Study on nonuniform microstructure and strength-toughness mechanism of thick-walled hot induction seamless bend

Juntai Hu1, Yu Liu1,2, Ge Wang1 and Qiang Li1,3,*

1 School of Mechanical Engineering, Yanshan University, Qinhuangdao 066004, People’s Republic of China
2 China Petroleum Pipeline Engineering Corp, Langfang 065000, People’s Republic of China
3 School of Materials Science and Engineering, Hebei University of Technology, Tianjin 300130, People’s Republic of China

* Author to whom any correspondence should be addressed.
E-mail: ysuliq@163.com

Keywords: hot induction bending, thick-walled, continuous transformation cooling, nonuniform microstructure, mechanical properties

Abstract
This paper investigated the continuous transformation cooling curve of thick-walled seamless pipe steel, microstructure nonuniformity of thick-walled hot induction seamless bend along wall thickness direction caused by hot induction bending, and their influences on the mechanical properties. The differences in microstructure, dislocation, precipitate characteristic, local misorientation angle, and grain boundary distribution were investigated according to scanning electron microscopy, optical microscopy, transmission electron microscopy, x-ray diffraction, and Electron back-scattering diffraction. The tensile tests and low-temperature impact tests were used to determine the strength and impact toughness of thick-walled hot induction seamless bend. The results demonstrate that thick-walled seamless pipe steel can obtain bainite structure in a wide range of cooling rate, which is beneficial to the adjustment of bending process. The nonuniform microstructure is mainly caused by the difference in heating temperature and cooling rate. The structure in the center of wall thickness is dominated by granular bainite, showing the excellent combination of strength and toughness. Among all the samples, dislocation strengthening and fine-grain strengthening are the two most important strengthening mechanisms. However, for the outer surface samples, precipitation strengthening mechanism also plays an significant role, which is attributed to the high heating temperature that causes a large amount of carbonitrides to dissolve and a large amount of precipitation during the subsequent tempering process. The high density dislocations and the large average local misorientation result in dislocation pile-up and strain concentration, which promotes crack initiation and reduces toughness. Large average misorientation and high density high-angle grain boundaries help to consume more energy in crack propagation and improve impact toughness.

1. Introduction

The global demand for oil and natural gas has increased dramatically. People focus on finding and developing new oil and gas fields in remote areas. Most of the oil and gas pipelines are built in areas with harsh weather, sparsely populated, and extremely complex geological conditions, including seismic areas, fault zones and other large displacement areas. Increasing the strength of the pipeline material can not only expand the pipeline throughput, but also significantly lesson the pipeline construction and installation cost [1, 2]. In the future, pipeline transportation will develop in the direction of high strength, large diameter, large transportation volume, and thick walls. Therefore, the development of high-performance and thick-walled pipeline steel pipes and bends has been promoted [3, 4].

The bend is one of the most important components for long-distance gas and oil pipeline transportation [5]. In the process of oil and gas transportation, these components often bear more complex stresses. The bend can not only transform the transmission direction of the pipeline according to the construction requirements of the
terrain, but also buffer the tensile and compressive stress and torque attached to the pipeline such as the movement of the pipeline stratum, earthquake and environmental temperature changes. As the requirements for pipeline transportation pressure and transportation safety increase, higher requirements are also put forward for the performance of the corresponding steel-grade bends [5]. For thick-walled and large-diameter bends, in order to decrease manufacturing expenses and make certain the shape of the bends, online hot induction bending, online cooling, and offline tempering are generally used. Modern low-carbon high-strength microalloyed pipeline steels combining with high strength and toughness are derived from complex phases such as acicular ferrite, granular bainite or bainite ferrite obtained by using thermo-mechanical control process (TMCP) [7]. Normally, the parent pipe of the bends adopts double-sided submerged arc welding [8]. During the thermal cycle of welding, due to the high temperature of the molten pool, a welding heat-affected zone is formed near the deposited metal. In the process of hot induction forming of the bends, in order to minimize the forming stress, the induction heating temperature is often greater than the phase transition temperature, so that the high strength-toughness structure obtained by the advanced microalloying, controlled rolling and controlled cooling processing technology has undergone significant changes [8–10]. In the hot induction bending process, although the weld is required to be on the neutral line, the deformation of the hot induction bending process can be ignored, but the weld metal and heat-affected zone will also undergo heat treatment such as heating, cooling, and tempering at the same time process. Compared with the steel pipe matrix, the heat-affected zone subjected to welding thermal cycling and the weld metal formed by solidification have a big difference in the microstructure evolution law after the pipe bending process and the subsequent tempering. Therefore, the seamless pipe is used as the parent pipe of the bend. Since the parent pipe does not have welds, the overall performance deterioration generated with the aid of the structural evolution of the weld metal and the welding heat-affected zone for the duration of the hot induction bending procedure is avoided. The improvement of a new kind of seamless bend that satisfies the requirements of large-diameter and thick-walled pipes is imminent, which can effectually make sure its microstructure and mechanical properties as the improvement direction of bends. For thick-walled bends, during the induction heating process, the outer surface can reach the required temperature quickly due to the characteristics of induction heating, while the inner surface must be reached by heat conduction [11]. The nonuniform distribution of induction heating temperature and cooling rate along the wall thickness direction caused by the large wall thickness size leads to the difference in the microstructure and performance along the wall thickness direction [6]. As the wall thickness increases, the nonuniformity of the cross-sectional structure and performance of the bend increases significantly. Especially for the bend pipe with a wall thickness of more than 30 mm, the performance difference between the inner and outer layers is more obvious. From the current research results, researcher pay more attention to the influence of heating temperature, and there is less research on the cross-sectional temperature difference caused by induction heating. Due to the high induction heating speed, short holding time and insufficient austenitization, the hardenability of the steel is reduced, phase transition temperature of the material is increased, and the cooling rate required is greater, which further causes nonuniform microstructure along wall thickness direction.

