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Microstructure Features and Mechanical Properties of Modified Low-Activation Austenitic Steel in the Temperature Range of 20 to 750 °C

Igor Litovchenko ^{1,2,*} , Sergey Akkuzin ¹ , Nadezhda Polekhina ¹ , Kseniya Spiridonova ¹,
Valeria Osipova ^{1,2} , Anna Kim ^{1,2} , Evgeny Moskvichev ¹ , Vyacheslav Chernov ³  and Alexey Kuznetsov ⁴

¹ Institute of Strength Physics and Materials Science of the Siberian Branch of the Russian Academy of Sciences (ISPMS SB RAS), 634055 Tomsk, Russia; s.akkuzin@ispms.ru (S.A.); nap@ispms.ru (N.P.); almaevakv@ispms.ru (K.S.); lera.linnik.1999@mail.ru (V.O.); kim.a.v@stud.tsu.ru (A.K.); em_tsu@mail.ru (E.M.)

² Faculty of Physics, National Research Tomsk State University, 634050 Tomsk, Russia

³ Joint-Stock Company "A. A. Bochvar High-Technology Research Institute of Inorganic Materials", 123060 Moscow, Russia; vmchernov@bochvar.ru

⁴ Institute of Engineering Science, Ural Branch of the Russian Academy of Sciences, 620049 Ekaterinburg, Russia; alekseikuz031082@rambler.ru

* Correspondence: litovchenko@ispms.ru; Tel.: +7-913-824-16-94

Abstract: A new low-activation austenitic steel with a modified composition and high austenite stability is proposed. The features of its microstructure after solution treatment (ST) and cold rolling (CR) are studied. The mechanical properties and features of the fracture behavior of this steel under tensile tests in the temperature range of 20–750 °C are discussed. After ST, an austenitic structure with stacking faults and dispersed carbide particles of the MC and M₂₃C₆ types is observed in the steel. After CR, the grains are refined, and the average grain size decreases from 41.4 μm (after ST) to 33.9 μm. High-density microtwin packets form in the material, and the dislocation density increases relative to that after ST. As the test temperature increases from 20 to 750 °C, the yield strength of the steel decreases by approximately two times, from ≈300 to 150 MPa (for ST) and from ≈700 to 370 MPa (for CR). In the studied temperature range, the steel demonstrates up to 2.6 times higher values of elongation to failure, ≈40–80% (for ST) and ≈13–27% (for CR), compared to steels of similar compositions and lower manganese content. Mechanical twinning contributes to the high steel ductility up to 300 °C. Signs of discontinuous flow in the tensile curves after ST in the temperature range of 500–600 °C and a decrease in the elongation to failure in the close temperature range indicate dynamic strain aging (DSA). Steel fracture after tension at all test temperatures mainly occurs via a ductile dimple transcrystalline mechanism with elements of ductile intercrystalline fracture. It is shown that cracks nucleate on clusters of dispersed second-phase particles. The mechanisms of plastic deformation, fracture, and strengthening of the proposed modified low-activation austenitic steel are discussed.

Keywords: low-activation chromium-manganese austenitic steel; electron microscopy; microstructure; mechanical twins; disperse particles; temperature dependence of mechanical properties; cold rolling; fracture



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1. Introduction

Structural materials in nuclear power plants, operating under hard conditions of radiation and high temperatures, are subjected to higher requirements for heat resistance, resistance to swelling, and embrittlement under radiation exposure, as well as corrosion resistance when interacting with aggressive coolants. Stainless chromium-nickel austenitic steels are widely used for these applications [1–5]. A critical problem with these steels at high radiation doses (up to 100 dpa) is their tendency to radiation swelling (more than

6%) [2,4,5]. At the same time, steels with lower austenite stability (e.g., type 321 steel) have a stronger tendency to swelling compared to more stable steels. Under radiation conditions, due to significant depletion of the matrix in austenite-forming elements, a $\gamma \rightarrow \alpha$ phase transformation with γ - α interphase boundaries may become preferred places for pore formation [6]. To increase austenite stability, determined by the nickel equivalent ($Ni_{eqv} = C_{Ni} + 0.5 \cdot C_{Mn} + 30 \cdot (C_C + C_N)$), and resistance to radiation swelling, structural austenitic steels of the latest generation need to have a higher nickel content (18–25 wt. %) compared to type 321 and 316 steels.

Low-activation chromium-manganese austenitic steels were earlier proposed as substitutes for high-activation chromium-nickel austenitic steels [6,7]. In these steels, the content of high-activation elements (Ni, Mo, Co, Nb, N) is minimized, and austenite is stabilized exclusively by manganese and carbon. It should be noted that manganese is a weaker stabilizer of austenite than nickel. To form a stable austenitic structure, the content of manganese and carbon should be increased compared to chromium–nickel austenitic steels. Compositions based on Fe–12Cr–(20–25)Mn–0.25C [7–10], with minor alloying additions of V, W, Ti, B, P, as well as various variants of steels with similar compositions [6,11–14], were developed. However, the austenite stability of these steels, determined by the nickel equivalent, is (17–20), which is lower than that of the latest generation of chromium–nickel steels (19–26). The formation of chromium–manganese steels with an addition of Ni and N for austenite stabilization, and with Mn and C, was also reported [15–18]. The induced radioactivity of such steels decreases faster than in high-activation steels, which allows them to be referred to as reduced-activation materials, though they are not classified as low-activation materials. Low-activation chromium–manganese austenitic steels can be used as cladding for fuel elements and other parts of the first wall in fast neutron nuclear reactors.

In our work we have developed and investigated a low-activation austenitic steel with an increased manganese content of up to 29 wt. %, and a carbon content of 0.25 wt. %, additionally alloyed with strong carbide-forming elements [19,20]. It has a nickel equivalent of 22.5, calculated from the actual composition. The steel has a stable austenitic structure with dispersed carbide particles after ST and CR [19]. However, in addition to improving austenite stability, an increased Mn content in the chromium–manganese austenitic steels increases their tendency to form the σ -phase during high-temperature annealing. It is known [21] that steels with high manganese content and low (lower than 0.1 wt. %) carbon content tend to form σ , α , and other phases under long-term (1000–30,000 h) high-temperature annealing. It should be noted that not only the content of manganese and carbon, as γ -stabilizing elements, but also the content of chromium and other α -stabilizing elements have a combined effect on the tendency toward phase transformations. The presence of a significant amount of different phases in any austenitic steel is highly undesirable, as it has a negative effect on its mechanical properties. For example, a significant amount of σ -phase can cause material embrittlement [22–24].

The purpose of this work is to develop and manufacture a low-activation austenitic steel with a modified composition and study the features of its microstructure and mechanical properties in the temperature range of 20–750 °C. The modification of the composition aims to achieve high values of the nickel equivalent while simultaneously reducing the tendency to form undesirable phases (mainly the σ -phase) during long-term exposure to high temperatures.

2. Materials and Methods

The basic experimental material was the Fe-29Mn-12Cr-W-Si-Ta-Ti-V-Zr-0.25C steel reported earlier [19]. In order to optimize the elemental composition, we calculated different variants of compositions with high manganese content using JMatPro[®] software [25] (version 7.0, Sente Software Ltd., Guildford, UK). Based on the results of these calculations, it was proposed to reduce the Cr content to 10.5–11.5 wt. % and Mn content to 26–28 wt. %, while increasing the C content to 0.25–0.35 wt. % in order to limit the tendency to form the σ -phase and expand the austenitic region, while maintaining high values of the nickel

equivalent. Low activation of the modified steel should be provided by minimum content (in wt. %): Ni, Cu, N < 0.1; Nb, Mo Co, Al < 0.01. To ensure the necessary level of corrosion resistance, the chromium content should be higher than 10.5 wt. %.

