Achievement of ultra-high strength in 45Si2MnCr2Mo steel via efficient and low-energy tempering

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Abstract
In this study, 45Si2MnCr2Mo steel was fabricated by casting, hot rolling, and normalizing, followed by tempering at different temperatures from 180 to 520 °C for 2 h. Special attention was paid to the effects of tempering on the microstructure and mechanical properties of the steel. The results indicate that microstructure of tempered steel consists of bainite, martensite, and retained austenite (RA), and it is not sensitive to tempering at temperatures below 320 °C. However, after tempering at temperatures of 420 and 520 °C, bainitic ferrite plates merged and their lath boundaries became vague. In turn, RA decomposed, whereas carbide particles appeared and then coarsened. With increasing tempering temperature, the volume fraction of RA achieved its peak value at 320 °C and carbon concentration within RA attained its peak value at 420 °C. The optimal balance of ultimate tensile strength of 2306 MPa and total elongation of 8.24% in the steel were achieved after being normalized at 950 °C for 0.5 h and subsequent tempering at 220 °C for 2 h. These values are found to exceed those of air-cooled and isothermal bainitic steels. Finally, the correlation between microstructure and mechanical properties of the steel under consideration was discussed.

1. Introduction
There is an increased need for protection of vehicle armor system, in particular, for armor personnel carriers or infantry fighting vehicles [1–3]. In this regard, significant research attention has been paid to systematically explore the ways to reduce the weight of armor, and correspondingly, to determine the appropriate materials with the minimum possible areal density. High-strength steels are commonly used in armors because of their low expense, excellent mechanical properties, good machinability, and superior performance. The impact resistance of steels depends on their microstructure and resulting strength, hardness, toughness, strain hardening rate, etc For instance, optimization of microstructure and properties to resist the impact of armor made of high-strength steel has always been of practical interest for military applications [4–7].

Among different types of steel that combine outstanding strength and high hardness along with good toughness and wear resistance, bainitic steel with excellent combination of mechanical properties has been considered as the most attractive for commercial applications. However, early bainitic steels failed to achieve simultaneous improvement in strength and toughness due to elevated carbon concentrations [8]. In recent years, isothermal quenching (austempering) processes have been carefully studied to acquire a good practice in synthesis of bainitic steels with required tensile properties and toughness [9–15]. However, such steels are usually obtained by isothermal oil quenching or salt bath heating, which involves sophisticated stages and high-energy consumption process. The air-cooled bainitic steel is hardened by air cooling method with the advantages of cost effectiveness, simple and feasible technique, short manufacturing period, and low energy...
consumption. However, unfortunately, air-cooled bainitic steels still possess insufficient strength and ductility to meet the stringent demands for armor protection [16–20]. Moreover, air-cooled bainitic steel exhibits unstable structure in which internal stresses are generated by phase transformations. In this respect, tempering can improve microstructural stability and eliminate internal stress, thus improving the performance and stabilizing the geometric sizes of steel workpieces [21–23]. Till date, numerous studies have been carried out to uncover the effects of tempering temperature on the mechanical performances of steels [24–26]. However, the accurate selection of tempering temperature requires the simultaneous and careful consideration of several factors such as, the adequate strength loss at a high tempering temperature and tempering embrittlement [11, 27] at a medium tempering temperature. Moreover, different tempering temperatures have different effects on microstructure and mechanical properties. For instance, Saastamoinen et al. [28] pointed out that low-temperature tempering (below 300 °C) was able to increase the tensile strength and hardness. Once the tempering temperature exceeded 300 °C, the softening process led to a decrease in the strength of steel. Wang et al. [29] investigated the retained austenite (RA) morphology on thermal and mechanical stability, the blocky M/A islands start to decompose after tempering at 280 °C but the thin M/A strips change slightly after tempering at 280 °C, showing high thermal stability. When the tempering temperature is above 350 °C, the thin M/A strips start to decompose. Furthermore, some carbides have been formed at higher tempering temperature, and the bainitic ferrite lath disappears and merges dramatically with the tempering temperature increases to 400 °C. The bainitic rail steels exhibit excellent strength and toughness combination after tempering at 280 °C (UTS of 1388 MPa and TE of 16%). Kanga et al. found that the thickness of a tempered bainitic ferrite plate did not increase significantly at temperatures below 360 °C. In turn, the RA content increased with the increase in the tempering temperature, reaching its maximum at 400 °C. A further increase in the temperature to 450 °C led to a noticeable coarsening of the bainitic ferrite plates and the emergence of a small amount of carbide precipitates. The optimum performance characteristics of the steel (ultimate tensile strength (UTS) of 1850 MPa and total elongation (TE) of 15%) were obtained at the tempering temperature of 340 °C [15].

