Enhanced mechanical properties of ultrahigh strength Mn–Si–Cr–C steels treated by a novel bainitic transformation plus quenching and partitioning process

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ABSTRACT

In this paper, we attempt to review our recent works on the ultrahigh strength Mn–Si–Cr–C steels treated by BQ&P and BQ–P–T processes. The core of these novel processes is to incorporate the formation of carbide-free bainite during the initial quenching step (rather than partitioning step) of Q&P process and to achieve a fine multiphase microstructure (bainite + martensite + retained austenite). Our studies showed that the enhanced ductility, toughness and fatigue properties were achievable in the ultrahigh strength BQ&P and BQ–P–T steels, which is mainly attributed to the refined multiphase microstructures.

KEYWORDS

Bainite; quenching and partitioning; ductility; toughness; fatigue

1. Introduction

Quenching and partitioning (Q&P) process has been proposed to produce high strength steels, which exhibit higher ductility than martensitic steels after conventional quenching and tempering (Q&T) treatment [1,2]. The Q&P concept involves an initial cooling step to a quenching tempering (Tq) between the martensite-start temperature (Mt) and the martensite-finish temperature (Mf), followed by an isothermal holding at Tq (or higher than Tq) with the purpose of promoting carbon transport from martensite into austenite, thereby obtaining more retained austenite at room temperature. Q&P steels with a multiphase microstructure of carbon-depleted martensite and carbon-enriched retained austenite exhibit an excellent combination of strength and ductility due to the transformation induced plasticity (TRIP) effect of the retained austenite. Many investigations into Q&P steels have focused on improving strength and ductility by adjusting alloying and processing routes, and also the mechanism of deformation [3–5].

The quenching–partitioning–tempering (Q–P–T) process was proposed by Hsu to pursue the potential of precipitation strengthening in Q&P steels [6]. The Q–P–T concept introduces a tempering step after quenching and partitioning so that the fine carbides can be precipitated during tempering step in the steels with carbide forming elements additions (such as Nb, Ti and V). The effect of fine carbides on mechanical properties of Q–P–T steels has been reported[7,8].

Besides martensite, austenite and carbide, bainite microstructures have also been recognized as a potential constituent in Q&P steels because the range of partitioning temperature normally enables the bainite formation from austenite [9–10]. Due to the addition of Si and Al, it is reported that bainite transformation in Q&P steels happens with absence of cementite (i.e. bainitic ferrite plus retained austenite). Thus, similarly to TRIP-assisted steels, bainite reaction in Q&P steels could contribute to carbon enrichment of the surrounding austenite [10]. However, the bainite formation including either carbide-free bainite or lower bainite (bainitic ferrite plus transitional epsilon carbide) that happens during a longer partitioning step was considered as a competing process against carbon partitioning from martensite into austenite and thereby reduces the fraction of austenite in the final product [9, 11]. Hence, it is a pity that the effect of bainite formation was not pursued in Q&P or P–T processes.

Based on the carbide-free bainite/martensite multiphase steels, we found that the optimization of properties can be achieved by designing Q&P and/or Q–P–T processes combined with bainite transformation [12,13]. The core of these novel processes is to incorporate the formation of carbide-free bainite during the initial quenching step (rather than partitioning step) of Q&P process (abbreviated as BQ&P or BQ–P–T) and to achieve a fine multiphase microstructure (bainite + martensite + retained austenite). The Schematic Q&P treatment is shown in Figure 1. Compared
with the conventional Q&P and Q–P–T processes, the BQ&P and BQ–P–T processes have the following unique advantages: (i) the prior-formed bainite plates divide the untransformed austenite; consequently, the martensite and retained austenite can be refined, leading to the formation of sub-micrometer-sized and nanometer-sized retained austenite films [12]. (ii) During the formation of bainite, the carbon atoms diffuse to untransformed austenite from bainitic ferrite, leading to the additional carbon enrichment of surrounding austenite [14]. (iii) Two types of retained austenite can be achieved: type I located between bainitic ferrite laths, type II located between carbon-depleted martensite laths [14,15]. (iv) The partitioning time can be prolonged in the range 30–120 min, which can be controlled easily for large-scale steel components (the partitioning time of conventional Q&P process is only tens of seconds) [16].

