Tempering influence on residual stresses and mechanical properties of AISI 4340 steel

Marcel Freitas de Souza1 · Luana Ferreira Serrão1 · Juan Manuel Pardal1 · Sérgio Souto Maior Tavares1 · Maria Cindra Fonseca1

Received: 4 November 2021 / Accepted: 3 February 2022 / Published online: 10 February 2022 © The Author(s), under exclusive licence to Springer-Verlag London Ltd., part of Springer Nature 2022

Abstract
The present work evaluated the tempering temperature influence on microstructure, mechanical properties, and residual stresses of AISI 4340 steel. The residual stresses were measured by X-ray diffraction (XRD) by the sin²ψ method and compared to magnetic Barkhausen noise (MBN). The residual stresses exhibited high tensile values after quenching, but a small relief was observed in tempering treatments at 300 °C and 400 °C, which also presented a hardness decrease compared to the as-quenched condition. XRD and MBN analyses indicated that residual stresses became compressive in tempering performed between 500 and 650 °C. Therefore, compressive residual stresses combined with appropriate hardness and toughness values (35 HRC and 33 J) obtained from 500 °C tempering temperature can be used to improve the mechanical properties of AISI 4340 steel components. Additionally, a mathematical model was established to estimate the tempered martensite hardness for different tempering temperature conditions. This model showed high accuracy ($R^2 = 0.99$) for a holding time of 90 min.

Keywords AISI 4340 steel · Heat treatments · Mechanical properties · Residual stresses · X-ray diffraction · Magnetic Barkhausen noise

1 Introduction
The increasing demand for components with high performance and the competitive market has promoted the improvement of techniques that guarantee better properties to the materials. Within these techniques, it is possible to highlight the heat treatments of steels, which are of particular importance in the metalworking industry, since the properties and characteristics of the product are largely modified [1].

The AISI 4340 steel is produced and employed in quenched and tempered condition. This steel is widely used in demanding structural applications such as axles, connecting rods, and transmission gears in the automobile industry and landing gear in the aeronautical industry due to a combination of high mechanical strength and toughness. However, such material is susceptible to tempering embrittlement phenomena at tempering temperature range 300–400 °C. This problem has been the subject of several studies conducted through the analysis of the microstructure and the mechanical properties of this steel under different tempering conditions [1–3].

The residual stresses state in the material subjected to heat treatment is function of the soaking temperature, phase transformation, and thermal gradient between the core and the surface of the part to be treated. Tempering heat treatments at different temperatures can be used as a stress relief technique, thus ensuring the reduction or even absence of tensile residual stresses, which may have a deleterious effect on the resistance to stress corrosion cracking and fatigue life of material. Despite the wide industrial use of AISI 4340 steel and in-depth study of the mechanical behavior of this material, there is little information in the literature regarding the influence of heat treatment parameters on residual stresses [4–8].

Considering these facts, it is important to evaluate alternative non-destructive and low-cost residual stress techniques, such as magnetic Barkhausen noise (MBN), which can be implemented in production lines by providing...
faster and high-quality results. Nevertheless, this technique presents some challenges due to the combined effect of residual stress and microstructure, which require careful pre-calibration and verification procedures to obtain reliable quantitative results [6].

The present study aims to evaluate the residual stresses generated in quenching and tempering heat treatments at different temperatures, as well as to analyze the tempering influence on the final stress state and the mechanical and microstructural properties of AISI 4340 steel, comparing these results with MBN. Additionally, an equation was established that relates hardness and tempering temperature with excellent accuracy and, thus, allows complementing the literature on the heat treatment applicable to 4340 steel, since there was no study considering a holding time of 1.5 h.

2 Materials and experimental techniques

The AISI 4340 steel studied was obtained from a disc of 10.5 mm thickness, cut from a forged and normalized bar of ∅ 228.6 mm. Chemical composition and mechanical properties (yield strength (σYS), ultimate tensile strength (σUTS), elongation, reduction of area (RA), and Rockwell hardness—HRC), according to the manufacturer, are shown in Tables 1 and 2, respectively.