A steel ingot with the composition Fe–26Mn–11Cr–W–Si–Ta–Ti–V–Zr–0.32C, wt. % was melted in a vacuum induction furnace. The elemental composition presented in Table 1. It was studied using a Shimadzu LAB CENTER XRF-1800 X-ray fluorescence spectrometer (Shimadzu, Kyoto, Japan) and an ISKROLINE 300K optical emission spectrometer (ISKROLINE industrial Group, St. Petersburg, Russia).

Table 1. Elemental composition of modified high-manganese austenitic steel (wt. %, base Fe).

Cr	Mn	Ni	Cu	W	V	Ti	Ta	Zr	Nb	Mo	Co	Si	S	P	C	B
11.2	26.1	<0.1	<0.1	1.2	0.3	0.11	0.08	0.07	<0.01	<0.01	<0.01	0.5	<0.01	<0.01	0.32	0.004

The nickel equivalent of the resulting composition is $Ni_{eqv} = C_{Ni} + 0.5 \cdot C_{Mn} + 30 \cdot (C_C + C_N) = 22.65$. Carbon is a strong austenite stabilizer (included in Ni_{eqv} with a coefficient of 30), and manganese is a weaker austenite stabilizer (included in Ni_{eqv} with a coefficient of 0.5). The chromium equivalent is $Cr_{eqv} = C_{Cr} + C_{Mo} + 1.5 \cdot C_{Si} + 0.5 \cdot C_{Ti} = 12$. Such a ratio of nickel and chromium equivalents determine the steel within the limits of single-phase austenitic region on the Schaeffler diagram. Considering the modified Schaeffler diagram for high manganese steels proposed by [9], the composition is also within the single-phase austenitic regions, far from the boundaries of the martensitic and ferritic regions.

After melting, the ingot was homogenized at 1100 °C for 2 h, followed by hot forging and hot rolling at 1100 °C. Then, it was solution treated (ST) at 1100 °C for 1 h, followed by air cooling. Since the reactor austenitic steels are mainly used after cold rolling (CR), this treatment was performed on a 2-roll mill at $T = 20$ °C to a strain degree of $\epsilon = 20\%$. The microstructure and phase compositions of the steel were studied after its ST and CR treatments.

The phase content of the steel was determined by X-ray diffraction analysis (XRD) on a DRON-8N X-ray diffractometer (Bourestnik, Russia), equipped with a Mythen 2R1D linear position-sensitive detector, in $CuK\alpha$ radiation ($\lambda = 0.15406$ nm). The voltage applied to the X-ray tube was 40 kV and the current was 20 mA. Scanning was performed in a Bragg–Brentano symmetrical geometry (2θ) in the angular range of 2θ 20–100°. The exposure time at each point was 10 s. Scanning was performed with a rotation of the sample. X-ray patterns were processed by the full profile analysis (Rietveld method). After CR, studies were carried out in the rolling plane.

The equipment and research methods used in this work include the following: Scanning Electron Microscopy (SEM) with energy dispersive X-ray (EDX) microanalysis, Electron Back-Scatter Diffraction (SEM EBSD), and Transmission Electron Microscopy (TEM), including Scanning Transmission Electron Microscopy (STEM) mode with EDX microanalysis, as described in detail elsewhere [19]. Here, we report some differences. EBSD data were obtained with step sizes of 0.75 μm for large and 0.1 μm for small selected areas. Boundaries with misorientation angles of $2^\circ < \theta \leq 15^\circ$ and $\theta > 15^\circ$ were taken as low-angle (LABs) and high-angle boundaries (HABs), respectively. The average grain size was taken as the equivalent circle diameter of a region with the same orientation and HABs.

Mechanical tension tests were carried at a strain rate of $\approx 2 \times 10^{-3} s^{-1}$ in the temperature range from 20 °C to 750 °C, using cylindrical dog-bone samples with a gage length of 25 mm and a gage section of 5 mm² on a UTS 111.2–100-52 universal testing machine (Test Systems, Ivanovo, Russia). At each temperature point, 3–5 samples were used in tensile tests. Samples were cut from the section parallel to the rolling plane and subjected to tension in the rolling direction. Fracture surfaces of the samples after tensile tests were studied in an Apreo 2 S SEM (Thermo Fisher Scientific, Waltham, MA, USA).

3. Results

3.1. Microstructure and Phase Composition of the Modified Steel

Figure 1 shows the XRD results of the phase composition of the steel after ST and CR. From the analysis of these data, it is clear that in both cases, only austenite is observed in the steel and no traces of other phases are detected. The absence of a (200) reflection of austenite after ST may indicate the preservation of the texture formed during the hot rolling process preceding the solution treatment.

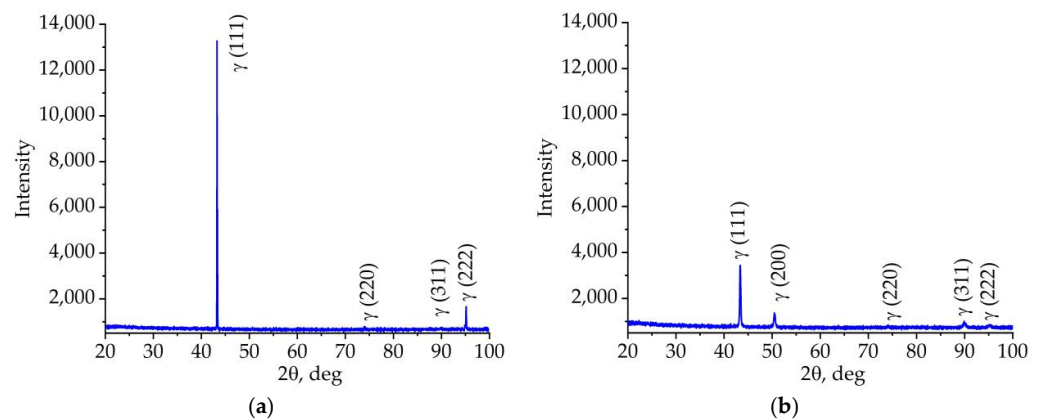


Figure 1. XRD patterns of the steel after ST (a) and CR (b).

SEM EBSD studies after ST confirm the single-phase austenitic state. In Figure 2a, a grain structure with annealed twins typical for quenched austenitic steels is shown. From the orientation map and grain size distribution, it is clear that sizes range from 6 to 271.3 μm, with an average grain size of 41.4 μm. Most grains are smaller than 50 μm in size, but there are individual large grains measuring hundreds of μm in size. The presence of a significant number of small grains (smaller than 10 μm) on the grain size distribution is apparently due to thin plates of twins. The presence of a large number of twins (special boundaries of the Σ3 type) and special boundaries of the Σ11 type associated with secondary twinning is confirmed by pronounced peaks near 60° and 39° on the grain boundary misorientation distribution. It should be noted that there are practically no low-angle boundaries (dislocation substructures) and carbide particles on the orientation map at a scanning step of 0.75 μm. A detailed analysis of carbide particles is presented by SEM EBSD after CR at a scanning step of 0.1 μm, and by TEM after ST and CR.

