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► To cite this version:

| K. Tsuzaki, T. Maki. Some Aspects of Bainite Transformation in Fe-Based Alloys. Journal de Physique
| IV Proceedings, 1995, 05 (C8), pp.C8-61-C8-70. 10.1051/jp4:1995807 . jpa-00254055

HAL Id: jpa-00254055

<https://hal.science/jpa-00254055>

Submitted on 1 Jan 1995

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Some Aspects of Bainite Transformation in Fe-Based Alloys

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Abstract. The mechanism of the bainite transformation in carbon containing steels was discussed after presenting our recent experimental results; (1) the bainite transformation has its own C-curve, (2) incomplete phenomenon is a general feature of the bainite transformation, (3) bainitic ferrite grows to a limited size, and (4) bainitic ferrite is highly supersaturated with respect to carbon. It was examined what kinds of interphase boundary structures and carbon concentration gradients in the vicinity of the transformation interface are possible to explain the experimental results. It was concluded that the bainite transformation should be regarded as the transformation that occurs by the movement of a glissile interphase boundary and requires diffusion of interstitial carbon atoms at least to a certain extent in the nucleation and growth stages.

1. INTRODUCTION

There have been many discussions regarding the nature of the bainite transformation in carbon containing steels and its mechanism has not yet been established despite much work both experimentally and theoretically [1–5]. There are two main schools of thought, i.e., a displacive and a diffusional mechanism. The displacive school proposes that the transformation occurs diffusionlessly as far as the solvent and substitutional solute atoms are concerned. In this view, austenite–ferrite interphase boundaries must have the nature of martensite, a glissile interface. The opposite school claims that the bainite is a diffusional transformation product formed by the ledge mechanism where it has been postulated that the area of the interface (the risers or kinks on the risers of growth ledges) has an incoherent or disordered structure [6–8]. A question thus arises; how the product phase with such a disordered interface exhibits a surface relief effect [8]. However, the consideration for the interfacial structure in the ledge mechanism is now changing. Howe [9] proposed that continuity of atomic planes across the growing interface leads to an atomic site correspondence even in diffusional transformations, resulting in a surface relief effect. More recently, Furuhashi *et al.* [10] have found that the interphase boundaries enclosing the bcc laths precipitated within a fcc matrix grain in a Ni–Cr alloy are all semicoherent; they consist of glissile and sessile antioherency dislocations accommodating a misfit strain in addition to coherency (transformation) dislocations. We now hence need to discuss the nature of the bainite transformation in consideration of these new findings.

In this paper, before presenting our view for the bainite transformation, we will briefly present our recent experimental results of the bainite transformation in carbon containing steels. The topics are 1) relation between time–temperature–transformation (TTT) diagram and microstructures of isothermally transformed products, 2) incomplete transformation to bainite, and 3) growth and carbon concentration of bainitic ferrite. To avoid confusion, it is to be noted that we first simply use the term of "bainitic ferrite" for the lath- or plate-shaped ferritic component of the products formed in a temperature range below that of grain boundary allotriomorphic ferrite formation and/or pearlite formation with disordered interphase boundaries (diffusional–reconstructive transformation) and above that of the martensitic transformation (diffusionless–displacive transformation). We also simply use the term of "bainite transformation" for the decomposition process of austenite to produce the bainitic ferrite.

2. FORMATION TEMPERATURE OF BAINITIC FERRITE

The kinetic definition for the bainite transformation in steels has been discussed as one of three principle definitions [1,2]. This definition is based on the fact that the bainite transformation has its own C-curve between the pearlite or proeutectoid ferrite and the martensite ranges on a TTT diagram, and becomes incomplete as the decomposition temperature approaches to the upper limiting temperature of this curve (the kinetic-Bs). It is however well known that the C-curves of bainite and pearlite overlap extensively in plain carbon steels and the kinetic-Bs appears usually in particular steels containing an appreciable proportion of strong carbide formers such as Mo and Cr. Aaronson and his co-workers [1,2,5] thus claim that the kinetic-Bs appears to be the result of the solute drag-like effect [11] of the strong carbide formers on the proeutectoid ferrite reaction, and that the appearance of the kinetic-Bs does not mean a different transformation mechanism for the bainite reaction. It has been proposed that the substitutional elements may segregate to the transformation interfaces and then reduce the activity of carbon in austenite resulting in the low growth rate of proeutectoid ferrite [11].

