Bainite Transformation Characteristics of High-Si Hypereutectoid Bearing Steel

Zhihui Chen¹ · Jianfeng Gu¹ · Lizhan Han¹

Received: 23 August 2017 / Revised: 23 November 2017 / Accepted: 6 December 2017 / Published online: 26 December 2017
© The Author(s) 2017. This article is an open access publication

Abstract

In this work, the bainite transformation characteristics of a high-Si hypereutectoid (1.3 wt.% Si, 1 wt.% C) bearing steel was analyzed by means of microstructural observation and dilatometric experiment. The results conclude that bainitic transformation occurs by the displacive mechanism at 250–350 °C. However, suppression of carbide precipitation from austenite, caused by the addition of 1.3 wt.% Si, leads to an increase in the effective carbon available for carbon enrichment in austenite. This reduces the bainite transformation kinetics. The incomplete transformation phenomenon can be explained in terms of the diffusionless growth of bainitic ferrite subunits and the suppression of carbide precipitation from carbon-enriched austenite.

Keywords Bainite · Transformation kinetics · Incomplete transformation phenomenon · Carbon enrichment · Carbide precipitation

Introduction

The hypereutectoid steel grade AISI 52100 (1% C, 1.5% Cr) (all units are given in wt.%) steel has been widely used for manufacturing high-performance bearings for over a century [1]. In more recent developments, the addition of silicon (Si), in combination with an increased alloying content, has been recognized as a way to improve bearing performance in particulate contamination environments [2, 3]. This improvement is thought to be related to the presence of stable retained austenite (RA) [3–5]. It has been reported in the literature that RA leads to cyclic hardening through martensitic transformation in service processes of bearings [3–5]. The advantages of stabilized RA in hypereutectoid bearing steels have recently been discussed in detail by Sidoroff [6]. In order to obtain more austenite in the final microstructure, it is necessary to reduce the martensite start transformation temperature (M_s) and avoid cementite formation. The main parameters that dictate the stability of austenite are the C and Si content of the austenite [7, 8]. The former determines the chemical driving force for the transformation of austenite to

martensite. Si has been reported in the literature to suppress cementite formation from austenite [7]. It has been shown that, whereas RA was fully destabilized during tempering at 200 °C for 1 h in AISI 52100 steel, the addition of 1.3% Si could bring the destabilization tempering to above 400 °C [8].

In addition, the addition of Si to hypereutectoid steels, under appropriate isothermal transformation conditions, leads to a distinctive microstructure consisting of a mixture of bainitic ferrite and carbon-enriched RA. This microstructure results in a unique combination of strength, toughness, and ductility [9, 10]. These outstanding mechanical properties are related to that deformation-induced transformation of the metastable RA to martensite in response to the applied load. However, during the isothermal transformation, the bainitic transformation may stop prematurely before the phase achieves its equilibrium compositions in the absence of carbides precipitation at ferrite/austenite boundaries. This is interpreted differently in relation to growth mechanism of bainitic ferrite. Two different growth mechanisms, i.e., diffusionless and carbon-diffusion-controlled mechanism, have been proposed [11–16]. According to diffusionless school, bainitic ferrite retains much of the carbon content of the parent austenite. The partitioning of carbon into the residual austenite occurs immediately after the bainitic ferrite formation. In that case, the bainite transformation is expected to cease as soon as the austenite carbon content reaches the value at which diffusionless

✉ Jianfeng Gu
gujf@sjtu.edu.cn

¹ Institute of Materials Modification and Modeling, School of Materials Science and Engineering, Shanghai Jiao Tong University, Shanghai 200240, China

transformation becomes thermodynamically impossible, since the free energies of the residual austenite become less than that of the bainitic ferrite of the same composition. Another view insists that bainitic ferrite growth process is controlled by carbon diffusion in austenite and there is no essential difference between Widmanstätten ferrite and bainitic ferrite. The incomplete transformation phenomenon was caused by additional energy dissipations resulted from Gibbs energy balance, coupled solute drag effect, and thermodynamic barrier for acicular ferrite growth [13–16].

