Randomization of Ferrite/austenite Orientation Relationship and Resultant Hardness Increment by Nitrogen Addition in Vanadium-microalloyed Low Carbon Steels Strengthened by Interphase Precipitation

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Interphase precipitation of nano-sized alloy carbides is recently used to strengthen low carbon steels for its excellent contributions to strength and formability. The effects of nitrogen addition on the hardness of vanadium-microalloyed low carbon steels were investigated by considering both the dispersion of interphase precipitation and the ferrite/austenite crystallography. Three-dimensional atom probe analysis reveals that interphase precipitation of vanadium carbide is hardly affected by increasing the nitrogen content, although the nanohardness of ferrite is slightly increased. Another important factor determining the overall hardness of ferrite is found to be the ferrite/austenite crystallography. At lower transformation temperature, nitrogen addition reduces the amount of Widmanstatten ferrite and bainite, which are formed in absence of interphase precipitation. Instead, relatively harder allotriomorphic and idiomorphic grain boundary ferrite without Kurdjumov-Sachs orientation relationship against austenite are formed extensively.

KEY WORDS: phase transformation; interphase precipitation; crystallography; atom probe tomography; nanoindentation.

1. Introduction

Strengthening of ferritic steels are usually realized by fine precipitation for industrial application. When alloy carbides are precipitated from ferrite, larger amount of precipitation strengthening can be obtained by decreasing the inter-particle spacing, i.e. with higher number density and smaller size under the same volume fraction. In particular, nano-sized alloy carbides formed through interphase precipitation were found to have great potential in strengthening ferritic matrix (~ 300 MPa) accompanied by reasonable contribution in ductility and thus formability. When steels are microalloyed with strong carbide-forming element, e.g. niobium (Nb), titanium (Ti) or vanadium (V), different from general precipitation by aging of supersaturated matrix, interphase precipitation occurs as the phenomenon that alloy carbides are precipitated in sheets by periodic nucleation at migrating ferrite/austenite (α/γ) interface during transformation. When α/γ interface moves forward, newly formed alloy carbides will be left in α and thus strengthen α matrix. As a result, steels with high strength can be produced through direct cooling without any further costly heat treatment. Due to its irreplaceable importance, interphase precipitation becomes increasingly popular in the production of practical steels for automotive application.

Through quantitative analysis by three-dimensional atom probe (3DAP), our group recently found that the dispersion of vanadium carbide (VC) formed by interphase precipitation is dominantly determined by the driving force for its precipitation. For example, higher V and C contents at α/γ interface as well as lower solubility of VC at lower temperature provide larger driving force, resulting in finer dispersion of VC and higher local hardness of α grains. However, the overall hardness of the fully transformed specimen becomes inversely decreased due to the extensive formation of Widmanstatten α (WF) or bainitic α (BF) at lower transformation temperature.

By further combining electron backscatter diffraction (EBSD) with 3DAP, the influence of interfacial coherency on interphase precipitation, as is evaluated by the orientation relationship between α and γ (α/γ OR), was also studied in a V-added low carbon steel. When α and γ hold closely Kurdjumov-Sachs OR (K-S OR: (111)α // (011)n, [101]α // [1T1]n, e.g., in the case of WF or BF, significantly reduced number density of VC precipitates was observed, indicating the suppression of its nucleation at interphase boundary. In contrast, the nucleation of VC interphase

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precipitation is facilitated by lower interfacial coherency of allotriomorphic or idiomorphic grain boundary $\alpha$ (GBF) largely deviated from K-S OR, which is suggested to be caused by more rapid interfacial diffusion of V and severer interfacial V segregation.\(^7\) This means that in addition to the driving force for precipitation, $\alpha$ morphology and essentially $\alpha/\gamma$ crystallography also play a key role in determining the overall mechanical properties. Accordingly, the strategy for further strengthening of precipitation-strengthened microalloyed steels is suggested to be the randomization of $\alpha/\gamma$ OR, especially at lower temperature.

