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Structure-properties relationship in TRIP type bainitic ferrite steel austempered at different temperatures

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Abstract

Background: Attractive properties of TRIP-type bainitic ferrite (TBF) steel ascribe to its unique microstructure of lath structure bainitic ferrite matrix and interlath retained austenite films. This work is concerned with obtaining ultra high-strength hot forged TBF steel with high elongation and excellent strength-elongation balance.

Methods: The effect of austempering temperature on the microstructure along with its retained austenite characteristics and tensile properties of a hot forged TBF steel was studied. A detailed investigation correlating the steel structure and its tensile properties was carried out.

Results: Tensile strength ranging from 1058 to 1552 MPa was achieved when the hot forged steel was austempered at (325 - 475 °C).

Conclusions: Ultra high tensile strength of 1058 MPa, large total elongation of 29% and excellent strength-elongation balance of 30 GPa % were attained when the steel was austempered at 425 °C. The large total elongation of this steel is mainly due to the uniform fine lath structure matrix and the pronounced TRIP effect of a large amount of retained austenite films which prevents a rapid decrease of strain hardening rate at low strain and leads to a relatively high strain hardening at high strain level. Rapid transformation of blocky retained austenite at low strain in the hot forged TBF steel austempered at higher temperatures results in a rapid increase of initial strain hardening. In addition, the coarse microstructure that contains large blocks of retained austenite / martensite and the insufficient numbers of bainitic ferrite lathes and retained austenite films deteriorate the total elongation and the strength-elongation balance of the TBF austempered at 475 °C.

Keywords: Ultrahigh-strength steels, TRIP-aided steels, Austempering heat treatments, Microstructure, Retained austenite characteristics, Tensile properties

Background

The reduction in weight of vehicle body can improve the fuel efficiency and environmental control. Therefore, there is an international attention to develop advanced high-strength steel (AHSS). The TRIP-aided multiphase (TMP) steel as a class of AHSS exhibits an excellent combination of strength and stretch-formability (Sugimoto et al. 1995), good deep drawability (Matsumura et al. 1992) and high fatigue strength (Sugimoto et al. 1997). The microstructure of this steel is mainly composed of bainitic ferrite (bf) and carbon-enriched retained austenite (γ_r) embedded in a matrix of polygonal ferrite (Sugimoto et al. 1992).

The ideal energy absorption behaviour of TMP steels, which can be attributed to the transformation of metastable retained austenite into martensite under stress and strain (TRIP effect) (Bleck 2002; Sugimoto et al. 2006), and the high work-hardening response improve the crashworthiness of a vehicle through good distribution of strain during crash deformation (Bleck 2002).The superior formability of TMP steel (Sugimoto et al. 2006) can also be attributed to its high work-hardening properties. TMP steel has been applied to some impact members (Ojima et al. 1998) due to its high ability of energy

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absorption under dynamic load. In spite of the excellent mechanical and technological properties of TMP steel, it cannot be applied to the automotive underbody press parts (e.g. lower arms and members) (Emadoddin et al. 2009) due to its poor stretch-flangability (Sugimoto et al. 1999). Also, this steel lacks sufficient performance in bendability and edge formability (De Cooman et al. 2004). Based on the fact that the bainitic steel exhibits an excellent stretch-flangability due to its uniform fine lath structure, (Sugimoto et al. 2000) have developed a new type of high-strength TRIP type bainitic ferrite (TBF) steel. The microstructure of TBF steel is characterized by bainitic ferrite lath matrix and interlath-retained austenite films. TBF steel is characterized by excellent balance between low edge crack susceptibility and high elongation (Sugimoto et al. 2002, 2006). It develops better stretchability compared to TMP steel with the same chemical composition and amount of retained austenite (Bhadeshia et al. 2001). It also exhibits good bendability and edge formability (Sugimoto et al. 2006). For these excellent properties, TBF steel can be applied to the applications that require high localized strain realization. Moreover, TBF steel shows high fatigue strength (Demeyer et al. 1999) and high impact energy (Hojo et al. 2008). Recently, (Sugimoto et al. 2010a) have developed a new type of hot-forged TBF steel with high hardenability. The present work is aimed at producing ultrahigh-strength hot-forged TBF steel with high elongation. Tensile properties of TRIP steel are affected by heat treatment conditions. The specific purpose of this investigation is to study the effect of austempering temperature (T_A) on the microstructure along with its retained austenite characteristics, and tensile properties of the investigated hot-forged TBF steel. The structure-properties relationship is also discussed.

