

Influence of heat treatment conditions of Hardox 500 steel on its resistance to abrasive wear

Research Article

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Abstract: The type, magnitude, and distribution of thermal stress arising during the hardening process of steel primarily depend – apart from the geometry of the treated component and the type of metallic material used – on the type of cooling medium employed. The difference in cooling rates between the core and the surface of the object affects the extent of deformation, particularly in large-scale components with complex shapes. Due to their high hardness, boron-alloyed steels are frequently used for components exposed to abrasive wear, and their heat treatment is typically performed on profiled structural elements such as plowshares or cultivator coulters. The critical cooling rate is primarily dependent on the chemical composition of the material and increases as its hardenability index decreases. Consequently, it is essential to investigate the feasibility of applying various heat treatment processes to medium-carbon steels, including the widely used martensitic steel Hardox 500. This study presents the results of abrasive wear tests conducted on the aforementioned steel, subjected to different heat treatment variants, which involved varying cooling rates following austenitization. For this purpose, the cooling media used included deoxygenated quenching water, mineral transformer oil, synthetic quenching oil, and air-blast cooling. For comparative purposes, equilibrium-cooled samples were also analyzed. The research findings demonstrated that the highest resistance to abrasive wear was achieved by samples quenched in water, whereas a gradual decrease in wear resistance was observed in samples cooled in media with lower cooling intensity coefficients. Furthermore, it was found that the traditionally employed roughness parameters are not suitable for evaluating the condition of surfaces subjected to abrasion. Therefore, for a comprehensive assessment of the tribological resistance of Hardox 500 steel, more advanced surface topography analysis methods must be employed.

Keywords: Martensitic steel • Heat treatment • Abrasive wear

1. Introduction

Among steels intended for mass application, metallic materials with a homogeneous martensitic microstructure or those designed for quenching in the stamping process – often referred to as boron steels for heat treatment – exhibit the highest possible tensile strength (R_m) values. Despite the lack of available data in manufacturers' catalog specifications, sources [1,2] indicate that their tensile strength (R_m) can significantly exceed 2,000 Mpa. These mechanical properties, while maintaining satisfactory ductility parameters, contribute to high resistance to abrasive wear as well as the ability to withstand both static loads and impact energy absorption.

The group of martensitic steels represents the most extensive category of materials designed for abrasive wear conditions, encompassing types such as Hardox and Raex (SSAB-Oxelösund), Perdur and TBL (ThyssenKrupp Steel Europe AG), Brinar and Durostat (Grobblech GmbH), Miilux (Miilukangas Group), Relia and Creusabro (Industeel), Abrazo (TATA Steel Group), and Dillidur (Dillinger Hütte GTS) [3].

A common characteristic of low-alloy martensitic steels is the presence of a microalloying addition of boron, which significantly increases the hardenability. Comparatively, its content in steel at a level of 0.001–0.003% by weight provides a hardening intensity equivalent to 0.6% Mn, 0.7% Cr, 0.5% Mo, or 1.5% Ni by weight [4]. Therefore, it can be concluded that an increase in hardenability occurs at boron concentrations that are considered trace levels for other alloying

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elements [5–8]. The production of a family of steels characterized by the aforementioned mechanical properties is justified by their high hardenability, which promotes deformation. Moreover, Hardox 500 steel, which exhibits a tensile strength (R_m) of 1,653 Mpa [9], is also classified by the manufacturer as a steel with limited structural applications [10]. The authors' experience indicates that quenching higher grades of Hardox steel in water increases the likelihood of cold cracking. It should be noted that data regarding the quenching methods of PHS steels are not commercially available and must be verified by the component manufacturer. Therefore, further research is necessary to examine the retention of specific functional properties in this material group while reducing the cooling rate after austenitization.

However, it should be noted that an increase in steel hardenability entails the risk of technological challenges. The amount of quenching-induced stresses is determined by the type of cooling medium used during the quenching process. The difference in cooling rates between the core and the surface of the object affects the extent of deformation, particularly in large-scale components with complex shapes. For example, the heat treatment of boron-alloyed martensitic steels is performed on profiled structural components such as plowshares or cultivator coulters. The critical cooling rate depends on the chemical composition of the material and increases as hardenability (most commonly defined by the carbon content) decreases. The optimal cooling rate minimizes the occurrence of potential deformations and cracks in the treated material, which is crucial for maintaining the required dimensions and shapes of finished products.

Water remains the most commonly used quenching medium. However, as the hardenability of the material increases, it is recommended to use synthetic or mineral oil for cooling. According to Rudnik [11], the cooling rate in mineral oil is four times lower than that in water, averaging 150°C/s within the temperature range of 650–550°C. A milder cooling medium is also provided by compressed air blast cooling.

In the study by Luo and Bai [12], it was demonstrated that air-cooled, medium-carbon, low-alloy MnCrB cast steel with a chromium content of 0.6% by weight exhibits a tensile strength (R_m) exceeding 2,000 Mpa, ductility (A) of 4–6%, and impact toughness in the range of 111–164 J/cm². Its wear resistance is 30% higher than that of Hadfield steel in a three-body contact under a high load. In terms of scratch resistance and achieved hardness levels, cryogenic cooling of Hardox 500 steel appears to be a beneficial process, as this heat treatment condition resulted in a 27% increase in wear resistance compared to the as-delivered material. However, the negative aspects of cryogenic cooling must be considered, including the presence of residual stresses, which

necessitate slower cooling to mitigate the risk of cracking [13]. A relevant example is Creusabro steels, characterized by a complex bainitic-martensitic microstructure with retained austenite and finely dispersed carbides. This microstructure is achieved through an appropriately tailored chemical composition and a controlled, slow quenching process [14]. Moreover, Creusabro 8000 steel exhibits a similar relative abrasion resistance coefficient (k_r) compared to TBL PLUS and XAR 600 steels, despite having a hardness lower by 57 and 92 Brinell units, respectively [15]. According to George et al. [16], Usibor 1500P steel ($C = 0.22\%$ by weight), used for B-pillars in automotive applications where zones of lower hardness are required, develops a bainitic microstructure when cooled in a die at a rate of 30°C/s after preheating to 400°C. Based on continuous cooling transformation (CCT) diagrams, the same steel exhibits a fully martensitic microstructure at cooling rates of 45–2,200°C/s, whereas at a cooling rate of 25°C/s, 95% martensite and 5% bainite are observed [17]. For 22MnB5 steel, the resulting microstructure varies depending on the cooling medium: martensite forms during water quenching, tempered martensite and bainite develop during oil quenching, and a multiphase microstructure is obtained with air-blast cooling at a rate of 10°C/s [18,19]. According to equilibrium phase diagrams, Hardox 450 steel cooled at a rate of 10°C/s from 900°C does not exhibit ferrite precipitation; instead, the resulting structure consists of bainite and martensite [20].

For the study on the influence of cooling methods after austenitization on abrasive wear resistance, Hardox 500 steel was selected. The choice was motivated not only by its widespread industrial application but also by findings from previous field investigations [21,22], which indicated a tendency for accelerated abrasive wear in the vicinity of its welded joints [21]. Namely, field data have repeatedly shown that the heat-affected zones suffer from accelerated material loss under abrasive conditions, compromising component durability. For this reason, we aimed to reproduce microstructural variations – similar to those occurring within the heat-affected zone of welded joints – by applying different cooling rates after austenitization. By doing so, we can systematically induce and characterize the same microstructural heterogeneities. Unlike higher-grade steels such as Hardox 600 and Extreme [23,24], Hardox 500 is classified as bendable and weldable, which enhances its potential for industrial applications [25]. Hardox 500 is used in products such as liners, wear-resistant bars, cutting edges, and excavator buckets [26–29]. Its resistance to abrasive wear in soil environments may be comparable to that of Hardox Extreme steel [30]. According to Szala et al. [31], Hardox 500 also exhibits the lowest mass loss compared to S355JR, S355J2, and AISI304 steels in tribological tests involving garnet and silicon carbide (carborundum). Its properties

may also be more favorable than those of 38GSA steel (hardness of 548 HBW), which is commonly used in Poland for plowshares [9,32]. Furthermore, the wear intensity of Hardox 500 is 12% lower than that of carburized 20MnCr5 steel [33].

2. Materials and methodology

The study utilized 10 mm thick sheets of Hardox 500 steel, supplied by the authorized distributor, STAL-HURT. The chemical composition analysis was performed using a spectral method with a Leco GDS500A glow discharge emission spectrometer. The following parameters were applied to ionize the inert gas: $U = 1,250$ V; $I = 45$ mA; 99.999% argon. The obtained results represent the arithmetic mean of at least five measurements.

The CCT diagram was obtained through computer simulations carried out using JmatPro software. All heat treatment operations were performed in gas-tight chamber furnaces (FCF 12SHM/R) manufactured by Czylok, using a protective atmosphere of 99.95% argon. The parameters of the heat treatment procedures are presented in detail in Table 1.

For microscopic examination, light microscopy (LM) was performed using a Nikon Eclipse MA200 light microscope. The samples were etched with a 5% HNO₃ solution, in accordance with ASTM E407. A Nikon DS-Fi2 digital camera, coupled with the microscope, and NIS Elements software from Nikon were utilized for recording and analyzing the captured images. Additional microstructure and worn surface images of the samples were conducted using a Phenom XL scanning electron microscope, employing secondary electron imaging or backscattered electron imaging at an accelerating voltage of 15 keV.

Hardness measurements of the base material were performed using a Zwick/Roell ZHU 187.5 universal hardness tester, applying the Brinell method, in accordance with ISO 6506-1:2014. A 2.5 mm diameter carbide ball was used under a load of 187.5 kgf (1838.7469 N), applied for 15 s.

Laboratory tests of abrasive wear resistance were carried out using a T-07 abrasive wear resistance tester in the presence of loose abrasive media, in accordance with GOST 23.208-79, a standard analogous to ASTM G65. The primary difference between the T-07 tester and the device described in ASTM standard G65 is that in the T-07 device, the tested material is positioned horizontally rather than vertically. The T-07 tribotester consists of a steel roller with a rubber ring of 50 mm (+0.2 mm) diameter and 15 mm (−0.1 mm) width, an abrasive feeder that allows for regulating the abrasive flow, and a lever with weights, which generates a vertical force pressing the sample against the roller. The hardness of the rubber coating on

No	Heat treatment parameters
1	As-delivered condition from the steel mill
2	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 880°C, 20 min and cooling in H ₂ O (−270°C/s) Tempering: 100°C, 120 min, and air cooling
3	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 900°C, 20 min and cooling in transformer oil (−25°C/s) Tempering: 100°C, 120 min, and air cooling
4	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 900°C, 20 min and cooling in Durixol W72 (−100°C/s) Tempering: 100°C, 120 min, and air cooling
5	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 900°C, 20 min, and cooling with 5 bar air blast (−5°C/s) Tempering: 100°C, 120 min, and air cooling
6	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 900°C, 20 min, and cooling with 3 bar air blast (−3°C/s) Tempering: 100°C, 120 min, and air cooling
7	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 900°C and 20 min cooling with 1 bar air blast (−1°C/s) Tempering: 100°C, 120 min, and air cooling
8	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 900°C, 20 min, and air cooling (−0.1°C/s)
9	Normalization: 880°C, 30 min, and air cooling (−0.1°C/s) Quenching: austenitization at 900°C, 20 min, and furnace cooling (−0.01°C/s)

Table 1. Parameters of the applied heat treatment procedures.

the roller is within the range of 78–85 ShA. The tests were conducted under a constant load of $F = 44$ N ($\Delta F = 0.25$ N). The abrasive medium used was electrocorundum #90, in accordance with PN-M-59115:1976. The test duration depended on the hardness of the tested material and was set to 30 min (1,800 rotational cycles). The selection of this testing methodology was dictated by its practical relevance – the analyzed group of low-alloy martensitic steels is commonly used in excavator bucket components and agricultural machinery parts, which operate under intense abrasive conditions during ore extraction and soil mass processing.

