Original Article

Effects of ausforming temperature on bainite transformation kinetics, microstructures and mechanical properties in ultra-fine bainitic steel

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\textbf{ARTICLE INFO}

Article history:
Received 6 November 2019
Accepted 28 November 2019
Available online 9 December 2019

Keywords:
Ultra-fine bainitic steel
Ausforming
Retained austenite
Plastic deformation
Mechanical properties

\textbf{ABSTRACT}

The effects of ausforming temperature on bainite transformation kinetic and plastic deformation mechanism were evaluated by thermal simulation method and warm rolling process. Results showed that entire process of bainite transformation austempered at 300 \(^{\circ}\)C was notably accelerated by ausforming due to the increased nucleation sites, and by diminishing ausforming temperature as well. However, ausforming would increase undercooled austenite stability, leading to the decrease of maximum attainable volume fraction of bainitic ferrite. Compared with traditional isothermal transformation, ausforming process could effectively refine microstructure, as well as improve mechanical properties. And with decreasing ausforming temperature, strength, hardness and ductility were all increased, which attributed to the thinner thickness of bainite lath and smaller dimension and proportion of blocky retained austenite, by contrast, the change trend of impact toughness was virtually opposite resulted from variant selection. When ausforming at 300 \(^{\circ}\)C, the ultra-fine bainitic steel exhibited the best comprehensive mechanical performance, which was almost 1850 MPa and with up to 23\% for ultimate tensile strength and total elongation, respectively.

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1. Introduction

Ultra-fine bainitic steel, which was derived from the nanostructured carbide-free bainitic steel developed by Caballero and Bhadeshia [1–3], has awakened increasing attention since its extraordinary ultrahigh strength along with excellent toughness. A great deal of scientific research has proved that ultra-fine bainitic steel not only exhibits almost the same hardness as maraging steel, but manifests more outstanding ductility than traditional lower bainitic steel. Due to the unique mechanical properties, ultra-fine bainitic steel has incredible potential in transport, construction and offshore industries, as well as defense applications.

Unsatisfactorily, the isothermal holding time of bainite transformation is very time-consuming, for example, it takes as long as dozens of hours to complete bainite reaction at a temperature range of 200 \(^{\circ}\)C–300 \(^{\circ}\)C [4,5], which restricts the wide application of ultra-fine bainitic steel. Numerous
works have been carried out in recent years for acceleration mechanism of phase transformation, including composition optimization [6,7], grain refinement [8], heat-treatment optimization [9], strain induced transformation [10–12] and prior phase transformation [13]. Thereinto, prior ausforming process under favorable condition is considered as the best method to accelerate bainite transformation. Gong et al. [14] claimed that bainite transformation of high carbon steel could be significantly accelerated after ausforming at 300 °C with 15% strain. Zhang et al. [15] reported that either decreased ausforming temperature or increased ausforming strain could shorten incubation period of bainite transformation. Kabir Mohammad et al. [16] investigated effects of ausforming strains on transformation kinetics in low temperature bainitic steel and showed that bainite transformation rate tended to be increased depending on applied strain level. Nevertheless, Xu et al. [17] indicated when ausforming at high temperature or large strain, that bainite transformation could be suppressed due to the retardation of bainite growth by dislocation debris. Accordingly, there are some disputing concerns influencing of ausforming temperature on bainite transformation mechanism in medium carbon or high carbon alloy steel.

In contrast to numerous researches concerning effect of ausforming parameters on transformation behavior and microstructures of ultra-fine bainitic steel, a few investigations have been conducted to analyze mechanical properties after ausforming. For instance, Golchin et al. [18] studied the effect of ausforming with 10% strain on impact toughness in high carbon steel, and it has been reported ausforming increased impact toughness energy by austempering for 2 h and 4 h. Wang et al. [19] revealed that extremely high tensile strength and appreciable total elongation were obtained through ausrolling at 500 °C and austempering at 200 °C due to the increased dislocation density. It has also been declared that multi-step ausforming can not only accelerate bainite transformation but also enhance tensile property of nanobainite steel [20]. However, the mechanisms behind acceleration of bainite transformation and improvement of mechanical properties remain unclear.