In this paper, we make an attempt to review our recent works on the improved ductility, toughness and fatigue properties of the ultrahigh strength Mn–Si–Cr–C series bainite/martensite steels treated by the BQ&P or BQ–P–T processes. The applications of BQ&P process in bainitic rail steels and hollow rock drill steels are also reviewed.

2. Enhanced ductility of BQ&P and BQ–P–T steels

Ongoing third-generation AHSS research is focused on increasing strength and/or ductility to higher levels without increasing the costs of these alloys [17–19]. A multiphase steel composed of austenite plus martensite is considered as a suitable candidate for the third-generation AHSSs. Their properties can be optimized once the volume fraction and stability of austenite are controlled properly.

We compared the BQ&P treatment with the conventional Q&T and bainitic austempering (BAT) treatments on a medium carbon steel. The chemical composition of the experimental steel was 0.4C–2.0Mn–1.7Si–0.4Cr (wt.%), with enhanced Si addition to suppress carbide precipitation during quenching, partitioning and tempering steps. The BQ&P treatment comprised austenitizing at 880°C for 45 min followed by cooling to 200°C (Ms = 252°C) at 1°C/s and subsequent partitioning step at 360°C for 45 min. The BQ&T (bainite-based quenching and tempering) treatment comprised cooling to ambient temperature at 1°C/s, and tempering at 280°C for 120 min. The BAT treatment was designed: after austenitization, the specimen was cooled down to 360°C at 1°C/s (without the occurrence of bainite transformation during cooling), and austempering for 120 min (to ensure a sufficiently long period for isothermal bainite transformation).

Figure 2 showed the secondary electron images of the BQ&T, BAT and BQ&P microstructures. The BQ&T microstructure is bainite/martensite multiphase (Figure 2a). The leaf-shaped bainite (width: 1–2 μm) is formed during the initial quenching (BQ) step. Previous studies showed that filmy retained austenite (width: 20–100 nm) may distribute within and/or between...
these leaf-shaped bainitic ferrite [16]. The BAT microstructure is composed of lath-shaped bainite and blocky martensite/austenite (M/A) islands (Figure 2b). The lath-shaped bainite is formed during austempering at 360°C, which also contains bainitic ferrite and filmy retained austenite. The BQ&P microstructure contains bainite, martensite and retained austenite (Figure 2c). Two types of bainite are found in the BQ&P sample: (i) the leaf-shaped bainite formed during the initial quenching step, (ii) the lath-shaped bainite formed during partitioning at 360°C. Here we consider a portion of the retained austenite is formed accompanying bainite transformation (i.e. considered as part of bainite). The other portion is stabilized during carbon partitioning (from martensite to austenite) and remains untransformed at room temperature. It is noted that the blocky M/A islands in BAT variant are not found in the BQ&P sample.

The total volume fraction of retained austenite (RA) and its average carbon content measured by XRD are shown in Figure 3. Compared to the BQ&T sample, the BAT and BQ&P samples contain a higher volume fraction of RA with higher carbon content. It indicates that carbon enrichment in austenite and the resultant austenite stabilization can be realized during isothermal bainite transformation and/or carbon partitioning from martensite to austenite. The TEM images of the BAT sample confirm the presence of the submicron-sized filmy RA (0.2–1 μm, Figure 4a) and micron-sized blocky RA (>1 μm, Figure 4b). It can be inferred from the submicron-sized (0.1–0.5 μm, Figure 4c) and nanometer-sized (20–100 nm, Figure 4d) filmy RA in the BQ&P sample that the filmy RA in BQ&P sample is much finer. Also, the micron-sized blocky RA is not observed by TEM in the BQ&P sample.

The mechanical properties of the BQ&T, BAT and BQ&P samples are listed in Table 1. Compared to the BQ&T sample with low RA (<10 vol.%), the BAT and BQ&P samples with relatively high RA (∼22 vol.%) show better ductility. This suggests the great contribution of RA to ductility. The product of strength and elongation (PSE) obtained for the BQ&P sample reaches to 42.6 GPa%, which is two to three times higher than that of the conventional martensitic steels [20].