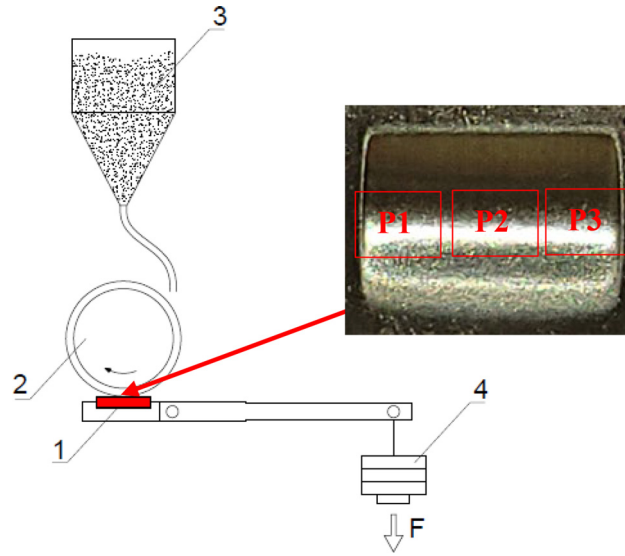


Figure 1. Schematic diagram of the T-07 tribotester. 1 – sample, 2 – rubber-rimmed steel wheel, 3 – abrasive, 4 – load, and P1, P2, and P3 – regions of samples subjected to surface topography evaluation.

The applied method has been confirmed to correlate with field tests [22,34]. Each test was performed in six repetitions, with sample dimensions of 30 mm × 30 mm × 3 mm. The relative abrasive wear resistance coefficient (k_b) was used as a measure of abrasion resistance, calculated according to equation (1).

$$k_b = \frac{Z_{ww}\rho_b N_b}{Z_{wb}\rho_w N_w}, \quad (1)$$

where k_b is the relative abrasive wear resistance coefficient [–], Z_{ww} is the mass loss of the reference material [g], Z_{wb} is the mass loss of the tested material [g], N_w is the number of counter-sample rotations during the reference material test, N_b is the number of counter-sample rotations during the tested material test, and ρ_w and ρ_b are densities of the reference and tested materials [g/cm³], respectively.

The reference sample was Hardox 500 steel, used in the as-delivered condition. Figure 1 presents the schematic diagram of the T-07 device, along with the positioning of the tested material sample.

For the quantitative evaluation of sample surfaces, a HITACHI TM-3000, the second scanning electron microscope

was used along with specific graphic-analytical software. Before roughness evaluation, the sample surfaces were cleaned with dry compressed air, inspected for mechanical damage, and marked at three locations: P1, P2, and P3 (Figure 1), where surface evaluation procedures were to be performed. All surface evaluation procedures were conducted at the same magnification, maintaining ten times the elementary segment length $l_r = 198.03 \mu\text{m}$, i.e., $1980.26 \mu\text{m}$, and with identical electron beam settings of 15 kV, ensuring precise observation and recording of surface details of steel samples. Before testing, a working distance (WD) calibration procedure was carried out using a roughness standard with a predefined profile.

The following roughness parameters were analyzed:

R_a – arithmetic mean deviation of the profile, R_p – height of the highest peak in the profile, and R_v – depth of the deepest valley in the profile.

Statistical analyses were performed using Statistica version 13. The key assumptions required for ANOVA are normality of distribution and homogeneity of variance (Table 2). In cases where these assumptions were not met, this did not necessarily prevent the use of parametric tests, as Lindman [35,36] demonstrated that the F statistic is robust to violations of variance homogeneity. Levene's

	Effect SS	Effect df	Effect MS	Error SS	Error df	Error MS	F	p
Mass wear per 1 m of sliding distance	0.002684	8	0.000335	0.003295	31	0.000106	3.156335	0.00987

Table 2. Results of Levene's test for homogeneity of variance.

C	Mn	Si	P	S	Cr	Ni	Mo	V	Cu	Al	Ti	Nb	B
0.29	0.74	0.28	0.007	0.001	0.61	0.06	0.018	0.012	0.010	0.054	0.003	—	0.0009

Table 3. Chemical composition of Hardox 500 steel (in % by weight).

test was used to assess the homogeneity of variance, while Shapiro–Wilk’s test was applied to evaluate the normality of distributions. All datasets satisfied the normality assumption except those obtained after 5 bar air cooling. To determine which means differed significantly, Duncan’s *post hoc* test was used. For statistical analysis, mass wear was normalized per 1 m of friction distance. For all roughness parameters, normal distributions with homogeneous variance were obtained.

3. Research results

3.1 Chemical composition

Based on the results, Hardox 500 steel can be classified as a medium-carbon steel ($C = 0.29\%$ by weight), with manganese (0.74% by weight), chromium (0.61% by weight), and a microalloyed addition of boron (0.0009% by weight) as the primary elements enhancing its hardenability (Table 3). Additionally, the low content of harmful elements such as phosphorus (P)

and sulfur (S) ensures the retention of high mechanical properties. The microalloying additions of aluminum and titanium are essential in low-alloy boron martensitic steels, as the production process of these steels must follow a specific sequence. Boron can only be introduced into the melt after it has been deoxidized and degassed using aluminum and titanium. These elements bind oxygen and nitrogen into intermetallic phases, preventing the formation of BN and B_2O_3 , which would otherwise reduce boron’s effectiveness in enhancing the hardenability [37]. The addition of niobium is also beneficial, as it prevents the formation of micrometric titanium-nitrogen compounds [38]. However, in the case of Hardox 500 steel, no detectable presence of niobium was recorded.

Figure 2 presents a CCT diagram for Hardox 500 steel, obtained through computer simulations. The assigned transformation temperatures for individual phases and structural constituents are as follows: ferrite – 795°C , pearlite – 736°C , bainite – 576°C , martensite start – $M_s = 366^\circ\text{C}$, 50% martensite – 331°C , and 90% martensite – 252°C . Particular attention should be paid to the relatively rapid formation of ferrite, which occurs concurrently with the onset of

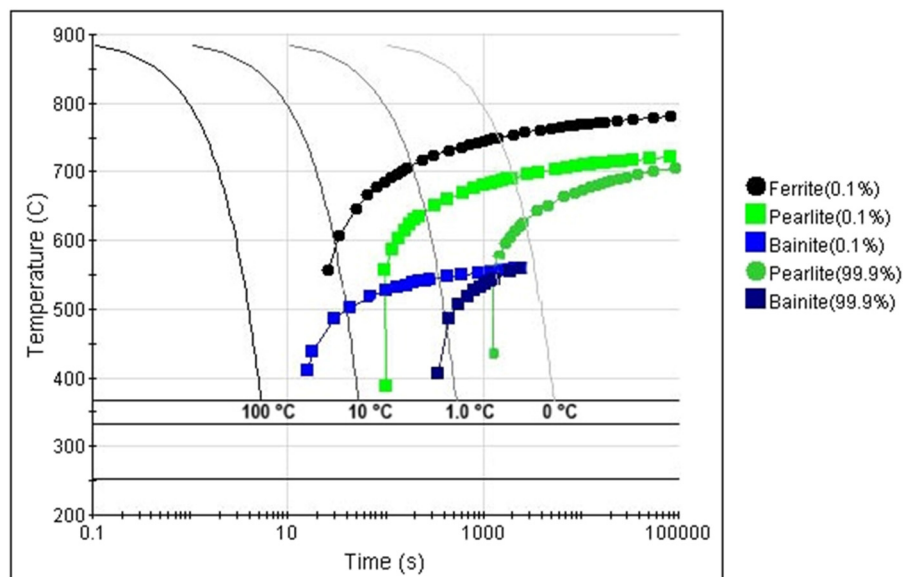


Figure 2. Time–temperature graph for Hardox 500 steel. Assigned temperatures for individual transformations, phases, and components of the structure: pearlite – 736°C , ferrite – 795°C , bainite – 576°C , martensite (50%) – 331°C , martensite (90%) – 252°C , and M_s – 366°C .

bainitic transformation, as well as to the predicted presence of bainite even at a slow cooling rate, indicating the strong stability of this phase under low cooling intensity.

3.2 Microstructure and hardness

The hardness measurements of Hardox 500 steel (Figure 3) showed that in the as-delivered condition, it exhibited an average hardness of 440 HBW. Considering the data provided by the manufacturer (470–530 HBW) and the experimentally determined mechanical parameters of the tested steel, it can be observed that the obtained hardness index corresponds to only 77–95% of the hardness declared by the manufacturer. To evaluate the effect of cooling rate on mechanical properties, the steel was first normalized, aiming to refine the microstructure and eliminate the influence of multiple recrystallizations and phase strain in steel sheets obtained through thermomechanical rolling. The study results indicate that the highest hardness levels were achieved after quenching in synthetic oil Durixol W72 (493 HBW), water (487 HBW), and transformer oil (479 HBW) (corresponding heat treatment process parameters are presented in Table 1, entries 4, 2, and 3). In all three cases, the hardness exceeded 470 HBW, which is the minimum value specified by the manufacturer. The use of a higher austenitization temperature before oil quenching is justified by the slower cooling rate of this medium. It is also worth noting that achieving high strength

parameters in heat-treated samples under the described conditions requires performing normalizing annealing in a slightly lower temperature range than the hardening process. A gradual decrease in hardness was observed with a reduction in the cooling rate, achieved by using compressed air at progressively lower pressures, ultimately reaching 253 HBW. The lowest hardness value (135 HBW) was recorded after furnace cooling. Interestingly, this value was lower than the expected hardness based on carbon content (0.29% by weight), which should be approximately 156 HBW. This places it between the hardness of normalized and fully annealed samples examined in this experiment. Further discussion will focus on the analysis of microstructural properties.