Notably, Si is a non-carbide forming element that serves as austenite strengthening agent to improve the tempering stability of the nano-bainite and refined bainitic plates [30–32]. Besides, Liu et al. [33] found that the solution-hardening effect of Si is beneficial to the yield strength, yield strength increases with Si content at a rate of 104.23 (low tempering temperature) and 137.33 (high tempering temperature) MPa per wt% in low-carbon steels. Increased Si contents can result in higher dislocation densities in martensite and bainite, also lead to finer cementite sizes in bainite [34]. In particular, use of up to 1.5 wt% of Si enables the prevention of cementite precipitation from austenite during the reaction of bainite under isothermal quenching and to form the carbide-free bainite [35]. Varshney et al. [11] investigated a low alloy multiphase steels of 1.71 wt% Si, and obtained super strength and highly ductile (UTS of 1994.9 MPa and TE of 10.5%). It is generally recommended that Si content should not exceed 3 wt%, otherwise it significantly reduces the plasticity and toughness of steel [36]. However, García–Mateo et al. [9] analyzed the tensile behavior of 3 wt% Si-doped nanocrystalline bainitic steel, and observed the unparalleled balance between strength and ductility (UTS of 2 GPa and TE of 21%), although under higher Si (>3 wt%) concentrations. Moreover, Cr allows for the enhancement of wear resistance and hardenability of low alloy steels [12, 13]. In particular, Long et al. [12] explored 34MnSiCrAlNiMo medium carbon steel, for which the UTs of 1616 MPa and the TE of over 15% could be obtained after adding 1 wt% Cr. Furthermore, the larger the added content of Cr, the higher the strength. For instance, introduction of 1.58 wt% Cr in a medium carbon steel resulted in the UTs of 1921 MPa and theYS of 1487 MPa [13]. Therefore, the use of elevated amounts of Si and Cr is an effective strategy to improve the mechanical performance of steels.

In the present study, 45Si2MnCr2Mo bainitic steel plate with increased Si (2 wt%) and Cr (2 wt%) contents was produced by successively carrying out casting, hot rolling, normalizing, and tempering processes. Special attention was paid to the effects of tempering temperature on the microstructure and mechanical properties of the steel. In particular, the optimal combination of mechanical characteristics, namely, UTS of 2306 MPa and TE of 8.24%, could be achieved, which were found to be better than those of traditional air-cooled bainitic steels and isothermal bainitic steels. The correlation between microstructure and mechanical properties of 45Si2MnCr2Mo steel was discussed as well. Noticeably, the normalizing process applied in this study was less laborious and less-energy consuming than conventional isothermal treatment of bainitic steels, which can thus be conducive to the large-scale industrial application.

2. Experimental

A 5-mm-thick 45Si2MnCr2Mo bainitic steel plate was produced by intermediate frequency induction melting and hot rolling. The chemical composition of the hot-rolled steel was determined via spectrochemical analysis using an ARL IRON + STEEL metal analyzer (see table 1 for details). The phase transformation temperatures were found using a DIL-805A dilatometer. Prior to the dilatometry experiments, cylindrical specimens with
diameters of 4 mm and lengths of 10 mm were cut from the hot-rolled plate. The specimens were then heated to 1000 °C at a heating rate of 0.1 °C·s⁻¹ and kept at this temperature for 10 min. Then, the specimen was cooled down at a rate of 20 °C·s⁻¹ to room temperature (RT). In the dilatometry curve plotted in figure 1(a), three points denoted as Ac₁, Ac₃, and Mₛ can be clearly distinguished. The first is the temperature at the beginning of the transformation from the original structure to the austenite phase in the steel during heating. The second refers to a final temperature at which the microstructure in the steel gets completely transformed into the austenite phase during heating. The third is initial temperature of martensitic transformation. According to the plot, the corresponding values are 773, 837, and 217 °C, respectively. For the subsequent heat treatment, samples with widths of 35 mm, lengths of 115 mm, and thicknesses of 5 mm were cut along the rolling direction. According to the dilatometry data, all the specimens were normalized at 950 °C for 0.5 h, cooled down to RT in air, and then tempered at different temperatures (180, 220, 320, 420, and 520 °C) for 2 h (figure 1(b)).