Moreover, we found that the ductility could be further improved through tempering after BQ&P treatment. The XRD results combined with CCE (constrained carbon equilibrium) calculation proved that carbon partitioning from martensite to austenite also occurred during tempering. Hence, the thermodynamic stability of retained austenite was enhanced. This ensured a considerable amount of retained austenite and the subsequent TRIP effect [13]. If the ε-carbide/cementite precipitation could be suppressed by alloying during tempering, tempering could be considered as a secondary partitioning step in the Q&P process. Figure 5 showed the carbon content of...
retained austenite after the Q&P and Q–P–T processes calculated by the CCE model taking account of tempering (secondary partitioning step). This idea can be utilized to design novel BQ–P–T process. Consequently, we obtained a high PSE of 44.5 GPa% in a medium carbon steel treatment by the BQ–P–T process. We also focused on the effect of bainitic transformation on the ductility of the BQ&P steels. Here we controlled the fraction, morphologies, size and distribution of the carbide-free bainite during the BQ step through controlling austempering at various bainitic transformation temperatures. During BQ&P treatment, the carbon enrichment of austenite is achieved through two mechanisms: carbon partitioning from bainitic ferrite to austenite during bainitic austempering, and carbon partitioning from martensite to austenite during partitioning step. The retained austenite in the final product comprised two types: type I located between bainitic ferrite laths, type II located between carbon-depleted martensite laths, shown in Figure 6. Hence, the volume fraction and stability of retained austenite in the final product depends on both the bainitic austempering and Q&P processes. In this context, we optimized the fraction and stability of retained austenite by controlling the BQ&P treatment and obtained an excellent combination of strength and ductility (tensile strength: 1500 MPa, total elongation: 32%) [15].

The total elongations versus tensile strength of AHSSs (advanced high strength steels) obtained by the existing approaches (combining alloy/processing design) have been summarized by Speer et al., as shown in Figure 7 [19,20]. The total elongations reported in the literature are adjusted according to the ASTM E8 standard specimen geometry [19,20]. The tensile test results of the BQ&T, BAT, BQ&P and BQ–P–T materials were also plotted in Figure 7. The elongation data of all the specimens were conversed based on ASTM E8 standard specimen geometry. It is revealed that the combinations of strength and ductility of BQ&P steels are superior to those reported until now. Therefore, the BQ&P steel is an attractive and promising candidate for the third-generation AHSSs.

### 3. Enhanced toughness of BQ&P and BQ–P–T steels

In the fields of railway, engineering machinery and pressure vessels, both ductility and toughness of AHSS have been underlined (e.g. impact toughness higher than 20 J/cm² at −20°C) to prevent sudden and brittle failure of the structures [21,22]. In this context, the carbide-free bainitic microstructures comprising of retained austenite films were mainly employed to achieve the yield strength of 1100–1300 MPa and impact toughness at −20°C of 2–40 J/ cm² [21,22]. Generally, these bainitic steels were produced by BQ&T or BAT treatments. Unfortunately, high cost elements like Mo, Ni and Co were indispensable in most cases, which limited the application of these steels. Some studies have been done to improve the toughness of steels by Q&P treatments [23–25]. For example, Hong et al. used a low carbon

| Samples | Rm (MPa) | Rp0.2 (MPa) | UEL (%) | TEL (%) | PSE (GPa%) |
|---------|----------|-------------|---------|---------|------------|
| BQ&T    | 1908 ± 5 | 1501 ± 8    | 4.1 ± 0.2 | 10.9 ± 0.3 | 20.9       |
| BAT     | 1505 ± 7 | 1218 ± 9    | 13.5 ± 0.9 | 21.2 ± 0.8 | 31.9       |
| BQ&P    | 1688 ± 4 | 1391 ± 7    | 17.2 ± 0.6 | 25.2 ± 1.1 | 42.6       |
steel to study the improvement of impact toughness by the Q&P processes [23]. Bagliani et al. showed a better combination of strength and toughness obtained by a Q&P process as compared to traditional Q&T processes in a low alloy medium carbon steel [24]. These results reveal the advantage of Q&P processes in improving the toughness of steels.