The present work is dedicated to systematically evaluate bainite transformation kinetics, microstructural characterization and mechanical performance in ultra-fine bainitic steel after subjected to isothermal heat treatment without or with ausforming, and explore the feasible method to rapidly produce ultra-fine bainitic structure without expense of outstanding comprehensive properties

2. Material and methods

The experimental ultra-fine bainitic steel has a chemical composition of Fe-0.7C-2.47Si-1.46Mn-0.87Al-0.02Nb (wt.%). Extremely high content of Al and Si was added to retard precipitation of cementite from austenite during heat treatment process [21]. The cast ingot was produced by vacuum induction melting furnace and forged into square bars (150 mm × 70 mm × 30 mm), subsequently subjected to homogenizing heat treatment at 1200 °C for 36 h in a high temperature resistance reheating furnace to eliminate macrosegregation. Two heat treatment procedures were conducted on a Gleeble 3500 thermomechanical simulator in light of processing schedules illustrated in Fig. 1. Specimens used for thermomechanical processing were machined into dumbbell shape with dimension of 44 mm × 10 mm for heating and ausforming section, and 410 mm × 30 mm for two holding sections. In schedule 1, samples were heated up to 950 °C for 10 min followed by cooling down to 300 °C at a cooling speed of almost 20 °C/s for isothermal bainite transformation, named after DIT. In schedule 2, after similar austenitization treatment, samples were immediately quenched to 950 °C, 500 °C and 300 °C, respectively, subsequently ausformed at those temperature with a 30% reduction and 5 s⁻¹ strain rate, and then austempered at 300 °C for same heat treatment time of DIT sample, which were denoted as AIT hereafter. The dilatation in diameter was monitored simultaneously over the whole range of heat treatment. However, since thermomechanical simulation samples were not large enough to prepare tensile and impact specimens, the warm rolling process with 30% thickness reduction instead of ausforming were conducted through one-pass rolling with a two-high mill. In order to eliminate thermal gradients, austenitic samples were held at 950 °C, 500 °C and 300 °C for 30 s prior to warm rolling, respectively. Meanwhile, the isothermal bainite heat treatments were carried out in molten salt huth furnace to provide a uniform heating environment.

After being polished and etched with 2% Nital solution, microstructural characterizations were observed by optical microscope (OM), scanning electron microscope (SEM) and transmission electron microscope (TEM). The mean bainite lath thickness (T) was determined by measuring mean linear intercept (L) through drawing perpendicular line to plate length in TEM micrographs, and can be revised by the Eq. (1) [22]:

\[
T = 2L/n
\]

X-ray diffraction (XRD) analysis was subjected to measure constituent phase quantification. XRD patterns were obtained using Cu-Kα radiation with a voltage of 40 kV and a current of

![Fig. 1 - Thermomechanical processing schedules. DIT: direct isothermal bainite transformation; AIT: ausforming at different temperatures and subsequent isothermal bainite transformation.](image-url)
50 mA. The diffraction data were collected over a 2θ range of 47° to 93° with a step size of 1°. The volume fraction of retained austenite and bainitic ferrite was calculated by direction comparison from integrated intensities.

Quasi-tensile tests were performed using a MTS 810 tensile testing machine at room temperature according to GB/T 228–2010, with a 100 kN load at a cross-head velocity of 0.1 mm/min. Dog-bone shaped tensile specimens were taken from the warm rolling slabs along the rolling direction and fabricated with a gage diameter of 5 mm and a gage length of 25 mm. The yield strength was taken as 0.2% offset stress. The Charpy impact tests were conducted at ambient temperature with standard U-notched samples of size 10 mm × 10 mm × 55 mm, which were oriented with the notch perpendicular to rolling direction. Fractographic features of tensile and impact fracture surface were investigated by SEM. The hardness measurement was conducted using Rockwell microhardness tester machine with a load of 10 kg. At least five different measurements were implemented to make sure the reproducibility of results and average values were reported in each case.