So far, the investigation on the incomplete bainite transformation mainly focuses on hypereutectoid steels with the addition of 2–3% Mn and 1–2% Si [9, 10, 17, 18]. There are only few published studies in the literature focusing on the bainitic transformation behavior in hypereutectoid steels with the addition of 1–2% Si. The aim of this work is to analyze the bainite transformation characteristics of a high-Si hypereutectoid (1.3% Si, 1% C) bearing steel.

Experimental Procedure

The chemical composition of high-Si hypereutectoid bearing steel, that is designed based on AISI 52100 bearing steel, is given in Table 1. The studied steel contains Si to prevent the precipitation of carbide during bainitic transformation. The initial microstructure of the studied steel exhibits a ferrite matrix with a certain amount of spherical cementite particles.

Dilatometric analyses of isothermal transformation have allowed the study of the transformation kinetics and the incomplete transformation phenomenon in the studied steel. A dilatometer with a resolution of 0.05 $\mu\text{m}/0.05\text{ }^\circ\text{C}$ has been selected for that purpose. Cylindrical dilatometric samples of 4 mm in diameter and 10 mm in length are austenitized at 1100 $^\circ\text{C}$ for 15 min and then isothermally transformed at temperatures range from 250 to 350 $^\circ\text{C}$ for different times before quenching.

Specimens for microstructural observation were mechanically ground, polished with 3- and 1- μm diamond polishing slurry, and then etched in a 10 vol% sodium metabisulfite solution. Optical microscope (OM) and scanning electron microscope (SEM) were used to observe the resulting microstructures. Specimens for electron backscatter diffraction (EBSD) examination were prepared by mechanical grinding and polishing with a 0.05- μm diamond polishing slurry and then vibrative polishing with a colloidal silica slurry for 2 h. EBSD measurements were performed using the following conditions: acceleration voltage of 20 kV, working distance of 5 mm, tilt angle of 70 $^\circ$, and step size of 50 nm. Transmission electron

microscope (TEM) samples (thin disks) with the diameter of 3 mm were mechanically thinned to 50 μm using 4000-grit SiC paper and then electro-polished with a twin-jet electro-polisher at $-20\text{ }^\circ\text{C}$ in a solution of 4% perchloric acid and 96% ethanol (by volume) at 50 kV until perforation occurred. The TEM machine operating at 200 kV was used to examine the microstructure at higher magnification.

Quantitative X-ray diffraction (XRD) analysis was used to determine the volume fraction and carbon content of the RA in the steel. After grinding and final polishing with 0.05- μm diamond polishing slurry, the samples were etched in hydrochloric acid for 30 s in order to minimize the possible errors originating from the polishing process. An X-ray diffractometer using unfiltered Cu $K\alpha$ radiation and operating at 40 kV and 30 mA was used. The RA amount was calculated from the integrated intensities obtained from the (200) α , (211) α , (200) γ , and (220) γ diffraction peaks according to Ref. [19]. The carbon content of RA was calculated using [20–23]:

$$a_\gamma = 3.556 + 0.045x_{\text{C}} + 0.00095x_{\text{Mn}} + 0.006x_{\text{Cr}} - 0.00057x_{\text{Ni}}, \quad (1)$$

where a_γ is in angstrom and x_{C} , x_{Mn} , x_{Cr} , and x_{Ni} are in wt.%. The presence of Si does not significantly influence the austenite lattice parameter within the experimental accuracy [24]. The average dislocation density in ferrite (ρ^a) was also evaluated using XRD line profile analysis (XLP) [25]. Selected reflection of (211) α peak was slowly scanned with a velocity of 0.2 $^\circ/\text{min}$ so as to obtain sufficiently high diffraction intensity and the good-shaped peak profile.

Experimental Results and Discussion

Microstructural Characterization

Figure 1 shows dilatations of the high-Si hypereutectoid bearing steel during holding at different temperatures. The transformation finished after holding at 250, 300, and 350 $^\circ\text{C}$ for 30, 8, and 5 h, respectively. However, the dilatometric amount in length induced by isothermal transformation increased with the decreasing transformation temperature, ensuring the transformation was finished. This implies that the deceleration of transformation kinetics is related to the decrease in transformation temperature. Figure 2 shows the microstructural evolutions of the high-Si hypereutectoid bearing steel transformed at 250, 300, and 350 $^\circ\text{C}$. After the transformation finished, the microstructure mainly consists of needle-type bainite and retained austenite (RA) with block morphology, as shown in

Table 1 Chemical composition (in wt.%) of high-Si hypereutectoid and AISI 52100 bearing steels

Element	C	Si	Mn	Cr	Ni	Mo	Fe
High-Si hypereutectoid	1.0	1.30	0.45	1.50	0.20	0.28	Bal.
AISI 52100	1.0	0.08	0.26	1.5	0.1	0.05	Bal.