We also compared the toughness of the steels treated by BQ&P with those treated by the BQ&T and BAT processes. Figure 8 shows the impact toughness of the BQ&T, BAT and BQ&P samples. It was found that the impact toughness was improved significantly after the BQ&P treatment, as compared to the BQ&T and BAT treatments. It was revealed that the BQ&P treatment was favorable for producing multiphase microstructures that exhibit excellent toughness even at low temperature (e.g. 48 J/cm² at −40°C).

Figure 9(a and b) shows the crack propagation paths in the BAT and BQ&P samples during impact tests. For the BAT sample, the lath-shaped bainite can hinder the propagation of cracks. However, the blocky M/A islands do not have such effect. Moreover, secondary cracks are prone to originate from the blocky M/A islands near prior austenite grain boundaries. The blocky RA in the M/A islands has low stability, which could transform easily to hard fresh martensite due to the stress concentration during impact testing. Unfortunately, the hard fresh martensite does not accommodate the deformation of its surrounding matrix effectively and hence promotes the formation of secondary cracks.

In contrast, the cracks may change their propagation paths at the submicron/nanometer-sized ‘film-like M/A islands’ in the BQ&P sample (as shown by arrows in Figure 9b). The submicron/nanometer-sized filmy RA located at the ‘film-like M/A islands’ has higher stability (compared to coarse blocky RA), which can hinder the crack propagation more efficiently. Consequently, the average length of unit crack paths is reduced to 1–2 μm, and hence the impact toughness is improved. Furthermore, the leaf-shaped bainite formed during the initial quenching step can also divide the prior austenite grains and reduce the average length of unit crack paths in the BQ&P sample, which has additional benefit for toughness.

As we know, the PSE reflects the energy absorption during deformation and has been taken as the design criterion of the third-generation AHSSs. The PSE ranging from 20 GPa% to 40 GPa% has been achieved in the third-generation AHSSs [19,20]. However, we found a potential problem: an increase in ductility or PSE may not always mean a concurrent enhancement of toughness [25]. In actually, for some Q&P and bainitic steels (especially with ultrahigh tensile strength ≥1200 MPa), the impact toughness decreases unexpectedly as the PSE increases, as shown in Figure 10.

In general, the high ductility or PSE of the AHSSs is mainly attributed to the TRIP effect of RA, which depends on the amount and stability of RA [26,27]. Sufficient RA is necessary for an adequate TRIP effect.
during deformation. In the third-generation AHSSs, the RA is usually stabilized through the enrichment of alloying elements in austenite [28,29]. Then, carbon is the cheapest element and used widely to stabilize the RA. Hence, increasing the carbon content in the steel is an effective way to increase the amount of RA and ensure a TRIP effect. As shown in Figure 10, the PSE of the high carbon (0.5–0.9 wt.% bainitic steels is relatively higher compared with the bainitic and Q&P steels with low or medium carbon (0.2–0.4 wt.%). However, for the ultrahigh strength bainitic and martensitic steels, increasing the content alloying element, either interstitial elements (e.g. C and N) or substitutional elements (exception of Ni and so on), always leads to the deterioration of toughness, which is due to the high lattice distortion and internal stress in the bainite/martensite lath [30]. Tanaka et al. pointed out the impact energy of lath martensite is very sensitive to carbon content, i.e. increasing the carbon content in martensite from 0.15 to 0.2 wt.% causes decrease in impact energy of more than 40% [30]. Consequently, an increase in ductility or PSE may not always mean a concurrent enhancement of toughness, as shown in Figure 10.

