Carbide Precipitation Behavior in Nb and Nb-Mo Microalloyed Steels During Heating: Postprint

Authors: Zhang Zhengyan, Li Zhaodong, Yong Qilong, Sun Xinjun, Wang Zhenqiang, Wang Guodong

Date: 2016-11-04T00:00:00+00:00

Abstract

The precipitation behavior of carbides during heating in quenched Nb and Nb-Mo steels was studied using a Gleeble thermal simulator, microhardness tester, scanning electron microscope (SEM), high-resolution transmission electron microscope (HRTEM), and differential scanning calorimeter (DSC). The precipitation kinetics of MC-type carbides in the quenched steels was calculated using classical nucleation and growth theory and the Avrami equation. The results show that Nb and Nb-Mo microalloyed steels, when heated at a rate of 20℃/min to various temperatures and subsequently quenched, exhibited hardness peaks at 300℃ and 700℃ due to cementite and MC-type carbide precipitation, respectively. MC-type carbides precipitated at approximately 650℃, resulting in increased hardness from precipitation strengthening, which is consistent with the theoretically calculated nose temperature for MC-type carbide precipitation of approximately 650℃. Analysis suggests that Mo entering into NbC reduces the lattice mismatch between NbC and the ferrite matrix, thereby decreasing the interfacial energy between the precipitates and the ferrite matrix, while also lowering the precipitation free energy of NbC. However, the reduction in interfacial energy is dominant, ultimately accelerating the precipitation kinetics of (Nb,Mo)C. Consequently, the precipitate particles in Nb-Mo steel are more densely distributed and finer in size, resulting in a higher precipitation strengthening effect.

Full Text

Study on Carbide Precipitation Behavior in Nb and Nb-Mo Micro-Alloyed Steels During Heating

Zhang Zhengyan[1,2]), **Li Zhaodong**[1]), **Yong Qilong**[1]), **Sun Xinjun**[1]), **Wang Zhenqiang**[3]), **Wang Guodong**[2])

[1]) Department of Structural Steels, Central Iron and Steel Research Institute, Beijing 100081, China
[2]) State Key Laboratory of Rolling and Automation, Northeastern University, Shenyang 110819, China
[3]) Shougang Research Institute of Technology, Beijing 100043, China

Correspondent: Yong Qilong, Professor, Tel: (010)62183616, E-mail: yongql@126.com

Supported by: National Basic Research Program of China (No. 2010CB630805) and National Natural Science Foundation of China (No. 12060980)

Manuscript received: 2014-08-02, in revised form 2014-12-04

Abstract

As a strong carbide-forming element, niobium plays an important role in steel. Precipitated Nb can restrain austenite grain growth during the soaking process and provide precipitation strengthening after the / phase transformation, while dissolved or strain-induced Nb can inhibit recrystallization of deformed austenite. Recently, both Nb and Mo have been added to steel to enhance the effect of Nb. However, most research has focused on the continuous cooling process of / transformation or isothermal tempering processes, with few reports on the precipitation behavior of MC-type carbides in Nb-Mo steels during reheating and the effect of Mo on NbC precipitation in ferrite.

Therefore, this work investigates the precipitation behavior of MC-type carbides and the synergistic effect of Nb and Mo in Nb and Nb-Mo micro-alloyed steels during reheating at 20 °C/min using Vickers hardness testing, SEM, HRTEM, and DSC. The results show that both steels exhibit hardness peaks at 300 °C and 700 °C, attributed to the precipitation of -carbide and MC-type carbide, respectively. The MC-type carbide precipitates at approximately 650 °C during reheating, which agrees well with the nose temperature of MC-type carbide precipitation calculated by the Avrami equation. In Nb-Mo steel, the (Nb, Mo)C particles have a smaller mismatch with the ferrite matrix compared to NbC, leading to decreased interfacial energy. Consequently, the precipitation kinetics of MC-type carbide in Nb-Mo steel are faster than in Nb steel, resulting in denser and finer MC-type carbides and a higher precipitation strengthening effect.