According to the view based on the solute drag-like effect, one can not expect the discontinuous changes of microstructures of the products formed at temperatures near the kinetic-Bs or the bay region in a TTT diagram. It was however found in an Fe-2%Mn-1%Mo-0.15%C alloy [12] that ferrite morphology dramatically changes at the kinetic-Bs and that the transformation above the kinetic-Bs is initiated by the formation of grain boundary allotriomorphic ferrite (GBAF), while that below the kinetic-Bs starts by the formation of lath-shaped bainitic ferrite. Typical light microstructures of those are shown in Fig.1. The TTT diagram consisted of two C-curves which are distinguished by a deep bay region, as is shown in Fig. 2(d). The upper limiting temperature of the lower C-curve (i.e., the kinetic-Bs) was 830K. This is lower by approx. 90K than that in an Fe-1%Mo-0.2%C alloy [13], indicating that Mn effectively lowers the kinetic-Bs. In order to confirm that the bainite transformation is regarded as a separate reaction with its own C-curve, we observed the microstructural change during the isothermal holding at 823K (just below the kinetic-Bs). At 823K the formation of a small amount of bainitic ferrite was observed at 64.8ks holding. The transformation rate was however very low and the significant progress of bainitic ferrite formation was not observed at 259.2ks (3 days) holding. As is shown in Fig. 2(a), only bainitic ferrite is seen and its amount is small at this stage. By prolonged isothermal holding of 518.4ks (6 days), the formation of small GBAF becomes evident (Fig. 2(b) and (c)). The structure of GBAF is the same as that formed above the kinetic-Bs (Fig. 1(a)(b)). This result clearly indicates that GBAF begins to form also at temperatures below the kinetic-Bs when the reaction time reaches the time for GBAF formation, of which time can be expected by the extrapolating of the upper C-curve, as is schematically illustrated in Fig. 2(d).

The TTT diagrams with the kinetic-Bs have been also reported in 9%Ni low carbon steels without strong carbide formers [14-17]. Since Ni does not show the solute drag-like effect, the appearance of the kinetic-Bs in these steels strengthens the view that bainite reaction has its own C-curve. In a high-purity Fe-9%Ni-0.28%C alloy [17], two C-curves still somewhat overlapped although the kinetic-Bs of 760K was distinguishable. Comprehensive microstructural observations showed clearly that the upper C-curves corresponds to the formation of GBAF and/or degenerate pearlite (DP) while the lower C-curves to the formation of lath-shaped ferrite (Fig. 3) [17]. The laths appear to nucleate on austenite grain boundaries, and at temperatures just below Bs, where individual laths are observed, the amount of carbide is very small (Fig.3(b)). As the reaction temperature is lowered, laths increasingly form adjacent to one another in parallel groups and the precipitation of carbide becomes significant (Fig.3(c)). This microstructural change with the reaction temperature was also found in other Fe-9%Ni-C alloys [17].

These results deduce the conclusion that the appearance of two C-curves is not a special phenomenon in particular steels containing strong carbide formers but generally observed in the cases where the transformation rates are significantly different between the proeutectoid ferrite or pearlite reaction and the bainite reaction at intermediate temperatures, and that the lower C-curve corresponds to the formation of lath-shaped ferrite. We consider that the solute drag-like effect suppresses the growth of GBAF with the movement of incoherent interphase boundaries where substitutional atoms can segregate but has no influence on the growth of bainitic ferrite with the movement of semicoherent interfaces, resulting in the distinguishable two C-curves of the steels with strong carbide formers.

Figure 1: Light microstructures in the early stage of transformation at temperatures near the kinetic-Bs (830K) in an Fe-2%Mn-1%Mo-0.15%C alloy [12]. (a) 873K 129.6ks (1.5d), (b) 843K 259.2ks (3d), and (c) 803K 28.8ks (8h).

Figure 2: Microstructural change during the isothermal holding at 823K (just below the kinetics-Bs) in an Fe-2%Mn-1%Mo-0.15%C alloy [12]. (a) 259.2ks(3d), (b) and (c) 518.4ks(6d), and (d) schematic drawing of TTT diagram. (c) is a high magnification photo. of (b). (AF: allotriomorphic ferrite, BF: bainitic ferrite)

Figure 3: Light microstructures in the early stage of transformation at temperatures near the kinetics-Bs (760K) in an Fe-9%Ni-0.28%C alloy [17]. (a) 780K 16.2ks, (b) 753K 420s, and (c) 688K 420s. (DP: degenerate pearlite)