Hence, we proposed the ductility and toughness could be improved concurrently by designing a multiphase steel that comprised refined ductile matrix (bainite and carbon-depleted martensite) and the nanometer-sized filmy retained austenite. The ductile matrix is generated by both the carbon partitioning from bainite/martensite to austenite and the higher degree of martensite/bainite tempering in the BQ–P–T condition. The ductile bainite and carbon-depleted martensite have relatively low carbon content and exhibit high plasticity and could absorb more energy during the growth of the voids and produce deeper dimples on the fracture surface. The filmy RA is beneficial to ductility and toughness by means of the TRIP effect, hindering the coalescence of voids or the propagation of cracks. Consequently, an excellent combination of strength, ductility and impact toughness at room temperature and low temperature was achieved after the BQ–P–T treatment (i.e. ultimate tensile strength: 1416 MPa, the PSE: ∼25.5 GPa·%, the CVN impact energy at 20°C and −40°C: ∼95 J/cm² and ∼45 J/cm²). The combined mechanical properties are positioned in a desirable region in the PSE versus CVN impact energy map for the bainitic and Q&P steels, as shown in Figure 10.

4. Enhanced fatigue properties of BQ&P and BQ–P–T steels

Recently, very high cycle fatigue (VHCF, i.e. enduring cyclic stress in excess of 10^7 cycles) behavior of metallic materials has become an important subject to ensure the long-term safety of the structural components, such as aircraft (gas turbine disks 10^10 cycles), automobiles (car engine 10^8 cycles) and railways (high speed train axles 10^9 cycles) [31–33]. Although a large amount of fatigue data has been published in the form of S–N curves, the data in the literature have been limited to fatigue lives up to 10^7 cycles. It is reported that the fatigue failures of metallic materials still happen beyond 10^7 cycles under a relatively low cyclic stress below the conventional fatigue limit [32]. The mechanism of crack initiation and early growth for VHCF is basically different from that for HCF [31–35]. For most of HCF cases, the fatigue failure is basically initiated by persistent slip bands from the surface of metallic materials. The fatigue crack initiation
within VHCF regimes is mostly from the subsurface or interior of specimens, which is characterized by so-called ‘fish-eye’ surrounding a ‘fine granular area’ (FGA) [35]. The region of FGA, as crack initiation characteristic region, consumes larger than 95% of total fatigue life [35]. To increase the knowledge about the formation mechanism of this crack initiation characteristic region, it is essential to understand the mechanism of crack initiation and early growth during VHCF regimes and develop advanced long-life metallic materials.

Here we have studied the VHCF properties of the BQ&P steels [36]. The heat treatment and consequent microstructure of the BQ&P steel was shown in Section 2 and Figure 2. Through the ultrasonic fatigue testing equipment and ‘up and down’ method, the fatigue limit strength enduring 10^9 cycles cyclic stress (σ_w9) was determined to be 875 MPa for BQ&P steel. As mentioned earlier, the tensile strength (R_m) and hardness of BQ&P samples were 1688 MPa and 545 HV, respectively. Thus the values of σ_w9/R_m and σ_w9/HV are ~0.52 and 1.61, respectively, which are significantly higher than those reported for steels in the literature (usually 0.2–0.4) [37–40], as shown in Figure 11.

We attributed the enhanced VHCF properties to the presence of ultrafine film-like retained austenite in the BQ&P steel (Figure 4). For most metallic materials, irreversible slip is the primary cause of failure. Slip accumulates during fatigue loading, as a result of dislocations movement: cross-slip, precipitates, transmitting through grain boundaries and pile up [41,42]. With more slips accumulating during cyclic loading, plastic deformation manifests into strain localization leading to crack initiation [41,42]. In the BQ&P steels, the dislocation absorption by retained austenite (DARA) effect [43] and the deformation-induced martensitic transformation (DIMT) from austenite (leading to TRIP effect) could effectively relax the stress concentration in the areas where dislocations accumulate and pile up. Hence, the strain localization and slip band formation are suppressed during cyclic loading, which potentially increase the cycle loading number before the formation of the micro-crack.

