Effect of Microstructure and Precipitates on Mechanical Properties of Cr–Mo–V Alloy Steel with Different Austenitizing Temperatures

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The mechanical properties of as-quenched and tempered steels are affected by austenitizing temperature. The present work has investigated the effect of austenitizing temperature on martensitic microstructure, carbide precipitates and mechanical properties of 30NiCrMoV12 alloy steel for the axle of high-speed train. The martensitic microstructure and carbide precipitates were studied using OM, FE-SEM, TEM, EBSD and EDS. Thermodynamic calculation of equilibrium precipitation were carried out by Thermo-Calc software. The results showed that the prior austenite grains, martensitic packets, blocks and laths were coarsening with increasing austenitizing temperature. Besides, with increasing austenitizing temperature, after tempering the amount of large size carbides precipitated at martensitic lath boundaries decreased while the amount of small size carbides precipitated in matrix increased. Meanwhile, phase transformation from M23C6 to M7C3 during tempering was enhanced with increasing austenitizing temperature. Coarse grains and wide martensitic laths were beneficial to reducing the amount of strip-like M23C6 carbides precipitated at martensitic lath boundaries due to the reduction of boundary area and thereby obtaining more fine precipitates in matrix. The strength and impact toughness could be improved to a certain extent by refining carbides in tempered steel with higher austenitizing temperature. However, the degree of favorable influence on impact toughness resulting from refining carbides was lower than the negative effect from coarse martensitic structures. Therefore, the toughness is deteriorated and the strength is improved with increasing austenitizing temperature.

KEY WORDS: microstructure; precipitation; mechanical property; austenitizing temperature; Cr–Mo–V alloy steel.

1. Introduction

Austenitizing temperature has significant effect on prior austenite grain (PAG) size, which influences the size of martensitic packets, blocks and laths, and thereby influences properties. Normally, one PAG is composed of several packets, which is further divided into blocks, and each block consists of several laths with parallel arrangement, as shown in Fig. 1. Extensive researches have been conducted to analyze the relationship between martensitic microstructure and impact toughness. In an early work by Tomita

![Fig. 1. Schematic diagram of martensitic structure.](image-url)
and Okabayashi,\(^5\) it was reported that the toughness was dominated by packet size, which is the minimum effective microstructure unit, and toughness could also be improved to a certain degree with decreasing martensitic lath width. Lately, Wang\(^3\) also found that the cracks propagation was strongly hindered by packet boundaries in 17CrNiMo6 steel, and the toughness has no obvious relationship with the lath width. However, Naylors\(^7\) found that the toughness (DBTT) was related to a logarithmic function of the packet diameter and lath width. Zhang\(^2\) studied the effect of martensitic morphology on impact toughness of 25CrMo48V steel and reported that the improvement of impact toughness was attributed to the refinement of martensitic microstructure with more high-angle grain boundaries, and the block was the minimum structure unit controlling impact toughness. Therefore, there is still dispute about the minimum effective structure of lath martensite dominating the toughness.

It was reported\(^{15–19}\) that not only the microstructure but also the carbides precipitated during tempering had an effect on properties. Some researchers\(^{15–17}\) pointed out that large size particles precipitated along boundaries of PAG, martensitic packet, block and lath deteriorated the toughness. Abet\(^18\) pointed out that spherical particles were better than bar shaped particles on improving toughness. More alloying elements can be dissolved in austenite with increasing austenitizing temperature, which contributes to the variation of precipitates characteristic during tempering, such as number density, size, type, shape and distribution. As a result, the strength and toughness can be strongly influenced by austenitizing temperature. In addition, the microstructure has an influence on the area of grain and sub-grain boundaries, which affects precipitation locations of carbides.\(^19\)

As mentioned above, austenitizing temperature has influences on both microstructures and precipitates, which significantly affects mechanical properties. So far, however, there is still dispute about the minimum effective structure of lath martensite dominating toughness. In addition, there are few reports about the effects of austenitizing temperature and martensitic microstructure on carbide precipitates during tempering in as-quenched and tempered Cr–Mo–V alloy steel. Therefore, the aim of present work is to clarify the relationship between martensitic microstructure and toughness, investigate the influence of lath martensitic microstructure on the behavior of carbides precipitation during tempering, and provide a heat treatment process for developing train axle steel 30NiCrMoV12 with excellent mechanical properties.