One possible way to realize that is through the addition of nitrogen (N), which was studied by some of the present authors in V-added steels. Vanadium carbonitride (V(C,N)) precipitate formed in $\gamma$ was found to act as the preferential nucleation site for subsequently formed intragranular $\alpha$ (IGF), due to its high coherency and low interfacial energy with $\alpha$.\(^6\) Such IGF grains were also found to hold Baker-Nutting OR with V(C,N) ($\langle100\rangle_{\gamma}/\langle001\rangle_{\gamma}$), but irrational OR with $\chi$.\(^1\) On the other hand, since alloy nitride has lower solubility in Fe than alloy carbide,\(^5\) another merit by N addition is the enlarged driving force for interphase precipitation, which was experimentally proved to refine its dispersion in literatures.\(^12\)-\(^14\)

In this study, variations in local dispersion of VC and resultant hardening in $\alpha$ grains with various $\alpha/\gamma$ ORs during isothermal holding were studied by means of 3DAP and nanoindentation. After that, the effects of N addition on $\alpha/\gamma$ crystallography and overall harden of V-microalloyed low carbon steels strengthened by interphase precipitation were examined.

2. Experimental Procedures

2.1. Preparation and Overall Microstructural Characterization

Two Fe-0.1C-1.5Mn-0.4V-0.05Si (mass%)–based alloys with less than 10 ppm and 150 ppm N additions, respectively, were used in this study. Hereafter they are defined as the N-free and N-added alloys, respectively.

As-received hot-rolled alloys were firstly cut into a block of $100 \times 25 \times 10 \text{ mm}^3$ and encapsulated into silica tubes and homogenized at 1423 K for 345.6 ks to eliminate the Mn segregation. Pure Ar gas and 70% Ar-30% N\textsubscript{2} mixed gas were used as the protective gas for the N-free and N-added alloys, respectively. The chemical compositions in Table 1 are those after homogenization, slightly lower in C and N contents than the as-received condition. Para-A\textsubscript{K}3 and solution temperatures in $\gamma$ (VC for the N-free alloy and V(C,N) for the N-added alloy, respectively) were calculated by using Thermocalc with TCFE5 database and shown correspondingly. Small pieces with $10 \times 5 \times 3 \text{ mm}^3$ were cut from the homogenized blocks and then austenitized in the electric vacuum furnace at different temperatures between 1473 K and 1523 K for 6 ks to dissolve all the added V and obtain similar $\gamma$ grain sizes about 200 µm. After austenitization, the specimens were soaked into salt bath and isothermally transformed at temperatures in the range from 848 K to 993 K for various times, followed by water quenching.

The transformed specimens were polished and etched with 3% nitral for optical microscope observation. EBSD measurements on field emission scanning electron microscope (FE-SEM; JEOL JSM-7001F) with a step size of 0.5 µm were carried out on the isothermally transformed specimens electro-polished by 6% perchloric acid in ethanol under the voltage of 30 V. The accelerating voltage was 25 kV and only the data points with a confidence index value higher than 0.2 were used for the following analyses on TSL OIM ver. 6 software. The overall Vickers hardness was measured at 10 randomly selected regions under a load of 9.8 N to evaluate the precipitation strengthening.

2.2. Local Microstructural Characterization

Figure 1 shows the experimental procedures for local microstructural characterization in this study, by taking the N-free alloy isothermally transformed at 923 K for 60 s as an example.\(^9\) Figure 1(a) is the $\alpha$ orientation map obtained by EBSD measurement, with its corresponding 001 bcc pole figure shown in Fig. 1(b). Due to near K-S OR between martensite and its parent $\gamma$ specific orientation distribution of K-S variants\(^15\) in black dots can be easily identified in the pole figure and used to reconstruct the orientation of prior $\gamma$\(^11\). Then the deviation angle from the exact K-S $\alpha/\gamma$ OR ($\Delta\theta$) of each $\alpha$ grain was calculated, while near K-S and non K-S ORs were defined by whether $\Delta\theta$ is less than 5 degrees or not following our previous work.\(^9\) Afterwards, microsampling method by using focused ion beam (FIB: FEI Quanta 3D) was applied to prepare the needle-shaped specimens from the $\alpha$ regions after the migration of $\alpha/\gamma$ interface with defined ORs, whose configurations were recorded in advance by SEM for reconstruction. 3DAP measurements of those specimens were performed on a CAMECA LEAP-4000 HR at 50 K with a pulse fraction of 20% and a pulse rate of 200 kHz. Figure 1(c) shows the typical V atom maps of near K-S region in WF ($\Delta\theta = 0.8$ deg.) and non-K-S region in GBF ($\Delta\theta = 19.2$ deg.) after reconstruction of the raw data on IVAS ver. 3.6 software. Cluster analysis by maximum separation method\(^17\) was performed to quantitatively evaluate the dispersion in terms of number density, average radius and volume fraction following our previous work.\(^7,8\)