Keywords: carbide, precipitation, Vickers hardness, interfacial energy

---

Niobium, as a strong carbide-forming element, strengthens the steel matrix through solid solution or by forming carbonitrides with carbon and nitrogen [1,2]. For instance, undissolved Nb carbonitrides prevent austenite grain growth during high-temperature soaking, dissolved or strain-induced Nb inhibits recrystallization of deformed austenite, and Nb precipitated during subsequent ferrite transformation provides effective precipitation strengthening [3–6]. Consequently, Nb is widely used in engineering machinery, tool, and fire-resistant steels [7-10]. To enhance the role of Nb, Nb-Mo composite addition is commonly employed. Cao et al. [11] reported that Nb-Mo composite addition in hot-rolled steel coils provides better precipitation strengthening increments from micro-alloy carbides and higher yield strength compared to Nb-Ti composites. Uemori et al. [12] found that in fire-resistant steels containing Nb and Mo, Mo segregation at NbC/matrix interfaces can inhibit NbC growth and coarsening for a period, thereby improving high-temperature mechanical properties.

In addition to Nb precipitation behavior during post-rolling cooling, researchers have extensively studied Nb precipitation in tempered martensitic steels to achieve a good balance of strength and toughness, taking advantage of the secondary hardening effect from extensive dispersion precipitation during tempering. Yu et al. [13] and Yang et al. [14] investigated the microstructure, properties, and Nb precipitation in quenched Nb-containing steels at different tempering temperatures. Liu et al. [15–17] further studied the nucleation, growth, and coarsening behavior of cementite and micro-alloy carbides during tempering of quenched martensite using three-dimensional atom probe. However, these studies were based on the same chemical composition at different tempering temperatures, with low Nb content (mass fraction below 0.05%) resulting in less significant secondary hardening effects. The mechanism by which Mo affects the precipitation kinetics of NbC in ferrite after Nb-Mo composite addition has not been elaborated in detail.

This work employs an alloy composition system with relatively high Nb content (0.1 mass fraction) to obtain higher precipitation strengthening increments. Samples were quenched from high temperature, reheated at a constant heating rate to various temperatures, and immediately water-cooled. Vickers hardness testing and microstructural/precipitate observations were performed to investigate hardness changes, microstructural evolution, and carbide precipitation behavior during heating. The precipitation kinetics of Nb precipitates in supersaturated ferrite were calculated using classical nucleation and growth theory, the Avrami equation, and a computational model established by Yong Qilong [18] for micro-alloy carbide precipitation kinetics.

Experimental

The experimental Nb and Nb-Mo micro-alloyed steels were melted in a vacuum induction furnace and cast into 50 kg ingots. Their chemical compositions are listed in Table 1, designated as Nb steel and Nb-Mo steel, both with micro-Ti treatment. After forging and controlled rolling/cooling, 11 mm thick plates were produced. Specimens were cut from the plate center, sealed in quartz tubes, soaked at 1200 °C for 5 h in a furnace, and then water-quenched. The quenched microstructure was observed using an S4300 cold field-emission scanning electron microscope (SEM). The Ac (pearlite-to-austenite transformation start temperature) and Ac (completion temperature for pro-eutectoid ferrite-to-austenite transformation) during heating at 20 °C/min were measured using a Formastor-F dilatometer on cylindrical specimens (Ø3 mm × 10 mm). Thermal cycling experiments were conducted on a Gleeble 3800 simulator by heating quenched specimens (Ø8 mm × 12 mm) at 20 °C/min to temperatures between 200–750 °C followed by immediate water cooling.

