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Microstructure, Hardness, and Tensile Properties of Vacuum Carburizing Gear Steel

Wu Chen ^{1,2}, Xiaofei He ³, Wenchao Yu ³, Maoqiu Wang ^{3,*} and Kefu Yao ¹

¹ School of Materials Science and Engineering, Tsinghua University, Beijing 100084, China; chenwu15@mails.tsinghua.edu.cn (W.C.); kfyao@mail.tsinghua.edu.cn (K.Y.)

² NCS Testing Technology Co., Ltd., Beijing 100081, China

³ Institute of Special Steel, Central Iron and Steel Research Institute, Beijing 100081, China; hexiaofei@nccast.com (X.H.); yuwenchao@nccast.com (W.Y.)

* Correspondence: maoqiuwang@hotmail.com; Tel.: +86-10-62186791

Abstract: We investigated the effects of the austenitizing temperature on the microstructure, hardness, and tensile properties of case-carburized steel after vacuum carburization at 930 °C and then re-austenitization at 820–900 °C followed by oil quenching and tempering. The results show that fractures occurred early with the increase in the austenitizing temperature, although all the carburized specimens showed a similar case hardness of 800 HV_{0.2} and case depth of 1.2 mm. The highest fracture stress of 1919 MPa was obtained for the experimental steel when the austenitizing temperature was 840 °C due to its fine microstructure and relatively high percentage of retained austenite transformed into martensite during the tensile tests. We also found that the stress–strain behavior of case-carburized specimens could be described by the area-weighted curves of the carburized case and the core in combination. The strain hardening exponent was about 0.4 and did not vary with the increase in the austenitizing temperature. We concluded that the optimum austenitizing temperature was around 840 °C for the experimental steel.

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1. Introduction

Gears are typically carburized to have very high hardness and wear resistance in the case and excellent toughness in the core. The most widely used carburizing method is gas carburizing [1]. However, this has been gradually replaced by vacuum carburizing due to the advantages of the latter, such as less near-surface intergranular oxidation and less quenching distortion [2]. As vacuum carburizing might result in coarse grain sizes when a high carburizing temperature above 930 °C is used, re-austenitizing and quenching are required instead of direct quenching after carburizing. Re-austenitizing and quenching after vacuum carburizing at around 930 °C are also used in certain cases, in particular for heavy-duty gears. However, the effect of the re-austenitizing temperature on the microstructure and mechanical properties of vacuum-carburized specimens has not been fully clarified, and the optimum re-austenitizing temperature should be determined for particular situations.

Due to the complex microstructure in the carburized case [1], which is typically composed of martensite, retained austenite, and carbides, the mechanical properties of vacuum-carburized specimens can vary largely and be influenced by many factors [3–5]. The mechanical properties of carburized specimens are usually characterized in terms of the hardness and fatigue resistance. However, hardness tests are too simple and fatigue tests are too complicated for carburized specimens. Tensile tests of carburized specimens are a good compromise for considering both aspects. Thus, it is necessary to study the tensile behavior of carburized specimens.

Previous studies showed some contradictory results regarding the tensile properties of carburized steels [6–12]. Luther and Williams compared the tensile properties of carburized and un-carburized low carbon mild steel and found an increase in tensile strength by 45% after carburizing and slant fracture instead of cup and cone [6]. Krauss observed the microstructure and fracture of carburized 8620 steel and found the transition of the fracture mode from microvoid coalescence (MVC) around carbides to transgranular fracture [7]. Furukawa and coworkers found that re-quenching after carburizing could improve both the strength and ductility of carburized specimens [8]. Li et al. investigated the influence of heat treatment after carburizing on the strength of carburized 18Cr2Ni4WA steel and found that the tensile strength was about 1200–1300 MPa and the fracture mode was intergranular [9].