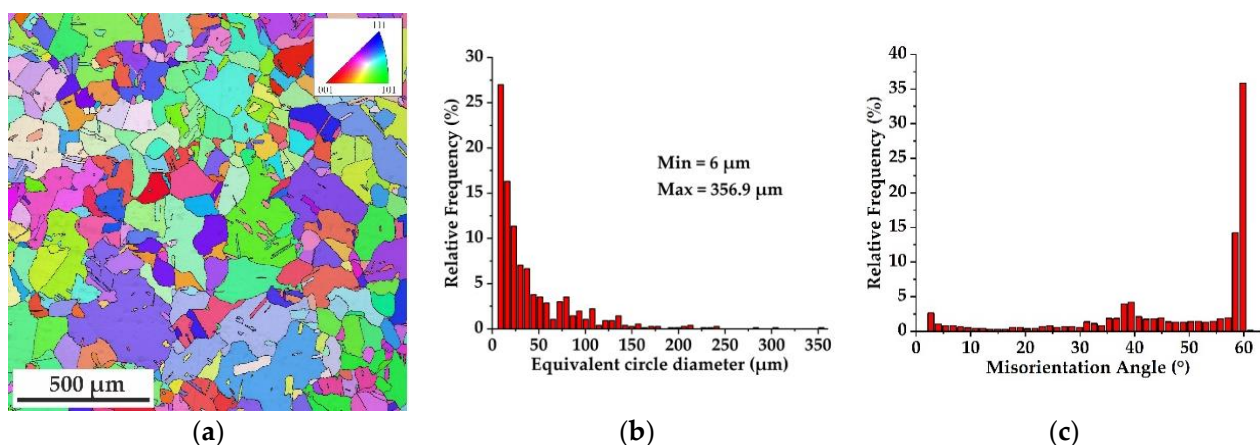


Figure 2. SEM EBSD orientation map after ST with HABs and LABs, denoted by black and white lines, respectively (a), histograms of grain size (equivalent circle diameter) distribution (b), and grain misorientation distribution (c).

Figure 3 shows SEM EBSD images of the microstructure after CR. From the analysis of these data, it follows that under the conditions of cold rolling, the initial austenitic grains are slightly elongated in the rolling direction, and a large number of low-angle misorientation boundaries are formed. The observed grain sizes vary from 5.6 to 356.9 μm . The average grain size decreases to 33.9 μm as a result of rolling, which is not significantly different from the quenched state. Similar to the ST case, grains smaller than 50 μm in size prevail. Kernel Average Misorientations (KAM) maps show the regions with significantly increased average misorientations, apparently related to the high density of dislocation substructures. The presence of such regions is also confirmed by an increase in the fraction of low-angle boundaries in the distribution of grain boundaries by misorientation angles. In this case, the distribution itself is smoothed, and the peak associated with twin boundaries is significantly reduced but does not disappear completely. This indicates the persistence of some initial annealing twins after rolling.

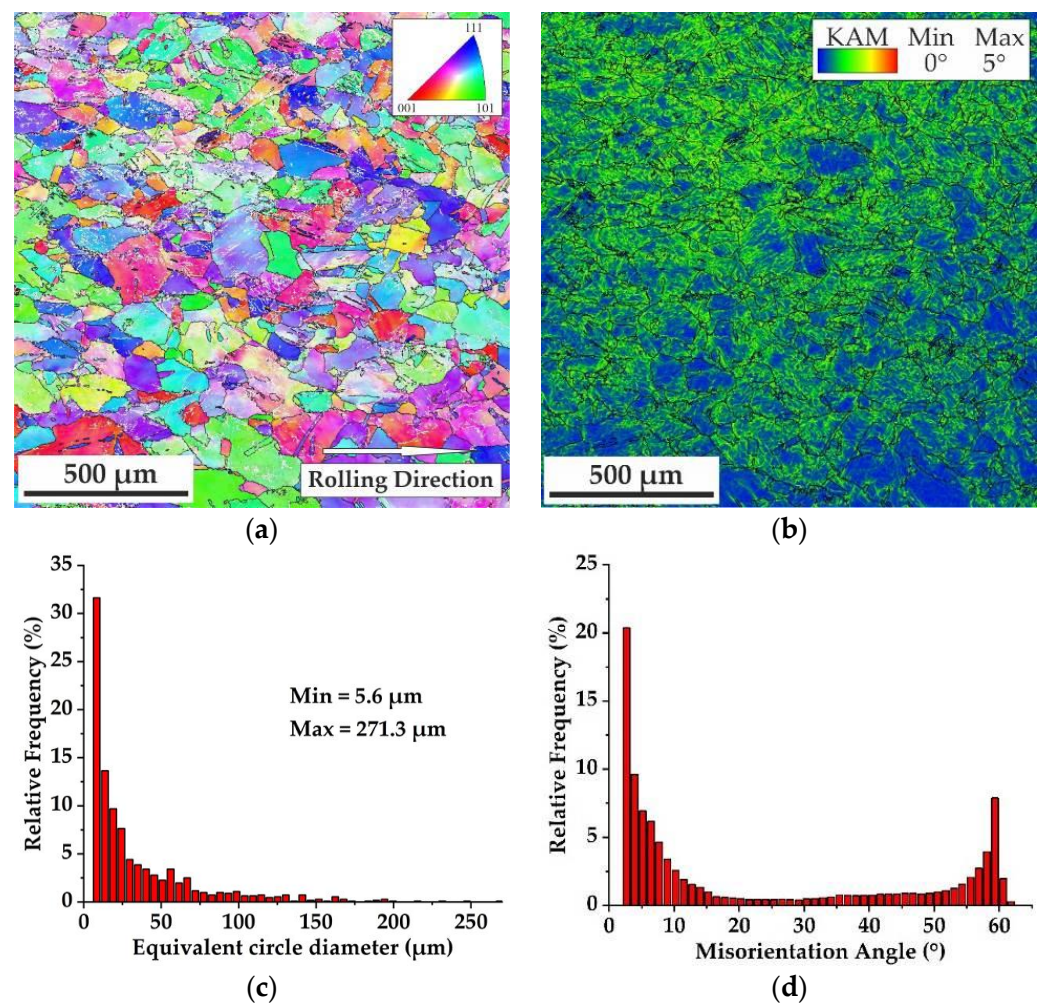


Figure 3. SEM EBSD orientation map after CR with HABs and LABs, denoted by black and white lines, respectively (a), KAM map (b), histograms of grain size (equivalent circle diameter) distribution (c), and grain misorientation distribution (d).

Figure 4 shows magnified images of the microstructure after CR, obtained by SEM EBSD. Deformation microtwins are observed to propagate in the structure containing dispersed carbide particles. The orientation map shows low-angle misorientation boundaries around the particles and near the microtwin boundary, while the KAM maps show an increase in the number of average misorientations associated with the formation of dislocation substructures under plastic deformation. Elemental microanalysis shows the presence

of two types of carbide particles: those enriched in Cr and Ti. Based on previous studies of a steel with a similar composition [19], we believe that these particles are carbides of the $M_{23}C_6$ type ($M = Fe, Cr, Mn$) and MC type ($M = V, Ti, Ta, Zr$). Their size can reach several microns and their volume fraction can reach several percent.

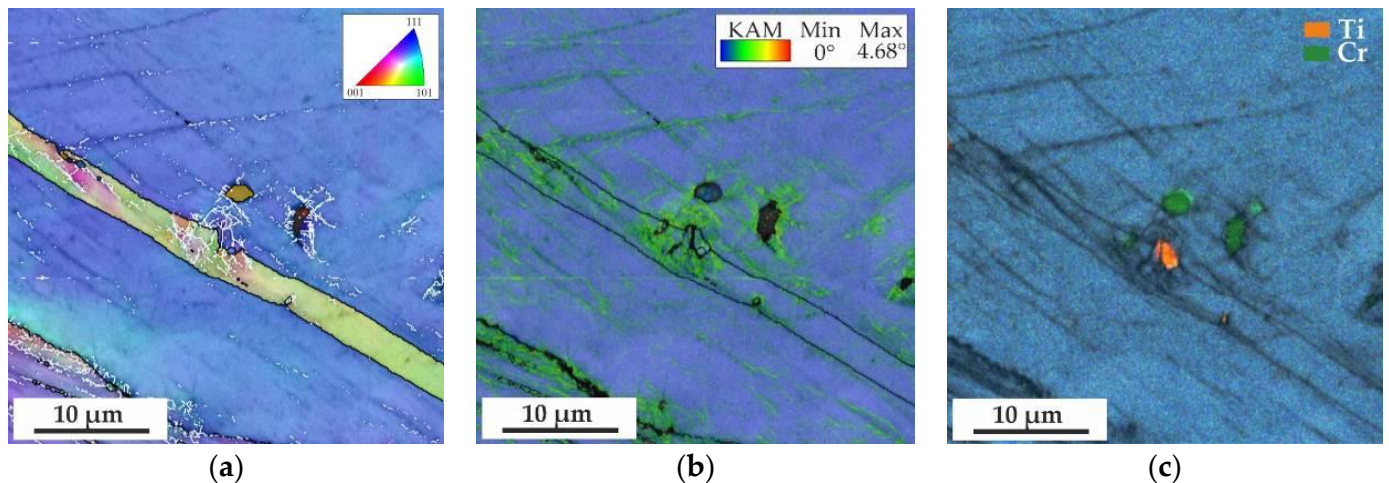


Figure 4. Magnified SEM EBSD. EBSD orientation map with HABs and LABs, denoted by black and white lines, respectively (a). KAM map (b), phase map with the matrix (austenite) in blue color, carbides enriched by Ti in orange color, and Cr in green color (c).