Methods

One hundred kilogramme Y-blocks of the steel alloy were produced in a medium-frequency induction furnace. Chemical composition of the steel is shown in Table 1. The charge was made up of steel scrap and the required Mn, Si, P, Mo and Nb were added as ferroalloys (Fe–80% Mn), (Fe–75% Si), (Fe–28% P), (Fe–70% Mo) and (Fe–65% Nb). The heats were cast into 300 mm × 200 mm × 50 mm Y-blocks. Each leg of Y-blocks was machined and sectioned into 200 mm × 60 mm × 40 mm specimens. The specimens were homogenized at 1250 °C for 2 h in a muffle furnace, and then furnace cooled.

The martensite start temperature was estimated to be 371 °C by the following equation (Tamura 1970):

$$\begin{aligned} M_S(^{\circ}C) = {} & 550-361x(\%C)-39x(\%Mn)-17x(\%V) \\ & -20x(\%Cr)-17x(\%Ni)-10x(\%Cu) \\ & -5x(\%Mo+\%W)-0x(\%Si)+15x(\%Co) \\ & +30x(\%Al) \end{aligned} \tag{1}$$

Table 1 Chemical composition of the investigated steel alloy

Element	C	Si	Mn	P	S	Mo	Ni	Al	Nb
Wt.%	0.35	0.6	1.5	0.02	0.008	0.25	1.5	1.1	0.05

Fee-forging as one of the forming technologies is commonly used to refine the microstructure and increase the strength values. The homogenized steel was reheated at 1200 °C for 30 min and then hot forged into rods of 16–18 mm in diameter. The hot forging was performed using Pneumatic Hammer (150 Kp).

Standard tensile specimens ASTM E-8 with 6 mm ϕ × 36 mm gauge length were prepared. For microstructure investigation, hot-forged round specimens of 6 mm ϕ were also prepared. In order to obtain TRIP type bainitic ferrite steel, the austempering treatments were performed by austenitizing of the hot-forged tensile and round specimens in an electrically heated furnace, followed by rapid quench into a salt bath at the austempering temperature. The specimens were austempered at 325–475 °C for 20 min, then quenched into oil and cooled to room temperature, as shown in Fig. 1.

The microstructure was identified by optical microscopy, scanning electron microscopy (SEM) and X-ray diffraction measurements. Specimens for optical microscopy were etched with 5% nital and rinsed with water followed by etching in a 10% sodium meta-bisulphite solution. With these etchants, retained austenite appears white, ferrite appears grey and bainite and martensite appear black (Jeong 1994). For microstructure investigation using SEM, samples were etched with 2% nital.

Volume fraction of the retained austenite (V_γ) was estimated by X-ray diffraction using a Cu target at 45 KV and 40 mA. V_γ was calculated using Eq. 2, where I_γ and I_α are the average integrated intensity obtained at the (200)γ, (220)γ, (311)γ and (200)α, (211)α diffraction peaks. K_α and K_γ are the reflection coefficients of the ferrite and austenite phases, respectively (Maruyama 1977).

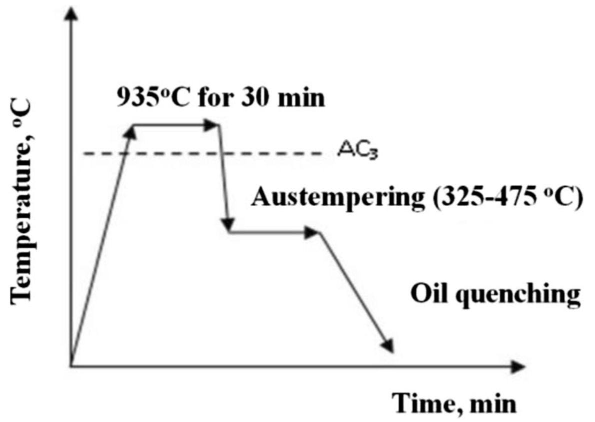

Fig. 1 Schematic representation of austempering treatment

$$V_\gamma = \frac{I_\gamma K_\alpha}{I_\gamma K_\alpha + I_\alpha K_\gamma} \qquad (2)$$

Carbon concentration of the retained austenite was estimated from the following equation (Dyson & Holmes 1970). The retained austenite lattice parameter (a_γ) was measured based on the (220)γ diffraction peak.

$$a_\gamma = 3.578 + 0.033\left(\text{wt\% } C_\gamma\right) \qquad (3)$$

Tensile test was carried out on a universal testing machine (1000 KN) at room temperature with a cross head speed of 0.5 mm min^{-1}.