Nanoindentation measurements on a Hysitron TI950 with a maximum load of 500 µN at room temperature were further performed to measure the local nanohardness of the specific $\alpha$ grains, of which typical examples are shown in Fig. 1(d). Such load gives an indentation size of about 300 nm, which is much larger than the inter-particle spacing of VC inter-

| Alloy     | C     | V     | Si   | Mn   | P, S, O | N    | Fe  | Para-A\textsubscript{K} | Sol. temp. in $\gamma$/K |
|-----------|-------|-------|------|------|---------|------|-----|----------------|-------------------------|
| N-free    | 0.074 | 0.43  | 0.05 | 1.49 | <0.002  | <0.001 | Bal | 1 103 | 1 170                    |
| N-added   | 0.085 | 0.42  | 0.05 | 1.52 | <0.002  | 0.015 | Bal | 1 088 | 1 502                    |
phase precipitation,\textsuperscript{4} and thus sufficient to evaluate its precipitation strengthening. Tip calibration and the calculation of nanohardness were conducted from the load-penetration depth curves according to Oliver and Pharr method.\textsuperscript{18} 10 measurements for each $\alpha$ grain were made to reveal the statistically average hardness.

2.3. Post-transformation Aging

The difficulty in studying the aging behaviors in $\alpha$ grains with known ORs is the absence of martensite for $\alpha/\gamma$ OR analyses, caused by the random OR between GBF and $\gamma$ grains.\textsuperscript{19} To solve this problem, special heat treatment was performed on the N-free alloy. The alloy was partially transformed at 923 K for 120 s to form GBF, WF and martensite after water quenching,\textsuperscript{9} and further isothermally held at the same temperature for various times up to 43.1 ks. Since 923 K is lower than $A_{e1}$ temperature, martensite is only tempered and remained even after prolonged holding. The diffusion distance of V at 923 K for equivalently 43.2 ks (120 s + 43.1 ks) in $\alpha$ was calculated to be less than 1 nm with the available diffusion coefficient data,\textsuperscript{20} so that partitioning of V from martensite to $\alpha$ is not likely to occur to influence the particle dispersion in $\alpha$.

3. Results

3.1. Optical Microstructure and Overall Hardness

Figure 2 shows the typical optical microstructure of the isothermally transformed specimens. The prior $\gamma$ grain boundary (PAGB) in the N-free alloy transformed at 923 K for 60 s is clearly decorated with allotriomorphic and idiomorphic GBF accompanied by a small amount of WF (Fig. 2(a)). After isothermal holding for 1.8 ks, $\alpha$ transformation is completed without change in its morphology, which is still mostly composed of polygonal $\alpha$ (Fig. 2(b)). The N-added alloy transformed at 923 K represents a very similar microstructure, while the amount of WF appears to be slightly decreased (Figs. 2(c) and 2(d)). When the temperature is lowered to 873 K, although some GBF grains can still be observed in the N-free alloy, WF/BF clearly becomes the dominant microstructure (Fig. 2(e)). The existence of martensite formed from $\gamma$ during quenching (denoted as M($\gamma$)) after 1.8 ks indicates that transformation is still not completed (Fig. 2(f)). In contrast, the formation of WF/BF is strongly suppressed in the N-added alloy, while GBF and idiomorphic IGF grains are instead extensively formed (Fig. 2(g)). After holding for 1.8 ks, a negligible amount of pearlite (P) with black contrast is also observed in addition to the polygonal $\alpha$ (Fig. 2(h)).