The carbide particles were investigated in more detail using TEM. From the analysis of Figure 5, it follows that the particle with characteristic dimensions of several microns is an $M_{23}C_6$ carbide, confirmed by the diffraction pattern (Figure 5b). A local stressed state is observed around the particle in the matrix, which is confirmed by the presence of multiple extinction contours in the bright-field image (Figure 5a) and azimuthal extension of matrix reflections in the diffraction pattern (Figure 5b). According to the elemental analysis results, this particle is mainly enriched in Cr; in addition, it has an increased content of Mn, V, Ta, W, and Si compared to the matrix.

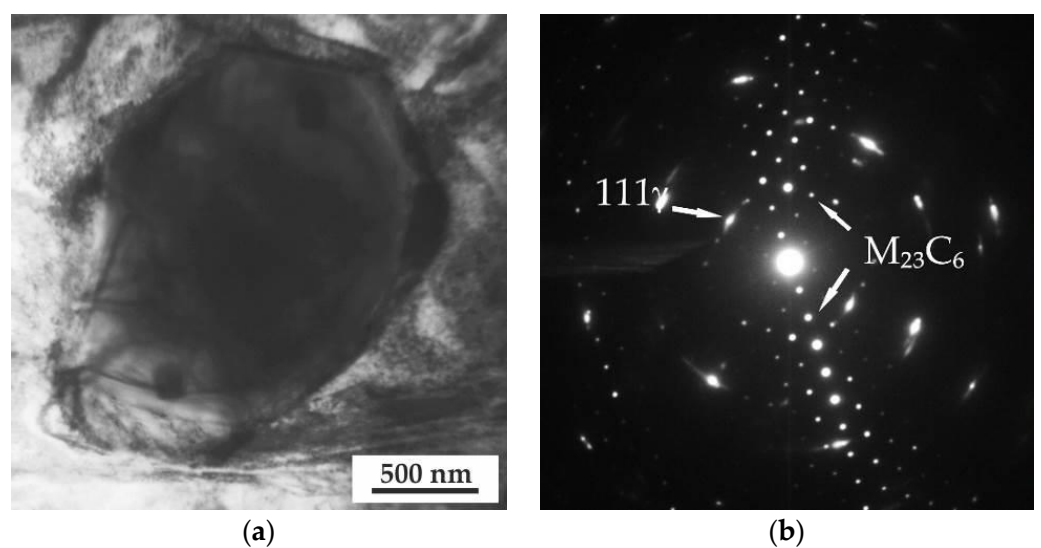


Figure 5. Cont.

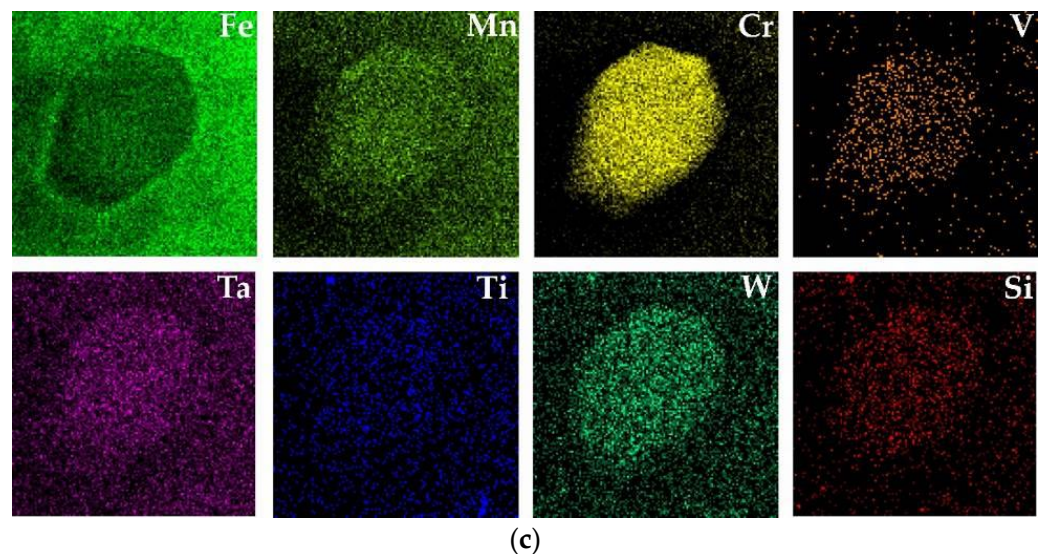


Figure 5. TEM Bright-Field (BF) image (a), corresponding Selected Area Electron Diffraction (SAED) pattern of dispersed particles of $M_{23}C_6$ type (b), EDX elemental maps from particles (c) after CR.

Other carbides characteristic of the studied steel both after ST and CR are the MC particles ($M = Ti, V, Ta, Zr$) with an FCC lattice. Figure 6 shows submicron and nanosized particles, the interplanar distances for which are closest to those in the TiC carbide. In addition to these particles, relatively large (micron-sized) particles of the same type were also observed in this steel. The formation of MC type particles is characteristic of many austenitic steels containing strong carbide-forming elements (Ti, V, Ta, Zr) [5,8].

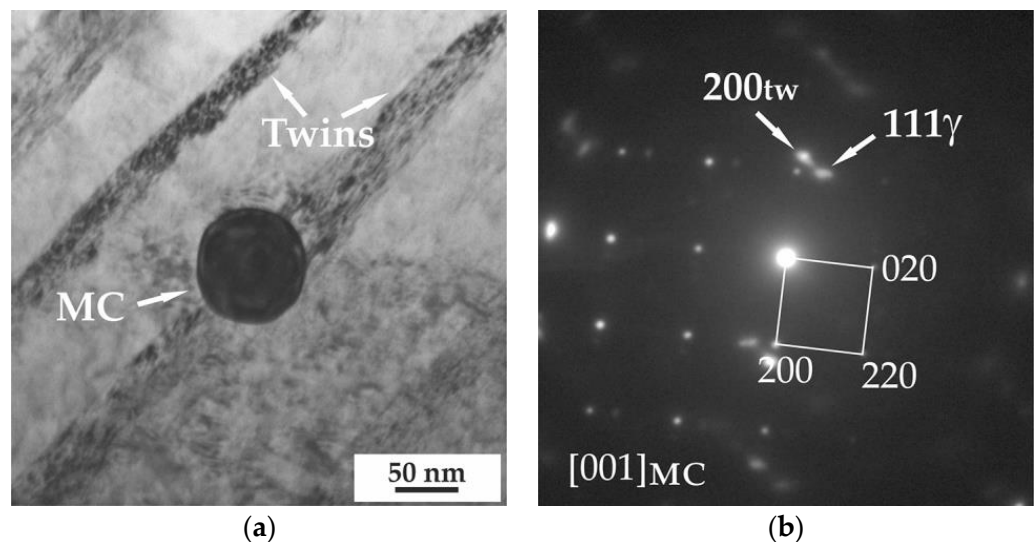


Figure 6. TEM BF image (a) and corresponding SAED pattern of dispersed MC type particles (b) after CR.