Ergoan and Tekeli investigated the tensile strength of carburized 8620 steel and found a much lower tensile strength of 1005 MPa with a fracture surface of the cleavage facet [10]. Murai attributed the intergranular fracture of carburized steel to carbide formation at the austenite grain boundaries [11]. Recently, Yu and coworkers investigated the influence of carbide formation on the tensile and fatigue properties of carburized steels and found that the fine carbides acted as resistance sites against crack propagation and improved the mechanical properties, while the networked carbides deteriorated the properties [12]. Other works focused on the effect of retained austenite in carburized steels and found that it could transform into martensite during stressing or after cryogenic treatment [13–18]. It is still controversial whether the tensile properties of carburized specimens can reflect the mechanical properties of gear steel since the carburized case has complex microstructural constituents.

The aim of this work is to study the tensile behavior of carburized specimens of gear steel after vacuum carburization and re-austenitization at different temperatures and to investigate the effect of the austenitizing temperature on the microstructure and tensile properties of vacuum-carburized gear steel. In the experimental part, the details including the experimental steel, processes, and methods will be described. In the results and discussion, experimental results including the microstructure, hardness, and tensile properties will be given and discussed. Based on the results, our conclusions will finally be drawn.

2. Experimental

Case-hardening steel 16CrMnH, which is specified in the Chinese National Standard GB/T 5216-2020, was used in this work. The experimental steel was industrially produced by the method of electric furnace melting, ladle refining, and vacuum degassing, followed by continuously casting and hot rolling. This was provided in the form of hot-rolled bars, 80 mm in diameter. The chemical compositions of the experimental steel are listed in Table 1 and were tested according to the serial Chinese National Standards numbered GB/T 223.

Table 1. Chemical composition of the experimental steel (wt.%).

C	Si	Mn	Cr	P	S	Fe
0.16	0.08	1.22	1.22	0.014	0.024	Bal.

Specimens with the size of 14 mm in diameter and 162 mm in length, as shown in Figure 1, were machined from the 1/2 radius position of the hot-rolled bars. Carburizing of the specimens was carried out in a WZST-45 type vacuum carburizing furnace, made by the Beijing Institute of Mechatronics (BIM, Beijing, China). The carburizing process and heat treatment process are shown in Figure 2. Carburizing was held first at 930 °C for 7 h at the expected equivalent carbon potential of 1.1%, and then at 930 °C for 1 h at the expected equivalent carbon potential of 0.85%, followed by air cooling to room temperature. The specimens were then re-heated to 820–900 °C in an air furnace (BIM, Beijing, China) for austenitizing for 1 h, followed by oil quenching and tempering at 180 °C for 2 h. After heat treatment, the specimens were finished with polishing before the tensile tests.

Figure 1. The shape and dimensions of the specimens for the tensile tests.

Figure 2. Schematic illustration of the vacuum carburizing and heat treatment processes of the experimental steel, where C_p shows the expected equivalent carbon potential during the carburizing process.

The microstructure in the carburized case and the matrix of the specimens, after cross-section machining, polishing, and etching in a 3% Nital solution, was observed on an optical microscope and a JSM-6400 type (JOEL, Tokyo, Japan) scanning electron microscope (SEM), operating at 15 kV. Retained austenite in the carburized case of the specimens was also measured by means of X-ray diffraction (XRD) on a BRUKER D8 ADVANCE X diffractor (BRUKER, Karlsruhe, Germany) at 35 kV with a 40 mA current using Co-K α radiation. The retained austenite amount was calculated according to the data collected containing (111), (200), (220), and (311) gamma and the (110), (200), and (211) alpha reflections.