Fig. 1 Dilatation of high-Si hypereutectoid bearing steel during holding at different temperatures after austenitizing at 1100 °C for 15 min

Fig. 2b, d, and f. However, after the partial transformation, the microstructure includes bainite, RA, and some martensite, as shown in Fig. 2a, c, and e. The experimentally determined martensite start temperature (M_s) at a rate of 20 °C/s after austenitizing at 1100 °C for 15 min and then isothermal transformation at different conditions is listed in Table 2. It indicates that the experimentally determined M_s is reduced as compared to that during directly quenching at a rate of 20 °C/s (120 °C), implying that the isothermal transformation of austenite contributes to the thermal stability of untransformed austenite improved.

Figure 3 shows the SEM and TEM micrographs of the high-Si hypereutectoid bearing steel transformed at 250, 300, and 350 °C for 12 h. The microstructures mainly consist of bainitic ferrite plates and RA with block and film morphology. Gong et al. [26] suggested that short-range diffusion of carbon atoms from bainitic ferrite plate to austenite film occurs. Hence, two populations of austenite with different carbon concentrations exist. It should also be emphasized

Fig. 2 OM microstructural images of high-Si hypereutectoid bearing steel transformed at 250 °C for 3 h (a) and 48 h (b), 300 °C for 1 h (c) and 12 h (d), and 350 °C for 1 h (e) and 12 h (f)

Table 2 Experimentally determined M_s (°C) at a rate of 20 °C/s after austenitizing at 1100 °C for 15 min and then isothermal transformation at different conditions

Temperature (°C)	250	300	350
Holding time (h)	5	1.5	1
M_s (°C)	54 ± 5	66 ± 5	92 ± 5

Fig. 3 SEM (a, c, d) and TEM (b) microstructural images of high-Si hypereutectoid bearing steel transformed at 250 °C (a, b), 300 °C (c), and 350 °C (d) for 12 h

that the TEM image (Fig. 3b) confirms the absence of precipitation at ferrite/austenite boundaries. Thus, it infers carbide precipitation is significantly suppressed during the isothermal transformation. As indicated by previous investigators, this results from the presence of Si, which leads to suppressed cementite precipitation in steels [27]. This is generally explained by the fact that the cementite, when it forms under paraequilibrium conditions, traps the Si as it grows. The resulting reduction in the free energy change of the reaction slows down the kinetics of precipitation. The cementite can then only form with the partitioning of Si [28]. Thus, it is suggested that the addition of 1.3 wt.% Si to hypereutectoid bearing steel suppresses the precipitation of carbide during bainite transformation. These results demonstrate that the bainite transformation during holding at 250–350 °C has incomplete transformation phenomenon, which is defined as the temporary cessation of bainitic ferrite formation before the fraction of austenite transformed to ferrite, allowed by the lever rule in the absence of carbides precipitation at ferrite/austenite boundaries, is reached [29].

RA carbon content (wt.%) and its fraction (vol%) determined by XRD analysis in the high-Si hypereutectoid steel transformed at different temperatures, ensuring the transformation was finished, are listed in Table 3. It is clear that the extent of transformation is a sensitive function of

temperature and this is totally characteristic of the incomplete transformation phenomenon. The volume fraction and carbon content of the RA gradually decreased and increased with the transformation temperature reduced, respectively. In the suppression of carbide precipitation, however, a simple mass balance estimate of the carbon suggests that the carbon content of bainitic ferrite should be significantly higher than the paraequilibrium value during bainite formation at low temperature.