The values of the stacking fault energy (SFE) determine the tendency of austenitic steels to split perfect dislocations into partial ones, forming planar dislocation substructures, stacking faults, mechanical twins, and ϵ -martensite. The SFE value is significantly influenced by the manganese content, and high-manganese steels are typically characterized by fairly low values. According to [26], at an SFE value of (20–25 mJ/m^2), stacking faults can spontaneously form in austenitic steels, and plastic deformation promotes intensive mechanical twinning. At lower SFE values (15–20 mJ/m^2), steels tend to form ϵ -martensite. According to calculations carried out in the JMatPro software version 7.0 based on the

actual composition (Table 1), the stacking fault energy for the modified steel at 20 °C is 24.8 mJ/m². Thus, planar dislocation substructures, stacking faults, and deformation twins are expected to be observed in the steel.

Our studies show that after CR, a high density of micro- and nanotwins is observed in the steel structure (Figure 7). Thin plates of twins form in grains with orientations favorable for twinning relative to the rolling plane. Such grains can consist almost entirely of twin/matrix packets. The bright-field (Figure 7a) and dark-field images in the matrix reflection (Figure 7b) show that a fairly high dislocation density is detected in microtwin structures, with estimates ranging within 10^{14} – 10^{15} m^{−2}. In local areas near large dispersed particles, the dislocation density is higher. The diffraction pattern (Figure 7b) shows a characteristic azimuthal extension of some twin and matrix reflections. This may be due to the formation of dislocation substructures, the incorporation of gliding dislocations into twin boundaries, and a corresponding change in the orientation of the boundaries from the exact twin position.

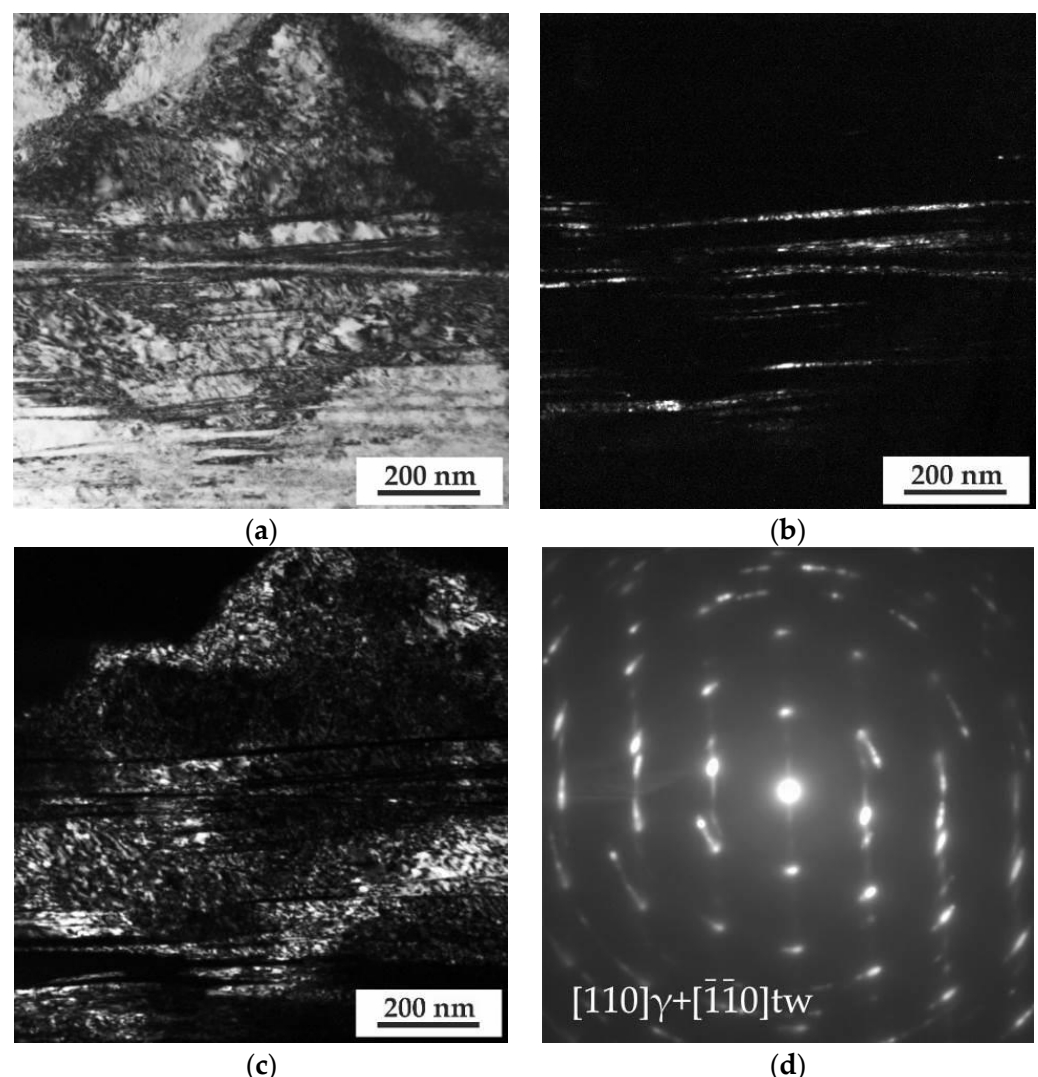


Figure 7. TEM images of microstructure after CR. BF image (a), dark-field (DF) images in twin reflection (b) and matrix reflection (c), SAED pattern, zone axes close to $[110]$ matrix + $[\bar{1}\bar{1}0]$ twins (d).

In addition to deformation twins, stacking faults, and individual ϵ -martensite plates were found after CR (Figure 8). The belonging of the plates to the ϵ -phase is confirmed by an analysis of the diffraction patterns (Figure 8b). Characteristic banded contrast is observed inside the plates, which may be associated with the dislocation substructure and

stacking faults in the hcp phase. The presence of stacking faults and individual plates of ϵ -martensite indicates relatively low SFE values.

