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The Effect of Cryogenic Treatment on a High Co Bearing DIN 1.2888 Steel Used as Die in Hot Forging Process

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ABSTRACT

Die steels used in metal forming processes require resilience in harsh conditions. Therefore, it becomes crucial to convert retained austenite into martensite and uniformly distribute fine carbide phases in these steels. Cobalt is known for its ability to enhance carbide formation in tool steels. In the present work, DIN 1.2888, a hot work tool steel characterized by high cobalt content, was subjected to the conventional HT and CT processes at -100, -140, and -180 °C for 6 h, then double-tempered. The properties of samples were investigated by XRD, SEM-EDS, and Microhardness analyses. The wear mechanism was investigated by a pin-on-disc wear testing device and the impact toughness of the samples was examined both at room temperature and the working temperature of the dies (350 °C). It was observed that decrease of the cryogenic treatment temperature resulted with a slight change on the hardness values of the samples from 507 HV to 529 HV, while impact toughness increased from 12.35 J to 23.44 J at 350 °C. Also, the wear rates of the samples exhibited a ~50% decrease. It can be deduced that cryogenic treatment had positive effect on the properties of steel, depending on the retained austenite decrement and homogenous fine carbide density increment with the decreasing cryogenic treatment temperature.

Key Words: Cryogenic treatment, DIN 1.2888 steel, Carbide distribution, Mechanical properties, Wear

1. Introduction

Hot work tool steels are frequently used in metal forming processes as dies in forging processes, because of their excellent properties under harsh working conditions. These steels are expected to have good wear resistance and toughness since they are subjected to high thermal shocks, high pressures, friction, and wear damage. There are varying processes to extend the lifespan of the die, like alloying, material selection, and modeling programs [1–3]. Yet, one key issue that reduces the operating life of tool steels is the presence of retained austenite (RA) in the steel after heat treatment (HT). RA, which is a soft phase, has the potential to reduce the die life even in very small amounts [4,5]. Considering this issue, it is critical transforming the retained austenite to the martensite phase in a controlled manner by cryogenic treatment (CT), where the amount of retained austenite reduces while the wear resistance of the work steel improves [6]. Cryogenic treatments are classified into two types: shallow cryogenic treatment (SCT) and deep cryogenic treatment (DCT). The SCT is applied between -50 °C and -100 °C, while the DCT is suitable below -125 °C [5]. There are numerous studies concerning the cryogenic treatment of varying tool steels. They concluded that

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although both SCT and DCT reduced the remaining austenite amount, DCT eliminated more retained austenite, and carbides were distributed more uniformly. As a result, it was determined that DCT provided more wear resistance than SCT [1,6–24].

Cryogenic treatment (CT) results in various microstructural changes compared to conventional heat treatment, such as decreased retained austenite (RA), refined martensite, altered precipitation kinetics of nano-sized carbides, and increased small globular carbides [25,26]. Quenching induces a diffusionless transformation from austenite to martensite, but not all austenite converts, leaving some RA [25]. CT stabilizes the remaining austenite, preventing further martensitic transformation by inducing higher compressive macro-stresses [25,27]. Martensite refinement at cryo-temperatures involves a thermally activated transformation process, resulting in high densities of defects like dislocations and twins, caused by differential thermal contraction and high compressive stresses [25,28,29]. Tempering reduces these stresses, allowing RA transition and subsequent volume increase through carbide precipitation [25,27].

CT enhances the precipitation kinetics of nano-sized carbides during tempering across various steel types, leading to more carbon clusters that accelerate cementite or ε-carbides precipitation [29,30]. However, CT has little impact on high-carbon steels due to the strong binding enthalpy of dislocations and carbon [25,28]. Higher quantities of uniformly dispersed carbide particles, typically less than 0.5 μm, are observed in some steels after CT [8,10,17,31–33]. Extra small globular carbides (SGCs) may form before tempering and reduce in quantity with increased tempering temperatures [25]. An alternative theory suggests SGCs form due to plastic deformation at cryo-temperatures, manifesting in the microstructure before tempering [25,31].

Numerous studies investigated the effect of cryogenic treatment on the microstructural, tribological, and mechanical properties of AISI H13 hot work tool steel. Gecü [7] examined the single and double tempering effect after SCT and DCT treated AISI H13 steel. After both SCT and DCT treatments, the retained austenite transformed into martensite, and finer carbide formation with higher density formed. Accordingly, the hardness and wear properties were positively affected especially by the DCT process. Thakurai et al. [34] subjected 40CrMoV5 (AISI H13) steel to HT and DCT for different times (12, 24, 36, and 48 hours). It was observed that DCT for 12 h lowered the hardness and wear resistance when compared to only HT and air quenching process, because of the formation of non-tempered martensite. However, increasing DCT time increased hardness and accordingly wear resistance. Shinde [35] also concluded the same phenomenon, however in that study it was observed that up to 16 h of the DCT process, the hardness of the steels increased due to the fine carbide precipitation but lowered over 16 h because of the growing particles according to the Ostwald ripening mechanism. Mokarian et al. [36] investigated the effects of cryogenic treatment and boronising treatment on the wear behavior and microstructure of DIN 1.2344 (AISI H13) hot work tool steel. They revealed that the hardness and wear resistance of cryogenically treated samples increased by 25% and 39%. Perez et al. [2] applied HT, CT, and up to 3 tempering processes to AISI H13 steel in different cooling environments (gas and oil). They claimed that the cryogenic treatment did not have much effect on the tensile strengths, but there was an increase of 22.5% and 24% in impact toughness for gas-cooled, and oil-cooled samples, respectively.