The microstructural features of the specimens were carefully characterized by scanning electron microscopy (SEM) and transmission electron microscopy (TEM). The specimens for the SEM analysis were prepared by mechanical grinding with 400 # to 2000 # grit silicon carbide papers followed by polishing and etching for 15 s with 4 vol.% nitric acid alcohol solution, and then they were scanned using an Auriga microscope. The specimens for the TEM measurements were made by cutting thin slices from the steel piece followed by the mechanical reduction of their thickness to ∼50 μm. The obtained foils were then electrochemically polished with 10 vol.% HClO₄-methanol electrolyte at a temperature of about −20 °C [37]. The TEM examinations of the samples were performed using a Tecnai G2 F30 S-TWIN transmission electron microscope (America FEI) operating at 300 kV.

The mechanical properties of the heat-treated specimens were evaluated via tensile and hardness tests. Prior to tensile testing, the plate specimens with the gauge lengths of 25 mm were cut along the rolling direction of the heat-treated steels. The tensile tests were conducted at RT using an electronic universal testing machine controlled with a WDW-200D-type microcomputer. The measurements were carried out at a strain rate of 6.6 × 10⁻⁴ s⁻¹ and an extensometer calibration interval of 25 mm according to the ASTM E8M standard [11]. A minimum of three trials were performed for each heated sample. The hardness testing was implemented using a Rockwell durometer (TH 320) under the load of 150 kgf at a dwell time of 3 s and a recovery time of 1 s.

In order to accurately determine the volume fraction of RA, the X-ray diffraction (XRD) experiments were carried out by utilizing a Rigaku TTR III X-Ray diffractometer with Cu Kα radiation operating at 20 KW. The XRD profiles were recorded at a scanning speed (2θ) of 1°·min⁻¹ in an angular range from 45 to 95° with unfiltered CuKα radiation. The RA volume fraction and the carbon content in austenite were calculated through

![Figure 1](https://example.com/figure1.png)

**Figure 1.** (a) Thermal expansion curve obtained by the dilatometry experiment. (b) Schematic diagram of different heat treatment processes for the experimental steel. Herein, Ac₁ is the temperature at the beginning of the transformation from the original structure to the austenite phase in the steel during heating. Ac₃ is final temperature at which the microstructure in the steel gets completely transformed into the austenite phase during heating. Mₛ is initial temperature of martensitic transformation.

| Table 1. Chemical composition (wt%) of the 45Si2MnCr2Mo steel under consideration. |
|------------------|--------|-------|-----|-------|-----|-----|-----|--------|
| Composition      | C      | Si    | Mn  | Cr    | Mo  | Ni  | Fe  |        |
| Nominal          | 0.45   | 2.0   | 0.8 | 2.0   | 0.25| 0.2 | Balance |        |
| Tested           | 0.43   | 1.71  | 0.89| 2.07  | 0.24| 0.22| Balance|        |

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a direct comparison method based on the integrated intensities of (200), (220), and (311) austenite (\(\gamma\)) peaks as well as (200) and (211) of ferrite (\(\alpha\)) signals [38].

3. Results and discussion

3.1. Mechanical properties

Figure 2 depicts the hardness-tempering temperature plots of the specimens. For each specimen, the hardness value was obtained by averaging data over five measurements, and the standard deviation was no more than 0.7 HRC. The highest value of hardness was approximately 57.8 HRC for the specimen tempered at 220 °C for 2 h. Notably, no drastic changes were observed in hardness within a tempering temperature range of 180°C–320°C, exhibiting only a small difference of less than 2 HRC. However, the hardness dropped sharply when the tempering temperature was above 420 °C, and the lowest value of about 32 HRC was recorded at 520 °C. This was attributed to the fact that recovery and recrystallization of \(\alpha\) phases along with carbide aggregation and spheroidization occurred at around 420 °C, which led to the significant growth of the crystal grains and the reduction in the defect density. As a result, the hardness decreased dramatically.

