Roles of transformation interfaces in the design of advanced high strength steels

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Abstract. Alloy partitioning and segregation or precipitation at transformation interfaces have been important in designing the mechanical properties of final products. Some representative examples in high-strength low-alloy steels are briefly discussed, with an emphasis on the transformation interface. For achieving better strengthening-ductility balance in TRIP-aided low alloy steels, it is essential to control carbon enrichment in austenite during ferrite and bainite transformations and the stability of retained austenite. Recent studies on alloy partitioning including carbon enrichment into austenite, provides a deeper insight into transformation mechanisms. We show that energy dissipation during interface migration and strain energy accumulation are important factors to control carbon enrichment into untransformed austenite during the ferrite/bainite transformation. As a counterpart of TRIP steels, interphase precipitation of alloy carbides during ferrite transformation leads to high yield strength and good elongation in low-carbon ferritic steels. A key for further strengthening is clarification of the conditions for refining the carbide dispersion. It is shown that incoherent ferrite/austenite interfaces are superior nucleation sites for carbides.

1. Introduction
A variety of phase transformation products, i.e., ferrite, pearlite, bainite and martensite, in low-alloy steels containing carbon, are used in steel products. In development of modern high strength steels, the importance of strength and the ductility/toughness balance is increasing and more sophisticated control of the microstructure formed during phase transformations is utilized. In order to control the microstructure and properties of steels, it is inevitable to use alloying elements in Fe-C alloys.

In the recent development of low-carbon high strength sheet steels, two distinctive categories have been extensively investigated [1]. One is multiphase steels, represented by dual-phase and TRIP steels produced by hot-rolling and subsequent interrupted ferrite/bainite transformation followed by rapid cooling or by intercritical annealing of cold rolled sheet accompanying reverse transformation to austenite and subsequent forward transformation. During forward or reverse transformation in the (ferrite+austenite) dual-phase condition, fast diffusing carbon enriches rapidly the austenite. After final cooling, carbon-rich martensite (for DP steel) or metastable high-carbon retained austenite (for TRIP steel) is obtained. For TRIP steels, the mechanical properties of the carbon-enriched second phase (retained austenite) are significantly influenced by the amount of carbon in solution [2, 3]. In addition, slow diffusing substitutional alloying elements may be also partitioned between the two phase in
complex manners. Recent systematic efforts studying on alloy partitioning during ferrite or bainite transformation reveal the importance of transformation interface [4]. Also substantial efforts are being made to investigate the partitioning behavior of carbon and other elements in Q&P steels for exploring higher strength and ductility balance [5,6].

Another is ferritic steels with nanoscale precipitation. In this case, ferrite transformation occurs during cooling after hot rolling. Addition of microalloying elements such as Ti, Nb and V, results in alloy carbide precipitation of which nucleation is promoted at growth interface of ferrite. This phenomena denoted as ‘interphase precipitation’ was extensively studied in 70’s as summarised by Honeycombe [7,8]. Recently, a group of microalloyed low-carbon steels is commercialized for automobile application as high strength, high ductility sheet steels produced by hot rolling and cooling [9]. It was shown that combined addition of alloying elements was effective for strengthening due to refined dispersion of carbide. This kind of steels shows superior ductility in terms of local elongation whereas high strength multiphase steels exhibit superior uniform elongation achieved by large work hardening.

It is obvious that the natures of transformation interface have direct impact on the behaviors of the alloy partitioning and interphase precipitation and resultant mechanical properties. We have been investigating on the roles of crystallography and interfacial structure of transformation products on microstructure development in steels [10]. In this paper, some representative examples in high-strength low-alloy steels will be discussed in this context.

2. Non-equilibrium alloy partitioning during transformation in low-alloy steels

2.1. Partitioning during ferrite/bainite transformation

Proeutectoid ferrite shows a variety of morphologies [11, 12]. Grain boundary ferrite (GBF), which is formed under low supersaturation, shows a characteristic morphology described as allotriomorphs elongated along austenite grain boundaries with smoothly curved interfaces. It has a near Kurdjumov-Sachs orientation relationship (K-S OR) with respect to one of the adjacent austenite grains and tends to grow toward the another austenite grain by motion of an ‘incoherent’ interface due to a non K-S and irrational OR. As temperature decreases, plate- or needle-shaped Widmanstätten ferrite (WF) is formed at austenite grain boundaries or on top of pre-existing allotriomorph and grows into the austenite grains by holding a more rigid OR and with partly coherent interfaces. As the transformation temperature further decreases typically below 600°C, bainitic ferrite (BF) with high interfacial coherency, similar to WF, starts to form both at austenite grain boundaries but also by more frequent intragranular nucleation.