Two different types of samples were made from the AISI 4340 steel. One type had Charpy dimensions and was used to assess residual stresses by XRD, MBN, and absorbed energy (AE) in the Charpy impact test, which also enabled fractured surface analysis. The other type, called reduced sample, was manufactured to analyse microstructure and HRC.

Quenching treatment was carried out in a tubular furnace with a controlled argon inert gas atmosphere to avoid decarburizing. The samples were placed in a furnace at 660 °C, which was heated gradually until the soaking temperature of 860 °C. After 60 min soaking time, the samples were removed and dipped in oil. A quenched sample of each type was selected for tempering, which was carried out using five different temperatures, 300 °C, 400 °C, 500 °C, 600 °C, and 650 °C, for 90 min and later air-cooled.

A Rockwell 2000 Series hardness testing machine with a conical diamond indenter tip was used to perform the indentations on the samples. HRC test was performed in as-received condition and after tempering treatment. Metallographic samples were etched with 2% Nital for 10 s.

Charpy impact test was performed at room temperature using a universal impact machine with 300 J capacity and 0.5 J accuracy and broken Charpy specimens were observed through a stereo microscope.

Fracture surfaces were characterized by scanning electron microscope (SEM) and energy-dispersive spectroscopy (EDS) was used to determine the chemical composition of the phases and inclusions in fracture surfaces. For this analysis, a scanning electron microscope NSPECT model produced by EI Company was used.

A StressRad stress analyzer manufactured by Radicon Company was used to evaluate the residual stresses of as-received and heat-treated condition. The residual stresses were measured by XRD using sin²ψ method, CrKα radiation and diffracting the plane {211} in the longitudinal direction (LRS), coincident with the largest length direction, and also in the transverse direction (TRS).

It is noteworthy that although the normalization treatment makes it possible to reduce the anisotropy, the residual stresses are modified when the samples are subjected to the manufacturing processes necessary for the production of Charpy specimens and, therefore, it is essential to consider both directions instead of considering the material as fully isotropic.

MBN was evaluated using Rollscan 200–1 analogue analyzer manufactured by Stresstech Group Company. It was used a sinusoidal magnetic field with 125 Hz frequency, 4.8 Vpp excitation magnetization amplitude and 0.6 V gain for signal amplification in the longitudinal (LMBN) and transverse (TMBN) directions.

The tests performed in this work are shown in Fig. 1.

3 Results and discussions

Figure 2 presents the results of residual stresses, MBN, Charpy impact toughness test, and HRC test.

Oil quenching provided higher hardness values than as-received material, which presented 30 ± 1 HRC. The tempering at 300 °C led to a reduction of less than 8% in the

Table 1 Chemical composition of AISI 4340 (% weight). Fe in balance

|        | C  | Mn  | Si  | P (max) | S (max) | Cr   | Ni  | Mo  | Al  | Cu  |
|--------|----|-----|-----|---------|---------|------|-----|-----|-----|-----|
| Manufacturer | 0.40 | 0.73 | 0.28 | 0.01    | 0.02    | 0.78 | 1.73| 0.21| 0.02| 0.11|
| SAE 4340 | 0.38–0.43 | 0.60–0.80 | 0.15–0.35 | 0.04 | 0.04 | 0.70–0.90 | 1.65–2.00 | 0.20–0.30 | -   | -   |

Table 2 Mechanical properties of AISI 4340

| Re (MPa) | Rm (MPa) | Elongation (%) | RA (%) | HRC |
|----------|----------|---------------|-------|-----|
| 765      | 960      | 14.8          | 39.5  | 32  |
hardness concerning quenched condition and, therefore, low tempering does not significantly change hardness [9].