### 2. Materials and Methods

The steel used for investigation was melted with 30 kg vacuum induction furnace, and then forged into bars with a diameter of 16 mm. An optical emission spectrometer (OES) was used to evaluate the chemical composition and the result is listed in **Table 1**. Firstly, the steel bars were normalized at 920°C for 0.5 h to homogenize the microstructure. Secondly, they were heated at different austenitizing temperatures of 880°C, 920°C, 960°C, 1000°C, 1040°C and 1080°C for 0.5 h, and then quenched in oil immediately. Finally, they were tempered at 600°C for 1 h. The schematic illustration of heat treatment procedures is shown in **Fig. 2**.

The quenched specimens were etched with supersaturated picric acid and sodium dodecyl benzene sulfonate aqueous solution at 65°C for 2 min after mechanical polishing for optical microscopy (OM, 9XB-PC) observation of PAG. The specimens were etched with 4% nital after PAG observation to observe the martensitic packets with cold field emission scanning electron microscope (FE-SEM, FEI MLA-250). The Zeiss Auriga focused ion beam field emission scanning electron microscope equipped with an Oxford Instruments Nordlys nano EBSD detector was used to reveal martensitic blocks and the scanning step was 0.1 μm. Martensitic lath was observed by scanning transmission electron microscopy (STEM, Tecnai G2 F20 S-TWIN). The size of PAG, martensitic lath width, carbides size and area fraction have been measured using Image J software.

The type of carbide precipitates was analyzed by energy dispersive spectrometer (EDS, SHD-FD-3022) and TEM. Carbon extraction replica samples were observed under TEM for carbide area fraction statistics. The sample used to prepare carbon extraction replicas for TEM observation were cut from tempered specimens, mechanically polished, and etched with 4% nital. Then a carbon film with 20–30 nm thickness was plated on the sample surface. The samples were then immersed in 4% nital to strip off the carbon film. Finally, the carbon films were collected with copper net.\(^{20}\) The equilibrium phase diagram of the investigated steel was calculated using Thermo-Calc software package with TCFE6 database. The EBSD specimens were electro-polished with 5% perchloric acid. Thin foil samples for the investigation of martensite laths under TEM were prepared by twin-jet electro-polishing with 5% perchloric acid.

The dimension of tensile specimens was \(\Phi 5 \text{ mm} \times 30 \text{ mm}\) and the tensile test was conducted according to Chinese standard GB/T 228.1–2 010. The dimension of Charpy impact specimens was \(10 \text{ mm} \times 10 \text{ mm} \times 55 \text{ mm}\) with a V-notch and the Charpy impact test was conducted according to Chinese standard GB/T 229–2 007. The tensile and Charpy impact tests of as quenched and tempered steels were conducted at room temperature. Five tests were performed for each test condition and the statistical tests of the data were analyzed using box-plot method. The outliers were removed and the average of the other data was noted.

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
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<th>S</th>
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<td>0.50</td>
<td>0.10</td>
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**Fig. 2.** Schematic illustration of the heat treatment procedures.
3. Results

3.1. Microstructure Observation

The typical morphologies of PAG in 30NiCrMoV12 steel quenched at different temperatures of 880–1 080°C are shown in Fig. 3. The relationship between the average size of PAG and the austenitizing temperature is shown in Fig. 4. The results show that the grain size increases with increasing temperature. It is observed that there is already obvious mixed structure of fine and coarse grains at 880°C. The grain size increases slowly when austenitizing temperature is below 1 040°C and coarsens significantly with increasing austenitizing temperature from 1 040°C to 1 080°C, as shown in Figs. 3 and 4.

The morphology of martensitic packets investigated by SEM is shown in Fig. 5. The crystallographic orientation and the effective grain size distribution of tempered lath martensite analyzed by EBSD are shown in Figs. 6 and 7. The size of martensite block is defined as the diameter of the equal-area circle with high angle grain boundaries (HAGB). The black lines represent HAGB above 15 deg. The boundaries of austenite, packets and blocks are high angle, while martensitic laths are separated by low angle grain boundaries. Block is the minimum unit with high angle grain boundary in martensite. Therefore, the effective

![Fig. 4. Effect of austenitizing temperature on PAG size.](image)

![Fig. 3. Austenite grain morphology quenched at different temperatures. (a) 880°C; (b) 920°C; (c) 960°C; (d) 1 000°C; (e) 1 040°C; (f) 1 080°C.](image)

![Fig. 5. SEM morphology of martensitic packets quenched at different temperatures. (a) 920°C; (b) 1 000°C; (c) 1 080°C.](image)

![Fig. 6. Crystallographic orientation of lath martensitic structure analyzed by EBSD of tempered samples. (a) 920°C; (b) 1 060°C; (c) 1 080°C.](image)
The grain size of martensite is the average size of blocks. The analysis results show that the effective grain size increases with increasing austenite grain size.