The growth of GBF in Fe-C alloys is controlled by volume diffusion of carbon in austenite. In the Fe-C system, ferrite growth with incoherent boundary is well explained by local equilibrium in a decarburization study [13]. The thickening of GBF is also C diffusion-controlled but is slower than the expected from calculation assuming the migration of incoherent boundaries. The reason for this was considered to be effects of the boundary structures mixed by partly coherent and incoherent ones [14].

In alloyed steel, GBF grows more slowly than the Fe-C case [15] with carbon partitioning. Alloy partitioning accompanied results where variations of phase stability and transformation kinetics are more complex. There are three distinctive local equilibrium growth modes in Fe-C-M (M: substitutional alloying element) system. These are para-equilibrium (PE), negligible-partitioning local equilibrium (NPLE) and partitioning local equilibrium (PLE). PE and NPLE are essentially C diffusion-controlled modes without macroscopic partitioning of M. The former is a local equilibrium in the (Fe, M)-C pseudo-binary system whereas the latter satisfies local equilibrium at the interface for all elements with a sharp spike of M near the interface. Recent efforts in decarburized experiments in the ternary systems reveal that the formation limit of ferrite in the Fe-Mn-C and Fe-Ni-C systems is below the PE lines and close to the NPLE/PLE limit [4]. In the beginning of transformation, para-equilibrium (PE) growth seems to occur and make a transition to NPLE growth mode and accordingly a model to treat the PE-NPLE/PLE transition has been proposed [16].

It has been recognized that there are various ‘frictions’ on the interphase boundary migration exerted from various components of the microstructure as represented in figure 1. Atomic displacement at the
interface causes intrinsic friction. Solute atoms segregated at the interface provide another dissipation known as 'solute drag'. Dislocations or fine second-phase particles may add further barriers for interface migration.

Hillert and his co-workers discussed the energy dissipation by such friction extensively [17]. Figure 2 shows the effect of the energy dissipation in boundary migration on partitioning of carbon in the Fe-C binary system. The same situation can be imagined for the (Fe, M)-C pseudo-binary case. (a) is for forward transformation to ferrite whereas (b) represents the case of reverse transformation to austenite.

Figure 1. Schematic illustration of microstructure components causing energy dissipation in interface migration.

Figure 2. Schematic diagrams of deviation of carbon partitioning from local equilibrium due to energy dissipation at a migrating interface; (a) austenite $\rightarrow$ ferrite forward transformation and (b) ferrite $\rightarrow$ austenite reverse transformation to austenite. White circles: carbon contents at local equilibrium, gray squares: carbon contents achieved with energy dissipation, Black circle: initial matrix composition.
Figure 3. FE-EPMA measurement of carbon concentration at ferrite/austenite interface in Fe-0.15C-2Mn (mass%) alloy [18].

Local equilibrium at the interface is achieved at carbon concentrations represented by white circles given by drawing the common tangent to free energy curves of ferrite and austenite phases. At this condition, there is no force exerted at the interphase boundary. When some friction accompanies boundary migration, an extra driving force to make the boundary move forward is necessary and a deviation from local equilibrium occurs, described as shift of the boundary concentration from the open circles to squares in the figure. Under partial equilibrium of carbon as a fast diffuser, the extra driving force or the energy dissipation by boundary migration is given by $\Delta G_{\text{mig}}$. The total driving force for transformation, $\Delta G$, is the sum of $\Delta G_{\text{mig}}$ and $\Delta G_{\text{diff}}$ which is consumed by long-range carbon diffusion. As seen in figure 2(a), such energy dissipation leads to lower carbon content in austenite for forward transformation and higher carbon content in austenite for reverse transformation (figure 2(b)). Furthermore, additional partitioning of substitutional alloying elements gives more complex effects on the stability of austenite.