Figure 3 shows the variations in overall Vickers hardness of these two alloys isothermally transformed at different temperatures for 1.8 ks. Although some of the specimens are still not fully transformed (e.g. Fig. 2(f)), the amount of M($\gamma$) is not so large to significantly influence the overall hardness. The hardness of both alloys are much higher than precipitate-free $\alpha$,\textsuperscript{21} reflecting the precipitation strengthen-
As the temperature becomes decreased from 993 K to 848 K, the hardness increment at first in the N-free alloy was explained by the refinement in dispersion of VC formed by interphase precipitation in the GBF-dominated specimens, while the reduction later was caused by the extensive formation of softer WF/BF with near K-S OR at low temperature. Such phenomenon can be observed in the case of the N-added alloy as well, whose hardness is relatively higher at all the temperature range especially at 873 K.

### 3.2. Effects of Isothermal Holding on Dispersion of Precipitates

Figures 4(a) and 4(b) show the variations in number density and average radius of VC in α grains with various
that at the early stage of $\alpha$ transformation, polygonal GBF with non K-S OR is comparatively harder than WF and BF with near K-S OR.

**Figure 5** shows the three-dimensional V atom maps superimposed by 2 at% V iso-concentration surface of $\alpha$ grains in the N-free alloy isothermally transformed at 923 K for (a)(b) 1.8 ks, (c)(d) 43.2 ks.

After analyzing several $\alpha$ grains, **Fig. 6** summarizes the variations in number density, average radius and volume fraction of VC and nanohardness of $\alpha$ grains against isothermal holding time, where the average values for near K-S and non K-S cases are connected by the dot lines. Number density of VC decreases continuously in the non K-S $\alpha$ grains by increasing the holding time, whereas number density in the near K-S $\alpha$ grains originally free of particle increases at first due to VC precipitation by aging and then becomes reduced by overaging (Fig. 6(a)). The size of VC for both cases simply increases with isothermal holding (Fig. 6(b)), while the volume fraction gradually approaches but still below the equilibrium value ($f_{\text{eq}}$, calculated by Thermocalc) and maximum value ($f_{\text{max}}$, when all the added V is precipitated) even after 43.2 ks (Fig. 6(c)). Continuous softening occurs in the non K-S $\alpha$ grains, while a peak hardness exists in the near K-S $\alpha$ grains (Fig. 6(d)). Non K-S $\alpha$ grains with interphase precipitation are relatively harder than near K-S $\alpha$ grains even after isothermal holding for 1.8 ks, while the difference almost disappears after prolonged holding for 43.2 ks.

### 3.3. Effects of N Addition on Dispersion of Interphase Precipitation and $\alpha$ Nanohardness

In order to explain the hardness increment by N addition shown in **Fig. 3**, interphase precipitation behaviors were checked by using 3DAP. **Figure 7** shows the dispersion of VC/V(C,N) interphase precipitation formed in $\alpha$ grains with non K-S OR in the alloys isothermally transformed at 993 K, 923 K and 873 K for 60 s, respectively. The dispersion of VC precipitates is clearly finer at lower temperature, whereas the effects of N addition at each temperature are not apparent from the atom maps.
Quantitative results for the dispersion and resultant nanohardness of α grains are summarized in Fig. 8. Higher number density, smaller radius and higher nanohardness can be obtained at lower transformation temperature for both alloys. The reported refining effects of N on the dispersion\textsuperscript{12–14} can be hardly observed (Figs. 8(a) and 8(b)), presumably due to the relatively smaller amount of N addition in this study. Regardless of transformation temperature, the estimated volume fraction of interphase precipitation in both alloys is less than half of the maximum volume fraction ($f_{\text{max}}$) when all the added V is precipitated (Fig. 8(c)). Higher nanohardness of the N-added alloy than the N-free alloy is presumably caused by slightly larger volume fraction (Fig. 8(d)). In addition, the slightly higher C + N content in the N-added alloy may also provide some extra solid solution strengthening in α.\textsuperscript{23} Such higher nanohardness of the N-added alloy explains the increment in overall hardness by N addition at 923 K to 993 K as shown in Fig. 3, but not the significantly increased overall hardness at 873 K. Therefore, additional reasons like the suppression of WF/BF formation with near K-S OR, whose nanohardness was found to be relatively lower as mentioned above, should be considered. In the aspect of WF/BF formation, Aaronson et al.\textsuperscript{24} systematically studied Fe-0.1C-based alloys with the additions of various alloying elements and concluded that alloying effects on the starting temperature of WF/BF formation ($W_S$) is closely related to those on $A_{\text{e3}}$ temperature. By considering the calculated $A_{\text{e3}}$ as shown in Table 1, $W_S$ temperature of the N-added alloy should be 15 K lower than the N-free alloy. Such small difference can hardly explain the large difference in α orientations between the two alloys transformed at 873 K.