Bainite Transformation Characteristic

In order to analyze the crystallographic orientation relationship (OR) between bainite and its parent austenite, the possible ferrite/austenite OR was applied. Table 4 shows the possible ferrite/austenite OR, including Kurdjumov–Sachs (K-S), Nishiyama–Wassermann (N-W), Pitsch and Bain ORs [30, 31]. Figure 4 shows the inverse pole figure (IPF) maps of bainitic ferrite (bcc) and austenite (fcc) and the OR maps of the interfaces between bcc and adjacent fcc in the high-Si hypereutectoid bearing steel transformed at 250, 300, and 350 °C. The tolerance angle in this analysis was set as 3°. The IPF images of bcc, as shown in Fig. 4a1, a2, and a3, clearly reveal that each bainitic ferrite sheaf has one single crystallographic variant, although it is possible to observe more than one variant within a single austenite grain. Remarkably, the OR of the interfaces between bcc and adjacent fcc is close to the N-W and K-S ORs, which is indicated by red and blue lines shown in Fig. 4c1, c2, and c3, respectively.

Figure 5 shows the TEM dark-field images on bainitic ferrite and austenite together with selected area diffraction pattern (SAD). SAD in Fig. 5c and dark-field images taken from bainitic ferrite and austenite in Fig. 5a and b, respectively, indicate $(110)\alpha//[(111)\gamma]$ and $[1\bar{1}\bar{1}]\alpha//[1\bar{1}0]\gamma$, which is described as K-S OR [30, 31]. Therefore, the isothermal transformation of austenite at 250–350 °C is inferred to be bainitic and take places by the displacive (diffusionless) mechanism. Again, experimental results on the temporary cessation of bainitic ferrite formation in high-Si hypereutectoid bearing steel confirm that the incomplete transformation phenomenon can be explained in terms of the diffusionless growth of bainitic ferrite subunits. In addition, carbon enrichment in the untransformed austenite occurs immediately after the bainitic ferrite

Table 3 RA carbon content (wt.%) and its fraction (vol%) determined by XRD analysis in the high-Si hypereutectoid bearing steel transformed at different temperatures, ensuring the transformation was finished

Temperature (°C)	250	300	350
Carbon content of RA (wt.%)	1.6 ± 0.1	1.5 ± 0.1	1.3 ± 0.1
Fraction of RA (vol%)	25 ± 2	45 ± 2	66 ± 2

Table 4 Possible orientation relationship (OR) between austenite (γ) and ferrite (α) [30, 31]

OR	Bain	K-S	N-W	Pitsch
Plane	$\{010\}_{\gamma} // \{010\}_{\alpha}$	$\{111\}_{\gamma} // \{110\}_{\alpha}$	$\{111\}_{\gamma} // \{110\}_{\alpha}$	$\{001\}_{\gamma} // \{\bar{1}01\}_{\alpha}$
Direction	$\langle 001 \rangle_{\gamma} // \langle 101 \rangle_{\alpha}$	$\langle \bar{1}\bar{1}0 \rangle_{\gamma} // \langle 1\bar{1}1 \rangle_{\alpha}$	$\langle 0\bar{1}1 \rangle_{\gamma} // \langle 001 \rangle_{\alpha}$	$\langle 110 \rangle_{\gamma} // \langle 111 \rangle_{\alpha}$

Fig. 4 bcc (a1, a2, a3) and fcc inverse pole figure (IPF) maps (b1, b2, b3) and OR maps (c1, c2, c3) of the interfaces between bcc and adjacent fcc in the high-Si hypereutectoid bearing steel transformed at 250 °C (a1, b1, c1), 300 °C (a2, b2, c2), and 350 °C (a3, b3, c3).

Red and blue lines in (c1, c2, c3) are represented as N-W and K-S OR of the interfaces between bcc and adjacent fcc, respectively (Color figure online)

formation. In the suppression of carbide precipitation from austenite, carbon enrichment in RA, as shown in Table 3, is a manifestation of the formation of bainitic ferrite with

a full supersaturation of carbon followed by carbon partitioning between bainitic and austenite. However, the driving force for the formation of bainitic ferrite plates decreases

Fig. 5 TEM microstructure of high-Si hypereutectoid bearing steel transformed at 300 °C, where (a) is the dark-field image taken from $[1\bar{1}1]\gamma$ reflection, (b) is the dark-field image taken from the $[10\bar{1}]\alpha$ reflection, and (c) is the SAD pattern showing the K-S OR: $(110)\alpha// (111)\gamma$, $[1\bar{1}1]\alpha//[110]\gamma$

as the carbon concentration in the untransformed austenite approaches a value, at which the free energy of ferrite and austenite phases of the same composition become identical, resulting in the displacive transformation becomes thermodynamically impossible [11].