By measuring the chemical composition, in particular the carbon content across the interface, it is possible to estimate the amount of such dissipation. We recently developed quantitative measurement method using field-emission electron probe micro-analysis [18]. Figure 3 shows an example applied to GBF in Fe-0.15C-2Mn. After we constructed calibration line (figure 3(a)) using as-quenched martensite specimens with a variety of carbon contents, measurement was made. In the analysis of a ferrite in (b), a clear concentration profile is obtained along a linescan with a spatial resolution of ca. 0.5µm in (c).

Figure 4 summarizes calculated phase boundaries and growth limits with measured carbon content in austenite at the interface in various Fe-C-Mn alloys [10]. Increase of Mn content stabilizes austenite and also shifts the NPLE limit toward the lower temperature region. The NPLE limit becomes almost coincident with the T₀ line, which is the growth limit for composition-invariant diffusional growth in massive transformation and diffusionless growth in martensitic transformation, at 2mass%Mn and finally surpasses the T₀ line at 3mass%Mn. The growth of GBF seems to correspond well to the NPLE mode for all the Mn-contained alloys, except at one temperature in the 1.5Mn alloy. On the other hand, it is interesting to see that the growth of WF at 923K deviates from NPLE in the 0.75Mn-0.1C alloy whereas GBF in the 1.5Mn-0.15C alloy grows in the NPLE mode at the same temperature. This difference might be affected by interfacial coherency. As the temperature is lowered, carbon enrichment at austenite/GBF deviates to a lower side than the NPLE case. In 2mass%Mn, those two limits both can explain the ferrite growth successfully. However, in 3mass%Mn alloy, carbon enrichment exceeds the NPLE limit and is coincident with $T'$ for undercooling of 400 J/mol from $T$, [19].

The solute drag effect [20] is another mechanism causing energy dissipation when alloying with substitutional atoms. It is considered that Mn has some tendency to segregate at the ferrite/austenite boundary and provides a dragging force. On the other hand, Si is thought to have weak interaction to ferrite/austenite boundaries. Thus, Fe-C-Si is a good candidate system to study dissipation mainly due to interfacial coherency.

Figure 5(a) shows measured carbon contents in austenite at the interface in a Fe-0.4C-3Si alloy [21, 22]. Similar to the Fe-C-Mn cases, carbon contents in austenite for GBF correspond well to the NPLE
whereas those for WF are lower than this. For BF, the deviation is much larger than for GBF and WF but carbon enrichment is still higher than the \( T_0 \) line. The estimated energy dissipation from para-equilibrium (PE) exceeds over 300J/mol for WF and 1500J/mol for BF (figure 5(b)). This result clearly indicates that interface coherency causes the friction in boundary migration.

As already stated, both WF and BF hold a near K-S OR so that they should grow by migration of partially coherent boundaries. Figure 6 shows that a periodic array of lattice dislocations exist at the BF/austenite boundary in a Fe-C-Si-Mn alloy [10]. Those dislocations at the boundary create a pinning force. Also, it is seen that plastic accommodation takes place during bainite transformation. Strong interaction between the boundary and dislocations in austenite exerts another dissipation in boundary migration. Furthermore, elastic strain energy associated with the transformation needs to be accommodated. Dislocation slip can accommodate shear strain but diffusion must play a major role in
accommodation of volumetric strain. It is concluded that bainite transformation accompanies a larger energy dissipation caused by the difficulty of strain accommodation for both components at lower transformation temperatures.

2.2. Effects of solute drag in energy dissipation during ferrite/bainite transformation

Addition of carbide-forming elements often retards the ferrite transformation. It has been proposed that segregation of alloying elements with strong affinity for carbon, such as Mo, at ferrite/austenite boundaries causes incomplete transformation [23]. Also, Nb is considered to be effective to exert a solute drag effect on the ferrite transformation [24]. However, the effects of solute drag are still controversial for bainite transformation since less potency of elemental segregation is expected and also reported [25, 26]. We have recently examined the energy dissipation for ferrite and bainite transformation in an Fe-C-Mo alloy [27]. Figures 7(a) and (b) show typical Mo and C concentration profiles across the interface between grain boundary allotriomorphic ferrite (AF) and austenite (γ) (transformed to martensite (M) by quenching). Extensive Mo segregation is seen at the non K-S interface in (a) but segregation at the near K-S interface is limited in (b), which indicates that Mo segregation is difficult at partly coherent interfaces. Figure 7(c) summarizes relations between the interfacial excess

![Figure 6](image1.png)

**Figure 6.** (a) HREM image of the BF/austenite(A) boundary with interfacial dislocations. (b) TEM image of BF and austenite with high densities of dislocations in Fe-0.6C-2Si-1Mn [10].