Figure 3(a) shows the engineering stress–strain curves of the steels subjected to normalizing and tempering at different temperatures. In turn, the evolutions of UTS, yield strength (YS), TE, and uniform elongation (UE) at different tempering temperatures were obtained as shown in figure 3(b). The results indicate that the UTS first increased to its highest value of 2306 MPa at a temperature of 220 °C and then gradually decreased with a further increase in the tempering temperature. This behavior coincides with that of hardness trend shown in figure 2. The same trend was also observed for the YS parameter with the only difference that it reached its peak value (1806 MPa) at 180 °C. In contrast, the TE and UE values both increased linearly within the studied tempering temperature range and attained their maxima of 10.3 and 6.5% at 520 °C, respectively. Figure 3(c) depicts the tensile properties of the steel under consideration in comparison with those of other high strength steels [9–20, 39–42]. Clearly, the best combination of the mechanical properties was achieved in the present study.

3.2. Microstructural evaluation

Figure 4 shows the SEM and TEM images of microstructure and the selected area diffraction pattern (SADP) of RA and twinned martensite of the steel normalized at 950 °C for 0.5 h. Figure 4(a) illustrates that the microstructure of the steel exhibits the existence of bainite (B), martensite (M), and blocky RA phases, which significantly distinct from the austempered microstructure in which only B and RA phases appear [43]. Besides, the morphology of B in the normalized steel is similar to that of lower bainite formed during austempering [44]. The microstructural characteristics of the normalized specimen were further evaluated by TEM. Figure 4(b) exhibits the existence of uniformly distributed bainitic ferrite (BF) plates with clear phase interfaces and widths ranging from 0.2 to \(~1 \mu m\). The RA film between the BF plates could be observed. Besides, numerous dislocations emerged in the BF phase. Moreover, the blocky RA in the BF matrix (figure 4(c)) were detected by the SADP analysis (inset in figure 4(c)). The SADP analysis also revealed the presence of the twinned martensite (the inset in figure 4(d)) within the plate ferrite (figure 4(d)). The formation of the twinned martensite was
correlated with a high carbon concentration that caused the partitioning of carbon atoms during the austenite-to-bainite transformation \[45\].

Figure 5 shows the SEM microstructures of steel specimens tempered at various temperatures after normalizing. The microstructures of specimens tempered at 180 and 220 °C still exhibited the presence of bainitic lath (BL) and M phases accompanied with a visible primitive austenite (PA) boundary, as shown in figures 5(a) and (b). It indicates the absence of any noticeable change in the metallic structure during low-temperature tempering (180 and 220 °C). In contrast, small carbide inclusions appeared at 320 °C, whereas the BL slats began to merge and the slat boundaries became blurry, as shown in figure 5(c). With the further increase in the tempering temperature to 420 °C, the BL slats overlapped and their boundaries could no longer be distinguished; in turn, a large amount of precipitated and coarsened carbide particles was observed (figure 5(d)). Finally, tempering at 520 °C led to the evolution of carbides into spherulites and long stripes \[46\] (figure 5(e)). At the same time, the slats were barely distinguishable, the PA grain boundaries were degraded, and plenty of carbides got precipitated at the grain boundaries, which is in accordance with the former investigation \[47\].

Figure 6 depicts the TEM images of the microstructures of different specimens tempered at 180, 220, 320, 420, and 520 °C. The microstructures of the steels tempered at 180 and 220 °C (figures 6(a) and (b)) consist of the M substructure along with RA and B with the packed block substructures in the BF, which is similar to the features of the as-normalized steel (figures 4(b) and (c)). Tempering at 320 °C led to the decomposition of the BF and RA into the ferrite matrix and carbides (figure 6(c)). Furthermore, the blocky RA almost disappeared at 420 °C and the thin RA film was still present at the ferrite lath boundary (figure 6(d)). An increase in the tempering temperature to 520 °C was conducive to the disappearance of both the blocky RA and the thin RA film (figure 6(e)). Noticeably, the blocky RAs in the steels tempered at 180, 220, and 320 °C were identified from the SADPs (the insets in figures 6(a)–(c), respectively). Moreover, although carbides were not observed in the BF after tempering below 320 °C (figures 6(a) and (b)), these were finely dispersed in the ferrite matrix when the tempering temperature climbed to 420 and 520 °C (figures 6(d) and (f)).