A large quantity of papers have reported on the manufacturing process of hot induction bend and the microstructure and mechanical properties of different positions in the bending zone along the circumferential direction [12–15]. Wang et al investigated the effect of bending process parameters of the X80 hot bending pipeline steel on the microstructure, mechanical properties and toughness, including the induction heating quenching temperature, cooling rate and tempering temperature [6]. Zhou et al described the effect of tempering process of heavy-wall hot induction bend pipe on the microstructure evolution, mechanical properties and strength-toughness mechanism. Wang et al revealed the microstructure evolution of X90 hot induction bend, including the influence of dislocation characteristics, precipitate behavior, grain orientation distribution, grain boundary distribution, recrystallization and texture on the mechanical properties of the inner arc side, neutral axis and outer arc side [13]. Wang et al expounded in detail that influence of quenching temperature on microstructure evolution and yield strength and the effect of welding heat input on microstructure and impact toughness in CGHAZ of X100Q seamless pipe steel [16, 17]. Godeforid et al investigated the microstructure and mechanical properties of seamless steel pipes API 5L type obtained by different processes of heat treatments, and believed the quenching and tempering process improved tensile mechanical properties and fracture toughness, but combined to reduce in fatigue crack growth resistance significantly [18]. However, few researchers have reported on microstructure nonuniformity along wall thickness direction of the X80 × D1422 mm thick-walled seamless bend during the heat-induced bending process and its influence on the tensile properties and low-temperature impact toughness.

In this article, the continuous cooling transformation (CCT) diagram of the seamless pipe steel was studied, and the microstructure, tensile properties, and low-temperature impact toughness of thick-walled hot induction seamless bend were investigated. The causes of the nonuniform microstructure and the relationship between the strength-toughness mechanism and the effective grain size, dislocations, precipitation behavior, misorientation angle, and local misorientation distribution were also systematically observed.
2. Experimental procedure

2.1. Materials and hot induction bending process

The material investigated in this study was thick-walled X80 × D1422 mm hot induction seamless bend with a wall thickness of 38.5 mm, an outer diameter of 1422 mm, and a pipe length of 6220 mm. The chemical composition of the hot induction seamless bend is presented in Table 1. The starting temperature (Ac₁) and completion temperature (Ac₃) for austenitization were measured as 776 and 863 °C, respectively. The hot induction bending process parameters such as the heating temperature, pushing velocity, and tempering temperature for heavy-wall hot induction seamless bend are exhibited in Table 2. The industrial production bending drawing and schematic diagram of the hot induction bending process is displayed in Figures 1(a)–(d).

Initially, the mother pipe was induction heated to 1050 °C at a medium frequency, which is greater than the austenitizing temperature, and was once pushed to plastic deformation at this temperature. Secondly, after hot bending, online water cooling must be operated. Lastly, the seamless bend was tempered at 580 °C to achieve a favorable combination of strength and toughness. Normally, the wall thickness on the outer arc side of bend is barely reduced, the wall thickness on the inner arc side is barely increased, and the wall thickness of the neutral axis is nearly unaffected. The primary motive of this article was to lookup the effect of the distinction in microstructure induced with the aid of the inhomogeneous temperature distribution along wall thickness direction of the heavy-wall bend on the mechanical properties for the duration of the hot bending process, therefore, a smaller bending angle and a larger bending radius was used in this experiment. The test sample was located on the outer arc side without considering the deformation factors during the bending process, as shown in Figure 1.
in figure 1(e), named as the inner surface (appointed as S1), the center position (appointed as S2), and the outer surface (appointed as S3), respectively.

2.2. Determination of CCT diagram

The continuous cooling transformation (CCT) diagrams of the X80 seamless pipe steel was determined by means of dilatability on a Gleeble 3800 hot simulator. The samples are cylindrical rods of 90 mm in length and 6 mm in diameter. In a vacuum state, the specimens were austenitized at 1000 °C for 3 min, held for 150 s and then cooled down to room temperature directly with a sequence of constant cooling rates from 1 to 60 °Cs⁻¹. The temperature-strain expansion curves of the sample during the cooling process were measured, the metallographic morphologies were observed, and then the start and end temperatures of various phase transitions at different cooling rates were determined.

2.3. Determine the critical phase transition temperature (Ar₃) at different heating temperatures

A round bar specimen of Ф10 × 75 mm was used in the Gleeble-3800 system to simulate heating at 930, 990, and 1050 °C, holding for 100 s, and cooling to 200 °C at 10 °C s⁻¹. The expansion curves were recorded automatically during cooling via a dilatometer, and the Ar₃ for each sample was finally determined via the tangent method.