3. INCOMPLETE TRANSFORMATION TO BAINITE

One of other important kinetic features of bainite reaction is the incomplete transformation; the transformation ceases before the volume fraction of ferrite reaches the value expected from the lever rule. Since the incomplete transformation is typically observed in some alloy steels with the strong carbide formers of Mo and Cr, Aaronson and his co-worker have been also trying to account for this phenomenon by the solute drag-like effect. According to this view, the incomplete transformation, or the transformation stasis they call, should appear at temperatures not only below the kinetic-Bs but also above it. However, Shiflet and Aaronson [18] have found no evidence for incomplete transformation above the kinetic-Bs in various Fe-Mo-C alloys. The absence of the incomplete transformation has been also found in the Fe-2%Mn-1%Mo-0.15%C alloy whose TTT diagram was shown in Fig.2 [12]. The incomplete transformation to ferrite in this alloy was outstanding at temperatures between 813K and 773K (below the kinetic-Bs; 830K). The absence of the incomplete transformation at temperatures above the kinetic-Bs weakens the view based on the idea of the solute drag-like effect.

In an alternative view, the incomplete transformation is considered as a general feature of the bainite reaction and is associated with the limited nucleation and growth of bainitic ferrite like the case of martensitic transformation [19]. The absence of the incomplete transformation in some steels such as Fe-C alloys and low-alloy steels is attributed to the masking of the incomplete portion of the bainite transformation by concurrent proeutectoid ferrite or pearlite formation. According to this view, the incomplete transformation must appear in all the cases where the concurrent reactions are suppressed. To clarify the generality of the incomplete transformation to bainite, we [20] have examined the isothermal transformation of an Fe-11%Ni-0.2%C alloy where the solute drag-like effect can not be expected. We needed 11%Ni to avoid the masking by the formation of GBAF or degenerate pearlite. The incomplete transformation was surely observed at temperatures between 723K and 693K below the kinetic-Bs of 730K, as is shown in Fig.4.

4. GROWTH AND CARBON CONCENTRATION OF BAINITIC FERRITE

For clarifying the mechanism of the bainite transformation, it is of great importance to examine the thickening process and the carbon concentration of bainitic ferrite. The measurements of those are however difficult for the bainite transformation of plain carbon steels because of the rapid precipitation of carbides and the absence of untransformed austenite within a bainite packet or sheaf. These problems can be settled by the addition of Si which suppresses the precipitation of carbide. We thus examined the isothermal bainite transformation in a high-purity Fe-2%Si-0.6%C alloy [21].

Figure 5 shows the fraction of constituents at room temperature in the specimens reacted at 723K for various periods of time followed by water quenching. The isothermal transformation to bainite started within several seconds and finished by 1ks. The amount of retained austenite became more than 10 vol.% at 50s holding. The microstructural change at 723K is schematically shown in Fig. 6. In the early stages of transformation (Fig.6(a)), the bainitic ferrite (BF1) was carbide free. The laths of bainitic ferrite within a packet were parallel to one another and separated by carbon enriched austenite (A). BF1 in a packet has the same crystallographic orientation (variant). In the later stage of transformation (Fig.6(b)), large carbide plates (CB) formed in the austenite between the laths (BF1), resulting in the decrease in the carbon concentration of the austenite. Subsequently, the ferrite (BF2) with a variant different from BF1 was decomposed for the completion of transformation (Fig.6(c)). No pearlite was observed at 723K.

Figure 7 shows the distribution of the lath widths of the primary ferrite (BF1) for different isothermal holding times at 723K. The primary ferrite could be easily identified even in the final stage of transformation, because it does not contain large carbide plates and has a crystallographic variant different from the secondary ferrite (BF2). The distribution hardly changes with the reaction time, and the average value of lath width is a constant of 0.3 μ m.

Figure 8 shows the representative TEM microstructure at the early stage of transformation (723K, 50s). In Fig.8(a), two bainitic packets are seen and a large black region marked M is the martensite formed during the quenching. Both the bainitic ferrite and retained austenite involve a large number of

Figure 4: Isothermal transformation curves in an Fe-11%Ni-0.2%C alloy where the incomplete transformation is observed at temperatures between 723K and 693K [20].

Figure 5: Variation of fractions of constituents at room temperature with isothermal holding time at 723K in an Fe-2%Si-0.6%C alloy [21].

Figure 6: Schematic illustration showing the microstructural change during the isothermal transformation at 723K in an Fe-2%Si-0.6%C alloy [21]. (a) early stage, (b) middle stage, and (c) final stage of the transformation. (CB: carbides)

Figure 7: Distribution of the lath widths of the primary ferrite for different isothermal holding times at 723K in an Fe-2%Si-0.6%C alloy [21].