3. Results and discussion

3.1. Effect of ausforming on transformation kinetics

Fig. 2a presents the dilatation as a function of temperature for various samples during cooling. The martensite temperatures (Ms) were determined from the dilatation curves by the tangent method [23]. Compared with DIT sample, the Ms values of ausforming samples drop down. Meanwhile, as ausforming temperature decreases from 950 °C to 300 °C, the Ms further drops from 184 °C to 174 °C, which is in accordance with the literatures [24,25]. Fig. 2b shows the flow curves of ausforming samples during compressive deformation prior to bainite transformation. It is observed that with decreasing ausforming temperature, the flow stress of austenite gradually increases. Since ausforming can generate substantial crystal defects, which in turn increase strength of undercooled austenite and accordingly intensify the shear resistance of austenite to martensite transformation [26]. At this time, the same undercooling degree is not enough to provide driving force for displacive transformation, hence the Ms should be
decreased to a certain extent to increase phase transformation driving force. Furthermore, higher temperature ausforming (950 °C) yields a great number of grain boundaries caused by austenite recrystallization, but lower temperature (500 °C and 300 °C) introduces high-density dislocation, which makes a larger contribution to austenite strengthening than former. Consequently, low temperature ausforming has a more pronounced effect on Ms than high temperature.

In order to evaluate phase transformation under different conditions, the dilatation of specimens was normalized using the formula \((d_i - d_0)/d_0\), in which \(d_i\) referred to the instantaneous diameter during austempering, \(d_0\) to the initial diameter after deformation. Fig. 3a presents the dilatation-time curves of samples treated by different processes. The incubation period and termination time of bainite transformation, which referred to the time reaching initial explosive expansion and maximum expansion from the starting point of thermal holding, were determined by tangential line method, as marked in Fig. 3a. It should be noted that incubation period and termination time of bainite transformation are notably shortened by ausforming and by decreasing ausforming temperature as well.

These observations may be closely associated with three major factors: ausforming can produce a great many of deformation bands, dislocations and subgrains in prior austenite grain, all of which facilitate for heterogeneous nucleation of bainite, not only resulting in increasing potential nucleation sites but also promoting carbon atom diffusion, and in turn shorten the incubation period [27, 28]; the deformed austenite primarily contributes to providing additional phase transformation driving force, which can accelerate bainite transformation and in turn decrease total bainite transformation time; a planner dislocation array in austenite brought by low temperature ausforming is believed to show a stronger acceleration effect, in comparison with introduction of a dislocation cell structure by high temperature ausforming [14].

Specifically, as shown in Fig. 3a, the incubation period of sample ausforming at 300 °C is only about 20s, which is
approximately two orders of magnitude less than DIT sample (about 1215s). Nerveless, as the temperature further increases to 500 °C and 950 °C, the extent to which ausforming promotes the bainite transformation decreases. According to the literature [29], the occurrence of dynamic recrystallization under ausforming at high temperature (950 °C) leads to the increase of austenite boundary area per unit volume, resulting in increasing nucleation rate. Conversely, grain refinement could increase impingement probability of bainite lath upon growth and accordingly decrease growth rate. Whereas ausforming at 500 °C, dynamic and static recovery of austenite decreases the crystal lattice distortion degree and dislocation density, which weakens the influence of ausforming on accelerating mechanism. When ausforming performs at a relatively low temperature of 300 °C, the great number of planar dislocation configurations generated in undercooled austenite are difficult to recover due to the relatively low stack fault energy [30,31], thus causing obvious accelerating effect. Consequently, it is evident that bainite transformation is promoted by ausforming at different temperatures, and low temperature ausforming will be more effective in assisting in transformation than high temperature under the same ausforming strain.