The PAG, martensitic packets and blocks are coarsening significantly with increasing austenitizing temperature, as shown in Fig. 8. Since there are many dislocations in quenched lath martensite, it is difficult to observe all laths boundaries clearly by TEM. In tempered steel, there are carbides precipitated at lath boundaries and the boundaries are easier to be observed. Besides, the mechanical properties are tested for as quenched and tempered steels. Hence, the laths width of tempered samples are observed under TEM. The laths width is measured to analyze the relationship between martensitic structure and properties. The TEM morphology of martensitic lath in tempered samples is shown in Fig. 9.

The density distribution and average width of martensitic laths are shown in Fig. 10. The lath width in tempered steels is from 100 nm to 600 nm, as shown in Fig. 10(a). The peak of the distributions curve moves to right with increasing...
ing austenitizing temperature, indicating that the lath width increases with increasing austenitizing temperature. The statistical result in Fig. 10(b) also shows that the martensitic lath width increases slightly with increasing austenitizing temperature.

Twin martensite which is the typical transformation microstructure from high carbon austenite is also observed in the samples quenched at 1,040°C and 1,080°C under TEM, as shown in Figs. 11(a) and 11(b). Since more carbon and alloying elements can dissolve in austenite at higher temperature, which decreases martensite transformation start temperature (Ms) and increases the amount of twin martensite in medium-carbon alloy steels.21,22) Meanwhile, coarsening of austenite grains at higher temperatures increases the size of twin martensite, which normally has micro cracks and is adverse to the impact toughness.

3.2. Carbide Precipitates during Tempering

SEM and TEM images demonstrating the morphologies of carbide precipitates in tempered steel are shown in Figs. 12 and 13, respectively. There are carbides at martensitic lath boundaries and in matrix. The strip-like carbides precipitated at martensitic lath boundaries are relatively larger than that in matrix. When the austenitizing temperature is 880°C, carbides mainly precipitate at PAG boundaries and martensitic laths boundaries, and seldom carbides precipitate in matrix. With increasing austenitizing temperature, the amount of large size carbides precipitated at boundaries decreases while the amount of small size spherical and round bar carbides precipitated in matrix increases, espe-

![Fig. 11. TEM morphology of twin martensite in thin foils of the quenched samples. (a) 1,040°C; (b) 1,080°C.](image1)

![Fig. 12. SEM morphology of carbides quenched at different temperatures and tempered at 600°C. (a) 880°C; (b) 920°C; (c) 960°C; (d) 1,000°C; (e) 1,040°C; (f) 1,080°C.](image2)

![Fig. 13. TEM morphology of carbides quenched at different temperatures and tempered at 600°C. (a) 920°C, (b) 1,000°C and (c) 1,080°C in carbon extraction replica; (d) 920°C, (e) 1,000°C and (f) 1,080°C in thin foil.](image3)
cially the round bar carbides.

TEM morphology, SAD pattern and EDS analysis results of carbide precipitates are shown in Fig. 14. The EDS signal of Cu element comes from copper mesh used for carbon extraction replica preparation. The results show that the round bar carbides are M7C3, whose diameter is about 50 nm and length is about 200 nm, as shown in Fig. 14(a). The shape of M23C6 is not only of one kind. M23C6 carbides precipitated at lath boundaries are strip-like and those precipitated in matrix are spherical, square and rectangular. MC carbide is normally smaller than 20 nm and precipitates in matrix and is not very easy to be analyzed using electron diffraction. Therefore, high-resolution transmission electron microscopy (HR-TEM) was employed, and the image taken is shown in Fig. 14(c). The interplanar spacing is 0.25 nm which is the (100) orientation index of MC carbide. EDS analysis result shows that MC mainly contains C, Mo and V elements.