2.4. Microstructure analysis

The samples for optical microscopy (OM) and scanning electron microscopy (SEM) observation were prepared by wire cut electrical discharge machining, mechanical grinding, polishing and then etched with 4% nital. Standard metallographic techniques were used to observe the microstructure by using Zeiss optical microscope and a scanning electron microscope Quanta 450 field emission gun (FEG) operating at 20 kV. For further microstructural analysis, transmission electron microscopy (TEM) observation was also conducted and the TEM samples were prepared by cutting thin wafers from small coupons, grinding them to 40–50 μm and then punching into Ф3 mm discs. They were twin-jet electrolytically-polished in a solution consisting of 10% perchloric acid and 90% glacial acetic acid at 50 V/−30 °C for about 2 min. These foils were examined by a JEM-2100(HR) TEM operating at 200 kV. The Image-Pro Plus software (Media Cybernetics, Rockville, MD, USA) was used to statistically calculate the average size (D_p) and the volume fraction (f_p) of the particles. Electron back-scattering diffraction (EBSD) with orientation imaging microscope system was employed on a Hitachi S-3400N SEM and was performed under the condition of tilt angle 70° with an acceleration voltage of 20 kV to investigate grain boundary distribution and grain misorientation characteristics. Samples for EBSD analysis were first ground and mechanically polished, and then electrolytically polished for stress relief in an electrolyte of 5% perchloric acid in alcohol at 50 V for 25 s. The size of scan region was used to be approximately 473 × 355 μm with a step size of 0.3 μm. Channel5-Oxford Instruments software was used for the post processing of the collected EBSD patterns. The local misorientation distribution component is used to calculate the average misorientation between every pixel and its surrounding pixels, then assigned the mean value to that pixel. The magnitude of local misorientation can...
be used to evaluate the magnitude of stress distribution in the tested steel because it is closely related to the dislocation density and the local strain. X-ray diffraction (XRD) analysis was carried out to determine the dislocation density ($\rho$). The XRD samples were examined over a 2$\theta$ range from 40°–110° with a step measurement of 0.02° and a scanning speed of 2° min$^{-1}$ in a Rigaku DMAX-RB x-ray diffractometer, operated at 150 mA and 40 V using a graphite monochromator and filtered Cu-K$\alpha$ radiation ($\lambda = 0.1506$ nm).

2.5. Mechanical properties
In order to evaluate the comprehensive mechanical properties, a tensile test and a Charpy V-notch impact test were carried out. The test of tensile property was performed with ASTM E-8 standard on an 810 Materials Testing System (MTS810) with a strain rate of 1.3 $\times$ 10$^{-3}$/s at room temperature. According to ASTM E-8 specification, the standard tensile specimen with the diameter of 8 mm and gauge length of 25 mm were processed. The low temperature impact toughness was measured on a Zwick/Roell 450 J impact tester with a test Xpert® II data acquisition system at –20 °C. The standard size of the samples was a V-shaped notch of 10 $\times$ 10 $\times$ 55 mm, which is separated from the transverse direction.

3. Results
3.1. Microstructures of transformed products during continuous cooling and CCT diagram
In order to systematically analyze the law of structural transformation of steel during the cooling process, the continuous cooling transformation curve (CCT) of X80 seamless pipe steel was determined. Figure 2 displays the microstructure of the thick-walled seamless parent pipe, which is mainly composed of granular bainite and part of bainite ferrite. Figure 3 indicates the metallurgical structure of the seamless pipe steel under different cooling rates. It can be seen from figure 3 that the X80 seamless pipe steel has high hardenability. During the continuous transformation and cooling process, the main microstructure formed is the low-temperature transformation structure, which includes a very small amount of martensite, a large amount of bainite ferrite and granular bainite, as well as part of polygonal ferrite and pearlite.

When the cooling rate is 40 °C–60 °C/s, the microstructure is typical bainite ferrite and a very small amount of martensite. The bainite ferrite is in the form of lath bundles, and the band-shaped M–A constituents are distributed in
parallel between the laths. Part of the original austenite grain boundaries are clearly visible. As the cooling rate decreases to 20 °C–30 °C/s, the microstructure is mainly bainite ferrite and part of granular bainite. The original austenite grain boundary partly disappears, and the island-like structure is distributed between the bainite ferrite laths in a rod shape. When the cooling rate is reduced to 5 °C–10 °C/s, the microstructure is mainly granular bainite, accompanied by a small amount of lath bainite, and the original austenite grain boundaries disappear. The island-like structure is mainly distributed on the bainite matrix in a granular or dotted shape. When the cooling rate is reduced to 2 °C/s, the microstructure is mainly granular bainite and irregular massive ferrite, accompanied by a very small amount of pearlite, and pearlite is mainly distributed between ferrite. The lath-like structure disappeared completely. The island structure is distributed in dots on the granular bainite matrix. When the cooling rate is reduced to 1 °C/s, the main structure is polygonal ferrite, quasi-polygonal ferrite, and pearlite.

Figure 4 displays the CCT diagram, which is drawn based on the phase transition point determined by the expansion method and the observation result of the metallographic structure. The percentage fraction of different microstructures at different cooling rates is also shown in figure 4. For example, when the cooling rate is 20 °C/s, the microstructure includes 95% granular bainite and 5% martensite. There are three main transformation zones in the CCT diagram: under the condition of low cooling rate, it transforms into ferrite and pearlite; under the condition of medium and high cooling rate, it transforms into granular bainite and lath bainite; under extremely high cooling rate, part of it transforms into martensite. At moderate cooling rates and high cooling rates, the polygonal ferrite and pearlite regions are significantly delayed, and the phase transition onset temperature is reduced. The X80 seamless pipe steel can obtain granular bainite and bainite ferrite in a wide cooling rate range, and the critical cooling rate for obtaining granular bainite is small, which is conducive to adjusting the bending process within a wider process window.
3.2. Mechanical properties

Figure 5 indicates the tensile properties and impact toughness of the heavy-wall X80 hot induction seamless bend at different sampling positions. The yield strength of S1, S2, and S3 were 563 MPa, 596 MPa, and 637 MPa, respectively. The tensile strength of S1, S2, and S3 were 686 MPa, 745 MPa, 829 MPa, respectively. The impact toughness of S1, S2 and S3 were 225 J, 215 J and 150 J, respectively. S3 shows the lowest impact toughness, and S2 and S3 have better impact toughness. S2 indicates an excellent combination of strength and toughness.