Results and discussion

Influence of austempering temperature on retained austenite characteristics

Figure 2a–c shows the variation of retained austenite characteristics with T_A in TBF steel austempered for 20 min. As shown in Fig. 2a, a small amount of austenite (3%) is retained at room temperature when the steel is austempered at 325 and 350 °C, lower than the M_s. This is due to transformation of large amount of residual austenite into bainitic ferrite/martensite mixed matrix. With increasing the austempering temperature up to 425 °C, the volume fraction of γ_r is considerably increased showing the maximum value of 16% at 400 °C. Further increase of the austempering temperature up to 475 °C leads to decrease of γ_r due to the transformation of low carbon residual austenite to fresh martensite upon cooling after austempering. At temperatures of 450 °C and higher, carbon begins to precipitate after certain overaging times resulting in a decrease of the amount of γ_r (Hausman 2013). Addition of Mn and Ni to TBF steel was found to increase the amount of γ_r (Kim et al. 2003) and the complex addition of Nb and Mo was found to increase the amount of γ_r as well especially when the TBF steel is austempered at 450–500 °C (Sugimoto et al. 2007). It is well known that the stability of γ_r is affected by a combination of many factors such as grain size, carbon content and morphology of retained austenite. The properties of the matrix and the alloying elements also influence the stability of γ_r. The addition of 0.05% Nb and 0.2% Mo to TBF steel was found to decrease the carbon concentration in γ_r especially at high austempering temperatures (Sugimoto et al. 2007).

Influence of austempering temperature on the tensile properties

Figure 3 shows the engineering flow curves of the TBF steel austempered at (325–475 °C). As shown in the figure, ultrahigh-strength hot-forged TBF steels exhibiting different tensile properties were produced. These TBF steels exhibit a continuous yielding behaviour. In dual-phase steel, this behaviour is mainly attributed to high dislocation density produced by martensite transformation (Saleh & Priestner 2001). In TRIP steel, the yielding behaviour may be explained not only by dislocation density but also by other factors due to the complex structure of this steel.

It is well known that the discontinuous yielding behaviour is produced in the steel due to the presence of interstitial atoms and the yield point phenomenon in the steel is a function of the composition and heat treatment of the material. In the investigated steel, the carbide formers (Mo and Nb) and nitride formers (Al, Si and Nb) reduce the level of the interstitial atoms of carbon and nitrogen, which are strong lockers of dislocation. Consequently, the continuous yielding is promoted as shown in Fig. 3.

Figure 4a–c shows the variation of the tensile strength (TS), total elongation (TEL) and strength-elongation balance (TSxTEL) with T_A in the TBF steel austempered for 20 min. As shown in Fig. 4a, b, increasing the austempering temperature from 325 up to 400 °C leads to a great decrease of TS and a slight increase of TEL. With increasing T_A to 425 °C, TS is slightly decreased while the TEL and TSxTEL balance are greatly increased to maximum values of 29%

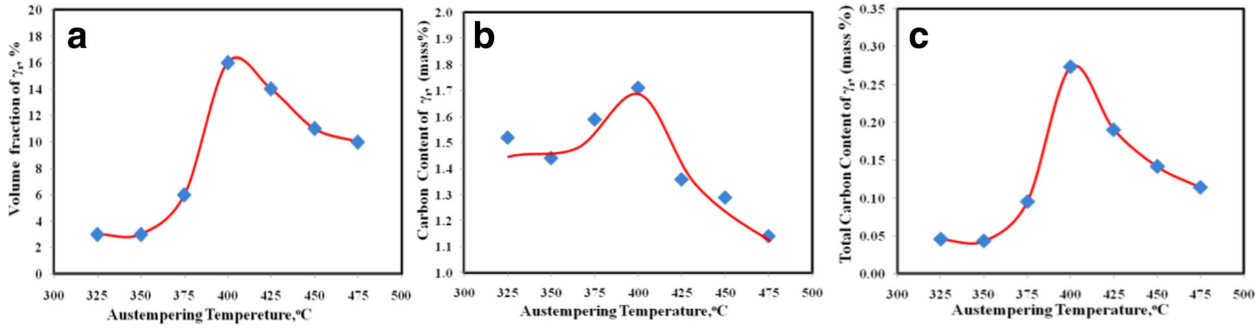

Fig. 2 Variation of **a** volume fraction, **b** carbon content and **c** total carbon content of retained austenite with T_A in TBF steel austempered for 20 min