![Figure 7](image2.png)

**Figure 7.** Ferrite orientation maps and Mo and C profiles across grain boundary allotriomorphic ferrite (AF)/austenite (γ) interface obtained by 3DAP analysis in the Fe-0.4C-0.5Mo (mass%) alloy transformed at 700 °C. (a) and (b) were taken from a non K-S interface and a near K-S interface, respectively. (c) Relation between amount of Mo segregation and energy dissipation at AF or bainitic ferrite (BF) / γ interfaces [27].
of Mo atoms and energy dissipation at the AF/γ interface with and without a K-S OR and the bainitic ferrite (BF)/γ interface at 650 °C. Mo segregation increases at longer holding time and energy dissipation becomes small mainly due to decrease in growth rate. It is noted that the amount of Mo segregation at BF/γ interfaces is much lower than that at AF/γ interface without a K-S OR. On the other hand, BF experiences much larger dissipation even at the same (but small) amount of Mo segregation as near K-S AF/γ interfaces. This indicates that the solute drag effect is small for bainite transformation, as described in the previous sub-section.

2.3. Austenite reversion behavior dictated by crystallography

The interfacial coherency also plays an important role during reverse transformation from ambient temperature microstructure to austenite. During reverse transformation, it is well known that two major morphologies of reverted austenite are formed: granular (γG) and acicular (γA) austenite [28]. Growth of γG takes place by migration of incoherent interface but γA grows with partly coherent interface [29]. Since the interfacial mobility is different between these two kinds of interface, as in the case of forward transformation, alloy partitioning is strongly affected by the crystallography at the interface. Energy dissipation and higher carbon content in austenite at the interface, described in figure 2(b) was detected in Fe-C and Fe-C-Mn systems [30]. However, since ferrite, in which diffusion is faster than in austenite, is the matrix phase, carbon partitioning reaches local equilibrium easily. Additionally, partitioning of substitutional alloying elements starts to take place even at the early stages of reverse transformation.

Figure 8 shows the partitioning behavior of substitutional elements in Fe-C-Mn-Si during intercritical annealing [31]. γG grows fast via the NPLE mode in the early stage of annealing, followed by slower growth kinetics with Mn and Si partitioning at the later stage. For γA, on the other hand, Mn and Si partitioning is already seen in the early stage of growth. Such a difference in partitioning kinetics also contributes to interface coherency of reverted austenite. The overall transformation kinetics were found to be much slower than para-equilibrium, and diffusional growth under low interfacial mobility was suggested from DICTRA calculation.

![Figure 8. Microstructure of reverted austenite and partitioning in Fe-0.3C-2Mn-1.5Si annealed at 1048K; (a) SEM image and (b) line profile obtained by EDS for γG. (c) STEM-BF image and (d) line profile obtained by EDS for γA [31].](image-url)
3. Strengthening by nanoscale precipitation at migrating interfaces

Interphase precipitation of alloy carbides during ferrite transformation establishes high yield strength and good elongation in low-carbon ferritic steels. A key for further strengthening is clarification of the conditions for refining the carbide dispersion, because high strength can be achieved via the Ashby-Orowan strengthening mechanism with moderate work hardening by introducing a high density of dislocations uniform [9, 32]. It was confirmed that a nano-sized carbide dispersion increases yield strength and work hardening in the early stage of deformation but later uniformly introduced dislocations interact each other and dynamic recovery is enhanced, resulting in suppression of inhomogeneous deformation.