3.3. Microstructural characteristics

3.3.1. OM and SEM

Figure 6 displays the microstructure of different sampling positions of heavy-wall X80 hot induced seamless bend, including S1, S2, and S3. S1 is composed of polygonal ferrite (PF) and degraded pearlite (DP), as exhibited in figures 6 (a), (d). Polygonal ferrite is a proeutectoid ferrite formed at a slower cooling rate and a high transformation temperature. It preferentially nucleates from the austenite grain boundary and has a regular grain shape. The growth of polygonal ferrite is manifested by the long-range diffusion of carbon atoms and the rapid migration of replacement atoms. Its growth can cross the original austenite grain boundaries, so that the contours of the austenite grain boundaries are covered, and the grain boundaries are relatively straight, clear and smooth. The interior of the polygonal ferrite is not
enough to form a local carbon concentration gradient. When the carbon content exceeds the solid solubility of the polygonal ferrite, a carbon-rich zone is formed next to the PF matrix. The carbon-rich austenite is relatively stable and cannot be transformed into PF. Due to the low cooling rate, it undergoes a complex transformation at a higher temperature than $M_s$, and then transforms into the degraded structure of martensite–austenite constituents (M–A constituents), that is, degraded pearlite, mainly distributed between the ferrite matrix. Under an optical microscope, the polygonal ferrite is white and bright, the crystal grains have an equiaxed or polygonal shape, and the degraded pearlite presents a black mass, as shown in figure 6(a). No M–A constituents was found in S1. S2 is mainly composed of granular bainite (GB). M–A constituents present a granular structure, distributed in ferrite grain boundaries or ferrite matrix, with larger size and less quantity than S3, as shown in figures 6(b), (e). The structure of granular bainite is ferrite with small island-like structures distributed on the matrix. Under an optical microscope, GB appears as irregular blocks, with M–A constituents visible on the inside and borders. The transformation temperature of GB is the highest among various bainite transformation processes. The diffusion coefficient of C is large, and C can diffuse in austenite for a long distance. The low-C area forms a ferrite $\alpha$ phase, and C is concentrated in the untransformed austenite area. With the continuous precipitation of ferrite, the C content in the original austenite structure increases. As the temperature further decreases, part of the carbon-rich austenite transforms into martensite, and the other part remains austenite at room temperature, transforming into the M–A constituents. S3 is mainly composed of lath bainite (or bainite ferrite), as shown in figures 6(c), (f). Lath bainite (LB) has a lower transformation temperature than GB granular bainite, and rapid cooling makes the ferrite appear in a lath-like parallel arrangement. Obvious austenite grain boundaries can be observed in S3, and the original austenite grain size is larger. The M–A constituents mainly presents a dotted-like or rod-like structure distributed between the bainite ferrite laths, and its size is small and the number is large.

Figure 7. Transmission electron micrographs of of thick-walled hot induction seamless bend in different position samples. (a)–(c) S1, (d)–(f) S2, (g)–(i) S3.
3.3.2. TEM observation

Although OM and SEM can analyze some microstructure characteristics, the substructure and M–A constituents morphology need to be further observed by TEM, as shown in figure 7. S1 presents an equilibrium structure, and the ferrite grain boundaries are relatively regular, as shown in figure 7(a). After the nucleation of polygonal ferrite, the diffusion rate of carbon atoms into untransformed austenite is reduced, and a large amount of degraded pearlite is found around polygonal ferrite, as exhibited in figures 7(b), (c). Figures 7(d)–(f) display the morphology of S2, which mainly presents a lath-like structure. The width of the lath is larger than that of S3, and the length of the lath is shorter than that of S3. The M–A constituents are distributed in strips between the laths and the size is larger than S3. Part of the granular bainite presents an irregular and featureless shape, and the M–A constituents are distributed among the ferrite matrix in a granular form. Figures 7(g)–(i) show that S3 exhibits a clear lath structure, and M–A constituents are distributed between the ferrite laths with film-like shape. Some laths are also recovered and recrystallized, the laths become wider and polygonal, the boundaries of adjacent laths are blurred, the subgrain boundaries in the laths move or disappear, and the M–A islands undergo transformation and decomposition. The structure and the evolution of the shape and size of M–A constituents are related to the tempering process. M–A constituents is the product of the last transformation during the cooling process, and its stability is the weakest. During the tempering process, the organizational morphology of the M–A constituents also changed. The morphology of the film-like M–A constituents between the bainite ferrite laths of S3 may be derived from the decomposition of striped-like M–A constituents during the tempering process. The M–A constituents in the granular bainite of S2 are partially decomposed or granulated during tempering.

3.3.3. Dislocation investigation and precipitation behavior

In order to further clarify the relationship between the microstructure and mechanical properties, the dislocation morphology and precipitation behavior were investigated in depth. The main forms of dislocations in S3 are high-density entanglement dislocations in the lath or accumulation of dislocations in the grain boundaries, as shown in figure 8. Most of the entanglement dislocations are difficult to eliminate during the tempering process, accompanied by a large number of precipitated phases, which are pinned together with the precipitated phases. Another part of the high-density entangled dislocations evolved into a cellular substructure during the tempering process, or the dislocations slipped to the grain boundary and disappeared. The energy spectrum analysis and the mapping distribution of the partially analyzed phases in figure 8(e) show that these precipitated carbonitrides are mainly (Nb,V)(C,N). There are usually two types of precipitated phases. The first is the larger-sized carbonitrides precipitated in high-temperature austenite grains, which are mainly used to refine the grains and restrain the recrystallization of austenite grains. The second is to precipitate in the ferrite during the cooling process and the tempering process. Its size is small and has a strong precipitation strengthening effect. Figure 9 shows that there are still high-density entangled dislocations and cellular substructures between the laths in S2. Some dislocations may be eliminated during tempering, and the dislocation density is lower than that of S3. In figure 9(e), A large amount of precipitated phase (Nb,V)(C,N) still exists, pinning dislocations and cellular substructures. The number of precipitated phases in S2 is less than that in S3, which may be related to the heating temperature. Due to the high heating temperature of S3, more carbonitrides are dissolved during the heating process, and more dispersed carbonitrides are produced during the subsequent tempering process.
Figure 10 shows that the polygonal ferrite in S1 exhibits a low dislocation density. The dislocation density near the grain boundary of the polygonal ferrite is higher than that in the crystal, and the dislocation strengthening effect is not obvious. There are dislocation accumulations near the ferrite grain boundaries, and the precipitated phases also exist on the dislocations and the grain boundaries, mainly Nb(C, N), as shown in figure 10(e). Use Image-Pro Plus software to perform statistics in a local area, and obtain the average size ($D_p$) and volume fraction ($f_p$) of the precipitated phase particles, as shown in table 3.