The dislocation density of bainitic ferrite formed at 250, 300, and 350 °C was calculated to be 4.1×10^5 , 3.5×10^5 , and $2.0 \times 10^5 \text{ m}^{-2}$, respectively. The relatively high dislocation density within bainitic ferrite is often attributed to the fact that the shape deformation accompanying the displacive transformation is accommodated at least partially by plastic relaxation [32, 33]. According to the tempering theory of Kalish and Cohen [34], however, carbon atoms are energetically favorable to remain segregated at dislocation compared with their partitioning into surrounding austenite or their presence in the carbide lattice. Hence, if the dislocation density within bainite ferrite is high, sufficient carbon can be captured by dislocation. Atom probe tomography results reported [35] elsewhere revealed that a substantial quantity of carbon (7.4 at% C) was trapped at dislocations in the vicinity of the ferrite/austenite interface in NANOBAIN steel. These results indicate that the bainite ferrite formed at lower temperature has higher carbon content. Thus, it is demonstrated that despite the significant fraction of bainite formed at 250 °C, the RA does not seem greatly enriched in carbon compared to that at 350 °C.

Effect of Si on the Deceleration of Bainite Transformation

Figure 6 shows the bainite transformation kinetics of the high-Si hypereutectoid and AISI 52100 bearing steels during holding at 300 °C. The transformation kinetics of the high-Si hypereutectoid bearing steel is much more sluggish compared to that of the AISI 52100 bearing steel. This implies that the bainite transformation is obviously decelerated by the addition of 1.3 wt.% Si in hypereutectoid bearing steel, consistent with the results presented in Ref. [36]. Figure 7 shows the SEM morphology of the AISI 52100 bearing steel transformed at 300 °C. The microstructural image reveals that carbide can significantly precipitate and austenite decomposes into ferrite and carbide during holding

Fig. 6 Bainite transformation kinetics of high-Si hypereutectoid and AISI 52100 steels during holding at 300 °C

Fig. 7 SEM microstructural image of AISI 52100 bearing steel transformed at 300 °C for 6 h

at 300 °C. However, the incomplete transformation phenomenon is not well pronounced in the AISI 52100 bearing steel, where carbide precipitation dominates.

Carbon enrichment of austenite leads to the precipitation of carbides which implies that the effective carbon enrichment of austenite is negligible. However, in the high-Si hypereutectoid bearing steel, the carbide precipitation is kinetically suppressed, leading to significant carbon enrichment of austenite associated with the carbon partitioning after bainite formation. Accordingly, once Si exerts a suppressing effect on the carbide precipitation in austenite, the formation of bainitic ferrite will lead to carbon enrichment of austenite adjacent to the ferrite plates. The higher carbon content of austenite lowers the driving force for the subsequent formation of ferrite plates, and the sympathetic nucleation rate of ferrite decreases [37]. The experimental result confirms that the deceleration of bainite transformation kinetics of the high-Si hypereutectoid bearing steel is related to the suppression of carbide precipitation from austenite. Therefore, the suppression of carbide precipitation is also interpreted as one of the necessary conditions for manifestation of incomplete transformation phenomenon because the consumption of carbon atoms to form carbides makes carbon enrichment in austenite difficult, and thus cessation of bainitic ferrite formation is also difficult.

Conclusions

The present study investigated isothermal transformation of austenite at 250–350 °C with a high-Si hypereutectoid (1.3% Si, 1% C) bearing steel. The results conclude that the isothermal transformation of austenite at 250–350 °C is bainitic and occurs by the displacive mechanism. Carbide precipitation during the transformation is significantly suppressed from austenite, resulting from the addition of 1.3 wt.% Si. The result leads to the effective carbon available for carbon enrichment in austenite increased and hence decelerates the bainite transformation kinetics as it progresses. As a result, a bainitic microstructure consisting of nano-bainitic ferrite plates and carbon-enriched RA films is obtained during the transformation. The experimental results on the temporary cessation of bainitic ferrite formation in the high-Si hypereutectoid bearing steel confirm that the incomplete transformation phenomenon can be explained in terms of the diffusionless growth of bainitic ferrite subunits and the suppression of carbide precipitation from carbon-enriched austenite.