Hardness gradually decreased with increasing tempering temperature, as also verified by Kashefi et al. [4], who affirmed that tempering temperature is significant in hardness reduction due to softening of martensitic matrix by carbon rejection. Manokaran et al. [10] reported hardness values similar to that observed in the present research considering different tempering temperature for AISI 4340. In this way, higher tempering temperatures provide sufficient mobility for the substitutional alloying elements to cause cementite growth, resulting in a continuous decrease in strength during tempering [11].
The absorbed energies of the specimens tempered at 300 °C and 400 °C were considerably low and below that found by Kwon et al. [2] and Chi et al. [3]. Tempering embrittlement occurs, according to Clarke et al. [1], with tempering in 200 to 400 °C probably due to the formation of intra-lath cementite from retained austenite. For this reason, this temperature range is critical, making it recommendable to use temperatures between 150 and 200 °C for low tempering treatment to obtain an adequate combination of strength and ductility.

For tempering temperatures above 500 °C, the absorbed energy increased considerably and, therefore, a sharp growth of energy absorbed with increased tempering temperature was verified, which is consistent with Jiang et al. [8].

Residual stresses in the as-received condition indicated a mean of 86 MPa in longitudinal (LRS) direction and −220 MPa in transverse (TRS) direction. The magnitude and nature of residual stresses are influenced by heat treatment and machining process, and probably the thermal effects of electrical discharge machining prevailed in the longitudinal direction, generating tensile residual stresses [12]. Quenching process in steels induces the generation of residual stresses which may be thermally generated because of high temperature gradients within the component and austenite–martensite phase transformations [13].

Relaxation of residual stress is expected if the stress exceeds the yield stress of the material or if density changes occur, such as decomposition of martensite or retained austenite, which makes it essential to carry out a microstructure analysis. Below 100 °C, no relaxation is expected, between 100 and 200 °C some could occur, and above 200 °C residual stress relaxation should take place. Temperatures above about 400 °C is used to create quench and tempered steels that have mechanical properties very different from the as-quenched condition [14].

Residual stresses after quenching were close to 300 MPa in the longitudinal direction and 230 MPa in the transverse direction. Tensile residual stresses are undesirable since they can decrease the fatigue life by promoting nucleation and propagation of surface cracks [15].

In tempering carried out above 500 °C, the stress relief was even more significant and provided the inversion of stresses nature, generating compressive residual stresses in both directions. Thus, the increase in the tempering temperature made the treatment more efficient in terms of reducing the tensile residual stresses resulting from the quenching.

The hardness profile is similar to the residual stresses behavior, since both were inversely proportional to tempering temperature increase and, therefore, opposite to the Charpy impact toughness result.

The MBN measured after tempering between 300 and 500 °C were low and without a significant difference concerning the direction. Tempering performed above 500 °C increased MBN, and this behavior corresponds to a compressive residual stresses state. The increase in tempering temperature caused the decrease of anchor points of the magnetic domain and, thus, increasing MBN.

The magnetization orientation is no longer favored at tempering temperatures above 400 °C and reverse domain nucleation and subsequent domain wall movement occurs at lower magnetic fields. All these factors promote the movement of the domain walls so that the amplitude of the MBN peak increases. In tempering between 500 and 650 °C, in parallel to the progressive coarsening of the microstructure, the
average size of the domains increases. These morphological changes and the compressive residual stresses state lead to a drastic increase in the MBN peak and a significant shift of the peak position to a lower external magnetic field by reducing the resistance to nucleation and movement of the domains [5].

Residual stresses and microstructure changes affect MBN, since significant changes can be observed during martensite transformation. Higher tempering temperature influences the precipitated carbides sizes, allowing dislocations to move more easily, causing a sharp decrease in hardness and a MBN increase. As described by Moorthy et al. [6], there is an increment in MBN magnitude with hardness reduction, because the domain wall displacement would be facilitated with microstructure softening.

Therefore, MBN can be applied to reduce inspection costs in production lines, using a faster and cheaper technique, avoiding destructive tests and ensuring that all samples could be quickly inspected. In ultra-high strength alloy steels, such as AISI 4340, the heat treatment in the range of 300–400 °C provides an evident reduction in mechanical properties and tensile residual stresses, and this behavior can be predicted using the MBN technique [7].