The thermally cycled specimens were axially sectioned into halves. From one half, 0.3 mm thick discs were cut, mechanically ground to 35 m thickness, punched into 3 mm diameter discs, and twin-jet electropolished in a 6% perchloric acid ethanol solution (volume fraction). The remaining material was etched in 3% nital (volume fraction) for SEM observation. The other half was ground, polished, deeply etched in 4% nital, carbon-coated in a HUS-5GB high-vacuum coater, and then carbon extraction replicas containing second-phase particles were collected on Cu grids. Both thin foils and extraction replicas were examined using a Tecnai F20 field-emission high-resolution transmission electron microscope (HRTEM) with energy-dispersive spectroscopy (EDS). Precipitation behavior of the quenched steels was studied using a DSC-60 differential scanning calorimeter on small discs (Ø8 mm × 1 mm) heated from 150–400 °C at 20 °C/min. Precipitate size distributions were statistically analyzed using Nano Measurer 1.2 software, with at least 200 particles measured per sample. Vickers hardness of both steels after heating to various temperatures and water cooling was measured using an FM-300 digital hardness tester with a 5 kg load, with final values averaged from five measurements.

Table 1 Chemical compositions of Nb and Nb-Mo micro-alloyed steels (mass fraction / %)

Results

2.1 Hardness Testing

Figure 1 [Figure 1: see original paper] shows the Vickers hardness of Nb and Nb-Mo micro-alloyed steels after quenching at 1200 °C and water cooling following heating to different temperatures. The as-quenched hardness of Nb steel is higher than that of Nb-Mo steel. Upon heating to 200 °C, the hardness of both steels decreases. At 300 °C, the hardness increases, forming a peak, then decreases again until temperatures exceed 600 °C, when hardness rises once more, reaching another peak at 700 °C. At 750 °C, hardness decreases again. Throughout the 300–700 °C range, Nb-Mo steel consistently exhibits higher hardness than Nb steel.

2.2 Microstructure

Figure 2 [Figure 2: see original paper] presents SEM images of Nb and Nb-Mo micro-alloyed steels after quenching at 1200 °C for 5 h. In addition to martensite, numerous elongated or granular bright white phases are distributed in both steel matrices, with a higher density in Nb-Mo steel. Figure 3 shows SEM images of both steels heated to 200–700 °C at 20 °C/min followed by water cooling. At 300 °C, most larger bright white phases disappear, while another phase precipitates within or at the boundaries of ferrite laths (Figure 3b [Figure 3: see original paper]), which gradually grows and coarsens at temperatures above 300 °C (Figures 3c–f). Figures 4a [Figure 4: see original paper] and 4b show bright-field and dark-field TEM images of the bright white phase in the quenched microstructure. Diffraction pattern analysis (Figure 4c) reveals a body-centered cubic structure with a different crystallographic orientation from the ferrite matrix. Literature [19] reported similar phases in rapidly cooled steels with comparable compositions, identifying them as martensite/austenite (M/A) islands [20]. Therefore, these elongated or granular phases are M/A islands. Figures 4d–f show TEM images, diffraction patterns, and EDS analysis of precipitates at 300 °C. The diffraction pattern indicates a hexagonal structure. Combined with characteristics of cementite and carbide precipitation during low-temperature tempering of martensite [15,21,22] and EDS results, this phase is identified as -carbide, which may evolve into cementite at temperatures above 300 °C [22].

Figure 5 [Figure 5: see original paper] shows TEM images of both steels heated to 500, 600, and 700 °C followed by immediate water cooling. With increasing temperature, some ferrite lath boundaries in Nb steel become blurred and gradually disappear, with parallel laths merging into wider ones and dislocation density decreasing. In contrast, Nb-Mo steel exhibits smaller ferrite lath widths, with most lath boundaries remaining distinct even at higher temperatures and significantly higher dislocation density than Nb steel, indicating that Mo addition retards ferrite lath recovery at elevated temperatures.