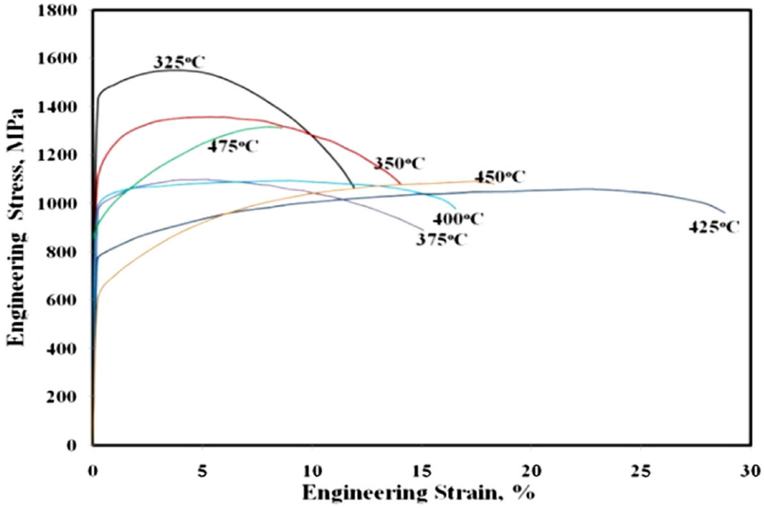

Fig. 3 Engineering stress-strain curves of the TBF steel austempered at different temperatures

and 30 GPa, respectively. Upon further increase of T_A, the TS is increased showing a considerable increase at 475 °C while the TEL and TSxTEL are decreased indicating serious decreases at 475 °C.

Influence of austempering temperature on strain-hardening behaviour

Figure 5 shows the variation of strain-hardening rate with the true strain in TBF steel austempered at different temperatures. The behaviour of strain hardening at high strain levels is related to the transformation rate of γ_r which is greatly affected by the stability of γ_r. Hence, the different transformation rate at high strain levels shown in Fig. 5 may be attributed to the location of the γ_r in the matrix, carbon content and size of the retained austenite grains (Chiang et al. 2011).

Microstructure and tensile properties

Figures 6 and 7 shows typical optical and SEM micrographs of the steel austempered at different temperatures. It is obvious that the austempering temperature has a great effect on the amount and morphology of the phases present in the steel microstructure.

TBF steel austempered at temperatures lower than M_s

Microstructure of the TBF steel austempered at 325 and 350 °C, lower than M_s, consists of bainitic ferrite/martensite mixed matrix of very fine lath structure and interlath-retained austenite films as shown in Figs. 6a, b and 7a. This fine microstructure clearly shown in Fig. 7a has been attributed to the increase in transformation driving force and suppression of coalescence of fine bainite laths (Bhadeshia et al. 2001). The fine prior austenite grain size shown in Fig. 7 can be attributed to the combined addition of Nb and Mo which has a better refining effect than single Nb addition (Huab et al. 2015). Recently, (Hojo et al. 2010) have shown that the complex addition of Al-Nb-Mo to TBF steel tends to refine the lath structure and γ_r films.

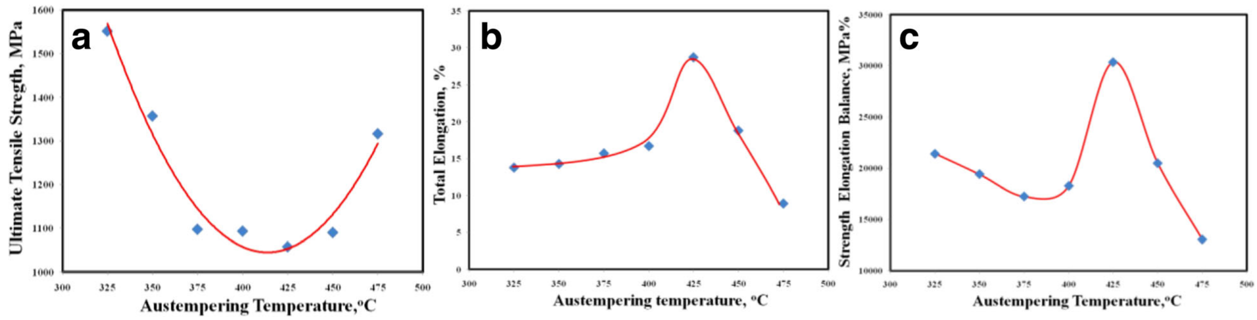

Fig. 4 Variation of **a** ultimate tensile strength, **b** total elongation and **c** strength-elongation balance with T_A