Furthermore, we obtained different types and fractions of RA (nanometer-sized film-like and micrometer-sized blocky morphology) by adjusting the BQ–P–T process parameters [44]. The effect of RA on the VHCF behaviors of the BQ–P–T steels was studied through S–N curves testing, fraction surface observation, failure mechanism and crack initiation analyses. We found that the morphology and size of RA have a significant influence on the VHCF behavior of the BQ–P–T steels. The micrometer-sized blocky RA is prone to suffer from local plastic deformation under cyclic loading, and hence the DIMT from RA occurs at lower strains. The martensite formed due to DIMT is relatively brittle and subjected to the stress concentration caused by impinging slips along the prior austenite grain boundaries, which finally induces the formation of quasi-cleavage cracks. From this perspective, the blocky RA is detrimental to the VHCF properties of BQ–P–T steel. In contrast, the nanometer-sized film-like RA does not easily become a fatigue crack initiation site under cyclic loading. Actually, the film-like RA is beneficial to the VHCF property to some degree. Compared with blocky RA, the film-like RA has relatively high mechanical stability. Hence, the DIMT from film-like RA occurs at a larger localized strain. Furthermore, the RA with FCC structure has more slip systems and higher work hardening rate compared with martensite and bainite. Hence, the RA films could absorb dislocations moving from adjacent martensite and bainite laths under cyclic loading, which enhance the deformation ability of martensite and bainite as well. With the accumulation of localized strain, the film-like RA could also transform to

![Figure 11. Summary of fatigue limit strength versus tensile strength of advanced high strength steels obtained by different approaches including bainite-based quenching and partitioning (BQ&P, present study). The cited studies were reviewed in [37–40]. (The value σ_w9/R_m of bainite-based quenching and partitioning steel is higher than other steels).](image-url)
martensite and contribute to TRIP effect, which could relieve the stress concentration and postpone plastic localization. The martensite transformed from film-like RA is small in size, and its deformation compatibility with adjacent phase is better than the blocky martensite transformed from blocky RA. Consequently, the BQ–P–T sample with film-like RA exhibits excellent VHCF property, namely, the fatigue limit in VHCF regime reaches 770 MPa while the tensile strength is 1410 MPa (the ratio of $\sigma_{w9}/R_m$ is about 0.55).

5. Conclusion

This paper makes an attempt to review our recent works on the ultrahigh strength Mn–Si–Cr–C steels treated by the bainitic transformation plus quenching and partitioning (BQ&P) processes. Our studies show that the optimization of properties (e.g. ductility, toughness and fatigue) can be achieved by designing the BQ&P and BQ–P–T processes.

(1) The PSE of the ultrahigh strength Mn–Si–Cr–C steels can be enhanced markedly (40–50 GPa%) after BQ&P and BQ–P–T treatments, which is twice of that of the conventional TRIP steels (~20 GPa%), and even falls into the PSE range (40–60 GPa%) of TWIP steels with high alloy content.

(2) The toughness at ambient and low temperature can be improved after BQ&P and BQ–P–T treatments, e.g. the CVN impact energy at 20°C and −40°C is ∼95 J/cm² and ∼45 J/cm², respectively, when the tensile strength is 1416 MPa. We also proposed a PSE versus CVN impact energy map to describe the concurrent improvement of strength, ductility and toughness.

(3) The very high cycle fatigue properties of the BQ&P steels can be optimized by adjusting the BQ&P and BQ–P–T process parameters. The ratio of $\sigma_{w9}/R_m$ is raised to 0.55 in the BQ&P and BQ–P–T steels with film-like retained austenite, which is much higher than that of conventional high strength tempered martensitic steels in the literatures (usually 0.2–0.4).

(4) The enhanced mechanical properties are mainly attributed to the refined multiphase microstructures achieved by the BQ&P and BQ–P–T treatments.