The mathematical model proposed by Hassan and Jabbar [16] can estimate the hardness of 4340 steel submitted to quenching and tempering treatments. The parameter of tempering (PT) considers that martensite softening is a phenomenon of diffusion controlled throughout the tempering of steels. The PT is a function of tempering temperature, T in Kelvin, and holding temperature, t in hours, according to Eq. (1) given by Hollomon and Jaffe [17].

$$PT = T \cdot \log(t) + G$$  \hspace{1cm} (1)

where $G$ is a constant, being equal to 10.3 for AISI 4340 [18].

Another equation for PT, according to Eq. (2), involves the parameters of the alloy elements of the material [15].

$$PT = T \cdot \left[ \log(T) + \left( H_0 + \sum_i H_i X_i \right) \right]$$ \hspace{1cm} (2)

where $H_i$ is the constant for the alloying element, $X_i$ is the mass amount % of the alloying element, and $H_0$ is a constant for tempered pure iron, as shown in Table 3.

Equation (3) describes the hardness of tempered martensite (HTM) in Vickers (HV) with the implementation of the exponential function factor correction, as performed by Hassan and Jabbar [16] in the original methodology proposed by Kang and Lee [18]. This exponential function factor correction was based on the values described in Fig. 3 using the conversion of HRC into Vickers hardness according to ASTM E140 [19].

$$HTM = \left( \frac{1542.9 - \frac{25.3}{X_C}}{X_C} \right) \cdot \exp(-1.64 \times 10^{-4} \cdot PT)$$ \hspace{1cm} (3)

A correlation factor was implemented in Eq. (3) to obtain a better fit of the data. This correction factor expressed in Eq. (4) is related specifically to a holding time of 1.5 h.

$$Factor of correlation = [85.5 \cdot \ln(T) - 564]$$ \hspace{1cm} (4)

Equation (5) can be written by adding the factor of correlation to Eq. (3).

$$HTM = \left( \frac{1542.9 - \frac{25.3}{X_C}}{X_C} \right) \cdot \exp(-1.64 \times 10^{-4} \cdot PT) \right]$$ \hspace{1cm} (5)

+ \left[ 85.5 \cdot \ln(T) - 564 \right]

Using the methodology developed by Hassan and Jabbar [16], the value of HTM can be expressed using the values of PT expressed in Eqs. (1) and (2).

$$HTM = 0.5 \cdot \left\{ \left( \frac{1542.9 - \frac{25.3}{X_C}}{X_C} \right) \right\} \cdot \left[ \exp(-1.64 \times 10^{-4} \cdot PT1) + \exp(-1.64 \times 10^{-4} \cdot PT2) \right]$$ \hspace{1cm} (6)

+ \left[ 85.5 \cdot \ln(T) - 564 \right]

| Parameter  | $H_0$  | $H_C$  | $H_{Mo}$ | $H_{Si}$ | $H_{Ni}$ | $H_{Cr}$ | $H_{Mo}$ |
|------------|--------|--------|----------|----------|----------|----------|----------|
| Value      | 17.396 | -6.661 | -1.604   | -3.412   | -0.248   | -1.112   | -4.355   |

Fig. 3 Variation of measured and calculated hardness for a holding time of 1.5 h

Table 3 Equation parameters for alloying effect according to Kang and Lee [18]
where PT1 and PT2 are tempering parameters that were calculated from Eqs. (1) and (2), respectively.

Figure 3 shows the variation of measured and calculated hardness for a holding time of 1.5 h with high accuracy ($R^2 = 0.99$).

The present research provided a specific model for 1.5 h and thus complemented the research of Hassan and Jabbar [16], who developed mathematical models for 2 and 48 h of holding time.

Figure 4 shows the microstructure of quenched and tempered samples.

Quenching generated, as observed in Fig. 4a, thin martensite laths, which were responsible for the high hardness of this sample. These laths are characteristic of martensite of low alloy steels, which are formed by a diffusionless shear mechanism when austenite is cooled rapidly to room temperature [11, 20].