Based on the above-mentioned experimental description of the arrangement and morphological change of RA in the steel specimens after normalizing and tempering procedures, the amounts of RA at various tempering temperatures were further analyzed by XRD, and the corresponding data are shown in figure 7. Clearly, the
intensities of austenite peaks ((200)γ and (220)γ) of the specimens tempered at 180 and 220 °C remained unchanged compared to those of the normalized steel, which indicated good stability of austenite at lower tempering temperatures. However, with the increase in the tempering temperature, the austenite peak monotonically decreased and finally disappeared at 520 °C, which is consistent with the microstructural evolution of RA shown in figures 4 and 6.

Volume fractions of the RA (Vγ) and carbon concentrations in the RA (Cγ) in the studied steels were calculated by using the lattice parameters of the RA extracted from the XRD patterns (figure 7) as follows [48, 49]:

\[
V_\gamma = \frac{1}{n} \sum_{j=1}^{n} \frac{I_j}{R_j} \left( \frac{1}{n} \sum_{j=1}^{n} \frac{I_j}{R_j} + \frac{1}{n} \sum_{j=1}^{n} \frac{I_j}{R_j} \right) \tag{1}
\]

\[
a(\text{Å}) = 3.578 + 0.033C_\gamma \tag{2}
\]

\[
R = \left( \frac{1}{\sqrt{V}} \right) \left[ |F|^2 P \left( \frac{(1 + \cos^2 2\theta) / \sin \theta \sin 2\theta} \right) \right] e^{-2M} \tag{3}
\]

where \( n \) is the examined number of peaks, \( I \) is the integrated intensity of the diffraction peak, \( R \) is the scattering factor of the material, \( v \) is the unit cell volume, \( F \) is the structure factor, \( P \) is the multiplicity factor, \( e^{-2M} \) is the temperature factor, \( 2\theta \) is the diffraction angle, and \( a(\text{Å}(\cdot)) \) is the lattice parameter of austenite, in line with the

\[\text{Figure 4. Cross-sectional micrographs of steel normalized at 930 °C for 0.5 h: (a) SEM micrograph, (b) bright-field TEM image, (c) TEM micrograph revealing blocky RA denoted with a white arrow (the inset displays the SADP of the marked blocky RA), (d) TEM micrograph with twinned martensite marked with a white arrow (the inset depicts the corresponding SADP). Herein, Blocky RA is blocky retained austenite, BL bundle is bainite lath bundle, M is martensite, Film RA is film retained austenite, and BF denotes bainitic ferrite.}\]
The method proposed in literature study \cite{15}. The corresponding results are illustrated in figure 8. According to these plots, the values $V_\gamma$ of the tempered specimens were much lower than that of the normalized one ($V_{\gamma0}$). In particular, the $V_\gamma$ parameter first increased with increasing tempering temperature; however, it then declined drastically once the temperature exceeded 320 \degree C (figure 8(a)). This behavior could be interpreted by the decomposition of the blocky RA at the temperature above 320 \degree C (figure 6(c)) and their further disappearance at 420 \degree C (figure 6(d)). The data agreed with the previously reported decrease in the amount of RA with increasing tempering temperature \cite{15, 50, 51}. According to the diffusivity expression $D = D_0 \exp(-Q/R\cdot T)$ (where $D_0$ is the diffusion constant, $Q$ is the diffusion activation energy, $R$ is the gas constant, and $T$ is the thermodynamic temperature) \cite{52}, the diffusion coefficients of solute atoms such as carbon, nitrogen, and microalloying elements increase exponentially with temperature. Therefore, $V_\gamma$ enhances with the increase in the tempering temperature from 180 to 320 \degree C. The strong carbide forming elements are completely dispersed under the tempering temperature above 320 \degree C, which induces the precipitation of different types of carbides along with the reduction of the carbon concentration in the RA, and therefore leads to the decomposition of the RA into ferrites and the corresponding decrease in $V_\gamma$.

Figure 5. Cross-sectional SEM micrographs of 45MnSi2Cr2Mo steel tempered for 2 h at (a) 180, (b) 220, (c) 320, (d) 420, and (e) 520 \degree C, after normalization at 950 \degree C. Herein, PA is primitive austenite boundary, BL bundle is bainitic lath bundle, and M denotes martensite.