4. Discussion - Effects of N Addition on αγ Crystallography and Resultant Precipitation Strengthening

In last section, it is clearly shown that αγ OR plays an important role in determining the overall hardness of the low carbon steels strengthened by interphase precipitation. The cause for the variations in αγ crystallography and the resultant higher hardness by N addition will be discussed in this section.

Figures 9(a) and 9(d) are the α orientation maps of the alloys fully transformed at 873 K. The regions surrounded by the black lines in these figures are transformed from single γ grains, whose 001bcc pole figures are shown accordingly in Figs. 9(b) and 9(e), respectively. Specific orienta-
tion distribution for near K-S OR\textsuperscript{13} as a result of extensive formation of WF/BF can be easily identified in pole figure of the N-free alloy (Fig. 9(b)), whereas the orientations in the N-added alloy are randomly distributed (Fig. 9(e)). According to Morito et al.\textsuperscript{15} the inter-variant misorientation angle among the 24 K-S variants should be smaller than 21 deg. or larger than 47 deg. The misorientation between those two values corresponds to the $\alpha$ grain boundaries without holding inter-variant relationship, which is caused by the formation of $\alpha$ without holding K-S OR, although it is not strictly true (sometimes two $\alpha$ grains without holding inter-variant relationship might have the misorientation angle smaller than 25 deg. or larger than 45 deg. by chance, but with the rotation axes and angles different from the K-S inter-variant case). By considering the error of orientation measurement in EBSD ($\sim 0.5$ deg.\textsuperscript{25}) and the deviation of BF from the exact K-S OR ($\sim 3.0$ deg.\textsuperscript{26}), the threshold is determined to be 25 deg.–45 deg. The positions of the $\alpha$ grain boundaries with such misorientation (25 deg.–45 deg.) are shown in thin black lines in the $\alpha$ orientation maps (Figs. 9(a) and 9(d)). It is clear that the amount of such $\alpha$ grain boundaries without K-S inter-variant relationship is much

![Fig. 8](image_url)

![Fig. 9](image_url)
higher in the N-added alloy, which is quantitatively represented in the misorientation angle profiles in Figs. 9(c) and 9(f), respectively. The fraction of these α grain boundaries without K-S inter-variant relationship is increased from at least 31% to at least 48% by N addition. Similar results were also confirmed by checking several other γ grains in these two alloys. Remember that these values are not directly related to the volume fraction of α with or without K-S ORs. However, such tendency clearly represents the randomization of α′/γ crystallography by N addition.

Figures 10(a) and 10(b) show the α orientation maps of the N-free and N-added alloys isothermally transformed at 873 K for 30 s and 60 s, respectively, with similar transformed fraction to show the difference in α′/γ crystallography at the early stage of α transformation. Each α grain was identified and differentiated from each other based on its orientation. Circles and triangles indicate near K-S and non-K-S ORs of α grains with adjacent γ grains. The fraction of α grains holding non-K-S OR with γ grains on both sides is relatively higher in the N-added alloy. After analyzing α grains formed at various γ grain boundaries following the same method in our previous report,8,9 more statistical results showing the distribution of ORs are shown in Figs. 10(c) and 10(d), respectively. Δθ1 and Δθ2 (Δθ1 < Δθ2) in these two figures represent Δθ between α grain formed at γ grain boundary and adjacent two γ grains on each side, while N values are the number of analyzed α grains. Table 2 summarizes the number fraction of analyzed α grains holding near K-S OR with both, either and neither of the two adjacent γ grains, corresponding to the red, green and blue regions in the graphs, respectively. The main difference is the fraction of α grains holding non-K-S OR with both sides, increased from 24% to 43% by N addition. In addition, about half of the α grains formed inside the γ grains hold non-K-S OR with the γ (Fig. 10(b)). Therefore, more α′/γ interface with non-K-S OR exists in the N-added alloy from the early stage of transformation. Table 3 summarizes the number fraction of migrating interface among all the interface with near K-S and non-K-S ORs (either side of α grains formed at PAGB is counted as one interface), as judged from the optical micrographs of the alloys transformed at 873 K. Both interfaces with near K-S and non-K-S ORs can migrate at this temperature and no large difference can be observed between these two alloys. Therefore, larger fraction of α grains with non-K-S OR in the N-added alloys should be mainly due to higher fraction of α′/γ interface with non-K-S OR, which should be determined during the nucleation of α grains.