Tempering at 300 °C promoted low-temperature-tempered martensite formation, as shown in Fig. 4b, not causing significant changes concerning hardness, due to the presence of fine carbides of Fe$_{23}C$, called ε-phase. This result is consistent with Clarke et al. [1], who also verified carbides after conventional quenching and tempering at 300 °C for 1 h.

Tempering at 400 °C, as evidenced in Fig. 4c, generated a microstructure similar to Fig. 4b; however, there was a noticeable change in the morphology of the martensite laths with the presence of larger laths.

Diffusion reactions occur more easily with increasing tempering temperature, and thus, there was a trend toward the formation of more stable carbides, such as Fe$_3C$, M$_7$C$_3$, and M$_{23}C_6$, providing martensite with less hardness, as observed in Fig. 4d.
For tempering temperatures at 600 °C, diffusional reactions are even more intense when compared to previous conditions. Thus, in addition to the formation of more thermodynamically stable carbides, their growth and coalescence can also occur. Additionally, in this tempering condition, based on the Fe-Fe₃C diagram, the temperature is close to the critical temperature Aₜ, favoring the formation of ferrite in the microstructure. In Fig. 4e, coarser morphological aspects of quenched martensite were observed with signs of precipitation of ferrite in the microstructure, which, combined, with the formation of more stable and coarser carbides, led to a decrease in hardness.

The sample tempered at 650 °C showed characteristics of martensite quenched at high temperatures, highlighting the formation of a microstructure of ferrite and coarse carbides, as evidenced in Fig. 4f. As expected, the hardness suffered a decrease in relation to the tempering condition at 600 °C and the values were very close to those of the material in the as-received condition.

Fracture surfaces using a stereo microscope with low magnifications are shown in Fig. 5.

Figure 5a, b presented brittle fracture surfaces with a low level of plastic deformation, which agrees with the low energy impact test verified in Fig. 2. The fracture occurred due to the cleavage mechanism, quite common for this type of fracture and characteristic of samples susceptible to tempering embrittlement phenomena.

Fracture surfaces showed for Fig. 5c, d, e presented ductile characteristics with a dull appearance and clear signs of plastic deformation with lateral expansion.

Figure 6 shows the microstructures obtained by scanning electron microscope (SEM), in the mode of secondary electrons (SE), for different tempering temperatures.

Figure 6a highlights a mixed-type martensite microstructure composed of laths and large plates, a feature confirmed by Lee and Su [21], who stated that the medium carbon steels have a complex microstructure. The martensite laths, previously highlighted by optical microscopy, were clear, as well as the contours of previous austenite. At this temperature, the hardness changed little because the fine carbides still provide good strength and counteract the carbon depletion in martensite, which is still small.

Furthermore, the presence of retained austenite was not detected, although it is more appropriate to characterize the presence of this phase by X-ray or magnetic diffraction measurements. Ajus et al. [22] did not detect the presence of austenite retained in AISI 4340 steel tempered at the same temperature used in this work. This is because the martensite finish temperature, Mₐ, is above the room temperature and the austenitizing temperature of 860 °C, which is well below the temperature at which retained austenite formation begins.

Figure 6b shows a mixed microstructure consisting of laths and some large martensite sheets. As the annealing temperature increased, conditions were more favorable for greater diffusion. Therefore, the carbon in supersaturation in the martensite precipitates in the form of more stable carbides, which have larger dimensions compared to tempering at 300 °C. In this sense, some carbides in the order of 200 nm were identified by the arrows in Fig. 6b. These particles tended to deplete the matrix in terms of precipitated carbon content, but the hardness remained high because the precipitated carbides were still very fine.

In Fig. 6c, a tempered martensite microstructure was observed and, as well as for the tempering at 400 °C, the precipitated carbides were still very fine. These precipitates hamper the movement of dislocation in the matrix; therefore, the hardness values were still high, but smaller compared to low tempering.