Figure 8(b) shows that the value $C_\gamma$ increases with increasing tempering temperature, achieving its peak value at 420 \degree C, and then it starts declining. Noteworthy, the $C_\gamma$ values of the tempered specimens were all larger than that of the normalized specimen ($C_{\gamma0}$). Furthermore, carbon could diffuse from a supersaturated M to the austenite when the tempering temperature was between 180 and 320 \degree C, which improved the thermal stability of austenite and thus resulted in the increase of $V_\gamma$ as presented in figure 8(a). However, once the tempering
Figure 6. TEM and HRTEM images revealing the microstructure of specimens tempered at (a) 180, (b) 220, (c) 320, (d) 420, and (e) 520 °C after normalization at 950 °C for 30 min. (f) HRTEM image of carbide. Herein, Blocky RA is blocky retained austenite, Film RA is film retained austenite, and M is martensite. The white arrows in images (a)–(c) denote the RA. The insets in images (a)–(c) display the diffraction patterns of blocky RA.
temperature reached 420 °C, the \( V_{\gamma} \) value decreased with the increase in the carbon concentration in the RA. This reverse variation could be attributed to the high temperature-induced instability of the RA and the corresponding \( \gamma \rightarrow B \) transformation \[50\] (figures 5(d) and 6(d)). As mentioned previously, no carbides were observed in this study at the temperature below 320 °C (figures 6(a)–(c)). However, the emergence of carbide precipitates at temperatures below 320 °C was reported \[24, 51, 53\] at the relatively low silicon content. This indicates that the alloy under consideration contained a large amount of Si, which suppressed the precipitation of carbides by analogy with literature studies \[31, 32, 35\]. In turn, tempering at a temperature above 420 °C caused the dispersion of numerous carbides in the ferrite matrix and at the ferrite boundaries, followed by their spheroidization and coarsening at the ferrite matrix and boundaries (figures 6(d) and (e)), which is consistent with the results of previous study \[54\]. Furthermore, besides the dislocation rearrangement, the TEM and XRD analysis revealed a competition between the RA-to-bainite transformation and the decomposition to carbides during the tempering process \[15\]. The improvement of the heat stability of AR after tempering at an appropriate temperature might be attributed to the transformation of a small amount of RA into B and secondary quenched martensite during tempering. This transformation often starts from the low-carbon zone, thus leading to the increase in the carbon concentration in the remaining RA \[55\]. The carbon atoms in the RA phase segregate, forming a Korotkoff gas mass, which strengthens the RA phase and increases the shear resistance of the martensitic transformation \[56\]. A further increase in the tempering temperature, on the one hand, neutralizes the effect of RA strengthening; however, on the other hand, it is conducive to the precipitation of a large amount of carbides in the carbon-rich RA.

**Figure 7.** XRD patterns of normalized and tempered 45Si2MnCr2Mo steels: (200) \( \gamma \) and (220) \( \gamma \) are fcc-Fe (austenite) peaks, (200) \( \alpha \) and (220) \( \alpha \) are bcc-Fe (martensite and bainite) peaks.

**Figure 8.** The evolution of (a) volume fraction of RA \( V_{\gamma} \) and (b) carbon concentration within RA \( C_{\gamma} \), with tempering temperature. \( V_{\gamma0} \) and \( C_{\gamma0} \) represent the volume fraction of RA and the carbon concentration in the normalized specimen, respectively.
The carbon content of the remaining RA phase itself decreases, leading to an increase in the value of $M_T$, which is manifested as a decrease in the heat stability of the RA phase.

### 3.3. Correlation between microstructure and mechanical performance of steel

After normalizing, the test steel structure consisted of martensite, bainite, and a small amount of RA. Tempering after normalizing led to the transformation of M, RA, and carbides. These transformations occurred through recovery and recrystallization, that is, microstructure disappearance, recombination, and reorganization.