Open Access This article is distributed under the terms of the Creative Commons Attribution 4.0 International License (<http://creativecommons.org/licenses/by/4.0/>), which permits unrestricted use, distribution, and reproduction in any medium, provided you give appropriate credit to the original author(s) and the source, provide a link to the Creative Commons license, and indicate if changes were made.

References

1. H.K.D.H. Bhadeshia, Steels for bearings. *Prog. Mater. Sci.* **57**(2), 268–435 (2012)
2. P. Daguer, O. Roesch, W. Trojahn, New development in through hardened bearing steel grades for use in contaminated lubrication. *J. ASTM Int.* **3**(6), 131–139 (2006)
3. P. Daguer, G. Baudry, J. Bellus, G. Auclair, J. Rofès-Vernis, G. Dudragne, D. Girodin, G. Jacob, Improved bearing steel for applications involving debris, higher loads and temperatures. *Bear. Steel Technol. ASTM Int.* **1419**, 320–329 (2002)
4. R.H. Richman, R.W. Landgraf, Some effects of retained austenite on the fatigue resistance of carburized steel. *Metall. Mater. Trans. A* **6**(5), 955–964 (1975)
5. C.A. Stickels, Plastic deformation of quenched and tempered 52100 bearing steel in compression. *Metall. Mater. Trans. A* **8**(1), 63–70 (1977)
6. C. Sidoroff, M. Perez, P. Dierickx, D. Girodin, Advantages and shortcomings of retained austenite in bearing steels: a review. *Bear. Steel Technol. ASTM Int.* **10**, 312–348 (2015)
7. E. Kozeschnik, H.K.D.H. Bhadeshia, Influence of silicon on cementite precipitation in steels. *Mater. Sci. Technol.* **24**(3), 343–347 (2008)
8. T. Sourmail, M. Millot-Méheux, Thermal and mechanical stability of retained austenite in 1%C bearing steels with varying Si contents. *Mater. Sci. Technol.* **32**(11), 1126–1132 (2016)
9. F.G. Caballero, H.K.D.H. Bhadeshia, Very strong bainite. *Curr. Opin. Solid State Mater.* **8**(3–4), 251–257 (2004)
10. F.G. Caballero, H.K.D.H. Bhadeshia, K.J.A. Mawella, D.G. Jones, P. Brown, Design of novel high strength bainitic steels: part 2. *Mater. Sci. Technol.* **17**(5), 517–522 (2001)
11. H.K.D.H. Bhadeshia, D.V. Edmonds, The mechanism of bainite formation in steels. *Acta Metall.* **28**(9), 1265–1273 (1980)
12. H. Bhadeshia, A rationalisation of shear transformations in steels. *Acta Metall.* **29**(6), 1117–1130 (1981)
13. H. Chen, K. Zhu, L. Zhao, S. van der Zwaag, Analysis of transformation stasis during the isothermal bainitic ferrite formation in Fe–C–Mn and Fe–C–Mn–Si alloys. *Acta Mater.* **61**(14), 5458–5468 (2013)
14. H.I. Aaronson, W.T.R. Jr, G.R. Purdy, Coupled-solute drag effects on ferrite formation in Fe–C–X systems. *Metall. Mater. Trans. A* **35**(4), 1187–1210 (2004)
15. H.I. Aaronson, W.T.R. Jr, G.R. Purdy, The incomplete transformation phenomenon in steel. *Metall. Mater. Trans. A* **37**(6), 1731–1745 (2006)
16. M. Hillert, L. Höglund, J. Ågren, Role of carbon and alloying elements in the formation of bainitic ferrite. *Metall. Mater. Trans. A* **35**(12), 3693–3700 (2004)
17. C. Garcia-Mateo, F.G. Caballero, Ultra-high-strength bainitic steels. *ISIJ Int.* **45**(11), 1736–1740 (2005)
18. F.G. Caballero, H.K.D.H. Bhadeshia, K.J.A. Mawella, D.G. Jones, P. Brown, Very strong low temperature bainite. *Mater. Sci. Technol.* **18**(3), 279–284 (2002)
19. B.D. Cullity, *Elements of X-ray Diffraction* (Addison-Wesley Publishing Company Inc., Massachusetts, 1956)
20. C.P. Scott, J. Drillet, A study of the carbon distribution in retained austenite. *Scr. Mater.* **56**(6), 489–492 (2007)
21. N.H. van Dijk, A.M. Butt, L. Zhao, J. Sietsma, S.E. Offerman, J.P. Wright, S. van der Zwaag, Thermal stability of retained austenite in TRIP steels studied by synchrotron X-ray diffraction during cooling. *Acta Mater.* **53**(20), 5439–5447 (2005)
22. E. Jimenez-Melero, N. Van Dijk, L. Zhao, J. Sietsma, S. Offerman, J. Wright, S. Van der Zwaag, Characterization of individual retained austenite grains and their stability in low-alloyed TRIP steels. *Acta Mater.* **55**(20), 6713–6723 (2007)