The classical theory of interphase precipitation was reviewed by Honeycombe [33]. There are two distinctive mechanisms in terms of ferrite/austenite interface characteristics: (1) a ledge mechanism and (2) a bowing/quasi-ledge mechanism. In the ledge mechanism, the ferrite/austenite interface, across which the K-S OR is held, consists of an immobile coherent plane (terrace) and a mobile incoherent plane (ledge or riser), and then carbides nucleate at the immobile terrace plane. In the quasi-ledge mechanism, bulging of an interface with irrational OR takes place between particles with relatively large interparticle spacing after the incoherent and mobile interface is immobilized by carbide pinning on the interface. Subsequent nucleation of carbides on the bulged interface takes place and produces the next sheet.

We have demonstrated recently that a larger driving force for precipitation results in a finer carbide dispersion with incoherent ferrite/austenite interfaces [34-36]. Figure 9 is statistical data for orientation relationships at ferrite formed at austenite grain boundaries [34]. At higher temperatures, grain boundary

![Figure 9](image_url)

**Figure 9.** Ferrite orientation maps of the Fe–0.1C–1.5Mn–0.4V alloy transformed at (a) 993 K and (b) 923 K. Circles represent near K-S interfaces and triangles indicate non K-S interfaces. (c) Schematic for angular deviations from K-S OR; (d, e) distribution of Δθ₁ and Δθ₂ at 993K and 923K, respectively [34].
allotriomorphic ferrite (GBF) is dominant and grows toward to the matrix grain with non K-S OR (figures 9(a), (d)). As temperature decreases, Widmanstätten ferrite starts to form at grain boundaries and near K-S interfaces are increasingly formed (figures 9(b), (e)). Figure 10 shows effects of incoherency at the interface (here represented as deviation angle from the K-S OR) on the VC particle dispersion and corresponding nanohardness [35]. As the deviation angle becomes larger, up to 5-10 degrees, the number density of VC particles increases drastically with slight increase in particle size, as shown in (a) and (b), respectively. Such an increase of particle density results in higher hardness through the dispersion hardening as confirmed previously. The reason why nucleation of VC is promoted at non K-S interfaces is as follows. V is thought to have an attractive interaction to ferrite/austenite interface and exert solute drag effect as in the case of Nb addition. Fig. 11 shows the result of 3DAP analysis on near K-S and non K-S interfaces [36]. It is clear that more V atoms segregate at non K-S interfaces and provide a large advantage for carbide precipitation thermodynamically and kinetically.

It is rather challenging to make incoherent ferrite/austenite interfaces to obtain fine carbide dispersion through interphase precipitation. One possible way is use of ferrite transformation from deformed austenite. Fine-grained austenite structure inhibits formation of Widmanstätten ferrite so that recrystallization of austenite can be effective to obtain granular ferrite structure. During ferrite transformation in unrecrystallized austenite can be promising since local orientation fluctuation in the deformed austenite will accomplish loss of interfacial coherency during growth even though initial nucleus had near K-S OR with austenite. Use of nucleation at particles is also a good way. We demonstrated that nucleation at grain boundary precipitate or intragranular inclusion effectively enhances the formation of granular ferrite grains with non K-S OR, rather than Widmanstätten or bainitic

![Figure 10](image1.png)

**Figure 10.** Variations in (a) number density and (b) average radius of VC precipitates, (c) nanohardness of α grains with Δθ in Fe-0.1C-1.5Mn-0.4V transformed at 923K for 60s. Solid and open symbols indicate WF and GBF, respectively [35].

![Figure 11](image2.png)

**Figure 11.** 1D concentration profiles of V and C obtained by 3DAP analyses of (a) near K-S and (b) non K-S ferrite (α)/austenite (γ) interfaces in Fe-0.1C-1.5Mn-0.1V alloy transformed at 923K [36].
ferrite without nanosized VC precipitation in as-transformed condition [34, 37]. This ferrite morphology change results in the significant increase of overall hardness in V-microalloyed carbon steels.

4. Concluding remarks

In this paper, the effects of transformation interfaces on two important phenomena, namely, alloy partitioning/segregation and nanoscale precipitation at migrating interfaces, in design of advanced high strength steels with good ductility have been summarized. Interfacial coherency plays major roles in the control of microstructure and mechanical properties. Since it is important to control crystallographic constraints during nucleation and growth in phase transformations, a variety of alloy design and thermomechanical processing is expected to contribute to further advances in the development of novel high strength steels.

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