Tempering at a lower temperature (220 °C and below) enlarged the driving force for the transformation of RA to lower bainite with a denser structure and higher hardness. Tempering at 320 °C resulted in direct nucleation of the carbon atoms from the segregated zone (near the dislocation line or at the slat boundary), causing the precipitation of carbides, as shown in figure 5(c). Moreover, the M phase decomposed with the formation of carbides. The amount of RA in the steel was low; therefore, the increase in the hardness caused by the RA decomposition was much slower than the hardness reduction induced by the continuous decomposition of tempered M. Therefore, the hardness of the steel decreased. Tempering at 420 °C with the precipitation, growth, aggregation, and spheroidization of carbides along with the recovery and recrystallization of the α phase resulted in a gradual decrease in the hardness of the steel. The hardness was gradually reduced with the decrease of solid-solution strengthening and the level of internal stress [57]. Tempering at 520 °C allowed the carbides to grow rapidly and the hardness to reach its lowest value. In general, tempering above 320 °C led to a decrease in the hardness with increasing temperature and no secondary hardening was detected. During the tempering process, the precipitated carbides could produce a certain degree of precipitation strengthening, which, however, was much smaller than the reduction degree of the solid solution strengthening due to the deposition of interstitial atoms from the solid matrix [43].

After low-temperature tempering at 220 °C, the thermal stability and mechanical stability of RA could be improved. Therefore, the structure composed of B, M, and RA phases combined the outstanding strength ($UTS = 2306$ MPa) and ductility ($TE = 8.24\%$), being even better than those of air-cooled bainitic steel and isothermal bainitic steel (figure 7(c)). The soft RA embedded in the interior of the hard bainitic ferrite matrix contributed to proper plasticity and toughness [58, 59]. The elevated RA volume fraction with a low carbon concentration provided the high TE. Moreover, nano or submicron-sized bainite ferrite plates and high density of dislocations yielded the high strength [60]. The specimen tempered at 320 °C showed the decrease of UTS and elongation along with the increased YS value. In turn, the B slats began to merge, the boundaries became blurred, and at the same time, fine and dispersed carbides appeared, as shown in figure 3(c). The low-carbon RA at 320 °C exhibited low thermal stability. The finely dispersed carbides interacted with dislocations, resulting in the refinement of the structure and improvement of the YS, as shown in figure 7(b). The carbides precipitated in the bainitic ferrite or along the ferrite lath interface to ensure higher tensile strength, but simultaneously exhibited tempered brittleness, resulting in the reduction of elongation. After tempering at above 420 °C, the slats further merged and the boundaries were barely distinguished. At the same time, a large amount of carbides precipitated and got coarsened, the crystal grains grew dramatically, and the defect density was reduced, resulting in a noticeable decrease in the hardness of the steel. Consequently, this softening effect resulted in significant increase in the elongation [11]. After tempering, the decrease in the hardness and UTS could be ascribed to the recovery of laths and the decrease in the content of interstitial carbon atoms in the matrix [61].

### 4. Conclusions

From the investigated heat treatment routes, mechanical properties, and microstructural analysis performed on a 45Si2MnCr2Mo ultra-high strength steel, the following conclusions are summarized as:

1. The 45Si2MnCr2Mo steel possessed the optimal combination of strength ($UTS = 2306$ MPa and $YS = 1530$ MPa) and ductility ($TE = 8.24\%$) after normalizing at and subsequent tempering at 220 °C for 2 h, which are even better than those of air-cooled bainitic steel, but comparable to the values reported for isothermal bainitic steel.

2. The tempered microstructure consisted of bainite (B), martensite (M), and retained austenite (RA) phases, which exhibited no sensitivity to tempering at temperature below 320 °C. With the increase in the tempering temperature to 420 and 520 °C, the bainitic ferrite plates merged and their lath boundaries got blurred, whereas the RA decomposed with the precipitation and coarsening of carbides.

3. With increasing tempering temperature, the volume fraction of the RA ($V_\gamma$) increased linearly at temperature below 320 °C and then reduced drastically. In turn, the carbon concentration in the RA ($C_\gamma$) parameter achieved its peak value at 420 °C and then reduced slightly with a further increment in the tempering temperature. At the tempering temperature of 420 °C, a significant increase in the diffusion
coefficient of carbon caused the decomposition of the RA with a low carbon content preferentially into carbides, resulting in a decrease in the volume fraction of the RA phase.

(4) The heat treatment realized in this study is less laborious and less energy-consuming, being simpler than traditional isothermal processing. Moreover, it allows to enrich the outstanding mechanical performance of the steel, which makes it suitable for the large-scale industrial applications.

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Data availability statement

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

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