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Reference

[1] Edmonds DV, He K, Rizzo F C, et al. Quenching and partitioning martensite—a novel steel heat treatment. Mater Sci Eng: A. 2006;438:44025–34.
[2] Speer JG, Assunção FC R, Matlock DK, et al. The “quenching and partitioning” process: background and recent progress. Mater Res. 2005;8(4):417–423.
[3] Santofimia MJ, Zhao L, Petrov R, et al. Microstructural development during the quenching and partitioning process in a newly designed low-carbon steel. Acta Mater. 2011;59(15):6059–6068.
[4] Liu H, Lu X, Jin X, et al. Enhanced mechanical properties of a hot stamped advanced high-strength steel treated by quenching and partitioning process. Scr Mater. 2011;64(8):749–752.
[5] Li H Y, Lu X W, Li W J, et al. Microstructure and mechanical properties of an ultrahigh-strength 40SiMnNiCr steel during the one-step quenching and partitioning process. Metal Mater Trans A. 2010;41(5):1284–1300.
[6] Hsu TY, Xu ZY. Materials Science Forum, Trans Tech Publ, 2007:2283–2286.
[7] Wang XD, Zhong N, Rong YH, et al. Novel ultrahigh-strength nanolath martensitic steel by quenching–partitioning–tempering process. J Mater Res. 2009;24(1):260–267.
[8] Zhong N, Wang XD, Wang L, et al. Enhancement of the mechanical properties of a Nb-microalloyed advanced high-strength steel treated by quenching–partitioning–tempering process. Mater Sci Eng A. 2009;506(1):111–116.
[9] Clarke AJ, Speer JG, Miller MK, et al. Carbon partitioning to austenite from martensite or bainite during the quench and partition (Q&P) process: A critical assessment. Acta Mater. 2008;56(1):16–22.
[10] Nishikawa AS, Santofimia MJ, Sietasma J, et al. Influence of bainite reaction on the kinetics of carbon redistribution during the quenching and partitioning process. Acta Mater. 2018;142:142–151.
[11] Li H Y, Lu X W, Wu X C, et al. Bainitic transformation during the two-step quenching and partitioning process in a medium carbon steel containing silicon. Mater Sci Eng A. 2010;527(23):6255–6259.
[12] Gao G, Zhang H, Gui X, et al. Enhanced ductility and toughness in an ultrahigh-strength Mn–Si–Cr–C steel: the great potential of ultrafine filly retained austenite. Acta Mater. 2014;76:425–433.
[13] Gao G, Zhang H, Gui X, et al. Tempering behavior of ductile 1700MPa Mn–Si–Cr–C steel treated by quenching and partitioning process Incorporating bainite formation. J Mater Sci Technol. 2015;31(2):199–204.
[14] Gao G, Zhang H, Gui X, et al. Enhanced strain hardening capacity in a lean alloy steel treated by a “disturbed” bainitic austempering process. Acta Mater. 2015;101:31–39.
[15] Gui X, Gao G, Guo H, et al. Effect of bainitic transformation during BQ&P process on the mechanical properties in an ultrahigh strength Mn–Si–Cr–C steel. Mater Sci Eng A. 2017;684:598–605.
[16] Gao G, Zhang H, Tan Z, et al. A carbide-free bainite/martensite/austenite triplex steel with enhanced mechanical properties treated by a novel quenching–
partitioning–tempering process. Mater Sci Eng A. 2013;559:165–169.

[17] Matlock DK, Speer JG. Third generation of AHSS: microstructure design concepts. Microstruct Text. Steels. 2009: 185–205.

[18] Aydin H, Essadiqi E, Jung IH, et al. Development of 3rd generation AHSS with medium Mn content alloying compositions. Mater Sci Eng A. 2013;564:501–508.

[19] Matlock DK, Speer JG, De Moor E, et al. Recent developments in advanced high strength sheet steels for automotive applications: an overview. Jestech. 2012;15(1):1–12.

[20] Speer JG, De Moor E, Findley KO, et al. Analysis of microstructure evolution in quenching and partitioning automotive sheet steel. Metall Mater Trans A. 2011;42(12):3591–3601.

[21] Caballero FG, Chao J, Cornide J, et al. Toughness deterioration in advanced high strength bainitic steels. Mater Sci Eng A. 2009;525(1):87–95.

[22] Caballero FG, Bhadeshia H, Mawella KJA, et al. Design of novel high strength bainitic steels: part 1. Mater Sci Technol. 2001;17(5):512–516.

[23] Hong SC, Ahn JC, Nam SY, et al. Mechanical properties of high-Si plate steel produced by the quenching and partitioning process. Met Mater Int. 2007;13(6):439–445.

[24] Bagliani EP, Santofimia MJ, Zhao L, et al. Microstructure, tensile and toughness properties after quenching and partitioning treatments of a medium-carbon steel. Mater Sci and Eng: A. 2013;559:486–495.