After heating to 700 °C and water cooling, the precipitate morphology, HRTEM images, and EDS spectra are shown in Figure 6 [Figure 6: see original paper]. The precipitates in Nb-Mo steel are more densely distributed and finer than those in Nb steel. Statistical analysis yields average precipitate sizes of 2.83 nm and 2.23 nm, respectively. EDS reveals that the dispersed second-phase particles in Nb steel are NbC, while those in Nb-Mo steel are (Nb, Mo)C with an average Mo/Nb atomic ratio of 0.56, indicating Mo occupies approximately 30% of the sites in the carbide. The lattice constants of NbC and (Nb,Mo)C are 0.446 nm and 0.430 nm, respectively. Literature [11,23] precisely measured the lattice constants of (Nb, Mo)C powder samples in hot-rolled Nb-Mo steels using XRD, finding that Mo incorporation reduces the NbC lattice constant. Jang et al. [24] studied (Ti, Mo)C stability in ferrite using first-principles calculations and experiments, finding that Mo (with a smaller atomic radius than Ti) partially substitutes for Ti in TiC to form (Ti, Mo)C, decreasing the lattice constant. Since NbC and TiC share the same FCC structure and both Nb and Ti have larger atomic radii than Mo, Mo incorporation similarly reduces the NbC lattice constant.

Discussion

3.1 Hardness Variations

Due to the low carbon content and resulting insufficient hardenability, the as-quenched microstructure of both steels consists of martensite plus bainite. Mo addition promotes bainite transformation, leading to more bainite in Nb-Mo steel (Figure 2) and consequently lower as-quenched hardness than Nb steel. After heating to 200 °C and water cooling, hardness decreases in both steels due to carbon atom diffusion, which reduces carbon supersaturation and lattice distortion in the matrix [15]. At 300 °C, M/A islands decompose [16] to form martensite and -carbide. The low carbon content and presence of strong carbide-forming elements Nb and Mo inhibit -carbide growth and coarsening [18,25], allowing fine -carbides to provide precipitation strengthening. Figure 7 [Figure 7: see original paper] shows DSC curves for both steels during heating. An exothermic peak near 300 °C, combined with microstructural changes during martensite tempering [22], confirms -carbide precipitation, consistent with SEM and TEM observations. At higher temperatures (600 °C), hardness decreases due to matrix recovery and coarse cementite formation (Figure 3). At elevated tempering temperatures, cementite dissolves and releases carbon atoms that form MC-type micro-alloy carbides, generating precipitation strengthening that slows the hardness decrease caused by matrix recovery and cementite coarsening. As temperature continues to rise, MC-type carbide precipitation increases, and when the strengthening increment exceeds matrix softening, hardness begins to increase. Thus, the hardness increase above 600 °C results from micro-alloy carbide precipitation. The Ac and Ac temperatures are 735 °C and 915 °C for Nb steel, and 715 °C and 905 °C for Nb-Mo steel. When temperatures exceed Ac , the small amount of austenite formed in the short time combined with ferrite matrix softening from recovery leads to decreased hardness after water cooling from temperatures above 750 °C.

Furthermore, Figure 5 shows that Mo addition inhibits substructure recovery. Although small amounts of Mo co-precipitate with cementite or MC-type carbides, most Mo remains in solid solution due to its high solubility in the matrix [18], enhancing atomic bonding [26] and retarding martensite recovery. Additionally, the denser and finer (Nb, Mo)C precipitation effectively pins dislocations [16,26].

Based on the MC-type carbide precipitation morphology (Figure 6), Nb-Mo steel exhibits denser and finer precipitates than Nb steel, resulting in significantly higher hardness, particularly at 700 °C where extensive dispersion strengthening causes a pronounced hardness increase.