Figures 4–12 present the microstructures of Hardox 500 steel in the as-delivered condition and those obtained after cooling at different rates. In the as-delivered state (Figure 4a and b), Hardox 500 steel exhibits a lath martensite microstructure, characterized by a three-level structural hierarchy, organized into laths, blocks, and packets. The martensite laths within a single block share the same crystallographic orientation, meaning they represent the same martensite variant. In turn, packets consist of blocks with an identical habit plane, corresponding to the {111} plane of the prior austenite [39–41]. A distinguishing feature is the presence of tempered martensite regions, where coalescence occurs due to the steel's lower tendency for spontaneous tempering processes. Coalescence results from the fact that martensite blocks sharing the

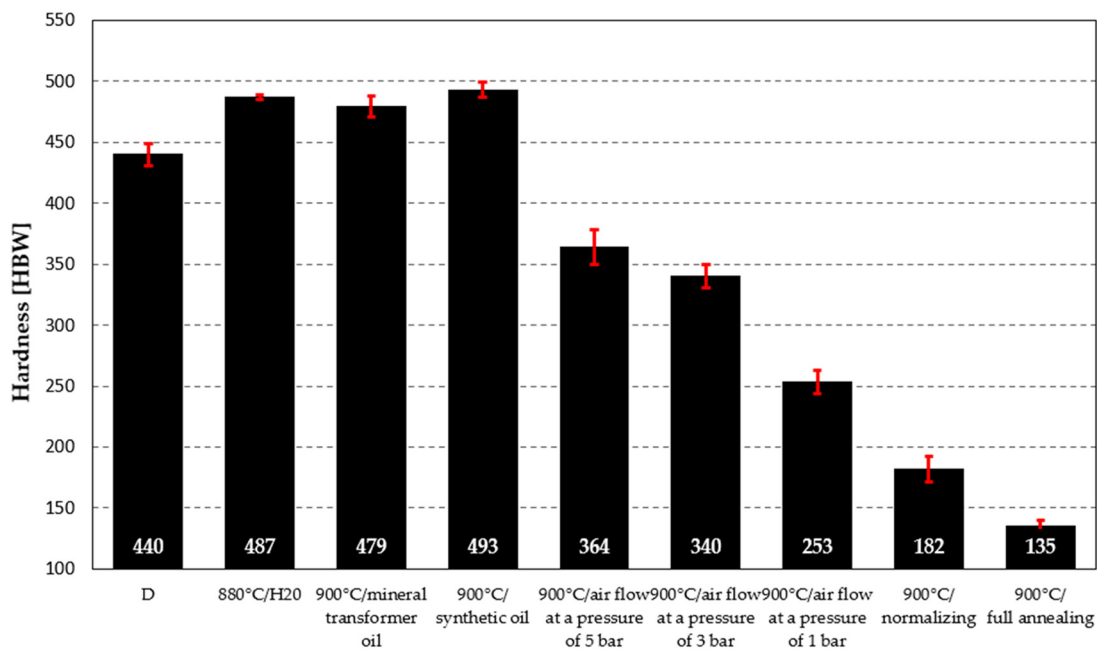


Figure 3. Hardness measurement results of Hardox 500 steel under different heat treatment conditions.

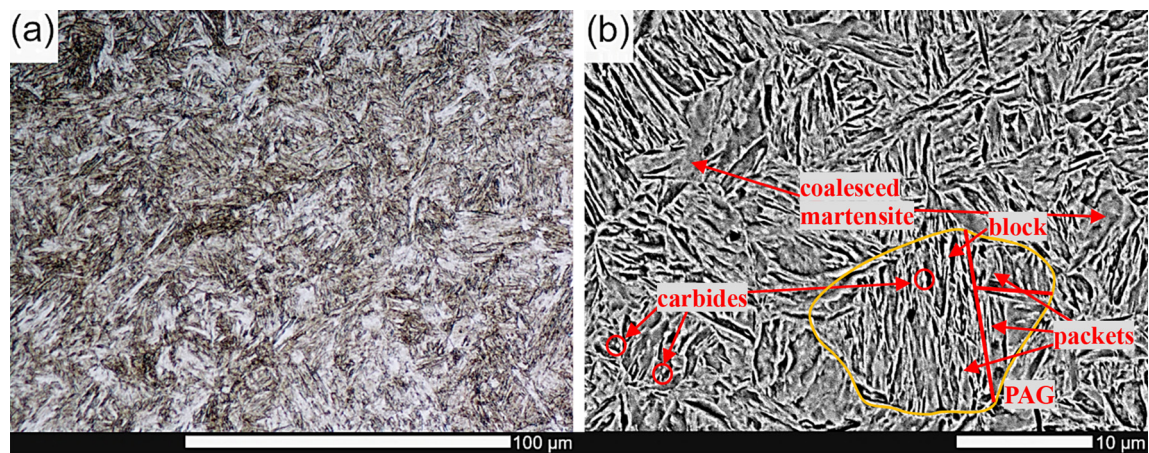


Figure 4. Microstructure of Hardox 500 in the as-delivered condition and etched with 5% HNO_3 : (a) LM and (b) SEM.

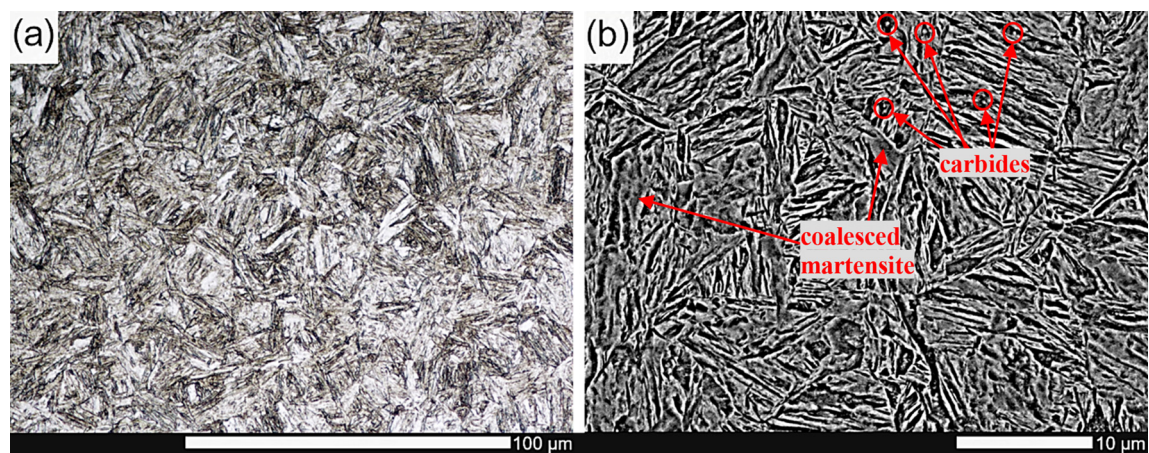


Figure 5. Microstructure of Hardox 500 after water cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

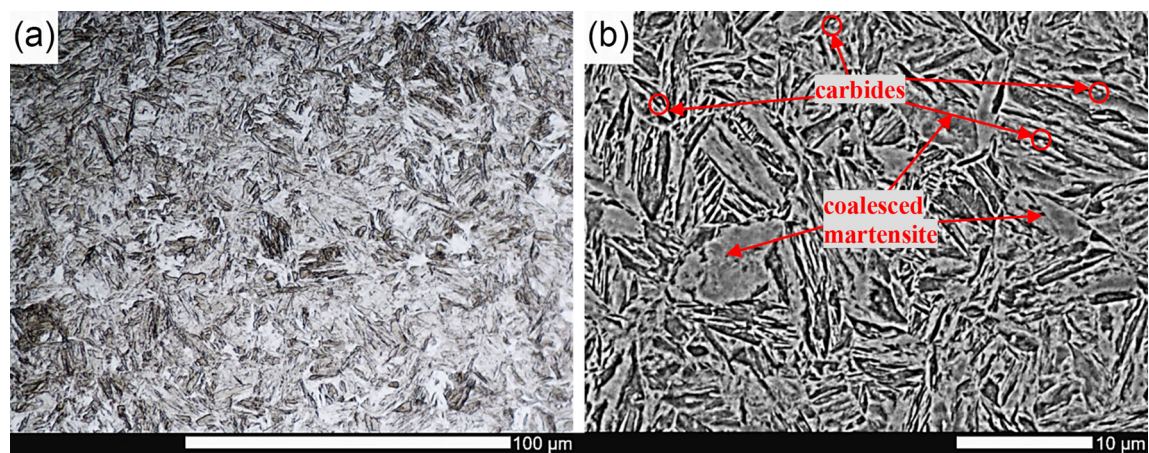


Figure 6. Microstructure of Hardox 500 after mineral oil cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

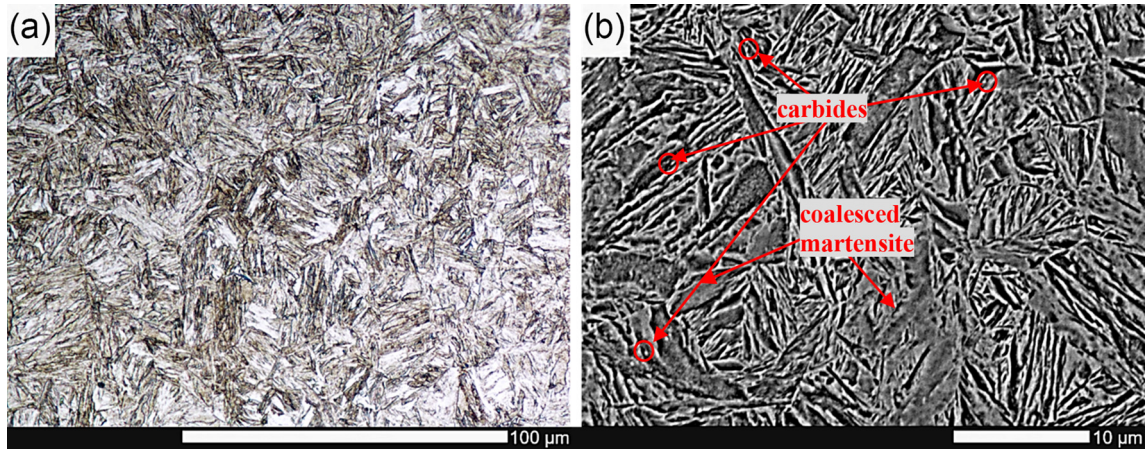


Figure 7. Microstructure of Hardox 500 after synthetic oil cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

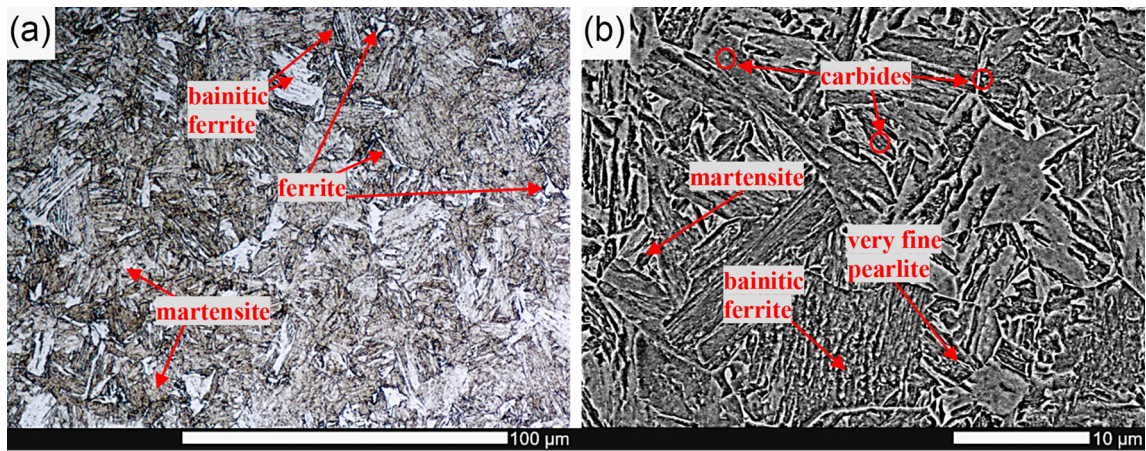


Figure 8. Microstructure of Hardox 500 after air at 5 bar pressure cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