The in-depth Vickers hardness distribution from the carburized case was obtained on a TUKON 2500-6 type (TUKON, Norwood, Massachusetts, USA) Vickers hardness tester with a load of 200 gf (1.96 N) for a holding time of 10 s. Each reported value of Vickers hardness was averaged one for five readings. The case-hardening depth was calculated from the results of the Vickers hardness distribution with the critical value of 550 HV_{0.2} according to the ISO 18203 standard. Using the specimens as shown in Figure 1, tensile tests at room temperature were performed on the machine of MTS 810 type (MTS, Minneapolis, Minnesota, USA) at a strain rate of 2.5×10^{-3} – 2.5×10^{-4} s^{−1} according to the Chinese National Standard GB/T 228.1-2010. For each austenitizing temperature, three specimens were tested and the results were averaged.

After the tensile tests, the fracture surfaces of the specimens were observed on a JSM-6400 type (JOEL, Tokyo, Japan) scanning electron microscope (SEM), operating at 15 kV.

3. Result and Discussion

3.1. Microstructure

Figure 3 shows the SEM micrographs in the carburized case of the experimental steel after carburization and austenitization at 820–900 °C, followed by oil quenching and tempering at 180 °C for 2 h. The carburized case consisted of plate martensite, carbides, and retained austenite, which are typical for high-carbon steels. When the austenitizing temperature was 820 °C, network carbides were distributed along with the prior austenite grain boundaries, as shown in Figure 3a. When the austenitizing temperature was 840 °C, network carbides were occasionally found along with prior austenite grain boundaries, as shown in Figure 3b. When the austenitizing temperature was 860 °C and above, blocky retained austenite could be distinguished, as shown in Figure 3c,d. Thus, the network carbides gradually dissolved and the fraction of retained austenite increased with the increase in the austenitizing temperature from 820 to 860 °C.

Figure 3. Scanning electron microscope (SEM) micrographs showing the carburized case of the specimens re-austenitized at (a) 820 °C, (b) 840 °C, (c) and (d) 860 °C, (e) 880 °C, and (f) 900 °C.

As listed in Table 2, the weight percentages of retained austenite were measured by XRD diffraction to be 18–30% for the specimens austenitized at different temperatures. At the same time, the microstructure did not coarsen evidently with the increase in the austenitizing temperature from 820 to 860 °C, with the average prior austenite grain sizes being about 5–7 µm. However, when the austenitizing temperature was higher than 860 °C, the microstructure did coarsen, with the average prior austenite grain sizes being about 13–17 µm, as shown in Figure 3e,f. Thus, the experimental steel showed a relatively fine prior austenite grain size, as the highest austenitizing temperature was no higher than 900 °C.

Table 2. Microstructural constituents in the near-surface carburized case of the steel.

Austenitizing Temperature, (°C)	Retained Austenite, V_γ (%)	Martensite Start Temperature, M_s (°C)	C Content in γ before Oil Quenching, (%)	Undissolved Carbides, (%)	α' , (%)
820	18	177	0.73	2.0	80.0
840	25	143	0.81	0.7	74.3
860	30	130	0.84	0.2	69.8
880	23	151	0.79	-	77.0
900	21	164	0.76	-	79.0

During carburizing, carbon can diffuse into the specimens from the surface, resulting in a carbon gradient distribution in the near-surface case. Since the carbon potential was 0.85% at the later stage of the carburizing process, we assumed that the average carbon content in the specimen surface was around 0.85% after carburizing and air cooling, although much of the carbon was in carbides. After austenitizing at 820, 840, and 860 °C, some of the carbides were dissolved into austenite, with the carbon content in the austenite depending on the austenitizing temperature. During oil quenching, some of the austenite transformed into martensite, while some was retained as austenite. Thus, the content of retained austenite in the specimen was dependent on the situation of austenite before oil quenching.