3.2 Precipitation Kinetics of NbC/(Nb,Mo)C [18]

The precipitation kinetics of micro-alloy carbonitrides in steel are widely described using classical nucleation and growth theory with the Avrami equation, typically applied to precipitation during cooling in austenite or ferrite. However, calculations of precipitation kinetics in supersaturated ferrite during heating are rarely reported. After solution treatment above the complete dissolution temperature and water quenching, all Nb can be considered dissolved in the supersaturated ferrite. Rapid heating at a constant rate to a certain temperature induces NbC precipitation. At lower temperatures, although the precipitation driving force (supersaturation) is large, element diffusion is limited. At higher temperatures, diffusion is favorable but the driving force is lower. The combined effect of these two factors should produce C-curve kinetics during heating. Additionally, since micro-alloy carbides have lower free energy than cementite (i.e., are thermodynamically more stable [18]) and precipitate at higher temperatures where cementite dissolves to provide carbon atoms, cementite effects can be neglected when using a sufficiently fast heating rate to avoid cementite formation. Using the computational method from literature [18], the critical nucleation energy and critical nucleus size versus temperature, as well as precipitation-temperature-time (PTT) curves, were calculated for micro-alloy carbide precipitation during heating. To compare Mo effects, the compositions used were 0.042%C-0.1%Nb for Nb steel and 0.042%C-0.1%Nb-0.19%Mo for Nb-Mo steel.

3.2.1 Effect of Mo on Interfacial Energy of NbC Precipitation During micro-alloy carbide nucleation, interfacial energy between the precipitate and Fe matrix is a barrier to nucleation. Interfacial energy comprises structural and chemical components, with the structural component typically exceeding the chemical component by more than an order of magnitude for semi-coherent interfaces [18]; therefore, only the structural component is considered. According to Vegard's law [28], when both MoC and NbC have cubic structures, the lattice parameter of the composite precipitate (NbxMo)C (0 x 1) varies linearly with Mo atomic fraction [29]. Using lattice constants of 0.4477 nm for NbC [24] and 0.4277 nm for MoC [30], the lattice parameter of (NbxMo)C as a function of Mo fraction is shown in Figure 8 [Figure 8: see original paper], demonstrating that the lattice parameter decreases with increasing Mo fraction.

MC-type micro-alloy carbides (NbxMo)C with NaCl structure typically have a Baker-Nutting orientation relationship with the ferrite matrix [18]: [001]*{(NbxMo)C} // [011]*{ }, [110]*{(NbxMo)C} // [001]*{ }. The mismatch along the [110] direction can be calculated using equation (1) [18]:

= (a_{(NbxMo)C} - a_) / a_

where a_{(NbxMo)C} and a_ are the lattice constants of (NbxMo)C and ferrite, respectively. Since the lattice constant of (NbxMo)C varies with Mo fraction, the mismatch also changes accordingly. Based on semi-coherent dislocation theory [18], the interfacial energy between (NbxMo)C and ferrite was calculated, with results shown in Figure 9 [Figure 9: see original paper] for 0.5 x 1. The interfacial energy decreases with increasing Mo fraction and varies with temperature due to the temperature dependence of the ferrite elastic modulus [18].

3.2.2 Critical Nucleus Size and Critical Nucleation Energy of (NbxMo)C

The free energy change ΔG for forming a nucleus of diameter d on a dislocation line in ferrite is:

$$\Delta G = - (d^3/6) \cdot \Delta G\_V + d^2 \cdot + (d^2/4) \cdot \Delta G\_d$$

where ΔG_V is the volume free energy change, is the interfacial energy, and ΔG_d is the dislocation core energy. ΔG_V can be calculated from the solubility product of the precipitate in ferrite [18]. The dislocation energy per unit length is ΔG_d = Gb² / [4 (1-)], where G is the shear modulus of ferrite, b is the Burgers vector magnitude, and is Poisson' s ratio. ΔG_V = ΔG_m / V_m, where ΔG_m is the molar free energy of (NbxMo)C precipitation and V_m is the molar volume. The critical nucleus size d_c and critical nucleation energy ΔG_c are then:

$$d\_c = -4 / \Delta G\_V$$
$$\Delta G\_c = (16 ^3) / (3\Delta G\_V^2)$$

3.2.3 Nucleation Rate of (NbxMo)C

For nucleation on dislocation lines where the nucleation rate rapidly decays to zero, the nucleation rate I can be expressed as:

$$I = K \cdot (T/k) \cdot \cdot \exp[-(Q + \Delta G\_c) / (kT)]$$

where K is a temperature-independent constant, T is thermodynamic temperature, k is Boltzmann' s constant, is dislocation density, and Q is the diffusion activation energy of Nb. Taking logarithms of both sides yields:

