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Quantitative measurements of phase equilibria at migrating α/γ interface and dispersion of VC interphase precipitates: Evaluation of driving force for interphase precipitation

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ABSTRACT

For better industrial application of nano-sized interphase precipitates formed at migrating ferrite (α)/austenite (γ) interface during α transformation, microstructural control on its dispersion is of great significance. Although the driving force for its precipitation was known to be one of the important factors, the estimation on its value in literature without experimental supports was still questionable. Therefore, the present study aimed to quantitatively estimate the driving force for interphase precipitation based on experimental investigation in a series of V-added alloys isothermally transformed at various temperatures. Field emission-electron probe microanalysis (FE-EPMA) on α/γ interfacial C content reveals that α/γ phase equilibria can be well explained by negligible-partition local equilibrium (NPLE) model. Both number density and average radius of VC interphase precipitates quantified by using three-dimensional atom probe (3DAP) show good correlations with the calculated driving force for interphase precipitation at migrating α/γ interface under NPLE condition. Larger size of VC than the expected one from the driving force implies the possible growth after its nucleation at migrating α/γ interface.

1. Introduction

Interphase precipitation is the phenomenon that when steels containing carbide-forming elements (M; M = V, Nb or Ti) are transformed from austenite (γ) into ferrite (α), alloy carbides (MC) are precipitated in sheets as a result of periodic nucleation at migrating α/γ interface [1,2]. Due to the resultant high strength with reasonable formability [3–5], steels strengthened by MC precipitates formed by interphase precipitation have recently been commercialized in automotive industry [3]. It was reported that the strengthening magnitude by nano-sized MC interphase precipitates follows Ashby-Orowan model [3–6] that relatively larger strengthening can be obtained by more dispersed precipitates under a certain volume fraction. Therefore, for better industrial application, it is undoubtable that microstructural control on the dispersion of interphase precipitates is of great importance to obtain the desirable mechanical properties.

Batte et al. [7,8] found that finer VC interphase precipitates with larger strengthening effects can be obtained by decreasing transformation temperature or increasing V and C contents. The temperature dependence of the dispersion was also confirmed in the Nb-added alloy [9] and Ti-added alloy [10], respectively. Since interphase precipitation is one type of the precipitation reactions, it was suggested that the variations in dispersion should be affected by the degree of supersaturation, i.e. the driving force for precipitation, although other factors like α transformation kinetics may also have some influence [8,9].

Fig. 1(a) is the schematic of Fe-M-C phase diagram where the bulk C content indicated by the cross mark. C is enriched into γ during α transformation and builds up a concentration profile as schematically shown in Fig. 1(b). Since interphase precipitation occurs at migrating α/γ interface, interfacial concentration instead of the bulk alloy content should be considered in estimating the driving force for interphase precipitation. The driving force in α as well as in γ in the vicinity of interface can be thus approximately regarded as the gaps between the phase boundaries and the solubility lines as indicated by the red arrow lines in Fig. 1(a) [11], although it is still unclear yet whether MC precipitates are

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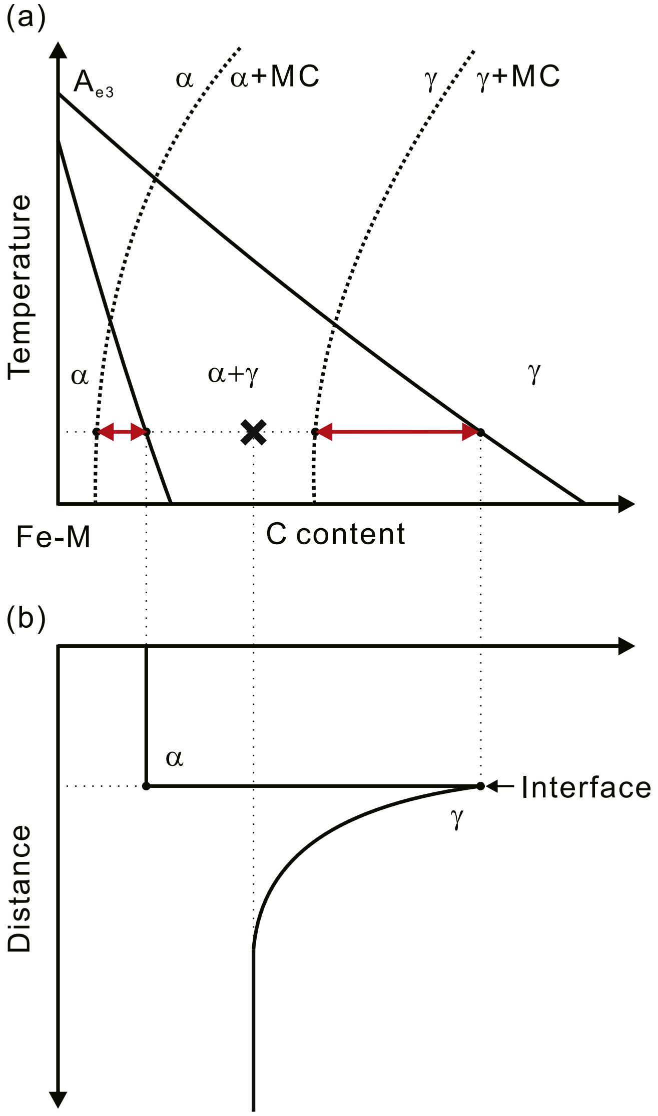

Fig. 1. Schematic of Fe-M-C phase diagram and C concentration profile across α/γ interface. Red arrow lines indicate the driving force for MC interphase precipitation [11]. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

nucleated in α or in γ. In addition, the interfacial M content at α/γ interface was also suggested to influence the driving force for interphase precipitation [12,13].