Figure 8: Representative TEM microstructure at the early stage of transformation (723K 50s) in an Fe-2%Si-0.6%C alloy [21]. (M: martensite, A: austenite)

dislocations (Fig.8(b)). There are no visual carbide precipitates at this stage. The retained austenite existed only between the primary ferrite laths within bainitic packets in the specimens quenched after 50s holding at 723K, where the carbon concentration of it was 1.61 mass% at the highest in several samples. The carbon concentration of austenite in paraequilibrium with respect to ferrite is 3.59 mass% at 723K in this alloy. It is thus deduced that the primary ferrite can not thicken during the subsequent isothermal holding (Fig.7), even though the "average" carbon concentration of the untransformed austenite within bainitic packets at 50s holding had not reached the paraequilibrium value.

Since in the specimens held for 50s at 723K carbides were not observed and the microstructure consisted of primary ferrite, martensite, and retained austenite, one can thus estimate the average carbon concentration of the primary ferrite from the mass balance of carbon (the lever rule) with appropriate assumptions. We obtained 0.19mass% as the lowest value from the highest value of average carbon concentration in retained austenite, 1.61 mass%. The carbon concentration of ferrite in equilibrium with austenite at 723K is 0.02mass%. This estimated value of 0.19mass% thus indicates that the bainitic ferrite is highly supersaturated with respect to carbon.

The bainitic ferrite supersaturated with respect to carbon is also presumed from the microstructures of an Fe-3%Ni-0.2%C alloy [22]. The bainitic ferrite laths form in contact each other and contain a large number of carbides (Fig.9(a)). A large number of carbides exist within a lath and there are at least three habit planes of the carbides. Figure 9(b) is a micrograph showing the bainitic ferrite laths formed separately to one another, micrograph which was taken from the same sample of Fig.9(a). The number of carbides within the laths is small and no carbides are seen in the interphase boundaries and the surrounded austenite which transformed to martensite by the quenching. This indicates that a large number of carbides observed in Fig. 9(a) is not the result of engulfment by the ferrite of carbides originally precipitated in the austenite. It is rather likely that the bainitic ferrite first forms without carbides and subsequently the precipitation of carbides occurs directly from the bainitic ferrite.

The results in this chapter support the view that the bainitic ferrite grows to a limited size and is at least partially supersaturated with respect to carbon; these are contrary to the view that the bainitic ferrite develops by the carbon diffusion-controlled mechanism with a local equilibrium condition.

Figure 9: TEM microstructures of bainitic ferrite formed at 765K 15s in an Fe-3%Ni-0.2%C alloy [22]. (a) multi-variant carbides within a ferrite lath, and (b) a few carbides in the laths formed separately to one another.

5. POSSIBLE MECHANISMS OF BAINITE TRANSFORMATION

In the nucleation stage of bainitic ferrite, there seem to be no controversy for the occurrence of carbon partitioning. The displacive school also considers that the carbon concentration of nuclei of bainitic ferrite is lower than the average value of austenite [3,4,17,23,24]. In the nucleation with carbon partitioning the maximum possible free energy change is given by ΔG_m , as illustrated in Fig.10.

To discuss the growth mechanism of bainitic ferrite, we should examine the possible interfacial structures and carbon concentration gradients for the lath or plate-shaped ferrite formed within an austenite grain, since these features define the transformation mechanism. According to the recent studies on interfacial structures [9,10] described in the Introduction, the particles formed within a matrix grain do not have incoherent interphase boundaries. One can thus consider two kinds of semicoherent interphase boundaries for the ferrite; one is a sessile interphase boundary consisting of

sessile anticoherency dislocations in addition to glissile ones (Fig. 11(a)), and the other is a glissile interphase boundary only with glissile anticoherency dislocations (Fig. 11(b)). The two semicoherent boundaries of course consist of coherency (transformation) dislocations in addition to anticoherency dislocations, although the coherency dislocations are not depicted in Fig. 11. It is apparent that atomic migration of substitutional elements, climb of sessile misfit dislocations, is necessary for movement of the former boundary but not for that of the latter one. There would be a fully-coherent boundary only with coherency dislocations as a possible interface when ferrite particles are very small and we deal it as the extreme case of glissile boundaries. Three possible carbon concentration gradients near the interphase boundary in Fe-C binary alloys are drawn in Fig. 12; (a) local equilibrium, (b) partial carbon supersaturation in ferrite, i.e., solute trapping, and (c) full carbon supersaturation, i.e., diffusionless.