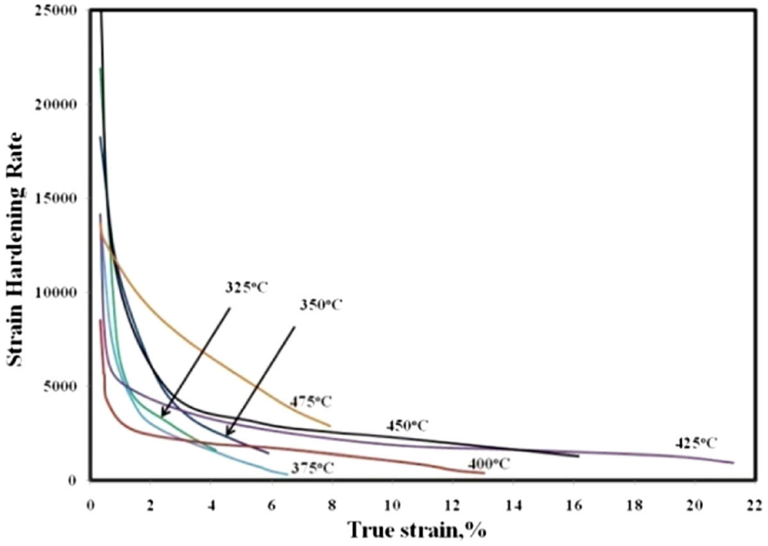

Fig. 5 Variation of strain hardening rate (dσ/dε) with true strain for the steel alloy austempered at different temperatures

TBF steel austempered at temperatures lower than its M_s indicates higher tensile strength “up to 1552 MPa” compared to the steel austempered at temperatures higher than M_s (Fig. 4a). This is due to its finer lath structure and its bainitic ferrite/martensite (α_m) mixed matrix. Tensile strength of this TBF steel is decreased with increasing austempering temperature, from 325 to 350 °C (Fig. 4a) due to the decrease of martensite lathes in the mixed matrix. The strengthening effect of martensite lathes is more dominant in this steel than a small TRIP effect of (3%) γ_r. TBF steel austempered at temperatures lower than M_s exhibits a smaller total elongation compared to that austempered at 425 and 450 °C, higher than the M_s (Fig. 4b). This is due to steep fall of strain hardening at an early stage which may be caused by small internal stress hardening and initial martensite hardening resulting from a small volume fraction of second phase (Sugimoto et al. 2000), followed by great decrease of strain hardening at low strain (Fig. 5) which in turn results from insignificant TRIP effect of small γ_r volume fraction.

TBF steel austempered at temperatures higher than M_s

TBF steel austempered at 400 °C The fine uniform microstructure of TBF steel austempered at 400 °C shown in Fig. 7d contains a large amount of γ_r ($\gamma_r = 16\%$). However, this TBF steel exhibits a smaller total elongation compared to TBF steel austempered at 425 °C (Fig. 4b). This can be attributed to very high stability of γ_r ($C_\gamma = 1.7$ wt%) which prevents high amount of γ_r from transforming during tensile deformation. This results in a considerable decrease of strain hardening at high strain levels (Fig. 5). Therefore, the total elongation and strength-elongation balance are greatly decreased (Fig. 4b, c). According to (Chiang et al. 2011), the very high levels of carbon in $\gamma_r > 1.8$ wt % C prevent the γ_r from transforming completely.

TBF steel austempered at 425 °C Figures 6e and 7d show the microstructure of TBF steel austempered at 425 °C. This fine uniform microstructure mainly consists of bainitic ferrite matrix of lath structure and interlath γ_r films. It is evident from the figures (Fig. 6d, e and 7c, d) that the matrix structure is changed in some zones from bf lathes to granular bainitic ferrite and the retained austenite films are changed to a few of coarse and island type γ_r. K. Sugimoto et al. (2010a) have also detected and reported these morphologic changes. It is also obvious from these figures that the pro-eutoctoid ferrite (α_{PE}) and upper bainite are developed. The grain refinement through Nb addition leads to promotion of transformation during cooling from austenitization which causes ~20% transformed phase fraction prior to the overaging step (Hausman 2013). Ultrahigh-strength TBF steel austempered at 425 °C exhibits high total elongation. As shown in Fig. 4b, c, the total elongation and strength-elongation balance were considerably increased at 425 °C to the maximum values of 29% and 30 GPa%, respectively. The large total elongation can be attributed to the uniform microstructure and fine lath structure matrix and to the pronounced TRIP effect of a large amount of stable retained austenite films which gradually transforms upon loading to martensite. The strain-induced transformation of γ_r simultaneously relaxes the localized stress concentration at the matrix/second phase interface to suppress the void