According to Magee's equation [19], the content of retained austenite (V_γ) in the as-quenched specimens can be calculated as follows:

$$V_\gamma = \exp [-\alpha(M_s - T_q)] \quad (1)$$

where M_s is the starting temperature of martensite transformation of the austenite before quenching, T_q is the temperature of the quenching, and α is a constant with the value of 1.10×10^{-2} . Since the contents of retained austenite in the specimens austenitized at 820, 840, 860, 880, and 900 °C were measured to be 18%, 25%, 30%, 23%, and 21%, the M_s temperatures were calculated according to Equation (2) to be 177, 143, 130, 151, and 164 °C, respectively. The carbon contents in austenite before oil quenching were calculated to be 0.73%, 0.81%, 0.84%, 0.79%, and 0.76%, respectively, according to the following equation [20]:

$$M_s(^{\circ}\text{C}) = 539 - 423\text{C} - 30.4\text{Mn} - 17.7\text{Ni} - 12.1\text{Cr} - 7.5\text{Mo}. \quad (2)$$

Considering the carbon balance between undissolved carbides and austenite at the austenitizing temperature, the contents of undissolved carbides and those of the quenched martensite can be also calculated. The calculated results are listed in Table 2, and the undissolved carbides decreased from 2% to 0.7% when the austenitizing temperature was increased from 820 to 840 °C, which is consistent with the micrographs in Figure 3. With the increase in the austenitizing temperature from 820 to 860 °C, the content of martensite in the near-surface carburized case was decreased from 80% to approximately 70%. According to theoretical calculations based on the iron-carbon phase diagram, the absolute dissolution of carbides could occur only at temperatures higher than 880 °C. This indicates that more austenite was retained at a higher austenitizing temperature. Because of the increased

number of carbides dissolved into austenite, the carbon content in the austenite is higher, and the M_s temperature is lower. It was easy to form retained austenite.

3.2. In-Depth Hardness Distribution

Figure 4 shows the in-depth distribution of the Vickers hardness of the experimental steel after carburizing, austenitizing at different temperatures, and tempering. There were three regions according to the hardness variations. In the region within the distance of 0.6 mm from the surface, the hardness was very high, with a value of around 750–800 HV_{0.2}, and it varied slightly within 50 HV_{0.2}. In the region with the depth distance from 0.6 to 1.5 mm (or 2.0 mm for the specimen austenitized at 820 °C), the hardness decreased linearly with the increasing distance. When the depth distance was more than 1.5 mm (or 2.0 mm), the hardness decreased slowly and finally tended to be constant, which was in the core matrix region. If a critical value of 550 HV_{0.2} was assumed, the carburized effective case depth was measured to be around 1.2 mm for all the specimens.

Figure 4. The in-depth hardness distribution of the specimens austenitized at different temperatures, as indicated by the serial number.

Comparing the in-depth hardness distributions of the specimens austenitized at different temperatures, we can see that the hardness in the carburized case (within 1.2 mm from the surface) varied very slightly with the austenitizing temperature. If we look carefully at the hardness variation in the near-surface region, we can find a slightly higher hardness of the specimen austenitized at 820 °C in comparison with the specimens austenitized at other temperatures. This can be attributed to the relatively higher content of martensite and lower content of retained austenite in the specimen austenitized at 820 °C, as listed in Table 2.

It can also be seen from Figure 5 that the core matrix hardness of the specimen austenitized at 820 °C was clearly lower than those of the specimens austenitized at other temperatures. As shown in Figure 5a, the microstructure in the core matrix of the specimen austenitized at 820 °C consisted of soft ferrite, pearlite, and martensite, which was different from those of martensite and bainite for the specimens austenitized at other temperatures, as shown in Figure 5b–e. The fully austenitizing temperature named the A_{c3} temperature of the experimental steel with the chemical compositions listed in Table 1 was calculated to be around 830 °C according to the following equation [20]:

$$A_{c3} (^{\circ}\text{C}) = 910 - 203C^{1/2} + 44.7\text{Si} - 15.2\text{Ni} + 31.5\text{Mo} + 104\text{V} + 13.1\text{W}. \quad (3)$$

Figure 5. Optical micrographs showing the microstructure in the core matrix of the specimens austenitized at (a) 820 °C, (b) 840 °C, (c) 860 °C, (d) 880 °C, and (e) 900 °C.