3.3.4. XRD analyses
Different samples were also investigated by XRD, and the typical XRD patterns are displayed in figure 11. The dislocation density, $\rho$, could be quantitatively calculated from the XRD pattern, and calculated via the following equation:

$$\rho = \frac{6\pi \varepsilon^2}{b^2}$$

Where $\varepsilon$ represents the non-uniform strain, and $b$ represents the Burst’s vector of the dislocation in $\alpha$-Fe. $\varepsilon$ and $b$ can be determined by XRD [19]. The dislocation densities of S1, S2 and S3 are $3.12 \times 10^{14}$ mm$^{-3}$, $4.57 \times 10^{14}$ mm$^{-3}$ and $6.26 \times 10^{14}$ mm$^{-3}$, respectively.
3.3.5. EBSD analyses

EBSD is appropriate for investigating and quantifying the characteristics of microstructures, in particular especially irregular grain boundaries. In the subject of crystallography, HAGBs acquired from EBSD could present the effective grain size (EGS)\(^1\). However, for S3 with a lath structure, a original austenite grains is divided into several packets. A packet is divided into blocks with the same habit plane, which consists of a number of laths in the same or similar orientation. Alloy steel material with lath structure has a crystallographic structural unit whose dimension is between the original austenite grain size and the block, that is, the effective grain size of the lath structure. Figure 12 shows the grain orientation distribution map and the grain size distribution map of different positions of thick-walled hot induction seamless bend. The area method by setting the critical misorientation angle was used on the grain orientation distribution maps to quantify the effective grain size. The effective grain sizes of S1, S2 and S3 are 4.05 µm, 4.76 µm, 6.32 µm, respectively. As we all know, grain refinement is benefical because it not only enhances grain boundary strength but also improves the impact toughness. Therefore, the smaller the grain size, the better mechanical properties.

4. Discussion

4.1. The main reason for the formation of inhomegeneous microstructure

Induction heating and subsequent tempering are two important links in the manufacturing process of the bend, which has an important influence on the final structure and mechanical properties of the bend. The nonuniform microstructure is attributed to the distinct heating temperature and cooling rate at different positions of the heavy wall during the heat-induced bending process. The heating temperature and cooling rate from low to high are S1, S2 and S3, respectively. Due to the high induction heating speed, short holding time, and insufficient austenite homogenization, it will also reduce the hardenability of the steel, increase the phase transition temperature, and require a greater cooling rate to obtain the ideal microstructure. Due to the low heating

![Figure 11. XRD patterns of thick-walled hot induction seamless bend in different position samples.](image)

| Table 3. The average size (\(D_p\)) and volume fraction (\(f_p\)) of the precipitated phase particles. |
|---------------------------------|---------------------------------|
| Average size (\(D_p\))/nm | Volume fraction (\(f_p\))/% |
| S1 82 ± 1.1 | 1.0 ± 0.08 |
| S2 80 ± 0.9 | 2.3 ± 0.11 |
| S3 65 ± 2.1 | 3.6 ± 0.10 |
temperature of S1, the carbonitride has not yet been dissolved, and the pinning effect on the grain boundary was more obvious during the austenitization process, so the austenite grains have not grown. The heating time of the induction heating bend was short, the C and alloying elements in the austenite have not been diffused and homogenized, and the austenite was in a non-uniform state. In addition, the cooling rate of S1 is low, which further reduces the hardenability of the steel, thus obtaining an equilibrium structure and mainly exhibiting polygonal ferrite.\(^{[21]}\) The austenite grain boundary also provided a large number of nucleation sites for polygonal ferrite, which makes the grain size of S1 smaller. As the heating temperature of S2 was higher than that of S1 during the bending process, the precipitates (carbonitrides) began to dissolve, the diffusion of C and alloying elements was accelerated, and the heterogeneous distribution of C and alloying elements in austenite was improved. In addition, since the cooling rate of S2 is higher than that of S1, the hardenability of the steel is increased, so an uniform mid-temperature transition structure is mainly obtained after cooling, which improves the strength and toughness. The pinning effect of carbonitride on the austenite grain boundary in S2 was reduced, and the austenite grains of S2 grew up, so the final grain size of S2 was larger than that of S1. Due to the high heating temperature of S3, the original austenite grains grew, carbonitrides were dissolved in a large

**Figure 12.** Grain orientation distribution maps and grain size distribution map of thick-walled hot induction seamless bend in different position samples. (a) (b) S1, (c) (d) S2, (e) (f) S3.
amount, and the density of carbonitrides pinning the original austenite grain boundaries decreased. The increase in solid solubility of elements that stabilize the austenite such C, N, Nb, Mn, and Ni may increase the thermodynamic stability of the $\gamma$ phase and increase the degree of subcooling of the phase transition. In addition, the cooling rate of S3 was the largest, which increases the hardenability of the material and reduces $\text{Ar}_3$, thereby providing a greater thermodynamic driving forces for nucleation and growth of LB. However, the coarse pre-austenite grains still resulted in a larger effective grain size of S3. Considering the starting temperature of the transformation from austenite to ferrite, it may have a great influence on the final microstructure [22, 23].