Fig. 3b shows corresponding dilatation rate curves of Fig. 3a. In the initial transformation stage, the transformation rate increases with the decrease of ausforming temperature, which is ascribed to the increased nucleation rate resulting from the high-density lattice defects. However, work hardening of austenite plays a negative role in growth of bainite lath, hence the transformation rate of ausforming sample slightly decreases in the later transformation stage, as confirmed by Bhadeshia et al. [32,33]. Corresponding to DIT sample, a larger maximum phase transformation rate and a shorter time reaching to the peak rate can be found in samples ausforming at 500 °C and 300 °C. In contrast, as increasing ausforming temperature to 950 °C, the maximum transformation rate of sample is markedly reduced and even lower than DIT sample, but the time reaching the maximum transformation rate is slightly shortened. This can be related with the fact that grain refinement by dynamic recrystallization during ausforming increases the mechanical stability of austenite, resulting in a retardation of subsequent bainite growth.

Fig. 4 shows OM, SEM and TEM micrographs of samples treated by various processes, and the rolling and normal directions are also indicated. In general, the microstructures are all consisted of lath bainite sheaves and retained austenite, with no carbides being clearly visible. It is obviously observed that the full morphology in high ausforming sample (950 °C) tends to be similar with that in DIT sample apart from the slightly shorter bainite plates. However, with decreasing ausforming temperature, the equiaxed austenite grains transfer into a thin pancake shape and no appreciable distinction can be clearly determined between block and packet boundaries. In addition, the growth direction of bainite lath exhibits an obvious discrepancy. Unlike the random alignment morphology in different crystallographic directions of DIT sample, the lamellar bainite structures with a strong texture can be found in AIT sample, and the normal direction of lath is parallel to compression direction.

Another extremely important feature is the morphology of retained austenite under different ausforming temperatures. Upon comparing the OM and SEM images shown in Fig. 4, retained austenite can be divided into two types, namely filmy retained austenite distributed between bainite lath, and blocky retained austenite inserted between bainite sheaves presenting a polygon shape in two-dimensional space. The dimension of blocky retained austenite was quantitatively measured from several SEM images and presented in Fig. 5. It is observed that blocky retained austenite size of AIT samples is smaller than that of DIT sample. Furthermore, with a decrease in ausforming temperature, the blocky retained austenite is gradually reduced and finally replaced by fine particles evenly distributing in bainite sheaves. Especially, the average size of blocky retained austenite in sample ausforming at 300 °C is below 0.3 μm, which is approximately fourth times lower than that in DIT sample. These results indicate that ausforming can largely reduce the amount and size of blocky retained austenite and increase the proportion of filmy retained austenite.

Ausforming process would produce nonuniform plastic flow in individual austenite grains, namely the operated slip system is limited for each grain [34,35]. When the primary slip plane occupies the dominate position, the majority of dislocations induced by plastic deformation would align along one (111) plane, which results in the numerous bainitic ferrite plates exhibiting a strong directionality and lying in one direction as shown in Fig. 4c, denoted by the blue arrow. However, with the increase of strength in undercooled austenite, it is difficult to coordinate plastic deformation with the limited independent slip system, two or more slip planes should be operated actively [14]. Afterwards, the growth of bainitic ferrite would distribute along two or three directions, as marked by the blue arrow in Fig. 4d, and the cross-growth phenomenon of bainite plates becomes pronounced following by formation of masonry microstructure. Due to the limitation of pre-existing bainite laths, subgrains and dislocation cell, the length of newly formed bainitic ferrite lath becomes shorter.

The measured average values of bainitic ferrite lath and filmy retained austenite thickness in various samples from
Fig. 6 – XRD patterns (a) and phase proportion (b) of different samples after isothermal treatment at 300 °C, where the abbreviation a.u. represents arbitrary units.