This indicates that the austenitizing was not fully complete at the temperature of 820 °C. Thus, the existence of ferrite in the core matrix of the specimen austenitized at 820 °C resulted in a much lower hardness in comparison with those of the specimens austenitized at other temperatures.

3.3. Tensile Properties

Figure 6 shows the tensile stress–strain curves of the experimental steel after carburizing, austenitizing at different temperatures, and tempering. All the carburized specimens show early fracture, although the fracture stress is relatively high. Since the carburized specimens had a gradient carbon content, the microstructure and hardness distribution along the depth and the tensile stress–strain curves were different from both high carbon steels with high brittleness and low carbon steels with good ductility. As shown in Figure 6, the engineering stress–strain curves and true stress–strain curves can be divided into two stages: the first was the elastic stage and the second was the plastic stage. In the elastic stage, the specimens extended uniformly and there was no evident difference among the specimens austenitized at different temperatures. In the plastic stage, the specimens

showed strain hardening, and all the specimens austenitized at different temperatures showed a similar hardening effect with a strain hardening exponent of approximately 0.4. The elastic-plastic behavior of the experimental steel did not vary with the austenitizing temperature.

Figure 6. Tensile stress–strain curves of the specimens austenitized at different temperatures as indicated by the serial number.

Table 3 shows the parameters obtained by the tensile tests on the carburized specimens with different austenitizing temperatures. The yield stresses and fracture stresses showed a decreasing trend with the increase in the austenitizing temperature, especially at the relatively low austenitizing temperatures. When the austenitizing temperature was 820 °C, the yield stress and the fracture stress were 1322 and 1713 MPa, respectively. When the austenitizing temperature was increased to 840 °C, the yield stress and the fracture stress were changed to 1249 and 1919 MPa, respectively. A further increase in the austenitizing temperature to over 860 °C resulted in the yield stress and the fracture stress at about 1230 and 1500 MPa, respectively. The highest fracture stress of the experimental steel was found to be 1919 MPa when austenitized at 840 °C. The tensile elongation of the carburized specimens was lower than 3.0%, indicating an early brittle fracture. Clearly, this was because the carburized case is high-carbon martensite with high brittleness. However, clear yield phenomena could be found for all the carburized specimens, as shown in Figure 6.

Table 3. The tensile properties of the carburized specimens of the experimental steel.

Austenitizing Temperature (°C)	Yield Stress (MPa)	Fracture Stress (MPa)	Tensile Elongation (%)
820	1322	1713	1.7
840	1249	1919	3.0
860	1227	1552	1.4
880	1231	1519	1.4
900	1244	1476	1.2

Pavlina and Tyne studied the correlation of tensile strength with hardness for steels [21]. According to their results, the average fracture stresses in the near-surface carburized case and in the transitional region of the experimental steel were calculated to be 2812 and 2170 MPa, respectively, which were independent of the austenitizing temperature. Those in the core matrix were calculated to be 1095 and 1524 MPa for the specimens austenitized at 820 °C and other temperatures, respectively. Since carburized specimens were multi-layered, the tensile fracture stresses of carburized specimens would depend on the tensile

fracture stresses of each layer. Assuming that the tensile fracture stresses of carburized specimens could be estimated through multi-layer stresses weighted by cross sectional area, the calculated results were 1830 and 1932 MPa for the specimens austenitized at 820 °C and other temperatures, respectively. Compared with the results listed in Table 3, the calculated tensile fracture stresses were not in agreement with the experimental results, indicative of a large error.

Figure 7 shows the calculated true stress–true strain curves of the carburized case and the core and the overall specimen, which was obtained in combination with the carburized case, the transition area, and the core. It is assumed that there exists an Ramberg–Osgood relation between the stress and the strain for the carburized case, the transition area, and the core [22]. The calculated overall curve is in good agreement with the experimental one, especially at the low strain stage. Thus, the tensile stress–strain behavior of the carburized specimens depended on both the carburized case and the core. For the experimental steel, the tensile stress–strain behavior did not vary with the austenitizing temperature, due to the similar in-depth hardness distribution.