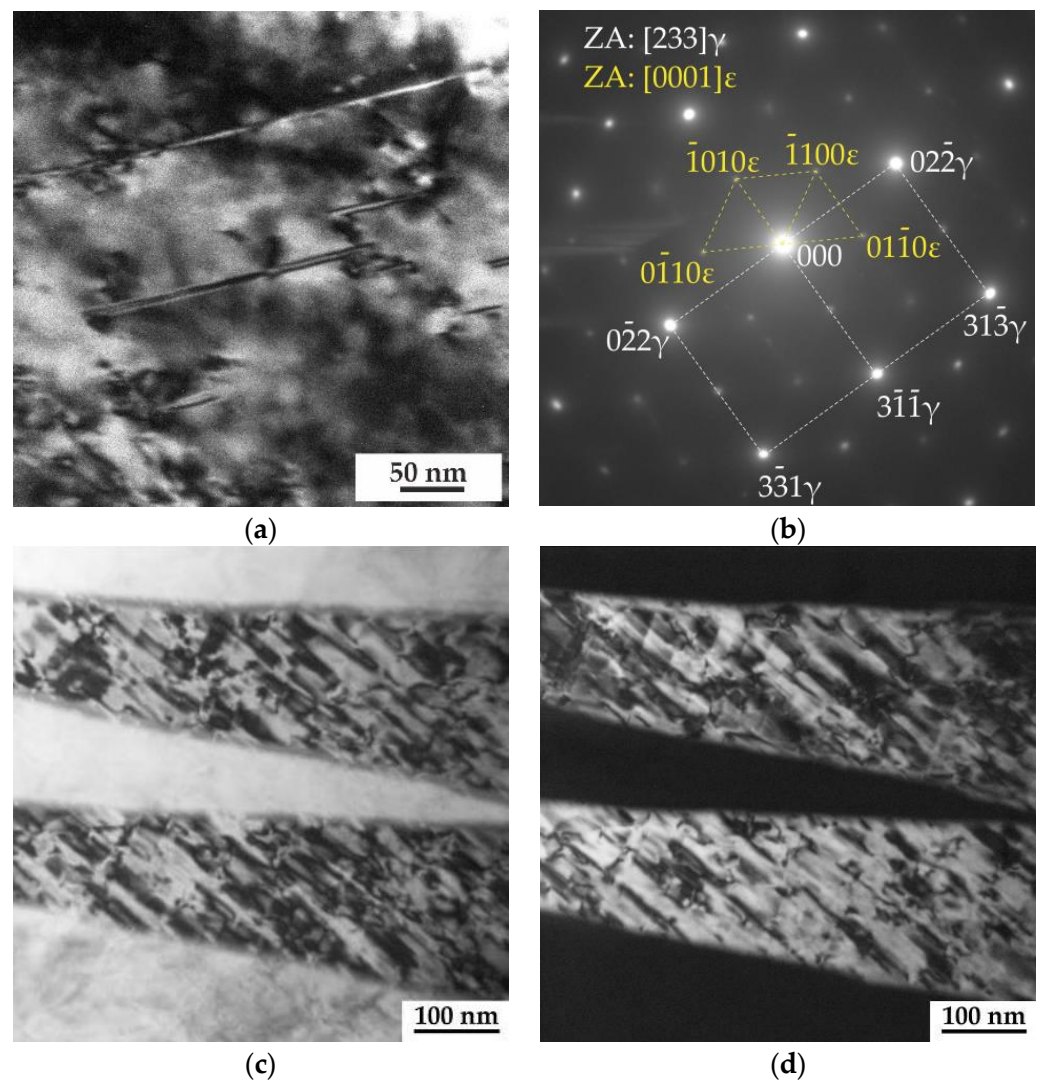


Figure 8. TEM images of microstructure after CR. BF image of stacking faults (a), SAED pattern with the scheme of analysis from ϵ -martensite plates (b), BF image of ϵ -martensite plates (c), DF image in the ϵ -martensite reflection (d).

After ST, there are flat dislocation pileups and stacking faults in the steel. Microtwins and/or ϵ -martensite plates can form near the dispersed carbide particles. However, both after ST and CR, ϵ -martensite plates are quite rare. Since no traces of this phase are detected by XRD, its content does not exceed several percent.

3.2. Temperature Dependence of Mechanical Properties

The tensile curves of the Fe–26Mn–11Cr–W–Si–Ta–Ti–V–Zr–0.32C modified high-manganese low-activation austenitic steel after ST and CR in the temperature range from 20 to 750 °C, and the curves depicting the temperature dependence of its mechanical properties, are shown in Figures 9 and 10, respectively. The standard deviation of the yield and tensile strengths does not exceed ± 20 MPa, and that of the elongation to failure does not exceed $\pm 5\%$. The general tendency of the strength properties of this steel decreasing with increasing tensile temperature is associated with a decrease in the values of Young's modulus. Higher strength values and lower ductility values after CR compared to those

after ST are due to microstructure features, specifically the high density of deformation twins and dislocations formed during rolling.

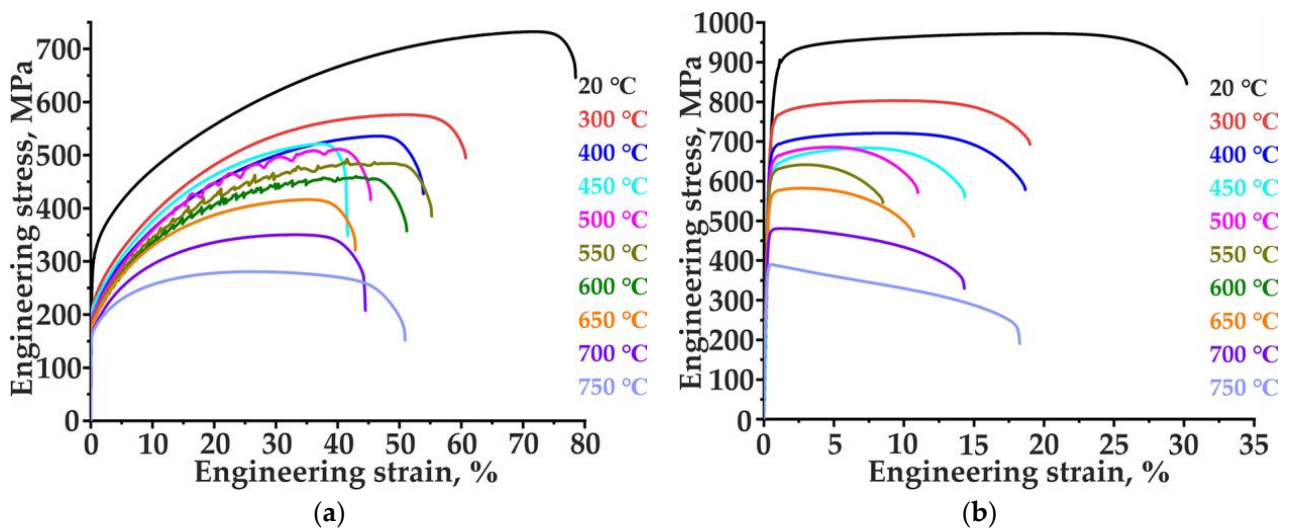


Figure 9. Stress–strain curves of the steel after ST (a) and CR (b) in the temperature range of 20 to 750 °C.

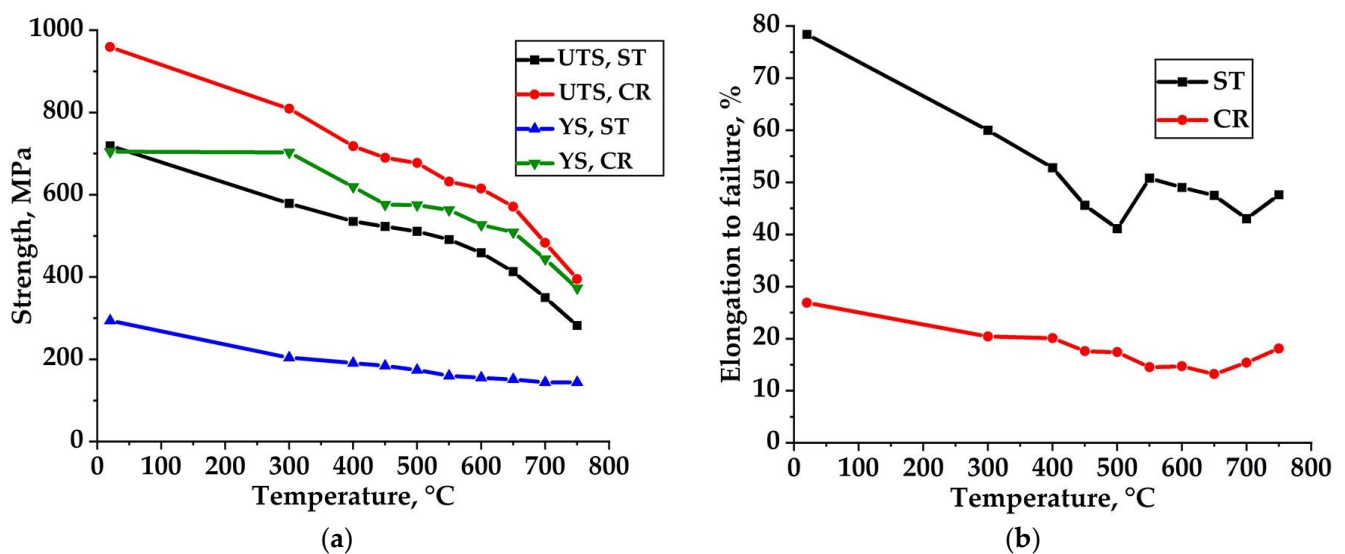


Figure 10. Variations of average values of mechanical properties of the steel with temperatures for ST (a) and CR (b).