First, we examine which carbon profiles are possible for the ferrite with sessile interphase boundaries. Taking into consideration the fact that the diffusivity of interstitial carbon atoms is much higher than that of iron atoms, nonequilibrium carbon concentrations at the interphase boundary are not realistic. During climbing of sessile anticoherency dislocations with a relatively slow rate, fast diffusion of carbon atoms makes the carbon concentrations the equilibrium values of the ferrite and austenite phases at a given temperature. The equilibrium carbon concentrations of the two phases may change from those for the stress-free equilibrium condition when the mobility of interphase boundaries is low compared with carbon diffusivity. In this case the equilibrium carbon concentrations of both the ferrite and austenite become smaller. This decrease of the equilibrium carbon concentrations also occurs when a capillarity effect at a tip of ferrite particles and/or a strain energy effect due to transformation strain is significant. In all the cases of the transformations with a local equilibrium condition with respect to carbon, the carbon concentration of ferrite is lower than that for the stress-free equilibrium condition at a given temperature.

Figure 10: Schematic illustration showing the maximum possible free energy change in the nucleation with carbon partitioning.

Figure 11: Semicoherent interphase boundaries possible for the ferrite particles formed within an austenite grain.

Figure 12: Three possible carbon concentration gradients near the transformation interface in Fe-C binary alloys.

This discussion leads us to the conclusion that sessile interphase boundaries are not the case for the bainitic ferrite described in the chapter 4, because at least partially supersaturated ferrite with respect to carbon was observed. The ferrite with sessile interphase boundaries is also inconsistent with the incomplete transformation to bainite. Even in the case that the transformation is controlled by long-range carbon diffusion in austenite with a local equilibrium condition, the incomplete phenomenon could be expected if there is a significant strain energy effect caused by the formation of ferrite particles, as was pointed by Enomoto [25]. However, a strain energy effect must be small for the ferrite with sessile interphase boundaries because atomic migration of substitutional elements (iron and solute atoms) necessary for dislocation climbing also reduces strain energy by their reconstructive movement.

On the other hand, in the cases of the ferrite laths or plates with glissile interphase boundaries, all of the three carbon concentration gradients in Fig. 12 are possible to occur. When the driving force is small (corresponding to small undercooling), the movement of the interphase boundaries is slow even though their mobility is high and hence the transformation is controlled by long-range carbon diffusion in austenite under a local equilibrium condition. In the case that the driving force becomes large but not enough for the diffusionless transformation, i.e., the driving force without carbon partitioning is not large enough to overcome the sum of surface energy and strain energy, partial carbon supersaturation in ferrite (Fig.12(b)) may occur because the carbon partitioning increases the driving force for interface movement. It is easily considered that less carbon partitioning is needed as the driving force increases by larger undercooling and finally diffusionless transformation occurs at low temperatures (Fig.12(c)).

Ferrite particle growth with a partial carbon supersaturation is of course unstable growth and its practical occurrence is still under discussions [23,26]. It is however theoretically possible to occur and Olson *et al.*[24] have calculated the carbon concentrations of a ferrite plate which is lengthening in an austenite grain in an Fe-0.4%C alloy. The supersaturation of carbon in ferrite is larger for the growth stage than the nucleation stage, and hence diffusionless growth becomes possible to occur even at temperatures higher than M_s (Fig.13). This is because the sum of elastic strain energy and surface energy per unit volume of a ferrite particle becomes smaller as the ferrite particle grows. The experimental result that the bainitic ferrite is significantly supersaturated with respect to carbon is in good agreement with their model. Their model also well rationalizes the microstructural change with decrease in the reaction temperature; the bainitic ferrite is carbide free in the initial stage of transformation at temperatures just below the kinetic- B_s and the carbide formation becomes significant as the reaction temperature decreases (Fig.3).