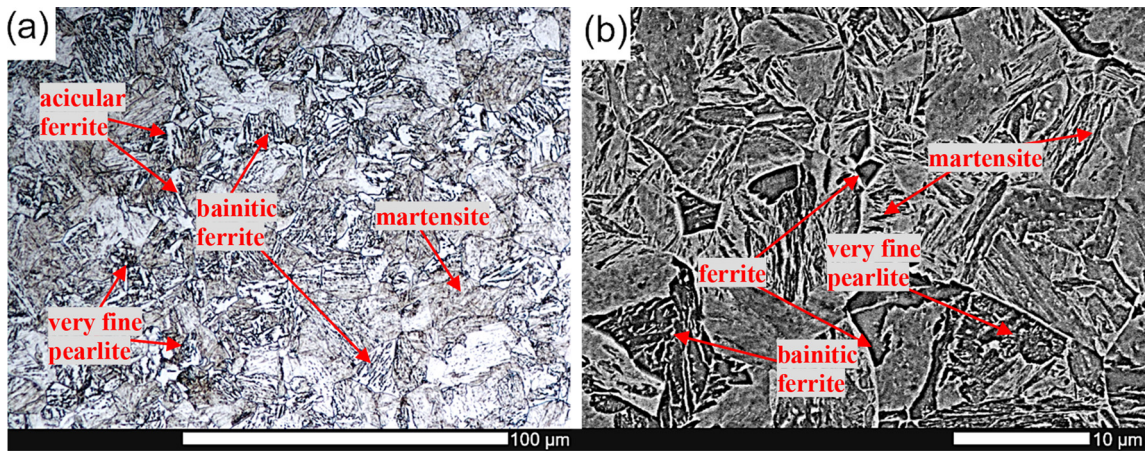


Figure 9. Microstructure of Hardox 500 after air at 3 bar pressure cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

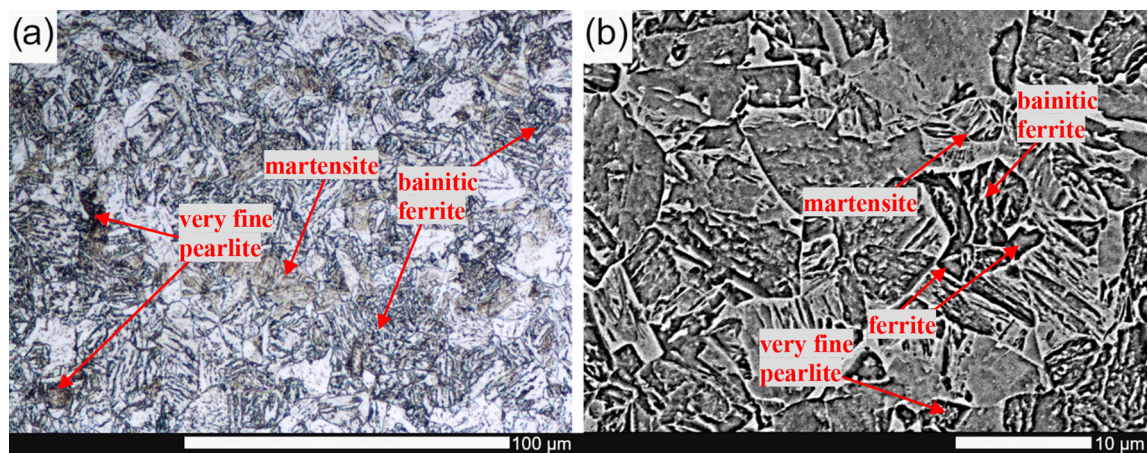


Figure 10. Microstructure of Hardox 500 after air at 1 bar pressure cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

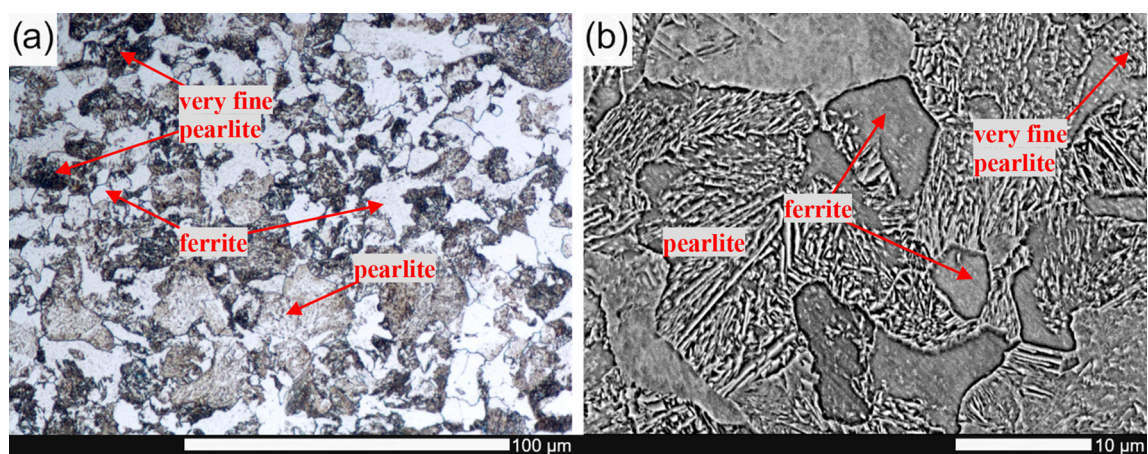


Figure 11. Microstructure of Hardox 500 after air cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

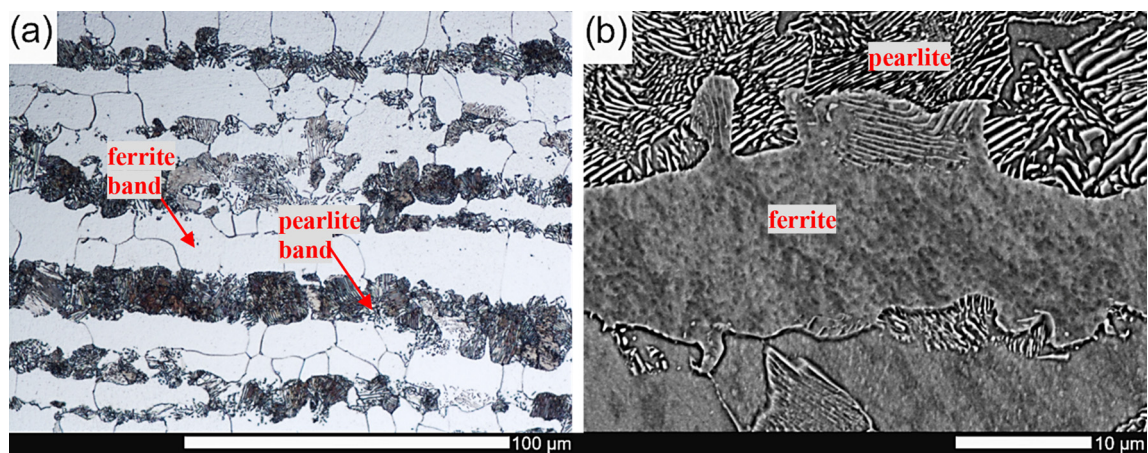


Figure 12. Microstructure of Hardox 500 after furnace cooling and etched with 5% HNO_3 : (a) LM and (b) SEM.

same habit plane and having similar crystallographic orientations relative to the prior austenite tend to merge without the involvement of intermediate phases, leading to the formation of thicker structures [42–46]. Additionally, some areas exhibit higher etching susceptibility, suggesting the presence of fine-dispersed phases such as lower bainite or tempered martensite. The fine-lath microstructure is distinguished by the presence of brighter bands, an effect of the thermomechanical processing performed by the manufacturer. Similar morphological features are observed in the microstructures of Hardox 500 in the as-delivered condition after cooling in water, mineral oil, and synthetic oil (Figures 5–7). SEM analysis revealed that in the case of synthetic oil-quenched Hardox 500, a limited number of martensitic regions undergoing coalescence were likely observed, as indicated by the presence of thickened lath areas. For microstructures obtained after compressed air cooling, it was found that with decreasing air pressure, the proportion of other structures, such as very fine pearlite, increased, while the fraction of martensitic regions decreased (Figures 8–10). Additionally, the presence of large bainitic ferrite (upper bainite) and acicular ferrite plates was noted. However, it should be emphasized that microstructures formed after compressed air cooling exhibit significantly higher tempering resistance than conventional hardened microstructures. Furthermore,

the presence of very few regions with acicular structural features indicates that substantial undercooling was still present. After normalization (Figure 11a and b), the microstructure mainly consists of nonequilibrium polygonal ferrite grains and pearlite (quasi-pearlite). Essentially, martensite is no longer observed. After furnace cooling, a banded ferritic-pearlitic microstructure was obtained (Figure 12a and b). The banded structure results from the initial hot rolling stage. Unlike steels produced by conventional methods, this rolling is carried out at relatively high temperatures. Consequently, Hardox steels often exhibit carbon segregation (dendritic segregation), which is not eliminated during subsequent processing stages. This ferritic-pearlitic microstructure is typical for low- and medium-strength steels that have undergone controlled furnace cooling, allowing for the gradual phase transformation of austenite into ferrite and pearlite.

3.3 Abrasive wear testing

Abrasive wear resistance tests (Figure 13) conducted in the presence of loose abrasive media revealed that Hardox 500 steel quenched in water exhibited the highest tribological resistance, with a relative wear resistance coefficient of $k_b = 1.05$. For samples cooled in mineral oil and synthetic oil, the k_b values decreased to 0.98 and 0.96. Additionally, the presence of large bainitic ferrite (upper bainite) and acicular ferrite plates was noted. However, it should be emphasized that microstructures formed after compressed air cooling exhibit significantly higher tempering resistance than conventional hardened microstructures. Furthermore,

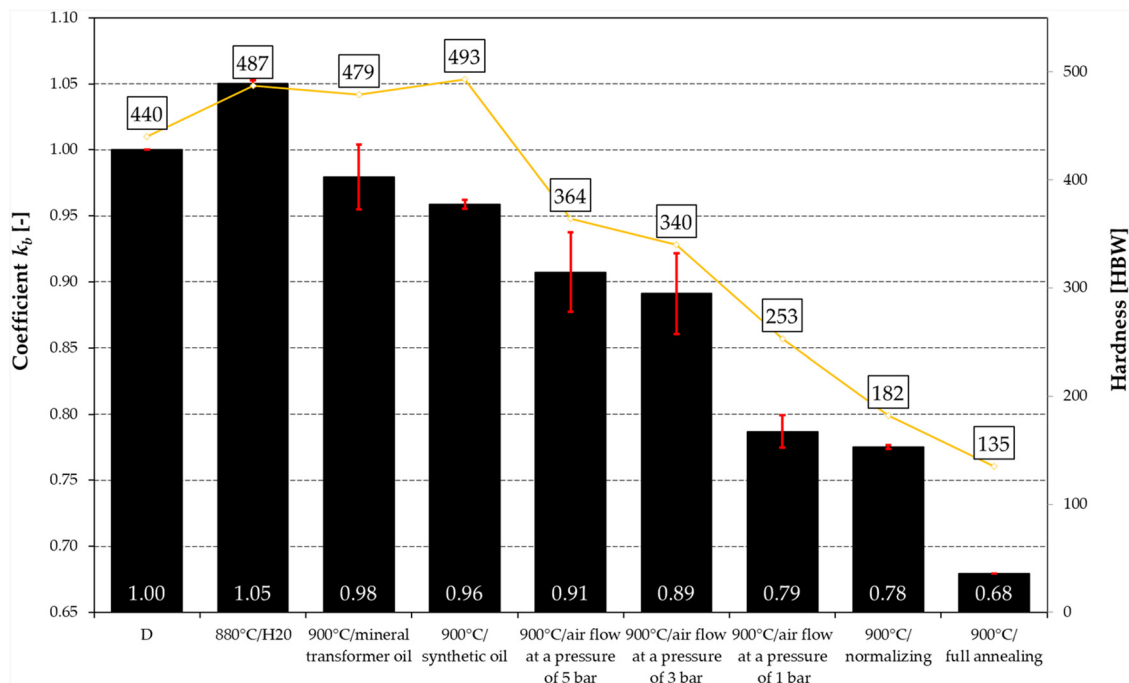


Figure 13. Relative abrasive wear resistance coefficient k_b and hardness of Hardox 500 steel under different heat treatment conditions.