An analysis of the temperature dependence of mechanical properties shows that the yield strength after ST decreases monotonically with increasing test temperature. The tensile strength after ST, as well as the yield and tensile strengths after CR, have a qualitatively similar dependence on the test temperature. The curves show several characteristic temperature intervals:

- 20 to 450 °C: The interval of gradual decline in strength properties. Elongation to failure in this range decreases by approximately 1.6–1.7 times relative to the values at 20 °C.
- 450 to 650 °C: The interval of a slight decrease in yield strength for CR and tensile strength for both treatments with increasing temperature. Elongation to failure has minimal values at 550 and at 700 °C after ST, and in the temperature interval of 550–650 °C after CR.

- 650 to 750 °C: The interval of a significant decrease in tensile strength after ST and the yield and tensile strength after CR. Elongation to failure starts increasing at 750 °C after ST and at 700 °C after CR.

The characteristic features of serrated flow are observed on the tensile curves after ST in the temperature range of 500–600 °C. After CR, there are no such features on the tensile curves. At the same time, in a close temperature range of 450–650 °C, the values of yield and tensile strength change slightly with increasing test temperature, and the values of elongation to failure are minimal. These features are characteristic of dynamic strain aging (DSA). This phenomenon is observed in many alloy steels [27–29] in a close range of deformation temperatures. It consists of the partial blocking of dislocations by atmospheres and impurity atoms under conditions of high-temperature plastic deformation, leading to an increase in flow stress and a decrease in ductility. When dislocations tear off from the atmospheres during deformation, characteristic serrations are formed on the flow curve.

3.3. Fractography

Fractographic studies of tensile samples after ST and CR show that the steel fracture in the investigated temperature range from 20 to 750 °C is predominantly the ductile dimple transcrystalline fracture (Figures 11 and 12). At room test temperature, multiple canyons of ductile intercrystalline fracture are also observed (Figures 11a,d and 12a,d). Moreover, their fraction after ST is higher than that after CR.

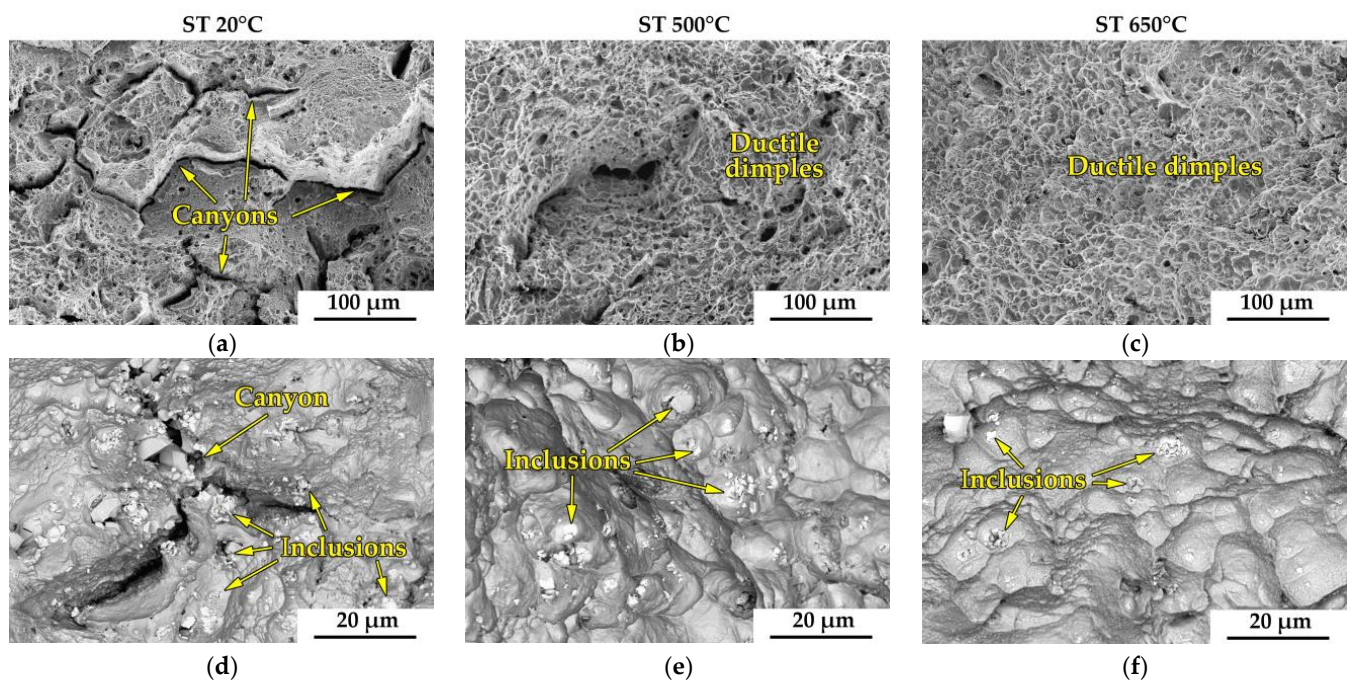


Figure 11. Fractographic images of the central part of tensile steel samples after ST, fractured at 20 °C (a,d), 500 °C (b,e), and 650 °C (c,f).

As the test temperature increases, the traces of intercrystalline fracture disappear. In addition, there is an increase in the size of the ductile fracture dimples. In particular, at $T = 20$ °C, their sizes range from ≈ 0.3 to ≈ 5 μm, while at $T = 650$ °C, they range from ≈ 1 to ≈ 15 μm (Figures 11a,c,d,f and 12a,c,d,f). In this case, in addition to the test temperature effect, we observed the impact of the size of the second phase particles on the size of the formed dimples. The largest particles (several microns in size) of MC and/or $M_{23}C_6$ type and their locally observed clusters are strong stress concentrators, contributing to the formation of the largest dimples or microcracks close to them (due to a combination of several dimples) when the samples are under tension.

It should be noted that in the interval of dynamic strain aging (500–600 °C), characterized by a considerable (≈ 1.5 –2 times) decrease in elongation to failure compared to room temperature tests, no signs of brittle fracture were found (Figures 11b,e and 12b,e).