23. E. Jimenez-Melero, R. Blondé, M.Y. Sherif, V. Honkimäki, N.H. van Dijk, Time-dependent synchrotron X-ray diffraction on the austenite decomposition kinetics in SAE 52100 bearing steel at elevated temperatures under tensile stress. *Acta Mater.* **61**(4), 1154–1166 (2013)
24. D.J. Dyson, B. Holmes, Effect of alloying additions on the lattice parameter of austenite. *J. Iron Steel Inst.* **280**, 469–474 (1970)
25. W. Li, W. Xu, X. Wang, Y. Rong, Measurement of microstructural parameters of nanocrystalline Fe–30 wt%Ni alloy produced by surface mechanical attrition treatment. *J. Alloys Compd.* **474**(1), 546–550 (2009)
26. W. Gong, Y. Tomota, S. Harjo, Y.H. Su, K. Aizawa, Effect of prior martensite on bainite transformation in nanobainite steel. *Acta Mater.* **85**, 243–249 (2015)
27. H. Bhadeshia, D. Edmonds, The bainite transformation in a silicon steel. *Metall. Trans. A* **10**(7), 895–907 (1979)
28. H.D. Wu, G. Miyamoto, Z.G. Yang, C. Zhang, H. Chen, T. Furuhara, Incomplete bainite transformation in Fe–Si–C alloys. *Acta Mater.* **133**, 1–9 (2017)
29. R.F. Hehemann, K.R. Kinsman, H.I. Aaronson, A debate on the bainite reaction. *Metall. Mater. Trans. B* **3**(5), 1077–1094 (1972)
30. P.P. Suikkanen, C. Cayron, A.J. DeArdo, L.P. Karjalainen, Crystallographic analysis of martensite in 0.2C–2.0Mn–1.5Si–0.6Cr steel using EBSD. *J. Mater. Sci. Technol.* **27**(10), 920–930 (2011)
31. P.P. Suikkanen, C. Cayron, A.J. DeArdo, L.P. Karjalainen, Crystallographic analysis of isothermally transformed bainite in 0.2C–2.0Mn–1.5Si–0.6Cr steel using EBSD. *J. Mater. Sci. Technol.* **29**(4), 359–366 (2013)
32. F.G. Caballero, C. Garcia-Mateo, M. Santofimia, M.K. Miller, C.G. De Andres, New experimental evidence on the incomplete transformation phenomenon in steel. *Acta Mater.* **57**(1), 8–17 (2009)
33. I.B. Timokhina, H. Beladi, X.Y. Xiong, Y. Adachi, P.D. Hodgson, Nanoscale microstructural characterization of a nanobainitic steel. *Acta Mater.* **59**(14), 5511–5522 (2011)
34. D. Kalish, M. Cohen, Structural changes and strengthening in the strain tempering of martensite. *Mater. Sci. Eng.* **6**(3), 156–166 (1970)
35. F.G. Caballero, M.K. Miller, S.S. Babu, C. Garcia-Mateo, Atomic scale observations of bainite transformation in a high carbon high silicon steel. *Acta Mater.* **55**(1), 381–390 (2007)
36. Y. Toji, H. Matsuda, D. Raabe, Effect of Si on the acceleration of bainite transformation by pre-existing martensite. *Acta Mater.* **116**, 250–262 (2016)
37. W.T. Reynolds, S.K. Liu, F.Z. Li, S. Hartfield, H.I. Aaronson, An investigation of the generality of incomplete transformation to bainite in Fe–C–X alloys. *Metall. Trans. A* **21**(6), 1479–1491 (1990)