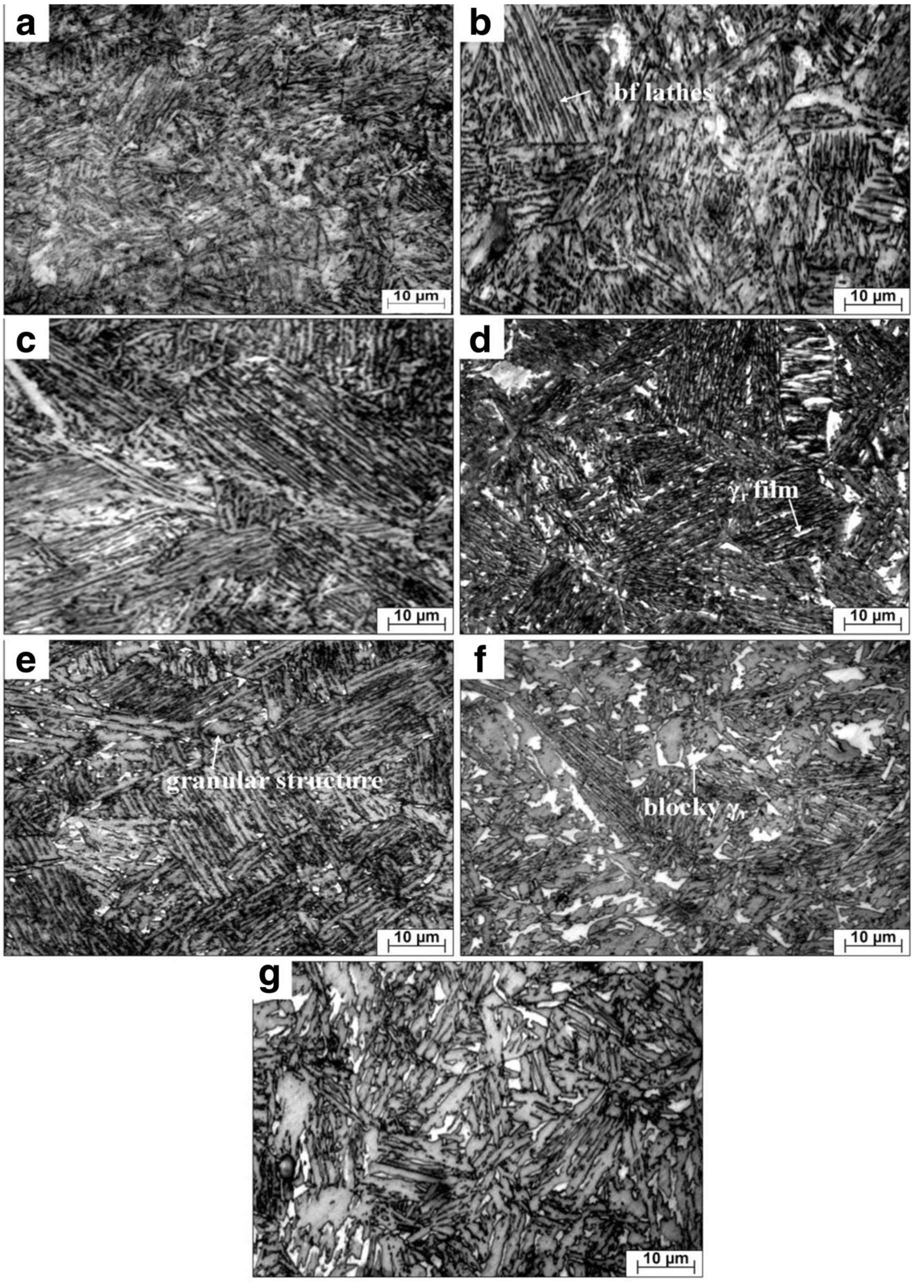

Fig. 6 Optical micrographs of the TBF steels austempered at **a** 325 °C, **b** 350 °C, **c** 375 °C, **d** 400 °C, **e** 425 °C, **f** 450 °C and **g** 475 °C for 20 min