Once the glissile interphase boundaries are accepted for the bainitic ferrite, the lattice deformations are the same in the martensite and bainite transformations. The incomplete transformation to bainite should be hence discussed based on the fact that the martensitic transformation is incomplete between M_s and M_f temperatures. There are two reasons for the cease of martensitic growth at a given temperature; one is the increase of elastic strain energy by thickening of a particle and the other is the loss of coherency of the interface boundary by interaction of the interface with accommodation dislocations in austenite [27]. Since plastic accommodation in austenite extensively occurs in the bainite transformation, the latter reason is likely for the limited growth of bainitic ferrite. In fact, Rigsbee and Aaronson [28] found sessile dislocations on the ferrite laths formed by the isothermal holding at 723K to 748K in an Fe-2%Si-0.62%C alloy. In such a case, the further growth of the ferrite laths needs the climb of the sessile dislocations, and hence it is not possible when the reaction temperature is low for migration of iron atoms. Moreover, the limited nucleation of martensite is now well established; the number of operational nucleation sites increases as the driving force increases [27]. We thus conclude that the incomplete transformation is a general feature of displacive transformations with glissile interphase boundaries.

Figure 13: Plot of calculated normalized supersaturation ($X_\alpha/\bar{X}$) of ferrite vs the isothermal transformation temperature for an Fe-0.4%C alloy. (After Olson *et al.* [24])

Figure 14 and Table 1 are the summary of our view for the γ - α isothermal transformations in carbon containing steels. In the transformations corresponding to the upper C-curve, the decomposition of austenite is controlled by long-range carbon diffusion in austenite and requires diffusional migration of iron and substitutional solute atoms for climbing of sessile antioherency dislocations. GBAF and pearlite with incoherent (disordered) interphase boundaries correspond to the transformation products for the upper C-curve. Lath or plate-shaped ferrite with semicoherent sessile boundaries can also form in the temperature range of the upper C-curve, and it may correspond to Widmanstatten proeutectoid ferrite forming at high temperatures. Widmanstatten ferrite can exhibit a surface relief effect because it has a semicoherent interface, as Howe [9] and Furuha *et al.* [10] have pointed out. In the transformation for the lower-C curve, long-range carbon diffusion in austenite may also occur to assist the interface movement but atomic migration of substitutional elements does not. Since the interphase boundaries have high mobility, the transformation can occur as far as sufficient diffusivity of carbon is held. We conclude that the bainite transformation should be regarded as the transformation that occurs by the movement of glissile interphase boundaries and requires diffusion of interstitial carbon atoms at least to a certain extent in the nucleation and growth stages. The bainite transformation can be thus distinguished by the interfacial structure, semicoherent glissile interfaces, from the transformations with disordered or semicoherent sessile interfaces for the upper C-curve.

According to the idea that the bainite transformation needs the carbon partitioning which assists the movement of a glissile interface, it is likely that the temperature range of the bainite transformation becomes smaller as the carbon content of a given steel decreases and the bainite transformation does not occur in a pure iron or alloy steels without interstitial elements. It is to be noted that after the halt of displacive growth of the bainitic ferrite by the loss of coherency, the reconstructive growth with the

Figure 14: Relationship between the C-curves and the occurrence of carbon diffusion and iron atom migration.

Table 1: Classification of γ - α transformations in carbon containing steels according to interfacial structures and the diffusion of atoms.

$\gamma - \alpha$ transformations in carbon containing steels			
Glissile interface		Sessile interface	
Semicoherent interface		Semicoherent interface	Incoherent interface
Martensite	Bainite	Widmanstatten ferrite	Grain boundary allotriomorphic ferrite + (Pearlite)
Presence of a specific orientation relationship			Absence of a specific O.R.
No diffusional migration of iron and substitutional atoms		diffusional migration of iron and substitutional atoms	
No partitioning of carbon	Partitioning of carbon		

climb of sessile dislocations, which were introduced by the interaction of a glissile interphase boundary with accommodation dislocations in the austenite, may follow the displacive growth when the reaction temperature is high enough for atomic migration of substitutional elements. This situation can be expected in very low-carbon low-alloy steels with high kinetic-Bs and makes the transformation mechanism obscure. The distinctive feature of the bainite transformation as a displacive transformation is obtained at temperatures where atomic migration of substitutional elements is negligible.

6. SUMMARY

We demonstrated that the bainite transformation in carbon containing steels should be characterized by the interfacial structure rather than by partitioning of interstitial carbon atoms, and concluded that the bainite transformation is a displacive transformation accompanying carbon diffusion at least to a certain extent in both the nucleation and growth stages. We should finally point out that the study on interfacial structures of the bainitic ferrite is extremely limited, nothing except for [28]. In order to clarify the present model we must confirm experimentally that the bainitic ferrite in the early stages of transformation has semicoherent interphase boundaries without sessile dislocations.

Acknowledgments: The authors would like to express their sincere thanks to Dr. T. Furuhashi (Kyoto University) for valuable comments and stimulating discussions.

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