Fig. 7 – EBSD micrographs of different samples after isothermal treatment at 300 °C. (a) DIT; (b) 950 °C; (c) 500 °C; (d) 300 °C.

the TEM observation are shown in Fig. 5. It is noteworthy that the average thicknesses of bainite plate and filmy retained austenite in AIT samples are much thinner than that in DIT sample, and the thickness value reduces slightly with decreasing ausforming temperature, showing a downward trend with ausforming temperature. The explanations therefore may predominantly comprise of the following two main constituents: (a) the formation of dislocation remaining on active slip plane, provides abundant heterogeneous nucleation sites, leading to the refinement of final bainite lath; (b) the introduction of
partial dislocation due to the decreased stacking fault energy with diminishing ausforming temperature is very difficult to annihilate by recovery, bringing about the enhancement of undercooled austenite strength and phase transformation resistance of austenite to bainite, which would lead to the decrease of critical nucleus size and then act an additional pivotal part in microstructure refinement. Because the decreased ausforming temperature tends to further increase austenite strength, thus the lower temperature corresponds to the thinner thickness of bainitic ferrite lath.

Fig. 6 shows the XRD patterns and phase proportion for DIT and AIT samples. Only ferrite (α) and austenite (γ) phase diffraction peaks, without diffraction peaks of carbide precipitation, are observable, which is consistent with the TEM observation. The amount of bainitic ferrite in all AIT samples is generally lower than that in DIT sample. Additionally, the volume fraction of retained austenite increases with ausforming temperature decrease to 500 °C and further slightly decrease as the temperature decrease to 300 °C. It has been demonstrated that ausforming increases the number of potential bainite nucleation sites, which contributes to increasing bainite fraction. Meantime, ausforming results in an increase in mechanical stabilization of undercooled austenite, hence retarding bainite displacive growth. Therefore, it exists a competitive effect between accelerating nucleation and suppressing growth. In this scenario, it is obvious that the latter has a decisive effect on the maximum attainable volume fraction of bainitic ferrite. Compared with ausforming at either 950 °C or 300 °C, the retarding action plays a more obvious role in determining the phase composition during bainite transformation for sample ausforming at 500 °C.

Fig. 7 shows the EBSD micrographs for DIT and AIT samples, different colors stand for different crystal orientations of bainitic ferrite. It can be observed that there is a significant difference between DIT and AIT samples. The bainite sheaves of DIT sample exist in multiple orientations, yet a strong orienta-

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Table 1 - Summary of mechanical properties of the investigated steel obtained from various tests performed at room temperature.

<table>
<thead>
<tr>
<th>Ausforming temperature (°C)</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>TE (%)</th>
<th>IE (J)</th>
<th>Hardness (HV)</th>
<th>PSD (GPa %)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DIT</td>
<td>1244 ± 18</td>
<td>1620 ± 20</td>
<td>16.3 ± 1.7</td>
<td>23.8 ± 1.0</td>
<td>500 ± 15</td>
<td>26.4</td>
</tr>
<tr>
<td>950</td>
<td>1320 ± 15</td>
<td>1700 ± 17</td>
<td>19.1 ± 1.4</td>
<td>32.2 ± 1.0</td>
<td>508 ± 13</td>
<td>32.4</td>
</tr>
<tr>
<td>500</td>
<td>1255 ± 16</td>
<td>1650 ± 14</td>
<td>17.9 ± 1.1</td>
<td>21.8 ± 0.67</td>
<td>514 ± 17</td>
<td>29.5</td>
</tr>
<tr>
<td>300</td>
<td>1500 ± 20</td>
<td>1860 ± 25</td>
<td>23.3 ± 1.1</td>
<td>17.1 ± 0.53</td>
<td>585 ± 21</td>
<td>43.4</td>
</tr>
</tbody>
</table>

YS stands for yield strength, UTS for tensile strength, TE for total elongation, IE for Charpy impact energy, PSD for production of tensile strength and ductility.

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Fig. 8 – (a) Engineering stress-strain curves and amplified view of the engineering curves (marked by a dashed-line rectangle in (a)); (b) Strength versus bainitic ferrite volume fraction/plate thickness; (c) work hardening exponent curves of the investigated steel by ausforming at different temperatures and isothermal treatment at 300 °C.
Fig. 9 – Fracture morphology from the center region of the fracture surface in different tensile samples. (a) DIT; (b) 950 °C; (c) 500 °C; (d) 300 °C.

Fig. 10 – Micrograph of longitudinal section of different tensile samples. (a) DIT; (b) 950 °C; (c) 500 °C; (d) 300 °C.