Figure 11(a) schematically shows the nucleation of WF/BF at γ grain boundary, which migrates into the γ grain with near K-S OR. When V(C,N) is formed at γ...
grain boundary, it should hold a near cube-cube OR with adjacent γ grain on one side and presumably an irrational OR with the other γ grain, which acts as the preferential nucleation site for α. 20) It should also be mentioned that the WF/BF formation is extremely sensitive to the α/γ OR and only slight deviation from the exact K-S OR is critical to suppress their growth. 27,28) No matter on which side of γ grain boundary is α nucleated as shown in Fig. 11(b), since α trends to hold near B-N OR with V(C,N), it will thus lose the near K-S OR with γ so that the formation of WF/BF becomes difficult. As a result, α morphology formed under this condition is changed into allotriomorphic or idiomorphic GBE. 20) The same mechanism also applies when α is nucleated on V(C,N) complexly precipitated on inclusion inside the γ grain, leading to the formation of IGF holding irrational OR with γ. 10) Therefore, the effects of N addition at 873 K should be the transition of α nucleation mode from direct nucleation at γ grain boundaries to that at V(C,N) precipitates.

By considering all the results, the strategy for better application of interphase precipitation in microalloyed low carbon steels should be the randomization of α/γ OR by suppressing the formation of WF/BF at lower transformation temperature, so that finer α matrix with better toughness as well as more finely dispersed interphase precipitation with larger strengthening potential can be formed globally in the transformed specimens. In this sense, the introduction of inclusion/precipitate, e.g. Al2O3 or V(C,N), MnS + V(C,N), MnS + CuS or Ti2O3 + TiN to act as the nucleation site for allotriomorphic or idiomorphic α, or the introduction of dislocation by deformation prior to α transformation to suppress the growth of WF/BF might also be the effective ways towards further hardening of steels strengthened by interphase precipitation.

5. Conclusions

The effects of N addition on interphase precipitation, α/γ crystallography as well as the resultant hardness in V-microalloyed low carbon steels were investigated in the present study. Main conclusions can be summarized as:

(1) After isothermal holding, VC formed by interphase precipitation are simply coarsened accompanied by the continuous softening of α grains with non K-S OR, while VC precipitation as well as hardening during post-transformation aging occurs in the α grains with near K-S OR.

(2) Particle dispersion of interphase precipitation is almost not influenced by N addition of 150 ppm at 993 K, 923 K and 873 K, although the nanohardness of α is slightly increased.

(3) The addition of N effectively enhances the formation of α grains with non K-S OR, rather than WF and BF absence of interphase precipitation at lower temperatures, which results in the significant increment in overall hardness of V-microalloyed low carbon steels.