	Effect SS	Effect df	Effect MS	Error SS	Error df	Error MS	<i>F</i>	<i>p</i>
Mass wear per meter of sliding distance	0.5748	8	0.0718	0.0157	31	0.0005	141.9316	0.0000

Table 4. Results of variance analysis.

respectively, which are lower than the reference value obtained for Hardox 500 in its as-delivered condition. A gradual decline in tribological resistance was observed as compressed air was used as the cooling medium, with lower air pressures leading to further reductions. Reducing the pressure to 1 bar resulted in only 78% of the wear resistance observed in the as-delivered state. Interestingly, normalization, characterized by slow air cooling, did not significantly reduce the k_b value compared to the previous condition. The lowest abrasive wear resistance was observed in Hardox 500 after full annealing. From a microstructural perspective, the progressive decrease in the relative wear resistance coefficient is associated with a declining fraction of martensite/lower bainite and other hardened structures, while the amount of ferrite increases as the cooling rates decrease. ANOVA (Table 4) confirmed that different heat treatment conditions resulted in statistically significant variations in wear resistance. To determine whether these differences were statistically significant, a Duncan post-hoc test (Table 5) was performed.

The results for the reference condition (as-delivered state) did not differ statistically from those obtained after quenching in water or mineral oil. However, for other heat treatment conditions, significant differences in wear resistance coefficients were observed. Notably, when considering only cooling methods that resulted in a hardened microstructure (water and oil quenching), Duncan's test did not show significant differences in wear resistance. Similarly, no significant differences were observed between samples

cooled in synthetic oil and those cooled with compressed air at 5 bar, between samples cooled with compressed air at 5 and 3 bar, and between samples cooled with compressed air at 1 bar and normalized samples. For furnace-cooled samples, which exhibited the lowest wear resistance, *p*-values ($p < 0.05$) confirmed statistically significant differences compared to all other cooling methods.

Figure 14 illustrates the dependence of mass loss on material hardness (HBW). Linear regression of the experimental dataset yielded a coefficient of determination (R^2) of approximately 0.92, confirming a strong linear correlation between these variables. Using hardness as a predictor of the durability of structural materials is particularly attractive, owing to the simplicity and speed of hardness-testing methods. Stachowiak's idealized single-particle abrasive wear model explicitly incorporates the material's surface hardness (H) and describes wear as the action of an individual abrasive grain [47,48]. His model rests on two fundamental assumptions: first, an abrasive particle penetrates to a depth governed by the surface hardness and the applied normal load (P) and second, the removed material volume equals the groove volume generated by that particle on the surface. Hardness-based wear predictions have also been advanced by Khrushchev and Babichev, who demonstrated a proportional relationship between hardness and wear resistance for pure metals and unalloyed steels in the annealed condition [49]. Likewise, Archard's sliding-wear equation (equation (2)) incorporates hardness, defining the wear volume as

State of heat treatment	{1} $M =$ 0.7908	{2} $M =$ 0.7911	{3} $M =$ 0.8231	{4} $M =$ 0.8388	{5} $M =$ 0.8838	{6} $M =$ 0.9098	{7} $M =$ 1.0293	{8} $M =$ 1.0449	{9} $M =$ 1.1922
1		0.9867	0.0595	0.0075	0.0000	0.0000	0.0000	0.0000	0.0000
2	0.9867		0.0499	0.0064	0.0001	0.0000	0.0000	0.0000	0.0000
3	0.0595	0.0499		0.3235	0.0008	0.0001	0.0000	0.0000	0.0000
4	0.0075	0.0064	0.3235		0.0074	0.0002	0.0001	0.0000	0.0000
5	0.0000	0.0001	0.0008	0.0074		0.1088	0.0001	0.0001	0.0000
6	0.0000	0.0000	0.0001	0.0002	0.1088		0.0001	0.0001	0.0001
7	0.0000	0.0000	0.0000	0.0001	0.0001	0.0001		0.3292	0.0001
8	0.0000	0.0000	0.0000	0.0000	0.0001	0.0001	0.3292		0.0001
9	0.0000	0.0000	0.0000	0.0000	0.0000	0.0001	0.0001	0.0001	

Table 5. Results of Duncan's test.

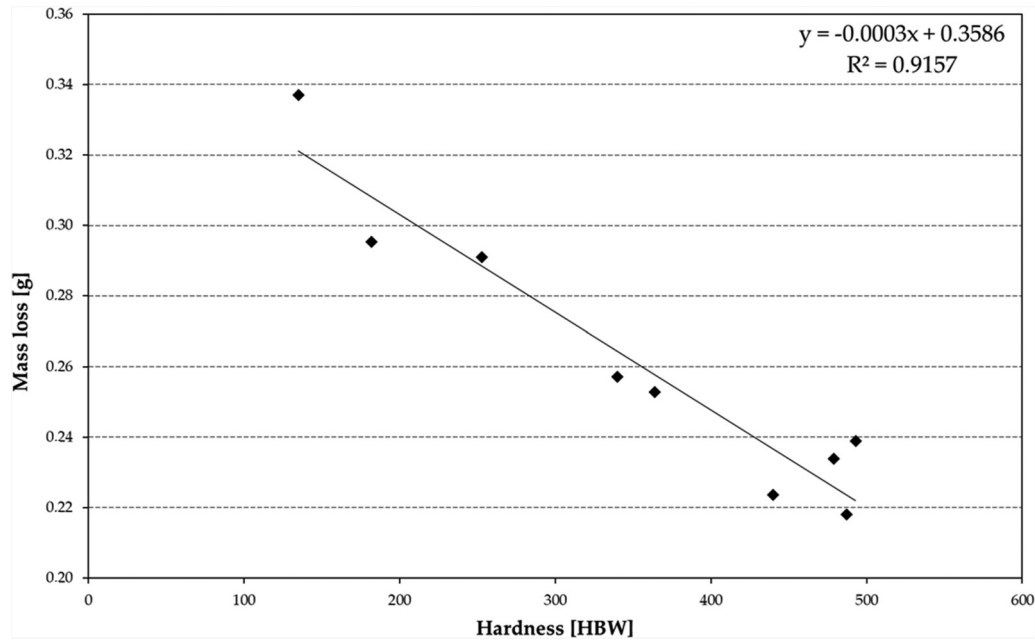


Figure 14. Effect of hardness on the mass loss of Hardox 500 steel under different heat treatment conditions.

the ratio of the product's wear coefficient (k), sliding distance (s), and load (P) to the hardness (H). This formulation is widely used to correlate analytically estimated wear resistance with that observed under real service conditions [27]:

$$I_z = \frac{k \cdot P \cdot l}{H}, \quad (2)$$

where I_z is the volumetric wear loss (m^3), k is the wear coefficient, P is the normal load (N), l is the sliding distance (m), and H is the hardness of the softest contacting surfaces (N/m^2).

In light of this finding, the Archard wear model was subsequently applied to quantitatively predict mass loss as a function of hardness under various heat-treatment regimes, thereby enabling a comparison between model forecasts and measured wear data. Based on the dataset collected for nine different states of heat treatment – each with known mass loss (converted to volumetric loss assuming a steel density of 7.8 g/cm^3), applied normal load ($P = 44$), and sliding distance ($l = 282.74 \text{ m}$) – a representative value of the wear coefficient k was determined. For this purpose, the average value of k was calculated across states 1–6 and equalled approximately 0.010272 and calculated across states 7–9 equalled approximately 0.005787. These values were then used to estimate the volumetric wear for each sample using Archard's equation. Determination of the two separate wear

coefficients k was driven by the distinct behavior of the slowest-cooled samples, in which abrasive wear mechanisms connected with plastic deformation became dominant.

The calculated wear volumes were compared with the experimentally determined values (obtained from measured mass loss), revealing satisfactory agreement for materials with the highest hardness levels (Table 6). The experimentally measured volumetric wear losses (I_{exp}) were compared with those predicted by Archard's model (I_z) using averaged wear coefficients (k). For the highest hardness states 1–4, predictions deviated by no more than $\pm 12\%$ from the measured values (errors ranging from -11.8 to $+5.6\%$), demonstrating that, in microstructures dominated by fine martensitic laths, hardness alone serves as a reliable first-order indicator of abrasive wear resistance. However, for the intermediate-hardness states 5–6, the model overestimated material loss by $+12.9$ and $+18.7\%$, respectively; this suggests that bainitic constituents' features were not captured by hardness, imparting additional resistance to wear. The greatest discrepancies appear in the lowest-hardness samples 7–9, where, even after calibrating a lower k for ductile microstructures, predicted wear errors span -21.7 to $+26.6\%$. These results indicate that a transition in dominant wear mechanisms – from predominantly micro-cutting into microplowing with plastic deformation – cannot be encompassed by a single, hardness-based parameter. These findings imply that Archard's model should be applied using a variable wear coefficient k that reflects the material's

microstructure, particularly for more ductile steels, in which wear mechanisms differ fundamentally from those governing abrasion in hard, quenched martensitic steels.

The surface analysis of worn samples, conducted using a scanning electron microscope, provided valuable insights into the micromechanisms of abrasive wear in the examined materials (Figures 15 and 16). The surfaces of Hardox 500 steel in the as-delivered condition and after cooling in water and oils exhibited similar morphological features (Figure 15a–d). The dominant wear characteristics included grooves and scratches aligned with the movement direction of the loose abrasive medium. Between the long scratches, only short, fine scratches oriented at different angles to the wear direction were observed, indicating localized variations in the nature of abrasive interactions. At the edges of these grooves, material buildup was evident, resulting from prior plastic deformation. This material, subjected to cyclic interaction of abrasive particles, was gradually detached, further intensifying the wear process. In addition to microplowing, microcutting played a significant role in the wear mechanism, occurring without significant plastic deformation of the material. In some cases, chips were observed at the edges and ends of the grooves, resulting from material shearing by abrasive particles (Figure 15c and f). The identified micromechanisms of wear – microplowing, microcutting, and asperity shearing – acted synergistically, leading to the progressive removal of material from the sample surfaces (Figure 15a–e). Among these mechanisms, microcutting is a more aggressive wear process than microplowing, which initially induces plastic deformation in the surface layer. Material detachment from the top layer may occur only after repeated passes of abrasive particles. Additionally, during the interaction of abrasive grains with the metal, the bottom of

the grooves undergoes work hardening, increasing the material's resistance to further wear. At lower cooling rates, in addition to long parallel grooves and scratches, shorter but deeper scratches appeared, oriented at different angles to the abrasive grain movement direction (Figure 15g–i). These effects led to material spallation, resulting in deep pits and significant mass loss. The spalled material exhibited irregular shapes, likely caused by the abrasive grains penetrating the top metal layer (Figure 15g–i). These grains were subjected to repeated interactions with other abrasive particles and were eventually pulled out along with fragments of the material. Around the spallation sites, plastically deformed regions were observed. These phenomena were particularly intense in samples cooled with compressed air at 1 bar, normalized, and fully annealed (Figure 15g–i) and were associated with the presence of soft ferritic phases in the microstructure. In complex microstructures containing phases with significantly different hardness, abrasive wear initiates in the softer phase. This phase gradually wears away, exposing regions containing the harder phase. As wear progresses, the hard phase begins to protrude above the surface, forming a natural protective barrier that partially shields the softer phase from further wear. This process continues until the hard phase protrudes sufficiently for it to be removed by the abrasive particles. For samples cooled at lower rates, the dominant wear mechanisms were material spallation leading to deep pits, grooving combined with plastic deformation, and wear caused by cyclic particle interaction.