Figure 6d shows a tempered martensite microstructure with ferrites and rounded carbides, about the size of 400 nm. In this way, dislocations could move more easily, causing a hardness fall.

Figure 6e presents a microstructure similar to Fig. 6d, but with even coarser and rounded carbides (500 nm), further reducing the hardness shown in Fig. 1.

Figure 7 shows the characterization using SEM by secondary electron (SE) and electron backscattered diffraction (EBSD) analysis of fracture surfaces. In some cases, energy-dispersive spectroscopy (EDS) was used as an elemental micro-analysis tool to qualitatively determine the chemical composition of certain regions of the fracture surface.

Figure 7a shows a quasi-cleavage fracture surface and small colonies of dimples or microcavities indicated by the white arrows. A faceted surface was also observed, corresponding to planes of lower atomic density in which the defects have propagated, named as cleavage planes. The dark arrow indicated river patterns, which would correspond to the crack propagation place.

Figure 7b reveals a quasi-cleavage mechanism similar to Fig. 7a, although with smaller microcavities or dimples in a larger proportion.

Figure 7c presents elongated inclusions in the direction perpendicular to the notch, having a more fibrous surface appearance than samples tempered at 300 °C and 400 °C. Thus, the largest proportion and size of dimples were consistent with the absorbed energy for this condition.

Figure 7d exhibits inclusions and a more fibrous-looking morphology compared to previous conditions. EDS of point 1 indicated that the inclusion was rich in S and Mn, giving sufficient evidence of this inclusion being a manganese sulfide (MnS), as shown in Table 4. The manufacturer indicated that the material in the as-received condition had 50% less S than the maximum allowable by the standard. However, the precise composition of S should be determined by more accurate methodologies, such as combustion, as this element has a significant harmful action in the drop of toughness.
Figure 7e shows a fibrous appearance with large sized dimples. An elemental microanalysis was also carried out in point 2, which corresponded to an inclusion particle that gave rise to a dimple, product of the shear stresses acting on this region. The results shown in Table 4 indicated that the particle was rich in Mn and Cr.

Figure 7f evidences elongated inclusions in the direction perpendicular to the notch. The fibrous aspect of the fracture was more marked when compared to the sample tempered at 600 °C.

Figure 7g-i, which were amplifications of Fig. 7f, show dimple colonies. EDS analysis was carried out in points 3, 4, and 5 to characterize the chemical composition of particles and/or inclusions.
Fig. 6 SEM of tempered samples at a 300 °C, b 400 °C, c 500 °C, d 600 °C, and e 650 °C

Point 3 showed the preponderant presence of Al, Mn, S, Ca, and Mg, highlighting once again the need for strict control of the elements of the studied alloy.

Point 4 corresponded to an elongated inclusion of MnS and point 5 was also rich in Mn and S, noting that, in general, the results were in accordance with what was exposed by Lee and Su. It is noteworthy that although this type of inclusion is not considered the most damaging in steels in general, the elongated sulfides influence the reduction of impact toughness, since voids are initially formed in the MnS inclusions during the fracture process.

The arrangement of the sulfides reduced the toughness, since the crack propagation occurred more easily along the grain boundary, clarifying the results below the expected obtained in the impact tests [23].

Additionally, the higher the tempering heat treatment temperature, the greater the amount and size of dimples observed on the fracture surface.
Table 4 Chemical composition of the points described in Fig. 7

| Point | Chemical composition (% weight) |
|-------|--------------------------------|
|       | Mg  | Al  | S     | Ca    | Mn    | Fe    | Cr    | Ni    |
| 1     | 0.00 | 2.01 | 31.80 | 0.43  | 59.33 | 6.44  | 0.00  | 0.00  |
| 2     | 0.00 | 0.02 | 0.03  | 0.11  | 8.29  | 89.87 | 1.68  | 0.00  |
| 3     | 3.61 | 36.04| 16.37 | 5.52  | 23.94 | 14.52 | 0.00  | 0.00  |
| 4     | 0.00 | 0.00 | 29.92 | 0.72  | 61.00 | 8.36  | 0.00  | 0.00  |
| 5     | 0.00 | 0.00 | 9.85  | 0.42  | 16.33 | 71.38 | 1.05  | 0.98  |