The relationship between austenitizing temperature and the size of different shaped carbides is shown in Fig. 15(a). The change of carbides size is not obvious with increasing austenitizing temperatures from 880°C to 1 000°C while the size of all kinds of carbides decreases significantly when austenitizing temperature is beyond 1 040°C, which is in agreement with the variation of microstructure with austenitizing temperatures. Besides, the area fraction of strip like carbides decreases and that of total carbides and small size carbides increases obviously with increasing austenitizing temperature, as shown in Fig. 15(b). The statistics of carbides area fraction is calculated using TEM carbon extraction replica photographs. The carbide area fractions in Fig. 15(b) are higher than the actual area fraction, since

![Fig. 14. TEM morphology, SAD pattern and EDS analysis of carbides quenched at 920°C and tempered at 600°C. (a) and (d) are SAD pattern and EDS analysis of round bar M7C3 in (g); (b) and (e) are SAD pattern and EDS analysis of square M23C6 in (g); (c) and (f) are the HRTEM and EDS analysis of MC; (h) and (i) are SAD pattern and EDS analysis of strip-like M23C6 carbide in (g).](image)

![Fig. 15. Relationship between austenitizing temperatures and precipitates in tempered samples. (a) Precipitates size; (b) area fraction of precipitates.](image)
the samples are etched before carbon coating. Therefore, all the carbides in the volume of the etched thickness were extracted by the carbon thin film. However, the variation trend of carbides area fraction with austenitizing temperature is still worthy of reference, because the etching time is the same for all samples and the etched thickness is almost the same.

3.3. Mechanical Properties

The mechanical properties of the samples quenched at different temperatures and tempered at 600°C are shown in Fig. 16. With increasing austenitizing temperature, the tensile strength and yield strength increase. However, the impact toughness decreases with increasing austenitizing temperature, especially when the temperature is beyond 1000°C. The experimental results show that fine and uniform PAG, packet, block and lath are obtained at austenitizing temperature of 920°C, and the impact toughness is optimized. The fracture morphologies of QT-treated 30NiCrMoV12 steel are shown in Fig. 17. The photos on the upper right corner of Figs. 17(a)–17(c) show the low magnification fracture morphologies. Shear lip and fiber regions have high area percentage and radiation region is not obvious. There is no macro-crack on fracture. Figs. 17(d)–17(f) show the micro-fracture morphologies. They are mainly composed of deformed shallow dimples, micro-porous concentrated dimples, dimple bands and a large number of short and curved tear ridges. Besides, there is a small amount of river patterns, which indicates that there is plastic deformation before fracture. Transgranular fracture occurs along PAG boundaries in some regions. However, the effective cleavage planes in grain interior are not clear since the true cleavage planes have been replaced by unclear and smaller cleavage planes. These small cleavage planes are named quasi-cleavage plane, as showed in Fig. 17(f). The impact fractures of test steel have mixed-rupture characteristics of quasi-cleavage and dimples.

The fine carbides precipitated during tempering affect cracks propagation, preventing the cracks propagating strictly along a certain crystallographic plane in the grain. Therefore, the distribution of fine carbides plays a key role in the cracks propagation path. Coarse carbides at PAG boundaries and martensitic lath boundaries can cause stress concentration, which is easy for cracks propagation.

4. Discussions

4.1. Effect of Austenitizing Temperature on Martensitic Structure

In order to study the influence of precipitates on the grain growth and the precipitation behaviors during tempering, Thermo-Calc is used to calculate equilibrium precipitates of tested steel. In normalized samples, the content of carbides and other precipitates might be lower than that of the equilibrium calculation, since the normalization process is a non-equilibrium process. Therefore, a superheating is required for dissolution of precipitates. In spite of that, the precipitates type and approximate precipitation temperature calculated by Thermo-Calc could provide guidance for actual production.