Figure 17 presents cross-sections of selected heat treatment conditions, including the as-delivered condition, mineral oil quenching, air cooling at 1 bar pressure, and furnace cooling. On the sample surfaces, pits and groove

State of heat treatment	Actual mass consumption (g)	Actual volumetric wear loss I_{exp} (m ³)	Wear coefficient k determined empirically	Wear coefficient k used in the Archard wear model	Theoretical volumetric wear loss I_z (m ³)	Relative difference (%)
1	0.2236	2.84841×10^{-8}	0.009880	0.010272	3.00644×10^{-8}	+5.55
2	0.218	2.77707×10^{-8}	0.010666	0.010272	2.71511×10^{-8}	–2.23
3	0.23884	2.97898×10^{-8}	0.011578	0.010272	2.76135×10^{-8}	–7.32
4	0.23385	3.04255×10^{-8}	0.011490	0.010272	2.68310×10^{-8}	–11.82
5	0.25272	3.21936×10^{-8}	0.009238	0.010272	3.63384×10^{-8}	+12.87
6	0.25722	3.27669×10^{-8}	0.008781	0.010272	3.89106×10^{-8}	+18.72
7	0.29104	3.70752×10^{-8}	0.007391	0.005787	2.90297×10^{-8}	–21.69
8	0.29543	3.76348×10^{-8}	0.005400	0.005787	4.03325×10^{-8}	+7.17
9	0.3371	4.29427×10^{-8}	0.004570	0.005787	5.43758×10^{-8}	+26.59

Table 6. Mass consumption and volumetric wear loss determined experimentally and predicted by the Archard model.

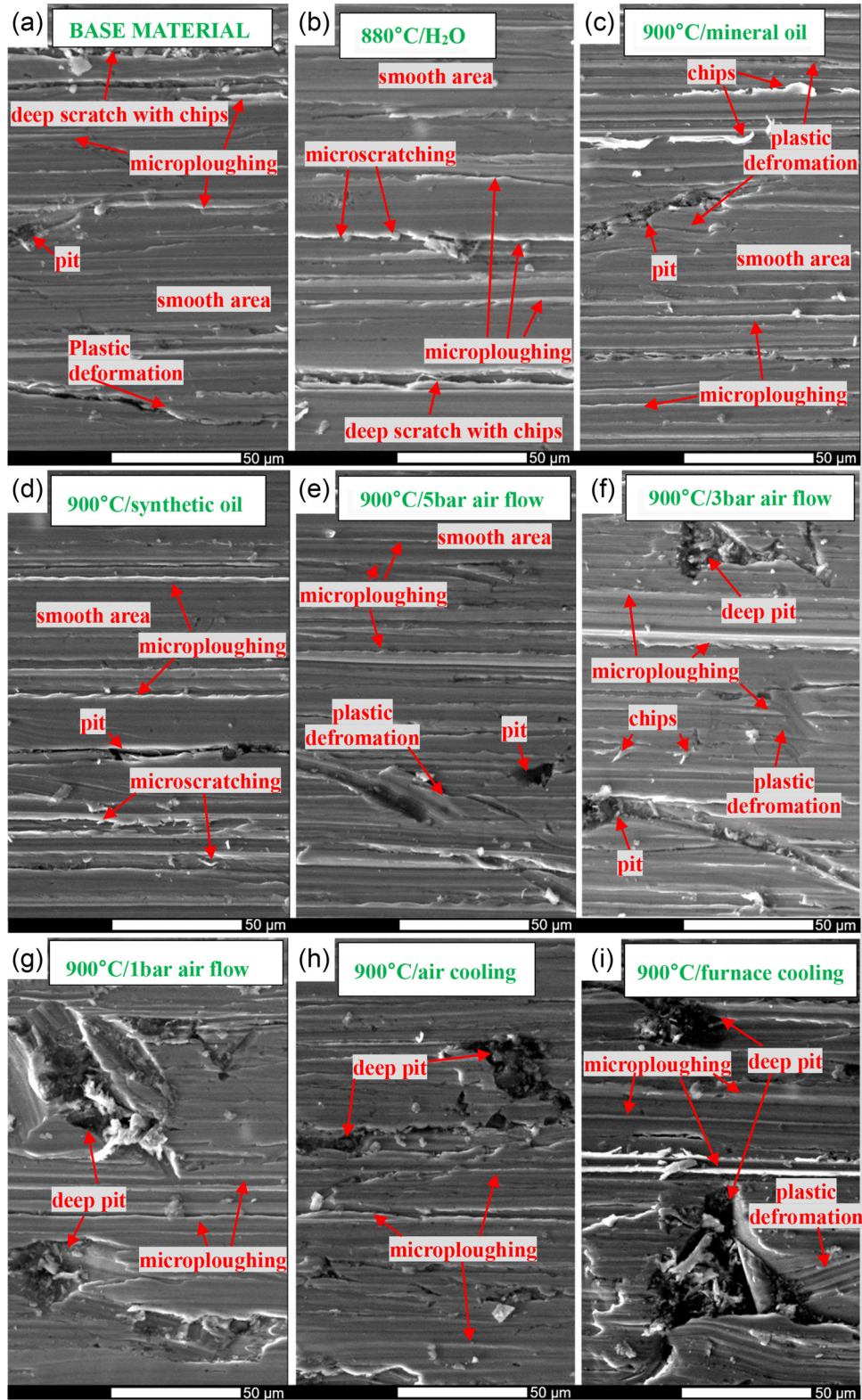


Figure 15. SEM analysis under unetched conditions of surfaces of Hardox 500 steel subjected to abrasive wear testing under different heat treatment conditions: (a) as-delivered condition, (b) after water cooling, (c) after mineral oil cooling, (d) after synthetic oil cooling, (e) after air cooling at 5 bar pressure, (f) after air cooling at 3 bar pressure, (g) after air cooling at 1 bar pressure, (h) after air cooling, and (i) after furnace cooling.

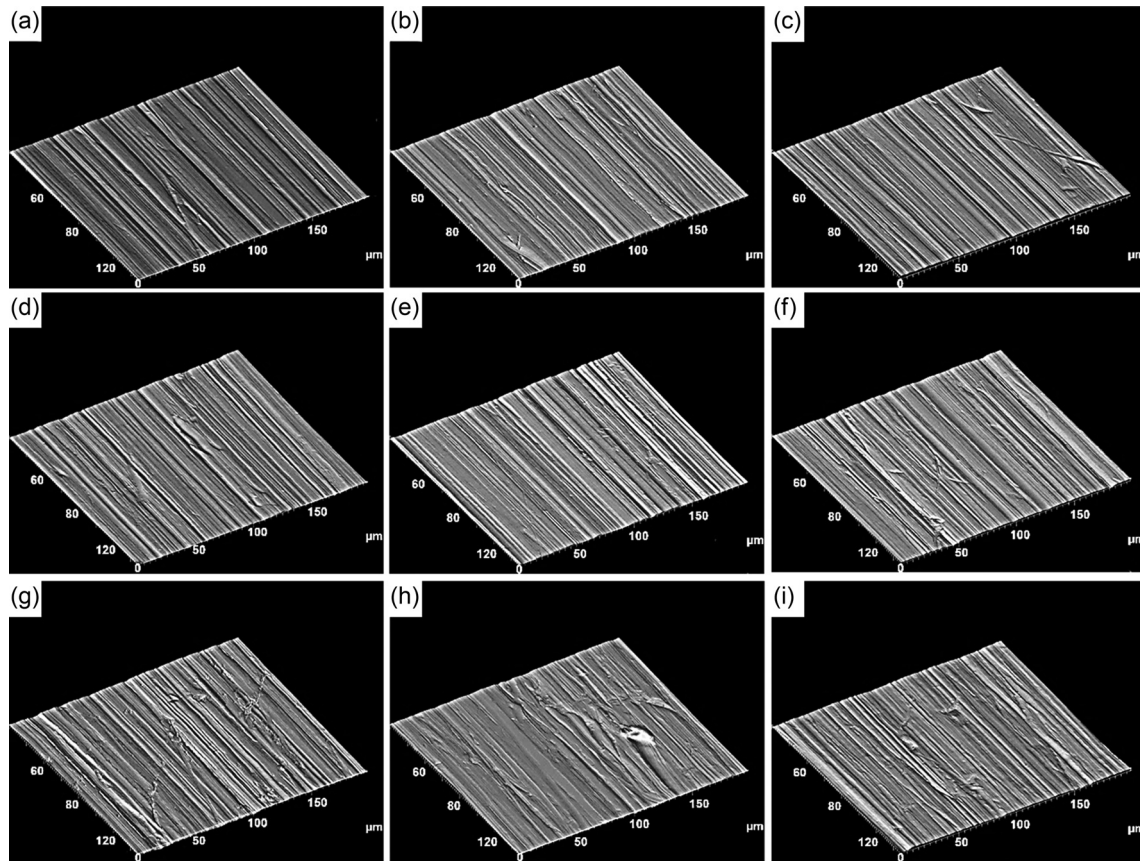


Figure 16. 3D images obtained by SEM analysis of sample surfaces subjected to wear testing along the longitudinal direction of abrasive movement: (a) as-delivered condition, (b) after water cooling, (c) after mineral oil cooling, (d) after synthetic oil cooling, (e) after air cooling at 5 bar pressure, (f) after air cooling at 3 bar pressure, (g) after air cooling at 1 bar pressure, (h) after air cooling, and (i) after furnace cooling.

bottoms of varying sizes are clearly visible (Figure 17a). Only small areas of plastically deformed material were observed around these features. However, the near-surface layer did not exhibit significant signs of plastic deformation. In the sample shown in Figure 17c, which has a complex multiphase microstructure, different structural components (bainite, ferrite, and martensite) wear at a similar rate, with pits distributed uniformly across these phases. In contrast, for the furnace-cooled sample (Figure 17d), deep pits predominantly form within the ferritic phase. This confirms the described wear mechanism of microstructures with components of differing microhardness, where the softer phase undergoes wear more intensively. This process leads to selective material degradation, contributing to localized pits and surface irregularities. Consequently, differences in the microhardness of individual phases significantly influence the service life of working elements made from the tested material (Table 7).