Figure 13 shows the $\text{Ar}_3$ of the phase transition curve at different heating temperatures. The results show that the higher the heating temperature, the lower the phase transition temperature ($\text{Ar}_3$), which is consistent with the above analysis. It can be obtained from the CCT diagram that through a greater cooling rate, a greater degree of sub-cooling can be obtained, and the phase transition temperature can be further reduced.

After high temperature heating and online cooling, the bends need to be tempered accordingly. The tempering process directly affects the final performance of the bend. Tempering can not only adjust the quenched microstructure to improve its comprehensive mechanical properties, but also eliminate the internal stress after hot bending and water cooling, and improve the fracture resistance and safety of the bend. Different matrix structures and M–A constituents formed in different temperature ranges during the cooling process have different tempering stability. Although during the tempering process, the transformation of the structure includes the evolution of the ferrite matrix, M–A constituents, and the precipitation of carbonitrides. However, tempering does not change the basic microscopic morphology of the matrix structure. The influence of tempering on the microstructure of the matrix structure, the precipitation of M–A constituents and carbonitrides has been described above. In summary, considering the discrepancy in heating temperature at different positions, the original austenite grain size, as well as the content and uniformity of C and alloying elements in the austenite of the thick-walled seamless bend were strongly affected. In addition, the different cooling conditions result in different hardenability of steel and different temperature ranges for phase transformation, showing nonuniform microstructure along the wall thickness direction.

4.2. The relationship between microstructure and yield strength

The inhomogeneous microstructure shows different yield strengths. The effective grain size, dislocations, precipitates and solid solution content of the microstructure contribute differently to the yield strength, resulting in different strengthening mechanisms, including fine-grain strengthening, precipitation strengthening, dislocation strengthening and solid solution strengthening, etc. The results are exhibited in Table 4.

When the material is subjected to external forces, the dislocation source is activated, and the dislocation slips and proliferates along a certain crystal plane. Due to the different orientations of adjacent grains, dislocations
slip to the grain boundaries and form dislocation accumulations, forming a stress field at the grain boundaries, which may stimulate dislocations in adjacent grains. The strength of the stress field of stacked dislocations is related to the number of dislocations, and the number of stacked dislocations is related to the size of the grains [24]. The finer grains, the greater the number of dislocations. Hall-Petch summarized the relationship between yield strength and grain size based on the above viewpoints [25]. Therefore, the contribution of grain size to yield strength $\sigma_g$ can be calculated according to the following formula:

$$\sigma_g = k \times d^{-1/2}$$  \hspace{1cm} (2)

where $k$ is a constant, 0.55 MPa$\cdot$m$^{1/2}$ is usually used for low-alloy high-strength steel, and $d$ is the effective grain size.

In addition to solid solution, the microalloying elements added to pipeline steel also combine with C and N atoms to form carbonitride precipitates. The strengthening effect of the precipitated carbonitrides is related to the characteristics, quantity and size distribution of the second phase. The contribution of the precipitated phase particles to the strength ($\sigma_p$) can be quantitatively evaluated by the following formula:

$$\sigma_p = \frac{11.3 \times f_p^{1/2}}{D_p} \ln (D_p/0.4963) \times 10^3$$  \hspace{1cm} (3)

where $D_p$ and $f_p$ represent the average size and volume fraction of precipitated phase particles, respectively [25, 26].

### Table 4. Various strengthening contributions from corresponding microstructural attributes in different position samples.

|     | $\sigma_g$/MPa | $\sigma_p$/MPa | $\sigma_s$/MPa | $\sigma_d$/MPa | $\sigma_s+\sigma_g$/MPa |
|-----|----------------|----------------|----------------|----------------|-------------------------|
| S1  | 563            | 273            | 71             | 146            | 73                      |
| S2  | 596            | 252            | 108            | 177            | 59                      |
| S3  | 637            | 218            | 160            | 207            | 52                      |

Figure 14. Local misorientation distribution maps. (a) S1, (b) S2, (c) S3, (d) Quantitative analysis of local misorientation angle.
The theory of metal defects believes that there are always dislocations in metal crystal grains, and the increase in dislocation density can effectively improve the strength. Metal materials are strengthened by plastic deformation, and the essence of plastic deformation is dislocation movement and proliferation. Therefore, dislocation strengthening is also one of the most effective strengthening methods in metal materials. The relationship between the contribution of dislocation strengthening to the yield strength \(\sigma_d\) and the dislocation density \(\rho\) can be expressed as:

\[
\sigma_d = \alpha M G b \rho
\]

where \(\alpha, M, G\) and \(b\) represent constant (0.15), Taylor factor (2.73), shear modulus (81.6 GPa) and Bragg's vector (0.248 nm), respectively [27, 28].

In addition to the above strengthening factors, the sum of lattice resistance \(\sigma_0\) and solid solution strengthening \(\sigma_s\) can be expressed as [29]:

\[
\sigma_s + \sigma_0 = \sigma_f - (\sigma_d + \sigma_p)
\]

Refinement strengthening and dislocation strengthening are the main strengthening mechanisms for seamless bends. The effective grain size of S3 is large, but its yield strength is high. This abnormal phenomenon must be caused by precipitation strengthening. Due to the high heating temperature of the outer surface, more carbonitrides are precipitated during the tempering process, which makes the precipitation strengthening play an important role, causing the outer surface to exhibit higher strength.

