening the variant selection to reduce the energy for nucleation. For bainite transformation at low temperature, V_1/V_2 variant pairs are formed on both sides of the Σ_3 grain boundary, and the coinciding variants formed here have different habit planes. With the increase of transformation temperature, coinciding variants tends to run through the entire grain.

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