3.2. Effect of ausforming on mechanical properties

The engineering stress-strain curves of DIT and AIT samples are shown in Fig. 8a, and the corresponding mechanical properties are tabulated in Table 1. All tensile curves are char-
Fig. 11 – Fracture morphology from the center region of the fracture surface in different impact samples. (a) DIT; (b) 950 °C; (c) 500 °C; (d) 300 °C.

characterized by the presence of continuous yielding behavior. Tensile strength, elongation and hardness are all improved to some extent by ausforming and by decreasing ausforming temperature as well, indicating that AIT samples possess a better combination of strength and ductility than DIT sample. These superior tensile properties are correlated with the refined microstructure.

When ausforming at 950 °C, the strength of sample increases by nearly 100 MPa. This enhancement of strength is mainly related to the reduction in the mean free path of dislocation glide owing to the finer bainitic ferrite lath by grain refinement [36]. However, sample ausforming at 500 °C performs basically similar tensile properties and slightly higher hardness relative to DIT sample. As is well known that strength and hardness of bainite primarily arise from the volume fraction of each phase and the scale of bainitic ferrite lath [37,38]. Fig. 8b shows the relationship between strength or hardness and ratio of volume fraction to lath thickness of bainitic ferrite. It can be seen that the yield strength, tensile strength and hardness all exhibit a good linear correlation with the ratio of bainitic ferrite volume fraction to plate thickness. That is, the strength of bainite can be functionally related to volume fraction and plate thickness, and the more volume fraction of bainitic ferrite and finer lath thickness, the higher strength and hardness can be obtained. Therefore, for sample ausforming at 500 °C, although the refined bainitic ferrite lath plays a positive role in improving strength, the decreased amount of bainite could bring about detrimental effect. Therefore, ausforming at high and intermediate temperature would not be of great benefit to the improvement of mechanical properties in ultra-fine bainitic steel.

Particularly, the sample which is subjected to ausforming at 300 °C, gives the highest tensile strength of almost 1850 MPa and total elongation of with up to 23%, and the product of strength and ductility (PSD) reaches as large as ~43 GPa %, in sharp contrast to DIT sample exhibiting a smaller PSD of ~26 GPa %. Consequently, the low temperature ausforming is an effective way to obtain ultrahigh strength Bainitic microstructure and will be a potential process for hot forming of ultra-fine Bainitic steel.

Fig. 8c shows the variation in instantaneous work hardening exponent (n) with true strain. The straight line refers to instability criterion \( n = n_{\text{cr}} \), where \( n_{\text{cr}} \) stands for the true strain at the beginning of necking. All samples display excellent uniform ductility and high work hardening capability, and no obvious necking can be observed in the whole tensile process until fracture failure. Compared with DIT sample, AIT samples exhibit a more enhanced and remarkably sustained work hardening phenomenon, which mainly results from the more effective transformation induced plasticity (TRIP) effect of retained austenite prior to the onset of necking instability. It has been proved that ductility properties were largely governed by retained austenite volume fraction, morphology and its metastability to mechanically induced martensite transformation [39–41]. Due to the lower carbon concentration and more inhomogeneous carbon distribution in large size of blocky retained austenite, the strain induced transformation may take place at the initial deformation, which would contribute limited effect to ductility improvement. However, the small size and film type retained austenite enriched with higher carbon content can transform into martensite in a progressive manner upon deformation, and hence the plastic deformation could accommodate in large strain range [42]. This phenomenon may result in the enhanced ductility and increased work hardening capacity by TRIP effect, which is greatly conducive to delaying necking behavior. Ausforming
process can effectively reduce the size and volume fraction of blocky retained austenite, and decrease the ratio of transformed retained austenite at the early stage of deformation, contributing to a sustainable and steady TRIP effect and providing additional strength and ductility.