[25] Gao G, An B, Zhang H, et al. Concurrent enhancement of ductility and toughness in an ultrahigh strength lean alloy steel treated by bainite-based quenching-partitioning-tempering process. Mater Sci Eng A. 2017;702:104–112.

[26] Zhou Q, Qian L, Tan J, et al. Inconsistent effects of mechanical stability of retained austenite on ductility and toughness of transformation-induced plasticity steels. Mater Sci Eng A. 2013;578:370–376.

[27] Liu H, Sun H, Liu B, et al. An ultrahigh strength steel with ultrafine-grained microstructure produced through intercritical deformation and partitioning process. Mater Des. 2015;83:760–767.

[28] Song R, Ponge D, Raabe D. Mechanical properties of an ultrafine grained C–Mn steel processed by warm deformation and annealing. Acta Mater. 2005;53(18):4881–4892.

[29] Xie ZJ, Yuan SF, Zhou WH, et al. Stabilization of retained austenite by the two-step intercritical heat treatment and its effect on the toughness of a low alloyed steel. Mater Des. 2014;59:193–198.

[30] Tanaka M, Choi CS, Kojima Y. Effect of Ms temperature on strength and toughness in martensitic Fe–Ni–C alloys. Trans Iron Steel Inst Japan. 1974;14(2):110–117.

[31] Shiozawa K, Morii Y, Nishino S, et al. Subsurface crack initiation and propagation mechanism in high-strength steel in a very high cycle fatigue regime. Int J Fatigue. 2006;28(11):1521–1532.

[32] Marines I, Bin X, Bathias C. An understanding of very high cycle fatigue of metals. Int J Fatigue. 2003;25(9):1101–1107.

[33] Murakam Y, Nomoto T, Ueda T. Factors influencing the mechanism of superlong fatigue failure in steels. Fatigue Fract Eng Mater Struct. 1999;22(7):581–590.

[34] Shiozawa K, Hasegawa T, Kashiwagi Y, et al. Very high cycle fatigue properties of bearing steel under axial loading condition. Int J Fatigue. 2009;31(5):880–888.

[35] Hong Y, Zhao A, Qian G. Essential characteristics and influential factors for very-high-cycle fatigue behavior of metallic materials. Acta Metall Sin. 2009;45(7):769–780.

[36] Zhao P, Zhang B, Cheng C, et al. The significance of ultrafine film-like retained austenite in governing very high cycle fatigue behavior in an ultrahigh-strength Mn–Si–Cr–C steel. Mater Sci Eng: A. 2015;645:116–121.

[37] Nie Y, Hui W, Fu W, et al. Ultra high cycle fatigue behavior of a medium-carbon high strength spring steel NHS1. Acta Metall Sin-Chin Ed. 2007;43(10):1031.

[38] Liu LB, Li SX, Li YD, et al. Factors influencing the GBF size of high strength steels in the very high cycle fatigue regime. Mater Sci Eng: A. 2011;528(3):935–942.

[39] Khan MK, Wang QY. Investigation of crack initiation and propagation behavior of AISI 310 stainless steel up to very high cycle fatigue. Int J Fatigue. 2013;54:38–46.

[40] Hui W, Zhang Y, Zhao X, et al. Very high cycle fatigue properties of Cr–Mo low alloy steel containing V-rich MC type carbides. Mater Sci Eng: A. 2016;651:311–320.

[41] Stein CA, Cerrone A, Ozturk T, et al. Fatigue crack initiation, slip localization and twin boundaries in a nickel-based superalloy. Curr Opin Solid State Mater Sci. 2014;18(4):244–252.

[42] Christ HJ, Fritzen CP, Köster P. Micromechanical modeling of short fatigue cracks. Curr Opin Solid State Mater Sci. 2014;18(4):205–211.

[43] Zhang K, Zhang M, Guo Z, et al. A new effect of retained austenite on ductility enhancement in high-strength quenching–partitioning–tempering martensitic steel. Mater Sci Eng: A. 2011;528(29):8486–8491.

[44] Gao G, Zhang B, Cheng C, et al. Very high cycle fatigue behaviors of bainite/martensite multiphase steel treated by quenching-partitioning-tempering process. Int J Fatigue. 2016;92:203–210.