Due to significant differences in the surface condition of the worn samples observed depending on the heat treatment state, the following roughness parameters were measured: R_a , R_p , and R_v , and their values were subjected to statistical analysis (Figure 18). The surface morphology evolved significantly depending on the cooling medium, which influenced the mechanism and intensity of abrasive wear. In the as-delivered condition, the surface exhibited a moderate roughness profile ($R_a = 0.41 \mu\text{m}$), with shallow valleys ($R_v = 1.62 \mu\text{m}$), whereas water quenching resulted in visibly increased surface irregularities, with deeper and more frequent grooves. Cooling in mineral and synthetic oil led to a noticeable improvement in surface smoothness, with reduced groove depth. Similarly, compressed air cooling at 5 bar pressure preserved relatively smooth surface features while reducing the pressure to 1 bar produced the most deteriorated surface profile ($R_a = 0.59 \mu\text{m}$; $R_v = 3.07 \mu\text{m}$), characterized by deep grooves and numerous pits. This is consistent with the wear

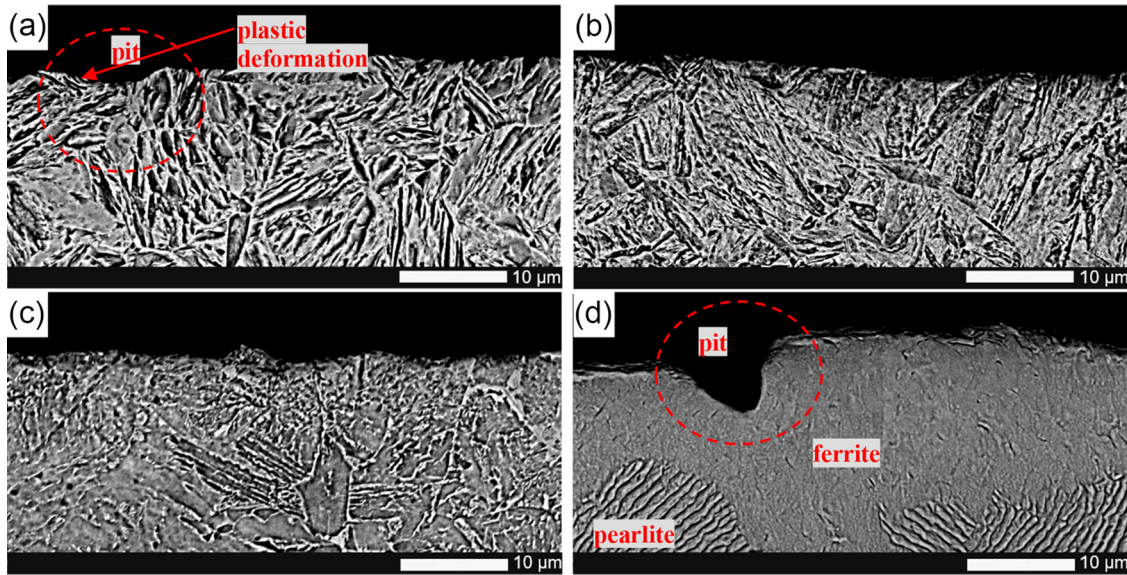


Figure 17. Cross-sectional SEM analysis under unetched conditions of selected samples subjected to abrasive wear: (a) as-delivered condition, (b) after mineral oil cooling, (c) after air cooling at 1 bar pressure, and (d) after furnace cooling.

mechanisms observed in SEM micrographs, where low-pressure air cooling failed to suppress the formation of large damage zones. Profilograms, as shown in Figure 19, provide a more intuitive and relevant reflection of wear mechanisms. Interestingly, although profilograms revealed substantial qualitative differences in surface morphology, the statistical analysis of roughness parameters indicated limited significance in differentiating between most heat treatment conditions, with the exception of R_v . The post-hoc Duncan test identified only the 1 bar air-cooled samples as significantly different, which also coincided with the most degraded surface profiles. These findings suggest that while profilometric measurements offer valuable visual and geometric insight, standard roughness parameters may not reliably capture the severity or type of surface degradation caused by abrasive wear. Therefore, interpreting wear resistance solely on the basis of R_a or similar indicators may be insufficient in the case of complex surface damage. Therefore, an extended Duncan post-hoc test was conducted, revealing significant statistical differences between the air-cooled (1 bar pressure) condition and all other heat treatment

conditions (Table 8). It should be noted that in this 1 bar air-cooled state, nearly all roughness parameters reached their highest values. Surprisingly, the increased number of pits associated with slower cooling did not significantly affect the quantitatively characterized surface condition, even though statistical analysis confirmed significant differences in abrasive wear resistance of Hardox 500 steel depending on the cooling rate. Consequently, correlating the surface condition (as characterized by roughness parameters) with abrasive wear resistance appears unjustified. The lack of significant differences in roughness parameters is unexpected, given the clear variations in wear mechanisms observed on worn surfaces and cross-sections of different samples.

4. Discussion

The article presents a comprehensive approach to the evaluation of heat-treated steels, with a particular focus on microstructural development aimed at achieving varied responses in terms of tribological resistance. In the course

	Effect SS	Effect df	Effect MS	Error SS	Error df	Error MS	<i>F</i>	<i>p</i>
R_a	0.1072	8	0.0134	0.1407	18	0.0078	1.7140	0.1629
R_p	0.9911	8	0.1239	4.6533	18	0.2585	0.4792	0.8551
R_v	4.4831	8	0.5604	3.9031	18	0.2168	2.5844	0.0450

Table 7. Results of variance analysis.

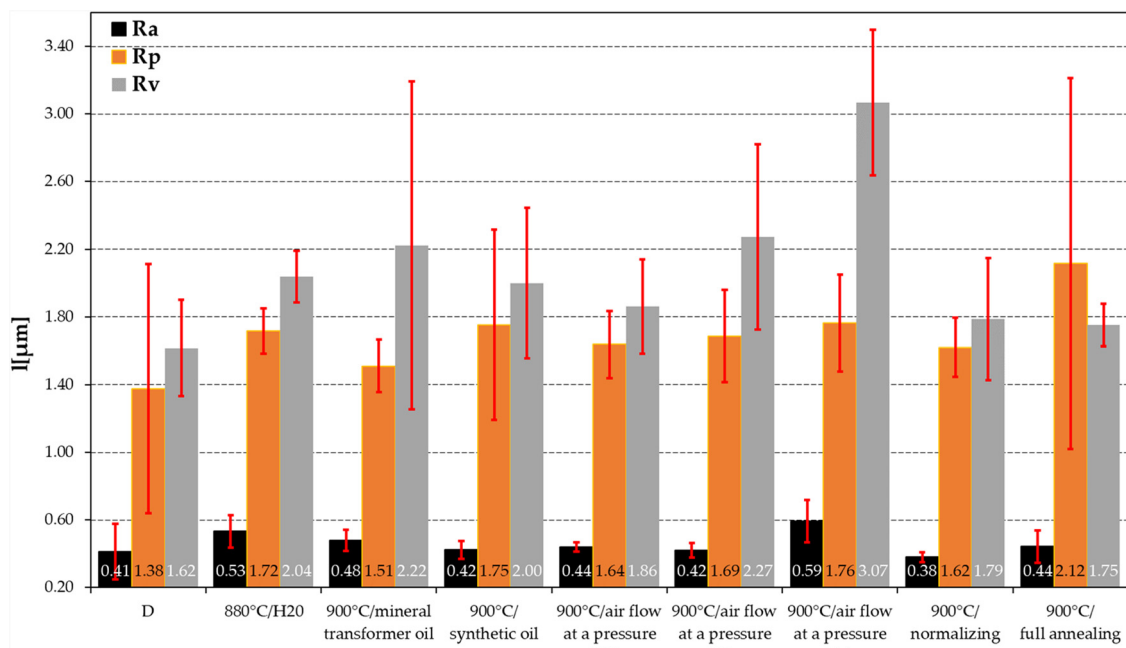


Figure 18. Roughness parameters R_a , R_p , and R_v of Hardox 500 steel under different heat treatment conditions subjected to abrasive wear testing.

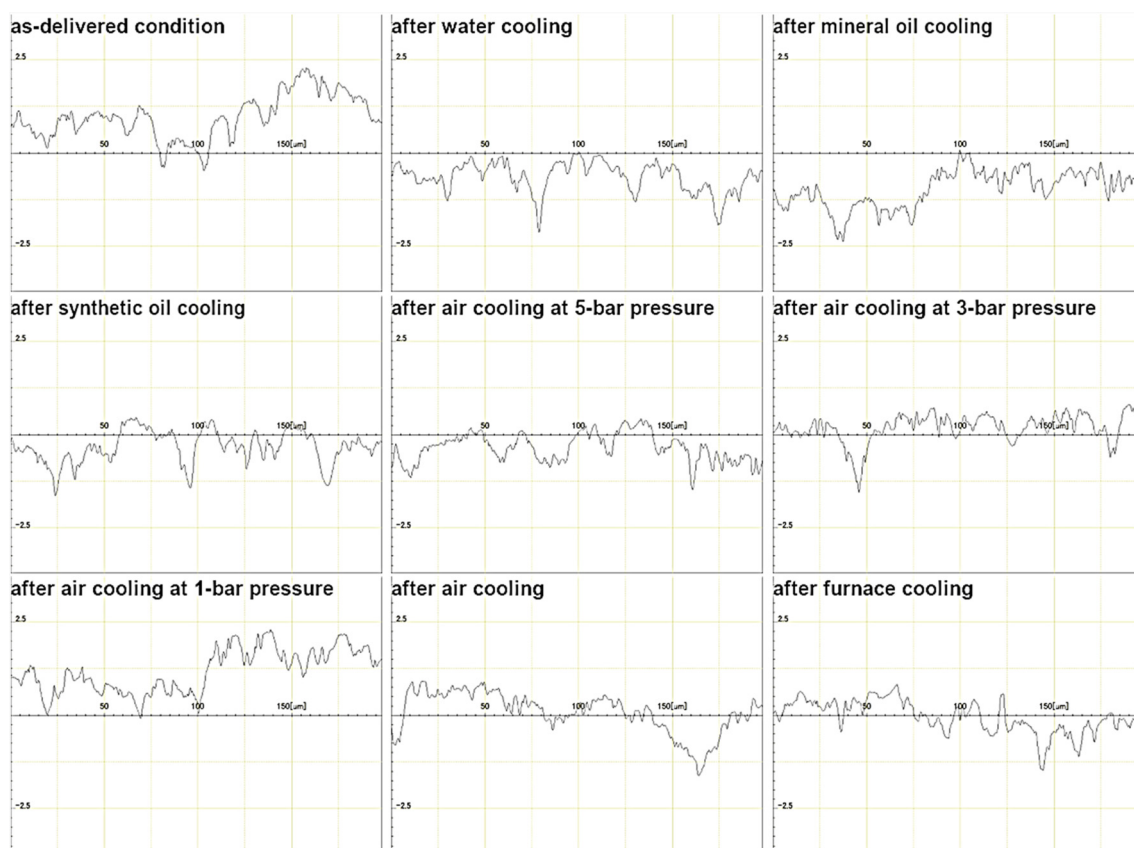


Figure 19. Profilograms of Hardox 500 steel under different heat treatment conditions subjected to abrasive wear testing.