Fig. 9 shows typical SEM fracture morphologies of tensile samples. It is quite notable that tensile fracture surface of DIT sample contains plenty of flat quasi-cleavage facets, a few tearing ridges and dimples, which is a quasi-cleavage type fracture. By contrast, the feature of cleavage fracture decreases while the number of tearing ridges and dimples increase for AIT samples. Ductile dimple is predominant in fracture of samples ausforming at 950 °C and 500 °C, which reveals that the major fracture micromechanism is ductile nucleation and microvoid growth. However, the quantity and depth of dimples in sample ausforming at 950 °C are slightly greater than that in sample ausforming at 500 °C, which indicates that the former has higher ductility. The fractography of sample ausforming at 300 °C shows many curved features connected together suggesting a ductile fracture due to its higher ductility approaching to ~ 26%. This may be associated with the refined morphology and increased retained austenite content with decrease in ausforming temperature.

Fig. 10 shows SEM images of longitudinal section of fracture surface for tensile samples. Some visible microcracks can be detected near tensile fracture in samples ausforming at 950 °C, 500 °C as well as DIT sample, which are all situated in blocky retained austenite, denoted by the arrow. Whereas, it is hardly to be observed in sample ausforming at 300 °C. Generally speaking, the blocky retained austenite exhibits lower mechanical stability, especially the central region, with low and inhomogeneous carbon concentration, is prone to be taken as an indicator of stain induced martensite at a relatively small plastic strain. In addition, due to the mechanical
heterogeneity (strength mismatch) between fresh martensite and blocky retained austenite, cracks are more likely to preferentially initiate at the interface of martensite and retained austenite, and propagate along them [26], which will degrade ductility.

The required time of carbon homogenization in retained austenite can be roughly calculated by the following Eq. (2) [43,44]:

$$t = \frac{I^2}{6D_0 \exp(-Q/RT)}$$

(2)

Where \( I \) stands for average diffusion distance, which refers to the half width of retained austenite, taking the size of blocky retained austenite in DIT samples and ausforming at 300 °C are respectively 1.1 \( \mu \)m and 0.28 \( \mu \)m; \( D_0 \) is a constant, 0.1 \( \times 10^{-4} \) m\(^2\)/s [45]; \( Q \) is activation energy for carbon diffusion, 135.7 kJ/m\(^3\) [45]; \( R \) is gas constant, 8.314 J/(mol·K); \( T \) is absolute temperature.

It has been calculated that it takes approximately 4 h to complete carbon homogenization in blocky retained austenite for DIT process, while ausforming can shorten required time to 0.4 h during austempering at 300 °C. Therefore, the distribution of carbon concentration inside the blocky austenite of ausforming samples is more uniform, and the reduced size of blocky retained austenite can effectively restrain crack initiation.

The impact toughnesses of various samples are listed in Table 1. The sample ausforming at 950 °C exhibits an impact toughness of 32 J/cm\(^2\), which is much higher than that of DIT sample, 23 J/cm\(^2\). However, with ausforming at 500 °C and 300 °C, the impact toughnesses of samples decrease to 21 J/cm\(^2\) and 17 J/cm\(^2\), respectively. These results clearly demonstrate that the impact toughness of sample can be slightly decreased by low temperature ausforming, but largely increased after ausforming at high temperature.

Fig. 11 shows typical SEM fractographs of center region of fracture surface in various impact samples. It is noticed that all samples are characterized by dimples and cleavage facets that are indicative of a mixed crack propagation mode of ductile fracture and brittle fracture. Thereinto, the fractographs of samples without or with ausforming at 950 °C are featured by plenty of dimples, typical of predominantly ductile fracture (Fig. 11a, b), furthermore, the dimples in ausforming samples are even finer and denser. It has been known that blocky retained austenite with poor mechanical and thermal stability tends to transform into brittle martensite during impact process, which is favorable for stress concentration, crack initiation and propagation, playing a negative role in impact damage resistance [46,47]. Therefore, decreasing the size of blocky retained austenite by microstructure refinement is beneficial for improving impact property. Whereas fracture surfaces of the samples ausforming at 500 °C and 300 °C (Fig. 11c, d) are covered with more quantity of cleavage facets and fewer small or shallow dimples, typical of predominant brittle fracture, which also correspond well with their poor impact toughness.