Figure 7. Comparison of the experimental and calculated true stress–true strain curves of the case, the core, and the overall specimens austenitized at different temperatures.

3.4. Fractography

The fracture stresses were dependent on the austenitizing temperature, which could be attributed to the occurrence of fractures. It can be seen from Figure 8 that fracture occurred at a relatively early stage with the increase in the austenitizing temperature. The fracture surfaces of the tensile specimens were observed using SEM with the emphasis on the crack initiation sites. As shown in Figure 8, the crack initiation sites were located in the near-surface region, being about 0.1–0.3 mm from the surface. Higher magnification micrographs show that the crack initiation sites had a primarily intergranular fracture appearance.

This can be attributed to the high carbon content in the carburized case, which might cause brittleness [9,11,12]. Transgranular facets were also observed at the crack initiation sites, which could be related to retained austenite in the carburized case [7,10,13]. Compared with the specimens austenitized at 820 °C, the specimens austenitized at other temperatures appeared to be more brittle, with secondary cracks observed, as shown in Figure 8c. It was not clear whether the secondary cracks were formed during quenching or tensile loading. In this specimen, the crack initiation site was just below the surface. Thus, an early failure occurred for the specimens austenitized at over 860 °C, resulting in the lower fracture stresses. The fine microstructure and relatively low content of the retained austenite of 18% in the specimen austenitized at 820 °C could be the reason for its high fracture stress.

Figure 8. Cont.

Figure 8. SEM fractographs showing (a–e) low-magnification overviews, and (f–j) crack initiation sites of fracture surfaces of the specimens austenitized at different temperatures of (a,f) 820 °C, (b,g) 840 °C, (c,h) 860 °C, (d,i) 880 °C, and (e,j) 900 °C.

The features of intergranular fracture at the crack initiation site are listed in Table 4. The crack initiation site moved toward the case surface and the intergranular area increased with the increase in the austenitizing temperature. This indicates that the coarser the austenite grain size, the more brittle the grain strength during tensile tests, resulting in early premature fractures. As also listed in Table 4, the estimated carbon content at the crack initiation site was around 0.8%, indicative of high carbon plate martensite with brittleness.

Table 4. Feature of the crack initiation site for the carburized specimens.

Austenitizing Temperature (°C)	Distance of Crack Initiation Site from Surface (mm)	Area of Intergranular Fracture (mm ²)	Carbon Content at Crack Initiation Site (%)
820	0.32	0.034	0.90
840	0.29	0.062	0.82
860	0.24	0.067	0.82
880	0.17	0.121	0.82
900	0.15	0.129	0.80

The fracture stress decreased with the increase in the crack length, in line with Griffith theory [23] as follows:

$$\sigma_c = \sqrt{\frac{2E\gamma}{\pi a}} \quad (4)$$

where σ_c is the fracture stress, E is the elasticity modulus, γ is the surface energy, and a is the crack length. Since an intergranular fracture mode was found at the crack initiation sites, the crack length could be related to the prior austenite grain size. The lower the austenitizing temperature, the finer the prior austenite grain size. Thus, a decreasing trend of fracture stress with the increase in austenitizing temperature can be expected.

3.5. Transformation of Retained Austenite

There was a high content of retained austenite in the carburized case, which can transform into martensite during tensile stress. A study reported that the fraction of retained austenite transformed into martensite depended on the extent of the plastic strain, which can be described as follows [24]:

$$f_\gamma = f_0 \exp(-k\varepsilon) \quad (5)$$

where f_0 and f_γ are the contents of the retained austenite at the initial time or time, respectively, ε is the deformation strain, and k is a constant indicating the mechanical stability of the retained austenite where the value of 34 is assumed. The calculated results of

the retained austenite after fracture of the specimens austenitized at different temperatures are listed in Table 5. Increasing the austenitizing temperature might result in a relatively high content of retained austenite after the fracture, indicative of an early fracture. The highest fracture stress of the specimen austenitized at 840 °C could be related to its high transformation percentage of 16% in the retained austenite, as shown in Table 5.