initiation (Sugimoto et al. 2000). Recently, (Zhao et al. 2014) have reported that the film type γ_r transforms into martensite and hardens the bf matrix. The resultant compressive stress on the bf matrix prevents crack propagation. Consequently, the gradual transformation of high amount of stable γ_r films in the TBF steel austempered at 425 °C prevents the rapid decrease of strain hardening rate at low strain and permits for the strain hardening to continue up to a high strain (Fig. 5) which results in a considerable increase of the total elongation (Fig. 4b).

The pro-eutectoid soft phases of ferrite and upper bainite contribute to enhanced elongation. The complex addition of Al and Nb to the investigated ultrahigh-strength TBF steel contributes also to enhanced total elongation (Sugimoto et al. 2010b). The reduced strain induced martensite formation at low strain levels in the steel austempered at 425 °C results in increasing the work hardening at high strain levels (Fig. 5). This result has been also detected and reported by (Hausman 2013).

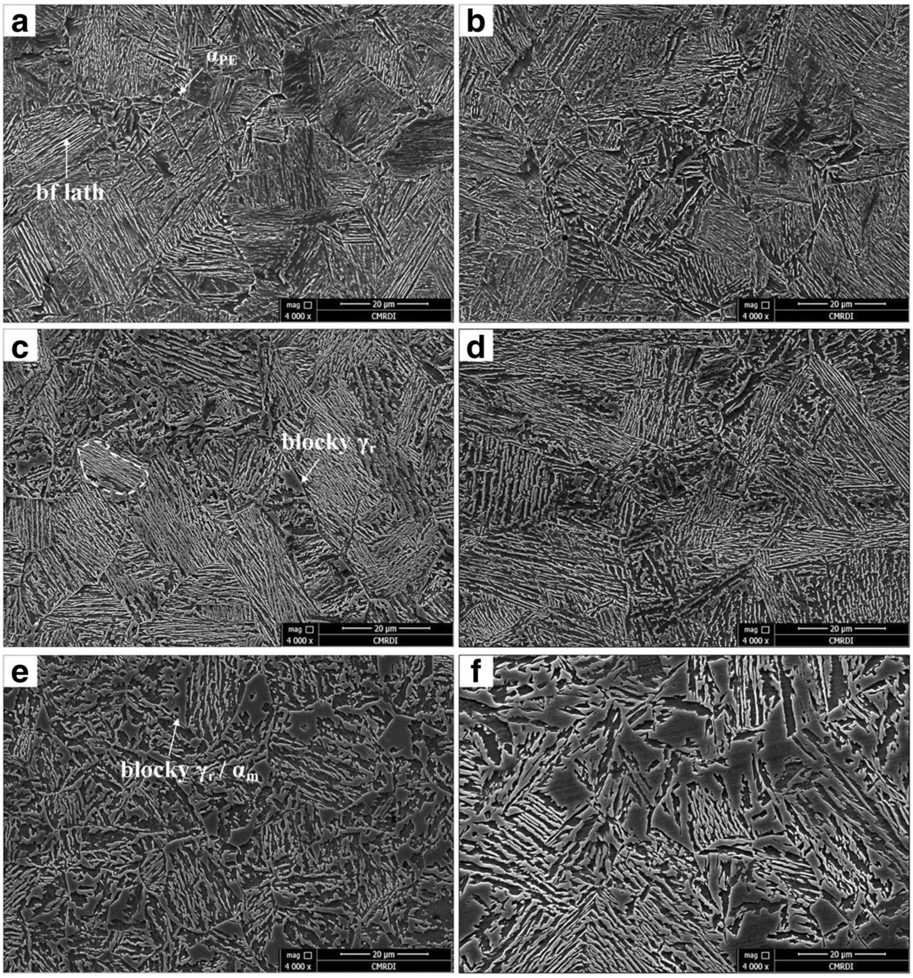

Fig. 7 SEM micrographs of the TBF steels austempered at **a** 350 °C, **b** 375 °C, **c** 400 °C, **d** 425 °C, **e** 450 °C, **f** 475 °C for 20 min