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REFERENCES

1) T. Gladman: Mater. Sci. Technol., 15 (1999), 30.
2) Y. Funakawa, T. Shiozaki, K. Tomita, T. Yamamoto and E. Maeda: ISIJ Int., 44 (2004), 1945.
3) N. Kamikawa, Y. Abe, G. Miyamoto, Y. Funakawa and T. Furuhara: ISIJ Int., 54 (2014), 212.
4) N. Kamikawa, K. Sato, G. Miyamoto, M. Murayama, N. Sekido, K. Tsuzuki and T. Furuhara: Acta Mater., 83 (2015), 383.
5) M.-Y. Chen, M. Goune, M. Verdier, Y. Brechet and J.-R. Yang: Acta Mater., 64 (2014), 78.
6) R. W. K. Honeycombe: Metall. Trans. A, 7 (1976), 915.
7) Y.-J. Zhang, G. Miyamoto, K. Shinbo and T. Furuhara: Acta Mater., 128 (2017), 166.
8) Y.-J. Zhang, G. Miyamoto, K. Shinbo, T. Furuhara, T. Ohmura, T. Suzuki and K. Tsuzuki: Acta Mater., 84 (2015), 375.
9) Y.-J. Zhang, G. Miyamoto, K. Shinbo and T. Furuhara: Scr. Mater., 69 (2013), 17.
10) T. Furuhara, J. Yamaguchi, N. Sugita, G. Miyamoto and T. Maki: ISIJ Int., 43 (2003), 1630.
11) T. Furuhara, T. Shinyoshi, G. Miyamoto, J. Yamaguchi, N. Sugita, N. Kimura, N. Takenuma and T. Maki: ISIJ Int., 43 (2003), 2029.
12) N. K. Balliger and R. W. K. Honeycombe: Metall. Trans. A, 11 (1980), 421.
13) N. K. Balliger and R. W. K. Honeycombe: Met. Sci., 14 (1980), 121.
14) S. Zajac, T. Siwecki and M. Korczyński: Int. Symp. on Low Carbon Steels for the 90’s, ASM/TMS Materials Week, TMS, Warrendale, PA, (1993), 139.
15) S. Morito, H. Tanaka, P. Konishi, T. Furuhara and T. Maki: Acta Mater., 51 (2003), 1789.
16) G. Miyamoto, N. Iwata, N. Takayama and T. Furuhara: Acta Mater., 58 (2010), 6393.
17) D. Vannousse, A. Cerezo and P. J. Warren: Ultramicroscopy, 95 (2003), 215.
18) W. C. Oliver and G. M. Pharr: J. Mater. Res., 7 (1992), 1564.
19) J. W. Christian: The Theory of Transformation in Metals and Alloys, Pergamon, Kedlington, (2002), 751.
20) H. Oikawa: Technol. Rep. Tohoku Univ., 48 (1983), 7.
21) N. Kamikawa, M. Hirohashi, Y. Sato, E. Chandiran, G. Miyamoto and T. Furuhara: ISIJ Int., 55 (2015), 1781.
22) K. Miyata, T. Kushida, Y. Onuma and Y. Komizo: Metall. Mater. Trans. A, 34 (2003), 1565.
23) F. B. Pickering: Physical Metallurgy and the Design of Steels, Applied Science Publishers, London, (1978), 64.
24) H. I. Aaronson, P. G. Boswell and K. R. Kinsman: Mechanical Properties and Phase Transformations in Engineering Materials, TMS, Warrendale, PA, (1986), 467.
25) F. Ram, S. Saefferer, T. Japel and D. Raabe: J. Appl. Crystallogr., 48 (2015), 797.
26) G. Miyamoto, N. Takayama and T. Furuhara: Scr. Mater., 60 (2009), 1113.
27) K. Arneymaya, G. C. Weatherly and K. T. Aust: Acta Metall. Mater., 40 (1992), 1835.
28) G. Spanos, A. W. Wilson and M. V. Kral: Metall. Mater. Trans. A, 36 (2005), 1299.
29) T. Furuhara, G. Miyamoto, H. Saito and T. Maki: Proc. Int. Conf. on Solid-Solid Phase Transformation in Inorganic Materials, Vol. 1, TMS, Warrendale, PA, (2005), 5.
30) J.-H. Shim, Y.-J. Oh, J.-Y. Suh, Y. W. Cho, J.-D. Shim, J.-S. Byun and D. N. Lee: Acta Mater., 49 (2001), 2115.
31) Y. T. Chi, Y. T. Tsai, B. M. Huang and J.-R. Yang: Mater. Sci. Technol., 33 (2017), 537.
32) C. Capdevila, F. G. Caballero, C. García-Mateo and C. G. De Andres: Mater. Trans., 45 (2004), 2678.
33) T. Hanamura, H. Shibata, Y. Waseda, H. Nakajima, S. Torizuka, T. Takenashi and K. Nagai: ISIJ Int., 39 (1999), 1188.
34) P. H. Shipway and H. K. D. H. Bhadeshia: Mater. Sci. Eng. A, 223 (1997), 179.
35) R. H. Larn and J. R. Yang: Mater. Sci. Eng. A, 264 (1999), 139.