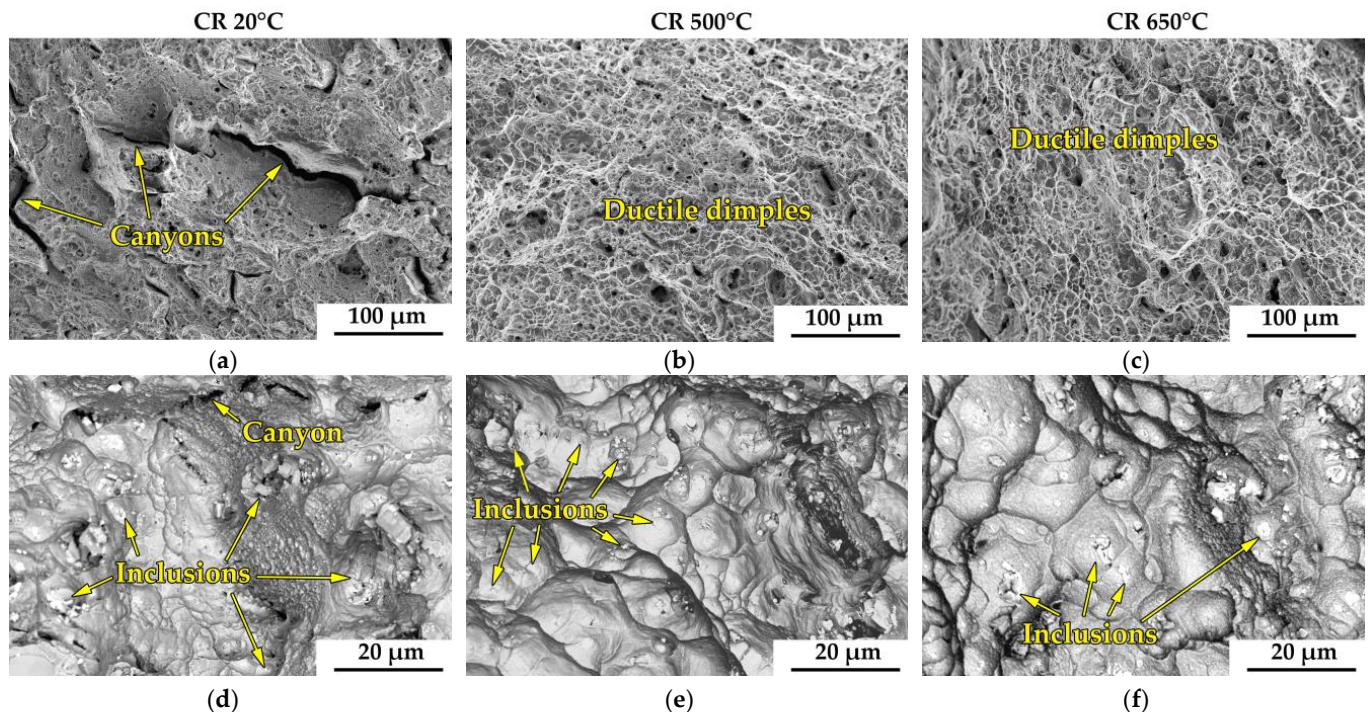


Figure 12. Fractographic images of the central part of tensile steel samples after CR, fractured at 20 °C (a,d), 500 °C (b,e), and 650 °C (c,f).

4. Discussion

A decrease in manganese content (by 3 wt. %) and an increase in carbon content (by 0.07 wt. %) in the modified steel, relative to the elemental composition of the Fe–29Mn–12Cr–W–Si–Ta–Ti–V–Zr–0.25C steel [19] after ST and CR, did not significantly affect the features of its microstructure. The differences were observed in the increased average grain size of the modified steel (1.6 times after ST and 2.2 times after CR). This may be due to the differences in elemental composition and the thermomechanical treatment of ingots of differing weights (10 kg vs. 2.5 kg) of the steels. The densities of dispersed carbide particles ($M_{23}C_6$ and MC) for the two steels are comparable.

The relatively high level of strength properties of the studied steel within the temperature range of 20–750 °C is ensured by a combined effect of substructural, dispersion, and solid solution strengthening mechanisms. The yield strength values at 20 °C are 30–50 MPa lower for the modified steel, probably due to its larger average grain size. At high tensile temperatures, the yield strength of the modified steel is 50–100 MPa higher than that of the previously studied steel. This may be due to increased efficiency in solid solution strengthening, namely a higher carbon content and increased tungsten content (1.2 vs. 0.9 wt. %).

When comparing the curves of the temperature dependence of mechanical properties for the modified steel with those of steels of the Fe–12Cr–(20–25)Mn–0.25C type [8], it can be noted that the modified steel, both after ST and CR, has higher values of elongation to failure and slightly lower values of strength properties in the temperature range of 20–600 °C. Moreover, at the maximum tensile temperature (750 °C), the modified steel after CR has fairly high yield strength values (≈ 370 MPa). Higher ductile properties are important under radiation near the operating temperature range, where the austenite solid solution decomposes with the precipitation of dispersed particles of $M_{23}C_6$ type carbides,

leading to a corresponding decrease in elongation to failure. The better ductile properties of the steels at high temperatures can increase their resistance to radiation embrittlement.

The SFE values at room temperature for the modified and previously studied steel [19] are approximately equal and reach $\approx 25 \text{ mJ/m}^2$. At this SFE value, mechanical twinning is one of the main mechanisms of plastic deformation at 20 °C. This determines the high elongation to failure values for both steels.

The estimation of the temperature M_s for the $\gamma \rightarrow \epsilon$ transformation, based on the elemental composition of the steel in accordance with [30], gave values of 82 K (−191 °C). Apparently, the studied steel has sufficiently low critical stresses for the start of the $\gamma \rightarrow \epsilon$ transformation, which could lead to the formation of ϵ -martensite plates near stress concentrators. However, at room temperature, the driving force ΔG for the $\gamma \rightarrow \epsilon$ martensitic transformation is quite low, which does not allow for the significant formation of the ϵ -phase.

The SFE value increases with the deformation temperature, and the role of mechanical twinning in plastic deformation decreases. According to the calculations carried out in the JMatPro program for the modified steel, at 300 °C, the SFE value reaches 60 mJ/m². At higher deformation temperatures, dislocation gliding becomes the dominant mechanism of plastic deformation. One of the reasons for the decrease in plasticity with increasing temperature of steel deformation is the gradual decrease in the contribution to plasticity from mechanical twinning (up to its complete disappearance), transitioning to the dislocation mechanism of deformation. In the temperature range of 450–650 °C, the decrease in ductility is probably determined by DSA. This effect manifests itself most clearly after ST in the form of serrated flow on the stress–strain curves. The blocking of dislocations by atmospheres and impurity atoms determines sufficiently high flow stresses, and helps reduce elongation to failure. At higher deformation temperatures (700–750 °C), the tendency for dislocation climb increases, making them more easy to tear off from atmospheres and impurities, and overcome obstacles in the form of dispersed carbide particles, leading to increased ductility.

Throughout the entire temperature range studied in this work, the modified steel predominantly fractures through a viscous transgranular mechanism. The presence of a large number of dispersed particles along grain boundaries can lead to ductile intercrystalline fracture at room temperature. At higher deformation temperatures, the signs of intercrystalline fracture disappear.

5. Conclusions

In order to reduce the tendency to form undesirable phases during long-term high-temperature annealing and ensure the high stability of austenite, determined by the nickel equivalent ($\text{Ni}_{\text{eqv}} = 22.65$), a modified low-activated austenitic steel, Fe–26Mn–11Cr–W–Si–Ta–Ti–V–Zr–0.32C, has been developed. The features of its microstructure after solution treatment and cold rolling have been studied. The temperature dependence of the mechanical properties of steel in the temperature range of 20–750 °C has been investigated. The main conclusions are as follows:

1. The formation of stacking faults, flat dislocation pileups, individual plates of ϵ —martensite after ST, and a high density of microtwins after CR is determined by the fairly low (24.8 mJ/m^2) value of the stacking fault energy of this steel.
2. The investigated steel has a good combination of strength and plastic properties over a wide temperature range. Its strength properties are ensured by the combined effects of substructural, dispersed, and solid solution strengthening. As the tensile temperature increases, the strength properties of the steel decrease by approximately two times. However, after CR at the maximum test temperature (750 °C), fairly high values of yield strength of at least 370 MPa, are observed. The modified steel has higher elongation to failure values compared to the corresponding values for steels of the Fe–12Cr–(20–25)Mn–0.25C type.
3. In the temperature range of 450–650 °C, DSA is observed in the modified steel. However, despite the appropriate ductility decrease associated with this phenomenon, the values of elongation to failure are no smaller than 13%.

4. Mechanical twinning plays a significant role in ensuring high ductility values at tensile temperatures up to 300 °C. At higher temperatures, the main deformation mechanism is dislocation gliding.

5. Predominantly, in the temperature range of 20–750 °C, ductile transgranular fracture is observed. Dispersed particles of the MC and $M_{23}C_6$ type carbides are responsible for intercrystalline fracture at room temperature.

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