$$\ln I = \ln[K \cdot (T/k) \cdot] - (Q + \Delta G\_c) / (kT)$$

3.2.4 Precipitation-Temperature-Time (PTT) Curves of (NbxMo)C

Assuming the time t for 5% precipitation represents the precipitation start time, the PTT equation is:

$$\lg(t/t) = 1.28994 \times 10 \cdot (^3 / T) \cdot (1/\Delta G\_V^2) - (2/3)\lg(Q/kT)$$

where n is the precipitation kinetics time exponent (n=1 when nucleation rate rapidly decays to zero) and t is a temperature-independent constant [18].

Equations (2-7) show that precipitation kinetics are governed by both the precipitation free energy (driving force) and interfacial energy. In ferrite, Mo has minimal effect on the precipitation driving force of NbC [18]. Experimental studies by Jang et al. [14] and Wang et al. [31] found that Mo fraction decreases as TiC grows and coarsens in Fe matrix. Liu et al. [16,17] observed using atom probe that Mo diffuses out of (Nb, V, Mo)C carbides during tempering, being replaced by Nb and V carbides, indicating that Mo is thermodynamically unstable in MC-type carbides and only has high occupancy during early nucleation stages. As MC-type carbides grow and coarsen and kinetics approach equilibrium, Mo fraction decreases. Since the calculated free energy represents values approaching equilibrium at a given temperature, Mo incorporation has minimal effect on precipitation free energy during early nucleation stages.

Calculated critical nucleation energy and critical nucleus size for (NbxMo)C precipitation in ferrite are shown in Figure 10 [Figure 10: see original paper]. At the same precipitation temperature, both critical nucleation energy and critical nucleus size decrease with increasing Mo fraction, indicating that (Nb, Mo)C precipitates more readily in Nb-Mo steel. This is consistent with TEM observations showing denser and finer precipitates in Nb-Mo steel.

The calculated PTT curves for (NbxMo)C precipitation in ferrite during heating are shown in Figure 11 [Figure 11: see original paper]. Mo incorporation reduces the mismatch between NbC and ferrite, thereby decreasing interfacial energy and shifting the PTT curves leftward and upward, indicating accelerated precipitation kinetics. The fastest precipitation nose temperature increases from 600 °C to 650 °C with increasing Mo fraction, which agrees well with the experimental precipitation start temperature of approximately 650 °C determined from hardness testing.

Conclusions

1. Quenched Nb and Nb-Mo micro-alloyed steels heated at 20 °C/min to various temperatures followed by water cooling exhibit hardness peaks at 300 °C and 700 °C. The 300 °C peak results from decomposition of martensite/austenite (M/A) islands into -carbide, while the 700 °C peak arises from nanoscale MC-type carbide precipitation strengthening.

2. The hardness decrease in both steels between 300–600 °C is caused by matrix recovery and cementite coarsening, with no MC-type carbide precipitation below 600 °C. When temperature exceeds 600 °C, nanoscale MC-type carbides precipitate, causing precipitation strengthening and hardness increase. Theoretical PTT curves for MC-type carbide precipitation during heating show C-curve characteristics with a nose temperature around 650 °C, consistent with experimental results.

3. Mo incorporation into NbC reduces the mismatch with the ferrite matrix, thereby decreasing interfacial energy between precipitates and matrix. This accelerates (Nb, Mo)C precipitation kinetics in ferrite, resulting in denser, finer precipitate distributions and higher precipitation strengthening in Nb-Mo steel.