4.3. Influence of EBSD-based microstructure on impact toughness

The local misorientation distribution map can be used to visualize the plastic deformation. Figure 14 shows local misorientation distribution at different locations. The local misorientation distribution can be used to evaluate the local strain gradient and dislocation concentration in the material [30, 31]. This is an effective way to define the stress concentration in the area with an orientation difference of 0° to 5°. In figures 14(a)–(c), the blue color, green, yellow, orange, and red respectively represent misorientations less than 1°, between 1° and 2°, between 2° and 3°, between 3° and 4°, and between 4° and 5°. The misorientations larger than 5° are not taken into consideration, because they may be due to the development of low angle grain boundaries. Mapping results show that most of the areas in blue represent local misorientation range between 0° and 1°, correspond to the

---

Figure 15. Grain boundaries distribution maps. (a) S1, (b) S2, (c) S3, (d) Misorientation angle distribution.
ferrite areas. Figure 14(d) shows the quantitative distribution of local strain. The area where the local misorientation is greater than 1 degree gradually increases from S1 to S3. The local misorientation image shifts to the right as a whole. The average local misorientation is 0.57°, 0.69°, and 0.81° in turn, indicating that the dislocation density and stress concentration gradually increase, which is consistent with the mapping distribution of local misorientation. In S3, the yellow area is mainly related to lath bainite. This is attributed to the high density accumulation of dislocations in the lath ferrite during the rapid cooling process during the bending process. It can also be obtained that in S1, there is also a partially orange area around the polygonal ferrite, which may be due to the accumulation of dislocations in the degraded pearlite. Reducing strain concentration and dislocation concentration is beneficial to toughness. High strain concentration and high density of dislocations can increase the tendency of dislocation accumulation and internal stress, and it promotes the nucleation, leading to decrease in toughness, which is consistent with the results of the TEM and XRD analysis.

The grain boundaries distribution maps of the heavy wall seamless bend pipe analyzed by EBSD are shown in figure 15. In figures 15(a)–(c), the black lines represent the high angle grain boundaries (HAGBs) that greater than 15°, and the red lines correspond to low angle grain boundaries (LAGBs) are characterized as boundaries with misorientation of 2°–15°. In the crystallographic theory, these two types of grain boundaries are usually used to represent the main grain boundaries [32, 33]. The proportions of the high-angle grain boundaries of S1, S2 and S3 are 41.1%, 38.3%, and 29.1%, and the average misorientation angles are 20.93°, 20.87°, and 16.71°, respectively, as presented in figure 15(d). Many studies have shown that in the fracture process, when the crack tip encounters a high-angle grain boundary, the crack propagation will be twisted or deflected. The more high angle grain boundaries, the more energy is dissipated during crack propagation. However, compared with high-angle grain boundaries, low-angle grain boundaries have little torsional effect on crack propagation. So S3 shows lower impact toughness. On the other hand, the microstructure of S3 is composed of bainite, and cracks are prone to propagate along the bainite laths. S1 and S2 exhibited higher impact toughness.

5. Conclusion

1. The X80 seamless pipe steel can obtain granular bainite and bainite ferrite in a wide cooling rate range, and the critical cooling rate for obtaining granular bainite is low.

2. In thick-walled seamless bends, the nonuniform microstructure is mainly caused by the difference in heating temperature and cooling rate. The heating temperature strongly affects the grain size of the original austenite, and the content and uniformity of C and alloying elements in the austenite, which in turn leads to different hardenability of the steel and different temperature ranges for phase transformation, resulting in the beginning of phase transformation The temperature (Ar3) decreases. In addition, the difference in the cooling rate further causes the nonuniform microstructure. Polygonal ferrite, granular bainite and lath bainite are mainly present on the inner surface, the center of the wall thickness and the outer surface, respectively.

3. Refinement strengthening and dislocation strengthening are the main strengthening mechanisms for thick-walled hot induction seamless bends. Due to the high heating temperature of the outer surface, more carbonitrides are precipitated during the tempering process, which makes the precipitation strengthening play an important role, causing the outer surface to exhibit higher strength.

4. Large average local misorientation, high-density dislocation distribution, low average misorientation and a small proportion of high-angle grain boundaries result in lower impact toughness on the outer surface, and the center of the wall thickness shows the best combination of strength and toughness.

Acknowledgments

The present work is supported by the National Natural Science Foundation of China (No.51761030).

Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).
Conflicts of interest

The authors declare no conflicts of interest.