Fig. 7 Fracture surface of tempered samples at a 300 °C, b 400 °C, c 500 °C, d 600 °C, e 600 °C with a selected-area amplification of d; f 650 °C, g 650 °C with a selected-area amplification of f; h 650 °C with a selected-area amplification of h; i 650 °C with a selected-area amplification of h.
4 Conclusions

The present study evaluated the influence of tempering temperature on residual stresses and mechanical properties of AISI 4340 steel and the findings can be summarized as follows:

1. Tempering temperatures above 500 °C caused a significant residual stresses relief, providing compressive residual stresses in longitudinal and transverse directions.
2. With tempering temperature increase, precipitated carbides become coarser and microstructure, in general, less refined, which explains hardness reduction.
3. The tempering temperature increase caused the decrease of the anchor points of the magnetic domain, facilitating the movement of these domains, thus, producing higher MBN.
4. The mathematical model obtained by regression analysis method allowed predicting the hardness of tempered martensite (HTM) with high accuracy for 1.5-h holding time.

Author contribution Marcel Freitas de Souza: investigation, writing—original draft, writing—review and editing. Luana Ferreira Serrão: conceptualization, formal analysis, investigation, methodology. Juan Manuel Pardal: investigation, methodology, resources, supervision, validation, writing—original draft. Sérgio Souto Maior Tavares: validation, writing—original draft. Maria Cindra Fonseca: conceptualization, formal analysis, funding acquisition, investigation, methodology, project administration, resources, supervision, validation, visualization, writing—review and editing.

Funding This study was financed in part by the Coordenação de Aperfeiçoamento de Pessoal de Nível Superior—Brasil (CAPES)—Finance Code 001. This study is also supported by the CNPq (304129/2018–6) and FAPERJ (E 26/211.14/2019 (250854) and 03–2017 of Young Scientist of Our State).

Declarations

Ethics approval Not applicable.

Consent to participate Written informed consent for publication was obtained from all participants.

Consent for publication Written informed consent for publication was obtained from all participants.

Conflict of interest The authors declare no competing interests.