The thermodynamic calculation temperature range is from 400°C to 1400°C, since the tempering temperature is above 400°C and there is no precipitate above 1400°C for 30NiCrMoV12 steel. Thermodynamic calculation results of equilibrium precipitation and composition of phases are shown in Fig. 18. The main elements of M7C3 and M23C6 are Fe, Cr, C and Mo, as shown in Figs. 18(b) and 18(c). FCC_A1#2 phase represents the MC type carbide with NaCl.
(B1) type structure which is composed of V, C, N, Fe, Cr and Mo elements, as shown in Fig. 18(d). The MCETA phase represents the MC type carbide whose elements composition are Mo, V and C, as shown in Fig. 18(e). The MCSHP phase also represents the MC type carbide whose elements composition are Mo and C, as shown in Fig. 18(f). The undissolved precipitates in austenite are MC (FCC_A1#2), AlN and MnS phases when austenitizing temperature is from 880°C to 1 080°C. MC starts to dissolve in austenite at about 800°C, and dissolve completely at about 920°C. The precipitates in quenched samples are observed by TEM, as shown in Fig. 19. There are MC carbides, mainly containing Fe, C, Mo and V, in sample quenched at 880°C, whose size is about 200 nm. AlN precipitate is observed in sample quenched at 1 000°C, whose size is about 300 nm. The EDS signal of Fe element mainly comes from matrix. It is difficult to find precipitates in sample quenched at 1 080°C. Therefore, the mixed structure with fine and coarse grains at 880°C has relationship with MC phase, since the grain boundary pinning effect is weak with dissolution of MC phase. AlN particles start to dissolve at about 1 000°C and completely dissolve in austenite at about 1 070°C according to Thermo-Calc calculation result. Therefore, the grains growth could be restrained effectively by AlN particles when austenitizing temperature is lower than 1 000°C. When austenitizing temperature is beyond 1 000°C, MnS precipitates could also pin grain boundaries and hinder grains growth, but the amount is not sufficient. Therefore, the grains grow fast and coarsening.

The statistical results of martensite laths in Fig. 10 show that the lath width increases slightly with increasing austenitizing temperature. The previous reports indicated that the lath width was determined by the nucleation rate and the martensite start temperature (Ms), and higher Ms may lead to coarser martensite laths. On one hand, increasing austenitizing temperature is beneficial to increasing alloy content.
in austenite and decreasing Ms. On the other hand, the Ms rises with increasing austenite grain size, which is proportional to the austenitizing temperature.\(^{28}\) When austenitizing temperature is high enough, the alloy content in austenite will be unchanging, but the austenite grain coarsens rapidly with increasing austenitizing temperature. Therefore, the Ms rises with increasing austenitizing temperature, meanwhile, martensite laths are coarsening.

4.2. Effect of Austenitizing Temperature on Carbide Precipitates during Tempering

According to the results of thermodynamic calculation, SAD pattern and EDS analysis, the main precipitates in as-quenched and tempered 30NiCrMoV12 steel are M\(_{7}C_{3}\), M\(_{23}C_{6}\) and MC. The main elements in M\(_{23}C_{6}\) and M\(_{7}C_{3}\) phases are Fe, Cr and Mo, and there is also Mn in M\(_{7}C_{3}\), as shown in Figs. 18(b) and 18(c). MC phase is composed of V, C, Cr, Mo and Fe elements. When temperature decreases, Cr and Fe atoms in MC phase are substituted by Mo. According to the classical nucleation theory, the nucleation rate of carbide precipitation can be expressed by formula (1):\(^{29,30}\)

\[
I = N_p \frac{kT}{h} \exp\left(\frac{-G^* + Q^*}{kT}\right)
\]

where \(I\) is the nucleation rate, \(N_p\) is the number of nucleation positions, \(k\) is the Boltzmann constant, \(h\) is the Planck constant, \(T\) is the Kelvin (K) degree, \(Q^*\) is the activation energy when diffusion atom through the interface between carbide and matrix, and \(G^*\) is the activation energy of carbide nucleation and can be illustrated by formula (2):

\[
G^* = \frac{16\pi\sigma^4V_m^2}{3\Delta G_m^2}
\]

where \(\sigma\) is the specific surface energy of carbide, \(V_m\) is the molar volume of carbide, and \(G_m\) is the precipitation driving force of carbide. It can be concluded that nucleation rate \((I)\) will increase with increasing the number of nucleation positions \((N_p)\) and temperature \((T)\), while it will decrease with increasing the activation energy of carbide nucleation \((G^*)\). Increasing precipitation driving force can reduce activation energy of carbide nucleation which will increase the nucleation rate of carbides. Therefore, a higher number of small size carbides can be obtained.