State of heat treatment	{1} $M =$ 1.6167	{2} $M =$ 2.0367	{3} $M =$ 2.2233	{4} $M =$ 2.0000	{5} $M =$ 1.8600	{6} $M =$ 2.2733	{7} $M =$ 3.0667	{8} $M =$ 1.7867	{9} $M =$ 1.7533
{1}		0.3384	0.1759	0.3759	0.5651	0.1470	0.0034	0.6780	0.7236
{2}	0.3384		0.6295	0.9243	0.6662	0.5640	0.0215	0.5545	0.5109
{3}	0.1759	0.6295		0.5860	0.3924	0.8970	0.0488	0.3146	0.2855
{4}	0.3759	0.9243	0.5860		0.7171	0.5185	0.0197	0.6028	0.5598
{5}	0.5651	0.6662	0.3924	0.7171		0.3405	0.0103	0.8493	0.7944
{6}	0.1470	0.5640	0.8970	0.5185	0.3405		0.0516	0.2694	0.2426
{7}	0.0034	0.0215	0.0488	0.0197	0.0103	0.0516		0.0075	0.0067
{8}	0.6780	0.5545	0.3146	0.6028	0.8493	0.2694	0.0075		0.9312
{9}	0.7236	0.5109	0.2855	0.5598	0.7944	0.2426	0.0067	0.9312	

Table 8. Results of Duncan's test for the parameter R_v .

of this study, a broad spectrum of surface roughness parameters was selected for analysis. This decision was motivated, in part, by the absence of clear literature guidelines that would support the selection of a narrower set of parameters. The authors were concerned that limiting the scope might reduce the likelihood of detecting potential microstructural differences arising from distinct heat treatment processes, which could influence wear resistance. Therefore, an extended set of roughness parameters – R_a , R_q , R_z , R_p , and R_v – was employed to ensure more comprehensive evaluation. These parameters served not only as the basis for data interpretation but also as a reference for previous studies [3,22,28,50] and as a foundation for future work aimed at identifying a tailored set of parameters capable of distinguishing between different steel grades and heat treatment strategies. The experimental plan, which included a detailed assessment of the obtained roughness parameters – each reflecting specific surface features formed during abrasive wear – was intended to enable the identification of even subtle differences in the resulting microstructures. Previous studies had demonstrated varied outcomes regarding the correlation between the roughness parameters and the hardness of tested materials, reinforcing the need for the broad approach adopted in this work. Studies on the effect of grain size on the abrasive wear resistance of Hardox 500 steel have shown that lower austenitizing temperatures – particularly 850 and 900°C – produce smoother surfaces with lower roughness, whereas austenitizing at 1,000 and 1,200°C increases roughness parameters, especially R_z , R_p , and R_v , compared to the as-delivered condition [28]. An analysis of roughness on worn plowshares made of pearlitic versus martensitic steel likewise revealed that the pearlitic steel exhibited higher roughness values – most notably in the maximum profile peak height (R_p) – than its

martensitic counterpart [50]. Roughness measurements on abraded Hardox 450 surfaces also demonstrated that microstructures with higher hardness correspond to the lower roughness parameter value [3]. In the as-delivered state, R_a was 0.393 μm , increasing after austenitization at 900 and 950°C; with a further increase in the austenitizing temperature, R_a decreased, a trend correlated with austenite grain growth and carbide-dissolution-induced microstructural changes. Similar trends were observed for R_p . It was found that as the austenite grain size increased, the wear weight-loss rate increased linearly – accompanied by a decline in abrasive wear resistance – yet no significant correlation between hardness and the wear-resistance index was detected. Finally, relationships between selected roughness parameters (R_p , R_v , and R_z) and surface hardness could be well approximated by a quadratic function.

In this study, despite conducting a comprehensive analysis of roughness parameters after various heat treatment conditions of Hardox 500 steel, it was observed that traditional roughness parameters do not fully reflect the surface states, which exhibit a high degree of variation. Therefore, the authors decided to address this issue in the discussion. Standard roughness parameters, such as R_a and R_q , provide only an average description of surface topography. However, for surfaces that are locally heterogeneous – containing deep pits, areas of intense wear, as well as relatively untouched regions or those formed by the wear of asperities – such averaging leads to a loss of critical information about localized defects. Statistical analysis of the obtained values did not show significant differences between the various heat treatment conditions, which contradicts microscopic observations, where clear local differences in wear mechanisms were noticeable. This discrepancy arises because roughness parameters

characterize only the global features of the surface, without considering the spatial distribution of anomalies such as deep pits, material spallation, or localized plastic deformations. A critical aspect is that averaging roughness values masks these local variations. This approach can lead to an inadequate assessment of the actual surface condition, particularly in applications where local defects play a key role in the durability of working components. The literature frequently emphasizes that classical roughness parameters are insensitive to localized topographical variations, which may result in misinterpretation of surface conditions when analyzing materials with complex structures. Studies by other research groups suggest that conventional roughness parameters are also insufficient for predicting the structure of worn surfaces under sliding friction conditions [51]. Moreover, Kovalev *et al.* demonstrated the lack of correlation between initial roughness and the parameters of the worn surface.

Explanations for the above observations, as well as valuable supplementation to the experimental studies, are provided by numerical analyses of surface topography evolution, as proposed by Garcia-Suarez *et al.* [52]. The simulation results indicate that, regardless of the initial surface condition, wear processes gradually “erase” the original topographic features. As a result, the surface tends toward a steady-state roughness, reached after a transitional period during which surface parameters oscillate around an average value. Comparing these results with the Hardox 500 study allows for several conclusions. First, both experimental studies and numerical analyses suggest that global roughness parameters are insufficient for a comprehensive description of surface conditions, especially when local defects play a crucial role. Second, the transitional period observed in simulations, during which the surface “forgets” its initial state, reflects dynamic wear processes that may also occur in steel subjected to different heat treatment variants. Numerical simulations have shown that initially smooth surfaces reach a steady-state roughness more quickly, whereas initially rough surfaces require a longer transitional period. This observation aligns with the findings for Hardox 500, where differences in wear mechanisms visualized through SEM were not fully captured by averaged roughness parameters. This indicates that, under real operating conditions, the influence of roughness on wear may be more complex and dependent on local micromechanical wear mechanisms than suggested by standard parameters. Therefore, integrating experimental studies on Hardox 500 with numerical analyses in future research would enable a more comprehensive understanding of wear mechanisms. Advanced techniques, such as three-

dimensional topography analysis, allow for a more precise identification of local defect distribution and their impact on overall surface conditions. Such a holistic methodology could contribute to the optimization of heat treatment processes, leading to increased wear resistance and improved durability of the material under operational conditions.

Additionally, analyses conducted by Bigerelle *et al.* [53] demonstrated that three primary wear mechanisms can be distinguished depending on the size of abrasive grains. For particles larger than 125 μm , wear occurs through microcutting, whereas for grains smaller than 10 μm , adhesion dominates. In the intermediate range, when the abrasive particle size is between 10 and 125 μm , a phenomenon known as the “grain size effect” is observed, where the extreme amplitude of peaks to valleys decreases with increasing abrasive particle size. This indicates the crucial role of grain size distribution in the surface degradation process. Bigerelle *et al.* also demonstrated that beyond a critical autocorrelation length (approximately 160 μm), the surface loses memory of its initial structure, meaning that its further evolution can be modeled using extreme value theory. Autocorrelation measures how roughness values at different points on the surface are related to one another. If there is a distinct correlation at a certain length, it indicates that the surface maintains an ordered structure on that scale. However, once the critical autocorrelation length (in this case, 160 μm) is exceeded, the surface loses this ordered relationship, and its further evolution becomes more random and independent of its previous state.

An extension of the above studies is provided by the results presented by Szala *et al.* [31], who investigated the wear mechanisms of S235JR, S355J2, C45, AISI 304, and Hardox 500 steels using three types of abrasives: garnet, corundum, and silicon carbide (carborundum). Their experiments demonstrated that the type of abrasive used has a decisive impact on wear mechanisms and local surface topography. For example, when silicon carbide – which is characterized by very high hardness and fine grain size – was used, the dominant wear mechanism for Hardox 500 steel was microcutting, resulting in lower values of R_k , R_{pk} , and R_{vk} parameters. In contrast, when using coarser abrasives such as garnet or corundum, microplowing and microfatigue mechanisms dominated, leading to higher averaged R_a and R_q values, which, however, did not fully capture the localized complexity of wear. The experimental results of Szala *et al.* [31] emphasize that both the steel microstructure and the properties of the abrasives used (e.g., hardness, particle size, and morphology) play a crucial role in wear processes and

the development of surface topography. Analogous to the results of numerical analyses, where the surface evolves toward a steady-state roughness, regardless of its initial condition, the experiments of Szala et al. show that the dynamics of wear result from the complex interaction between the material's microstructure and the abrasive properties. This effect highlights that traditional roughness measurement methods, which rely solely on averaged values, are insufficient for a comprehensive assessment of the surface condition, particularly in cases where local defects have a critical impact on the functionality of the worn component. In the context of Hardox 500 steel, this means that traditional roughness parameters do not allow for an accurate prediction of surface degradation mechanisms during friction, and their application in wear resistance evaluation may lead to incorrect conclusions. Therefore, it is necessary to employ more advanced surface topography analysis methods, which take into account not only global roughness characteristics but also the actual contact and microstructural transformations occurring during operation.

In conclusion, although traditional roughness parameters provide valuable information about the average surface condition, their limited sensitivity to local topographical variations makes them insufficient for analyzing materials with complex wear structures. Thus, it is essential to expand the research methodology to include techniques that enable a more precise description of local surface features, leading to a better understanding of wear mechanisms and the more effective adjustment of heat treatment parameters and material selection for specific operating conditions.

5. Conclusions

The obtained results emphasize the importance of selecting appropriate heat treatment parameters for components made from the analyzed material. The statistical analysis of the results confirmed significant differences between the examined heat treatment variants, indicating the material's sensitivity to changes in cooling intensity, which directly influences microstructure formation. The research, conducted using the described methodology, allowed for the formulation of the following conclusions:

1. The use of high-intensity cooling media, such as water and oils, resulted in the formation of a martensitic microstructure, characterized by high hardness, which directly translated into significantly increased tribological resistance ($k_b = 0.96\text{--}1.05$). In contrast, cooling in

lower-intensity media led to the formation of other hardened structures such as very fine pearlite, bainitic ferrite, and ferrite. The ferrite initially appeared acicular and later transformed into a polygonal form. These microstructures exhibited lower hardness, leading to a decline in tribological resistance ($k_b = 0.68\text{--}0.79$).

2. The application of the Archard model to describe the volumetric wear of steel with different heat treatment conditions requires adjusting the wear coefficient k based on the material's hardness, microstructure, and wear mechanism. While using a single averaged coefficient may be useful for general comparisons, achieving higher predictive accuracy necessitates differentiating k as a function of specific material properties. Thus, although the Archard model requires calibration, it remains an effective tool for tribological evaluation. A better fit of the model was obtained for harder microstructures, where the relative difference between theoretical and actual volumetric wear was max. 11.82%.
3. The dominant wear mechanisms included microplowing, microcutting, and asperity shearing, particularly in high-hardness samples. In samples cooled at lower intensities, the primary wear mechanism was cyclic particle interaction leading to material spallation. In complex microstructures, wear typically initiated in the softer phase, leading to the formation of deep pits, while the harder phase temporarily acted as a protective barrier. However, continued abrasion eventually led to its detachment, accelerating surface degradation.
4. Variance analysis indicated no statistically significant differences between roughness parameters depending on heat treatment conditions ($R_a = 0.38\text{--}0.59$; $R_p = 1.38\text{--}2.12$), except for the R_v parameter ($R_v = 1.62\text{--}3.07$). In this case, Duncan's test revealed differences between the air-cooled (1 bar pressure) condition ($\sim 1^\circ\text{C/s}$) and other heat treatment conditions.
5. Given the above findings, in addition to traditional roughness parameters, it is essential to apply more advanced surface analysis methods. Techniques such as three-dimensional topography analysis allow for a more precise assessment of local defect distribution and their influence on the overall surface condition. This holistic approach can contribute to enhancing the material's wear resistance and improving its durability under operational conditions.

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Author contributions

Martyna Zemlik: Conceptualization, methodology, validation, formal analysis, investigation, resources, data curation, visualization, writing – original draft preparation, supervision. Beata Białobrzaska: Conceptualization, methodology, validation, formal analysis, investigation, resources, data curation, visualization, writing – original draft preparation, writing – review and editing, supervision. Mateusz Stachowicz: methodology, validation, formal analysis, investigation, data curation. Lukasz Konat: formal analysis, resources, writing – review and editing.

Conflict of interest statement

Authors state no conflict of interest.

Data availability statement

Data sharing is not applicable to this article as no datasets were generated or analysed during the current study.

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