Specifically, in sample ausforming at 300 °C, cleavage facet and dimple appear to be distributed alternately, which may result from the nonuniform microstructure caused by deformation. It has been recognized that bainitic ferrite lath has a propensity for arresting crack propagation and deflecting crack path, and hence bainite lath with various crystal orientations can increase the resistance of crack growth [48,49]. Nevertheless, the experimental results indicate that ausforming will reduce the number of variants and lead to a very strong variant selection, which is unfavorable for suppressing crack propagation and can deteriorate impact toughness. For sample ausforming at 300 °C, the decreased degree of impact toughness caused by variant selection exceeds the promotional effect due to the increase of stable retained austenite, which appears the lowest impact toughness value.

The results of tensile and impact tests under different processes show that though ausforming can improve strength and ductility of ultra-fine bainitic steel, yet weaken impact performance. The major reason for this phenomenon is the discrepant mechanisms of retained austenite under different stress states and different strain rates. It has been recognized that ausforming can effectively refine bainite microstructures and eliminate large-scale blocky retained austenite, which has been proved to be effective in TIRP effect under plastic deformation, and thereby delaying the occurrence of necking and offering transformation toughening. However, the strain affected zone in the front of crack tip of AIT samples during impact loading, is thinner compared to DIT sample, and the strain induced martensite ahead of the crack tip will promote crack propagation [46]. Meanwhile, ausforming reduces the variant number of bainitic ferrite, which could not effectively change the crack propagation path, resulting in a worsening of impact toughness.

The affecting mechanisms of various processes are summarized in Fig. 12. For DIT sample, bainitic ferrite mainly nucleates at prior austenite grain boundary and grows into austenite grain interior. As a result of impingement effect of different packet laths, some large-scale blocky retained austenite is reserved, which takes responsibility for the lower strength and ductility. However, the multiple orientations of bainite laths can reduce stress concentration and impede crack propagation, contributing to a higher impact toughness.

For intermediate ausforming sample (500 °C), there is an evenly matched competition between microstructure refinement resulted from the increased nucleation site and decreased volume fraction of hard phase bainitic ferrite caused by mechanical stability of undercooled austenite, which keeps strength and ductility almost unchanged. With decreasing ausforming temperature to 300 °C, the multiple regions of prior austenite grains divided by deformation band, and the increased crystal defects and nucleation driving force result in shorter bainite sheaves and microstructure refinement. Low temperature ausforming significantly increases strength and ductility by superfine bainite laths together with the considerable amount of appropriately stable retained austenite giving rise to enhanced and continuous strain hardening. However, the variant selection formed during ausforming tends to deteriorate impact toughness, and with the decrease of ausforming temperature, the worsening effect becomes more obvious.
4. Conclusion

The effects of different ausforming temperatures on isothermal bainite transformation behaviors, microstructures, mechanical properties and plastic deformation mechanisms of ultra-fine bainitic steel have been conducted. The main results are as follow.

(1) As compared with DIT process, ausforming as well as decreasing ausforming temperature tends to shorten bainite incubation period and phase transformation completion time, plays a critical role in accelerating bainite transformation when austempering at 300 °C.

(2) AIT process effectively reduces the thickness of bainitic ferrite plates and filmy retained austenite, the size and amount of blocky retained austenite, thus refining the microstructure and improving tensile properties, yet increases the stability of undercooled austenite, leading to the decrease of final volume fraction of bainite.

(3) AIT process reduces the number of variants in a single austenite grain, and with decreasing ausforming temperature, the orientation selection becomes more pronounced, which slightly deteriorates impact toughness.

Conflict of interests

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper. The authors declare the following financial interests/personal relationships which may be considered as potential competing interests:

Data availability statement

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

Acknowledgments

The research was funded by the National Natural Science Foundation of China (Nos. 51774033 and U1560107) and Technology development Program of Weifang (Nos. 2019GX077). The financial support of the State Key Laboratory of Development and Application Technology of Automotive Steels is also gratefully acknowledged.

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