At lower austenitizing temperatures, some large carbides cannot be dissolved fully. As a result, the content of carbon and alloying elements in austenite is less and the precipitation driving force is at a lower level. Since there are more defects at boundaries than in matrix, the activation energy for precipitation nucleation at boundaries of grains and sub-grains is much lower than that in matrix. Therefore, it is much easier for carbides to precipitate at boundaries. With increasing austenitizing temperatures, structure coarsening causes less boundary area, which leads to fewer positions \((N_p)\) for carbides nucleation. Meanwhile, the precipitation driving force of carbides \((\Delta G_m)\) also increases with more alloying elements dissolving in austenite when the austenitizing temperature increases from 880°C to 1 000°C. As a result, the activation energy of carbide nucleation \((G^*)\) decreases according to formula (2). Hence, the nucleation rate \((I)\) increases according to formula (1). Therefore, the amount of small size carbides and area fraction of carbides increase with increasing alloying elements dissolving in austenite at higher austenitizing temperatures.\(^{31}\) When the austenitizing temperature is higher than 1 000°C, the carbides dissolve completely in austenite according to Thermo-Calc calculation and the alloying elements content will not increase with increasing austenitizing temperature. The refinement of carbides is thereby completely results from structure coarsening which causes less boundary area and promotes small carbides precipitation in matrix, as shown in Fig. 15.

4.3. Effects of Microstructure and Precipitates on Mechanical Properties

Impact toughness is deteriorated obviously with microstructure coarsening. Refinement of austenite grain, martensitic packet and block is the key factor to improve impact toughness, since the boundaries for PAG, martensitic packets and blocks are high angle. The amount of high angle grain boundaries increases with microstructure refinement. Cracks change its propagation direction when they meet high angle grain boundaries, which will absorb more energy. Therefore, high angle grain boundaries could hinder cracks propagation.\(^{25,30}\) Some researches indicated that refining martensite lath improved toughness.\(^{27,31}\) However, the total area of lath boundaries increases with structure refinement, and the precipitation positions for large size strip-like M\(_{23}C_{6}\) carbides increase, which are adverse to toughness improvement.\(^{18}\) As shown in Fig. 15(b), the amount of strip-like carbides, which are harmful for toughness, decreases with increasing austenitizing temperature and that of small size carbides in matrix increases. Therefore, the coarse martensite lath is beneficial to obtaining fine precipitates in matrix and improve strength and toughness. However, the favorable influence on impact toughness resulting from refinement of carbides is lower than the negative effect resulting from structure coarsening. Therefore, the impact toughness is deteriorated with increasing austenitizing temperature.

With increasing austenitizing temperature, the increment of alloying elements in austenite leads to lattice distortion in martensite and increases dislocation density. Therefore, the strength increases. Moreover, carbide precipitates also have an influence on strength and toughness according to the particle reinforcement theory.\(^{31}\) The precipitation strengthening effect of the steel material is inversely proportional to the average spacing of particles or the \(fr\) \((f\) is the particle volume fraction and \(r\) is the particle radius). Therefore, refining the carbide precipitates and decreasing the distance between carbides improve strength at higher austenitizing temperature.

5. Conclusions

The present work has investigated the effect of austenitizing temperature on martensitic microstructures, carbide precipitates and mechanical properties of 30NiCrMoV12 alloy steel. The results are as following:

1. Martensitic microstructure of 30NiCrMoV12 steel is coarsening with increasing austenitizing temperature. The
amount of large size strip-like M$_{23}$C$_6$ carbides precipitated at martensite lath boundaries decreases in tempered steel and the number of small size spherical and round bar carbides precipitated in matrix increases. Meanwhile, the area fraction of carbides increases with increasing austenitizing temperature. The phase transformation from M$_{23}$C$_6$ to M$_7$C$_3$ is enhanced and the amount of M$_7$C$_3$ carbides precipitated in matrix increases with increasing austenitizing temperature. It indicates that coarse martensitic structure is beneficial to obtaining more fine precipitates in matrix and reducing the amount of large size strip-like M$_{23}$C$_6$ carbides precipitated at boundaries.

(2) Increasing austenitizing temperature improves strength and impact toughness to a certain extent by refining carbides in tempered steel. However, the degree of favorable influence on impact toughness resulting from carbides refinement is lower than the negative effect resulting from structure coarsening. Therefore, the impact toughness is significantly deteriorated by coarsening of PAG and final martensitic microstructure at higher austenitizing temperature.

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