References

1. Clarke AJ, Klemm-Toole J, Clarke KD et al (2020) Perspectives on Quenching and Tempering 4340 Steel. Metall Mater Trans A 51:4985–5005. https://doi.org/10.1007/s11661-020-05972-1
2. Kwon H, Cha JC, Kim CH (1988) The effect of grain size on fracture behavior in tempered martensite embrittlement for AISI 4340 steel. Mater Sci Eng 100:121–128. https://doi.org/10.1016/0025-5416(88)90247-9
3. Chi YC, Lee S, Cho K, Duffy J (1989) The effects of tempering and test temperature on the dynamic fracture initiation behavior of an AISI 4340 VAR steel. Mater Sci Eng A 114:105–126. https://doi.org/10.1016/0921-5093(89)90085-0
4. Kashefi M, Rafaianjani A, Kahrobae S, Alae K (2012) Magnetic nondestructive technology for detection of tempered martensite embrittlement. J Magn Magn Mater 324:4090–4093. https://doi.org/10.1016/j.jmmm.2012.07.029
5. Davut K, Gür CH (2007) Monitoring the microstructural changes during tempering of quenched SAE 5140 steel by Magnetic Barkhausen Noise. J Nondestr Eval 26:107–113. https://doi.org/10.1007/s10921-007-0025-x
6. Moorthy V, Shaw BA, Evans JT (2003) Evaluation of tempering induced changes in the hardness profile of case carburized EN36 steel using magnetic Barkhausen noise analysis. NDT & E Int 36(1):43–49. https://doi.org/10.1016/S0963-8695(02)00070-1
7. Santa-aho S, Sorsa A, Honkanen M, Vippola M (2020) Detailed Barkhausen noise and microscopy characterization of Jominy end-quench test sample of CF53 steel. J Mater Sci 55:4896–4909. https://doi.org/10.1007/s10853-019-04284-z
8. Jiang B, Wu M, Zhang M, Zhao F, Zhao Z, Liu Y (2017) Microstructural characterization, strengthening and toughening mechanisms of a quenched and tempered steel: Effect of heat treatment parameters. Mater Sci Eng A 707:306–314. https://doi.org/10.1016/j.msea.2017.09.062
9. ASM Handbook Volume 4A (2013) Steel heat treating fundamentals and processes. ASM International, Materials Park
10. Manokaran M, Kashinath AS, Jha JS, Toppo SP, Singh RP (2020) Influence of tempering in different melting routes on toughness behavior of AISI 4340 steel. J Mater Eng Perform 29(10):6748–6760. https://doi.org/10.1007/s11665-020-05164-3
11. Ning D, Dai CR, Wu J, Wang YD, Wang YQ, Jing Y, Sun J (2021) Carbide precipitation and coarsening kinetics in low carbon and low alloy steel during quenching and subsequently tempering. Mater Charact 176:111111. https://doi.org/10.1016/j.matchar.2021.111111
12. García Navas V, Ferreres I, Marañón JA, García-Rosales C, Sevilla JL, Wang YD, Wang YQ, Jing Y, Sun J (2019) The effects of immersion rate on the distortion and residual stresses in quenched SAE 5160 steel using FEM 9:5557–5571. https://doi.org/10.1016/j.matchar.2019.09.024
13. Ericsson T (2014) Residual stresses produced by quenching of martensitic steels. Comprehens Mater Process 12:271–298. https://doi.org/10.1016/b978-0-08-096532-1.01209-7
14. Withers PJ, Bhadeshia HKDH (2001) Residual stress. Part 1 – measurement techniques. Mater Sci Technol 17:355–365
15. Hassan AD, Jabbar MA (2021) New prediction model of tempered martensite hardnesses for quenched and tempered low-alloy steel. Mater Perform Charact 10(1):267–277. https://doi.org/10.1016/j.matchar.2020.04.012
16. Hollomon JH, Jaffe LD (1945) Time-temperature relations in tempering steel. Trans Metall Soc AIME 162:223–249
17. Khang S, Lee S (2014) Prediction of tempered martensite hardness incorporating the composition-dependent tempering parameter in low alloy steels. Mater Trans 55(7):1069–1072. https://doi.org/10.2320/matertrans.M2014004
18. ASTM E140 (2019) Standard hardness conversion tables for metals relationship among Brinell hardness Vickers hardness, Rockwell hardness, superficial hardness, Knoop hardness, scleroscope
hardness, and Leeb hardness. https://doi.org/10.1520/E0140-12BR19E01

20. Thompson S (2018) Further observations of linear arrays of transition-iron-carbide precipitates in tempered 4340 steel. Metall Microstruct Anal 7:680–691. https://doi.org/10.1007/s13632-018-0492-8

21. Lee W, Su T (1999) Mechanical properties and microstructural features of AISI 4340 high-strength alloy steel under quenched and tempered conditions. J Mater Process Technol 87(1–3):198–206. https://doi.org/10.1016/S0924-0136(98)00351-3

22. Ajus C, Tavares SSM, Silva MR, Corte RRA (2009) Magnetic properties and retained austenite quantification in SAE 4340 steel. Revista Matéria 14(3):993–999. https://doi.org/10.1590/S1517-7076200900300011

23. Tavares SSM, Pardal JM, Souza JA, Pereira OC, Luz TS (2016) Failure of alloy steel socket-head cap screws used in offshore oil production. Eng Fail Anal 70:16–21. https://doi.org/10.1016/j.engfailanal.2016.07.004

Publisher's Note Springer Nature remains neutral with regard to jurisdictional claims in published maps and institutional affiliations.