ORCID iDs

Qiang Li https://orcid.org/0000-0002-6376-1822

References

[1] Guo A, Misra R D K, Xu J, Guo B and Janisato S G 2010 Ultrahigh strength and low yield ratio of niobium-microalloyed 900 MPa pipeline steel with nano/ultrafine bainitic lath Mater. Sci. Eng. A 527 3886
[2] Sharma S K and Maheshwari S 2017 A review on welding of high strength oil and gas pipeline steels J. Nat. Gas Sci. Eng. 38 203
[3] Sun J Q, Dai H and Zhang Y C 2011 Research on mathematical model of thermal deformation resistance of X80 pipeline steel Mater. Des. 32 1612
[4] Yoo J Y, Ahn S S, Seo D H, Song W H and Kang K B 2011 New development of high grade X80 to X120 pipeline steels Mater. Manuf. Process. 26 154
[5] Wang X, Zhou J and Liang Q 2014 Multi-objective optimization of medium frequency induction heating process for large diameter pipe bending Procedia Engineering 81 2255
[6] Wang X, Xiao F R, Fu Y H, Chen X W and Liao B 2011 Material development for grade X80 heavy-wall hot induction bends Mater. Sci. Eng. A 530 539
[7] Xu J, Misra R D K, Guo B, Jia Z and Zheng L 2013 Understanding variability in mechanical properties of hot rolled microalloyed pipeline steels: process-structure-property relationship Mater. Sci. Eng. A 574 94
[8] Wang X, Liao B, Wu D Y, Han X L, Zhang Y S and Xiao F R 2014 Effects of hot bending parameters on microstructure and mechanical properties of weld metal for X80 hot bends J. Iron. Steel Res. Int. 21 1129
[9] Stepanov P P, Zikerev V N, Efremov L I, Frantov I I and Morozov Y D 2011 Improvement in steel weldability for large diameter thick-walled gas pipelines by optimizing chemical composition Metallurgist 54 767
[10] Ren D L, Xiao F R, Tian P, Wang X and Liao B 2009 Effects of welding wire composition and welding process on the weld metal toughness of submerged arc welded pipeline steel Int. J. Min. Met. Mater. 16 65
[11] Wang Z T and Zhong H 1990 Theory of pipe-bending to a small bend radius using induction heating J. Mater. Process. Tech. 21 275
[12] Zhou T, Yu H, Hu J L and Wang Y 2014 Study of microstructural evolution and strength–toughness mechanism of heavy-wall induction bend pipe Mater. Sci. Eng. A 615 436
[13] Wang B, Wang L, Jiang Y, Xu M, Lei B B, Hu Y, Wu D, Luo Z and Liu L 2016 Microstructure and mechanical behavior of X90 bend using local induction bending Trans. Indian Inst. Met. 70 115
[14] Wang L, Wang B and Hu Y 2021 Effect of hot induction bending on microstructure, precipitate characteristics, and mechanical properties in thick-walled X90 bend J. Mater. Eng. Perform. 30 2267
[15] Wang L, Wang B and Zhou P 2018 Misorientation, grain boundary, texture and recrystallization study in X90 hot bend related to mechanical properties Mater. Sci. Eng. A 711 588
[16] Wang H, Wang F, Shi G, Sun Y, Liu J, Wang Q and Zhang F 2018 Effect of quenching temperature on microstructure and yield strength of Q-T-treated X100Q bainitic steel Mater. Res. Express 5 066509
[17] Wang H B, Wang F L, Shi G H, Sun Y, Liu J C, Wang Q F and Zhang F C 2019 Effect of welding heat input on microstructure and impact toughness in CGHAZ of X100Q steel J. Iron. Steel Res. Int. 26 637
[18] Godfredso I B, Sena B M and Trinidad Filho V B D 2017 Evaluation of microstructure and mechanical properties of seamless steel pipes API 5L type obtained by different processes of heat treatments Mater. Res. 20 514
[19] García–Mateo C, Caballero F G and Bhdashia H K D H 2005 Mechanical properties of low-temperature bainite Mater. Sci. Forum. 500–501 495
[20] Dáaz-Fuentes M, Iza-Mendia A and Gutiérrez I 2003 Analysis of different acicular ferrite microstructures in low-carbon steels by electron backscattered diffraction Study of their toughness behavior Metall. Mater. Trans. A 34 2505
[21] Duan L N, Wang J M, Liu Q Y, Sun X J and Cao J C 2010 Austenite grain growth behavior of X80 pipeline steel in heating process J. Iron. Steel Res. Int. 17 62
[22] Ming L, Wang Q, Wang H, Zhang C, Wei Z and Guo A 2014 A remarkable role of niobium precipitation in refining microstructure and improving toughness of A QT–treated 20CrMo17NbV steel with ultrahigh strength Mater. Sci. Eng. A 613 240
[23] Ishihara A N, Sugahashi H and Borhani G 2011 The effect of heat treatment on mechanical properties and corrosion behavior of AISI 3420 martensitic stainless steel J. Alloys Compd. 509 3931
[24] Sylwestrowicz W and Hall E O 1951 The deformation and ageing of mild steel Proc. Phys. Soc. Sect. B 64 495
[25] Hansen N 2004 Hall–petch relation and boundary strengthening Scr. Mater. 51 801
[26] Fan L, Wang T, Fu Z, Zhang S and Wang Q 2014 Effect of heat-treatment on-line process temperature on the microstructure and tensile properties of a low carbon Nb–microalloyed steel Mater. Sci. Eng. A 607 559
[27] Bandypadhyay P S, Ghosh S K, Kundu S and Chatterjee S 2011 Evolution of microstructure and mechanical properties of thermomechanically processed ultra-high-strength steel Metall. Trans. A 42 2742
[28] Morrison W B 2009 Microalloy steels—the beginning Mater. Technol. 25 1066
[29] Yakubtsov I A, Poruks P and Boyd J D 2008 Microstructure and mechanical properties of bainitic low carbon high strength plate steels Mater. Sci. Eng. A 480 109
[30] Sun J, Yu H, Wang S and Fan Y 2014 Study of microstructural evolution, microstructure-mechanical properties correlation and collaborative deformation–transformation behavior of quenching and partitioning (Q&P) steel Mater. Sci. Eng. A 596 89
[31] Zaefferer S, Ohlert J and Bleck W 2004 A study of microstructure, transformation mechanisms and correlation between microstructure and mechanical properties of a low alloyed TRIP steel Acta Mater. 52 2765
[32] Zhang C, Wang Q, Ren J, Li L, Wang M, Zhang F and Sun K 2012 Effect of martensitic morphology on mechanical properties of an as-quenched and tempered 25CrMo17V8 steel Mater. Sci. Eng. A 534 339
[33] Zhang J M, Huo C Y, Ma Q R and Feng Y R 2018 NiC–TiN co-precipitation behavior and mechanical properties of X90 pipeline steels by critical-temperature rolling process Int. J. Pressure Vessels Pip. 165 29