References

[1] Yong Q L, Ma M T, Wu B R. Micro-alloyed Steels-Physical Mechanical Metallurgy. Beijing: China Machine Press, 1989: 30
[2] DeArdo A J. In: Fu J Y, Wang W Z, eds., Niobium Science & Technology. Beijing: Metallurgical Industry Press, 2003: 271
[3] Rainforth W M, Black M P, Higginson R L, Palmiere E J, Sellars C M, Prabst I, Warbichler P, Hofer F. Acta Mater, 2002; 50: 735
[4] Cao Y B, Xiao F R, Qiao G Y, Huang C J, Zhang X B, Wu Z X, Liao B. Mater Sci Eng, 2012; A552: 502
[5] Zhang Z H, Liu Y N, Liang X K, She Y. Mater Sci Eng, 2008; A474: 254
[6] Park D B, Huh M Y, Shim J H, Suh J Y, Lee K H, Jung W S. Mater Sci Eng, 2013; A560: 528
[7] Weng Y Q. Ultra-Fine Grained Steels. Beijing: Metallurgical Industry Press, 2008: 58
[8] Huang G J, Kong X L, Guan J, An X G. Microalloying Technol, 2009; 3-4;293
[9] Chi H X, Ma D S, Liu J H, Chen Z Z, Yong Q L. Microalloying Technol, 2009; 3-4; 427
[10] Chijiiwa R, Tamehiro H, Yoshida Y, Funato K, Uemori R, Horii Y. Nippon Steel Techni Report, 1993; 58: 47
[11] Cao J C, Yong Q L, Liu Q Y, Sun X J. J Mater Sci, 2007; 42: 10080
[12] Uemori R, Chijiiwa R, Tamehiro H, Moriawa H. Appl Surf Sci, 1994; 76/77: 255
[13] Yu H, Zhang D D, Xiao R T, Zhuo P, Li C M. J Univ Sci Technol Beijing, 2011; 33: 715
[14] Yang G W, Sun X J, Li Z D, Li X X, Yong Q L. Mater Sci Technol, 2013; 21: 118
[15] Liu Q D, Liu W Q, Wang Z M, Zhou B X. Acta Metall Sin, 2009; 45: 1281
[16] Liu Q D, Peng J C, Liu W Q, Zhou B X. Acta Metall Sin, 2009; 45: 1288
[17] Liu Q D, Chu Y L, Peng J C, Liu W Q, Zhou B X. Acta Metall Sin, 2009; 45: 1297
[18] Yong Q L. Secondary Phase in Steels. Beijing: Metallurgical Industry Press, 2006: 361
[19] Olasolo M, Uranga P, Rodriguez J. M, Lopea. Mater Sci Eng, 2011; A528: 2559
[20] Duan L N, Chen Y, Liu Q Y, Jia S J, Jia C C. J Iron Steel Int, 2014; 21: 227
[21] Wang L J, Cai Q W, Wu H B, Yu W. J Univ Sci Technol Beijing, 2010; 32: 1150
[22] Pan J S, Tong J M, Tian M B. Fundamentals of Material Science. Beijing: Tsinghua University Press, 2011: 660
[23] Cao J C, Yong Q L, Liu Q Y, Sun X J. Trans Mater Heat Treat, 2006; 27: 51
[24] Jang J H, Lee C H, Heo Y U, Suh D W. Acta Mater, 2012; 60: 208
[25] Sakuma T, Watanabe N, Nishizawa T. Trans JIM, 1980; 21:159

[26] Zhao P, Xie F Z, Sun Z G. Materials Science Essentials. Harbin: Harbin Institute of Technology Press, 2009: 226
[27] Wang Z Q, Yong Q L, Sun X J, Yang Z G, Li Z D, Zhang C, Weng Y Q. ISIJ Int, 2012; 52:1661
[28] Denton A R, Ashcroft N W. Phys Rev, 1991; 43A: 3161
[29] Qadri S B, Fuller W W, Kihlstrom K E, Simon R W, Skelton E F, Van-Vechten D, Wolf S. A. J Vacum Sci Technol, 1985; 3A: 664
[30] Willens R H, Buehler E, Matthias B T. Phys Rev, 1967; 159: 327
[31] Wang Z Q, Sun X J, Yang Z G, Yong Q L, Zhang C, Li Z D, Weng Y Q. Mater Sci Eng, 2013; A573: 84

Note: Figure translations are in progress. See original paper for figures.