TBF steel austempered at 450 and 475 °C When the steel is austempered at higher temperatures of 450 and 475 °C, the numbers of bainitic ferrite lathes and the fraction of interlath γ_r films are greatly reduced while the amount of granular bainitic ferrite and both of blocky and island type γ_r/α_m are increased due to the morphologic changes (Fig. 6f, g), which can be attributed to Nb addition (Sugimoto et al. 2006; Hausman 2013; Bleck et al. 2001). Increasing austempering temperature leads to an increase in the thickness of bainitic ferrite lathes (Fig. 7e, f) due to the reduced nucleation kinetics of bainite formation at higher T_A. Increase in lath thickness is also attributed to low dislocation density due to the high temperatures (Zhao et al. 2014). Austempering at these higher temperatures leads to a small degree of super cooling that reduces the driving force of bainite formation. Consequently, the numbers of bainitic ferrite lathes are reduced (Fig. 6f, g) as well as the carbon content of the residual austenite. Upon cooling the steel after austempering, a high amount of lower carbon content residual austenite could not be stabilized at room temperature and both blocky and island type retained austenite/fresh martensite are formed (Figs. 6f, g and 7e, f).

As shown in Fig. 4b, c, the total elongation and strength-elongation balance of TBF steel are decreased with increasing austempering temperature from 425 to 450 °C due to the morphologic change of high amount of γ_r from film type to a less stable blocky one (compare Fig. 6e and f). Presence of high amount of initial martensite and reduced numbers of bainitic ferrite lathes (compare Fig. 7d and e) also lead to the decrease of total elongation of TBF steel austempered at 450 °C.

Enhanced strain hardening rate of TBF steels austempered at higher temperatures (450 and 475 °C) over a small strain range (Fig. 5) is attributed to the high long-range internal stress resulting from both the initial martensite hardening of original martensite and the difference in strength between the granular bainitic ferrite soft matrix and the γ_r hard second phase (Sugimoto et al. 2006). The total elongation and strength-elongation balance of TBF steel austempered at 475 °C are seriously decreased (Fig. 4b, c). The decrease of total elongation is firstly attributed to the rapid transformation of blocky γ_r at low strain which results in a rapid increase of initial strain hardening (Fig. 5). In addition, the coarse microstructure consisting of high amount of large blocks of γ_r/α_m and insufficient numbers of bainitic ferrite lathes (Fig. 6g) results also in the decrease of total elongation. Martensite blocks act as the source of voids initiation and insufficient numbers of bf lathes lead to the decrease of γ_r stability (Sugimoto et al. 2002).

Conclusions

The effect of T_A on the microstructure, retained austenite characteristics and mechanical properties of hot-forged TBF alloyed steel were investigated. The main results are as follows:

- Ultrahigh-strength hot-forged TBF steels with tensile strength ranging from 1058 to 1552 MPa were attained when TBF steel is austempered at 325–475 °C.
- Tensile strength of the hot-forged TBF steel was considerably increased, when austempered at temperature lower than M_s due to its hard fine lath-like bainitic ferrite/martensite mixed matrix. Relatively, small total elongation at these temperatures is due to both the steep fall of strain hardening at an early stage and the great decrease of strain hardening at low strain which resulted from insignificant TRIP effect of small amount of γ_r.
- The considerable decrease of strain hardening at high strain levels and the great decreases of total elongation and strength-elongation balance in TBF steel austempered at 400 °C are due to very high stability of γ_r ($C_\gamma = 1.7$ wt%) which prevents high amount of γ_r from transforming.
- Total elongation and strength-elongation balance of ultrahigh-strength hot-forged TBF steel were greatly increased to 29% and 30 GPa%, respectively, when austempered at 425 °C. Increase of total elongation is due to uniform fine bainitic ferrite lathes, interlath γ_r films and pronounced TRIP

  effect of large amount of γ_r which prevents rapid decrease of strain hardening rate at low strain and leads to a relatively high strain hardening at high strain levels. Pro-eutectoid soft phases and granular bainitic ferrite contribute also to enhanced total elongation.
- Rapid transformation of blocky γ_r at low strain in hot-forged TBF steels austempered at 450 and 475 °C resulted in rapid increase of initial strain hardening. Additionally, the large blocks of martensite and insufficient numbers of bf lathes and γ_r films resulted in serious decreases of the total elongation and strength-elongation balance, especially in TBF steel austempered at 475 °C.

Authors' contributions

All authors read and approved the final manuscript.

Competing interests

The authors declare that they have no competing interests.

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