Two models have been proposed in literature to predict the α/γ phase equilibria without macroscopic partitioning of substitutional elements before the system reaches the full equilibrium (Ortho), which are called as Paraequlibrium (PE) [14] and negligible partitioning local equilibrium (NPLE) [15], respectively. They vary with each other in terms of whether local partitioning of the substitutional alloying elements, i.e. concentration spike in γ exists or not. Fig. 2 shows the isothermal section of Fe-Mn-C for example. No partitioning of substitutional alloying elements was assumed in PE model so that the compositions of both α (point A) and γ (point B) are simply the intersection points of bulk Mn content line and PE phase boundaries determined by the same chemical potential of C. As for the NPLE model, α inherits bulk Mn content (point D) in equilibrium with γ of spike composition (point E) as connected by the tie line. C iso-activity line passing point E with bulk Mn content (point F) is the NPLE growth limit. It is obvious that the C content in γ for NPLE limit tends to be lower than that for PE model.

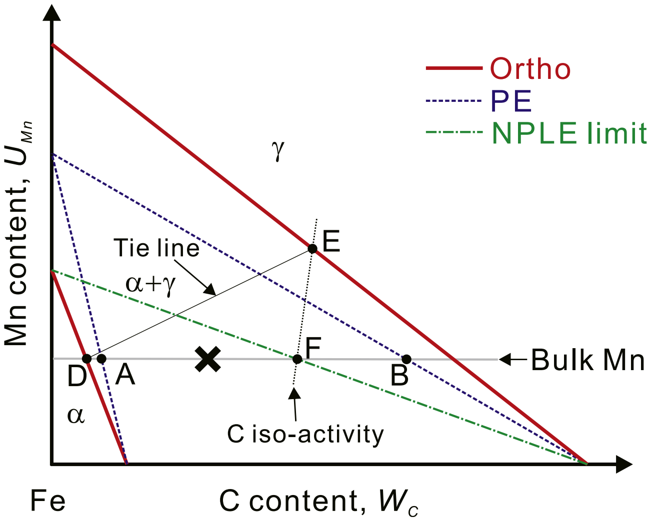

Fig. 2. Schematic isothermal section of Fe-Mn-C ternary phase diagram.

Lagneborg et al. [12] firstly tried to estimate the driving force for interphase precipitation. Based on the measurements on the α growth rate by Hillert [16], they assumed NPLE condition for higher temperature and PE condition for lower temperature, respectively. However, one may easily imagine that once M atoms can diffuse sufficiently to form MC interphase precipitation, α/γ phase equilibria should not be controlled by PE anymore. Therefore, the possibility of PE model to work in the alloys with interphase precipitation is quite low. On the other hand, α growth rate is known to be influenced by various factors including interfacial mobility, soft impingement and solute drag effects, etc. [17,18]. In this aspect, the validity of assumption by Lagneborg et al. is still questionable. Liu et al. [19] recently succeeded in directly measuring C content in γ by using field emission-electron probe microanalysis (FE-EPMA). They reported that phase equilibria at migrating α/γ interface in Fe-Mn-C system can be well explained by NPLE model in the temperature range down to 873 K. However, the case of M-added alloys with MC interphase precipitation still remains unclear yet.

Therefore, the objective of this study is to quantitatively estimate the driving force for interphase precipitation based on the experimental results of α/γ phase equilibria. The reasonability of the estimation will be further discussed by considering the resultant dispersion of interphase precipitates quantified by using three-dimensional atom probe (3DAP).

2. Experimental procedures

Four Fe-1.5Mn-0.05Si-V-added low carbon steels with different V and C contents, whose chemical compositions are shown in Table 1 were used in the present study. For simplicity, hereafter they will be called in shortened names, e.g. 0.1C-0.1 V, 0.1C-0.2 V, etc. All the as-received hot-rolled alloys were encapsulated into a silica tube with pure Ar gas for homogenization at 1423 K for 345.6ks to eliminate the Mn segregation. The homogenized blocks were further cut into small pieces with 10 × 5 × 3 mm^3 for further heat treatment. The C and N contents in Table 1 correspond to those after homogenization, with Para-Ae3 calculated by using *Thermo-Calc* with *TCFE5* database. Since the N content is only about 10 ppm, the solution temperatures of VC instead of V(C,N) were calculated and shown.