Table 5. Retained austenite fraction of the carburized specimens of the experimental steel during tensile test.

Austenitizing Temperature (°C)	Initial Retained Austenite Fraction f_0 (%)	Retained Austenite Fraction after Fracture f_γ (%)
820	18	10
840	25	9
860	30	19
880	23	14
900	21	14

The transformation of the retained austenite in the carburized case into martensite after the tensile tests was proven by the measured results of the retained austenite. As an example, Figure 9 shows the XRD patterns for the carburized case of the specimens austenitized at 900 °C before and after the tensile tests. The heights of the austenite peaks relative to the heights of the martensite peaks decreased after the tensile test, indicating the transformation of the retained austenite into martensite of about 7% from 21% to 14%.

Figure 9. The X-ray diffraction (XRD) patterns for the carburized case of the specimens austenitized at 900 °C before and after the tensile tests.

Figure 10 shows the Vickers hardness distribution for the carburized case of the specimens austenitized at 900 °C before and after the tensile tests. The transformation of the retained austenite might result in an increase in hardness, which could contribute to the high fracture stress of the experimental steel compared with the references [6–12].

Figure 10. The Vickers hardness distribution for the carburized case of the specimens austenitized at 900 °C before and after the tensile tests.

4. Conclusions

We investigated the microstructure, hardness, and tensile properties of case-hardening gear steel after vacuum carburization and re-austenitization at 820–900 °C, followed by oil quenching and tempering. The following conclusions can be drawn.

- (1) The content of retained austenite in the carburized case of the specimens of the experimental steel was measured to be 18–30%, which was dependent on the austenitizing temperature. This is related to the carburized carbon content and the carbide dissolution at the austenitizing temperature.
- (2) The tensile stress–strain curves of carburized specimens of the experimental steel showed a similar work hardening exponent of 0.4 and did not vary with the austenitizing temperature. The stress–strain curves were described by considering the carburized case and the core in combination.
- (3) Although the case hardness of 800 HV_{0.2} and the carburized depth of 1.2 mm were similar, the fracture stresses of the carburized specimens showed a decreasing trend with the increasing austenitizing temperatures. Observations of fracture surfaces indicated that cracks initiated at the near-surface carburized case and propagated first in an intergranular mode. The high fracture stress of the 840 °C austenitized specimen was related to the fine microstructure and high percentage of retained austenite transformed into martensite during the tensile tests.
- (4) The optimum austenitizing temperature for the experimental steel was determined to be around 840 °C when the fracture stress could be expected to be as high as 1900 MPa.

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Abbreviations

MVC	Microvoid coalescence
SEM	Scanning electron microscope
XRD	X-Ray diffraction
C_p	Carbon potential during carburizing
V_γ	Volume fraction of retained austenite
M_s	Start temperature of martensitic transformation
T_q	Quenching temperature
α	A constant of martensitic phase transformation with the value of 1.10×10^{-2}
α'	Martensite
A_{C3}	Transformation temperature of ferrite into austenite for hypoeutectoid steel
σ_C	Fracture stress
E	Elasticity modulus
γ	Surface energy
a	Crack length
f_0	Content of retained austenite at the initial time
f_γ	Content of retained austenite at a certain time
ε	Deformation strain
k	A constant indicating mechanical stability of retained austenite
C	Carbon
Mn	Manganese
Si	Silicon
P	Phosphorous
S	Sulfur
Fe	Iron
Cr	Chromium
Ni	Nickel
Mo	Molybdenum
V	Vanadium
W	Tungsten

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