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There are a few studies that investigate the effect of CT on cobalt-bearing tool steels [37–40]. Gogte et al. [40] investigated the influence of DCT on an AISI T42 high-speed steel (HSS), which contains 10 wt.% Co. They observed that 8 hours of DCT soaking improved the wear properties while higher DCT periods lowered the wear resistance due to the carbide detachment. Moreover, in a study conducted by Dhokey and Dandawate [37,38], AISI M35 HSS, containing 4.7 wt.% Co was subjected to DCT up to 48 h and it was noted that increasing the soaking period increased the fine carbide formation and relatively higher hardness, impact toughness, and lower wear rates. On the other hand, Jovičević-Klug et al. [39] claimed that a low amount of Co (0.69 wt.%) improved the carbide formation after DCT, but a high Co amount (4.50 wt.%) did not affect the carbide precipitation density significantly.

DIN 1.2888 is a high Co-containing (~10 wt.%) hot work tool steel which is widely used as a die in the hot forging process. As mentioned before, no studies are present in the literature that investigates the effect of CT on DIN 1.2888 steel. This study aimed to examine the influence of CT at different temperatures (-100, -140, and -180 °C) on the 1.2888 steel, which would reduce the cost of the die by increasing the life of the die steels used in the hot forging process. To achieve this goal, DIN 1.2888 die steels were subjected to the traditional HT and CT processes, then double-tempered. The properties of samples were investigated by XRD, SEM-EDS, and Vickers hardness analyses. The wear mechanism was investigated by a pin-on-disc wear testing device and the impact toughness of the samples was examined both at room temperature and the working temperature of the dies (350 °C).

2. Experimental

Annealed DIN 1.2888 steel with a hardness of 25 HRC was used for the experimental procedure. Initially, the steel was obtained from the manufacturer with the dimensions of 100x100x100 mm. The spectrometric analysis of the average composition obtained from different sections is given in **Table 1**. Then the steel was cut into pieces with varying dimensions for the microstructural and mechanical tests.

Table 1. Average chemical composition of DIN 1.2888 hot work die steel (wt.%)

C	Si	Mn	Cr	Mo	W	Co	Fe
0.19	0.26	0.52	9.54	1.99	5.56	10.08	Bal.

All the steel samples were heated to an austenitizing temperature of 1150 °C in a vacuum furnace for 2h. Then different samples were cooled to room temperature, -100, -140, and -180°C and labeled as HT (Heat treated), -100CT (cryogenic treated at -100 °C), -140CT (cryogenic treated at -140 °C), and -180CT (cryogenic treated at -180 °C), respectively. Then the samples were double tempered at 650 °C, to lower the retained austenite and improve the carbide precipitation, which is generally made before the utilization

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of 1.2888 steel. Schematic T–T diagram illustrating the processing steps employed on DIN 1.2888 steel samples is given in

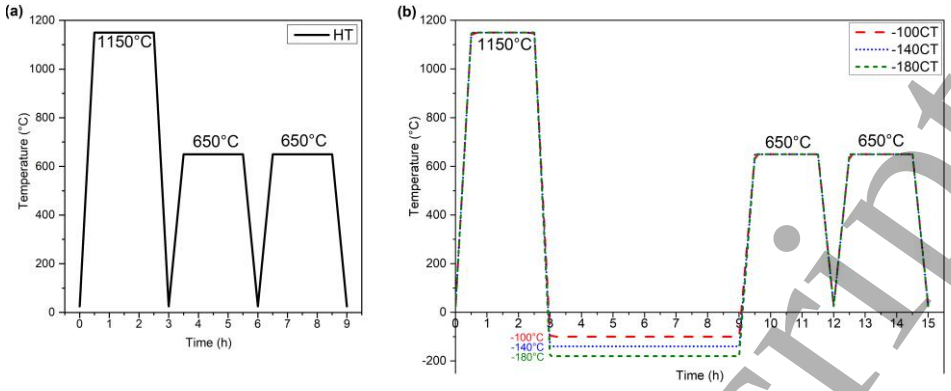


Fig. 1.

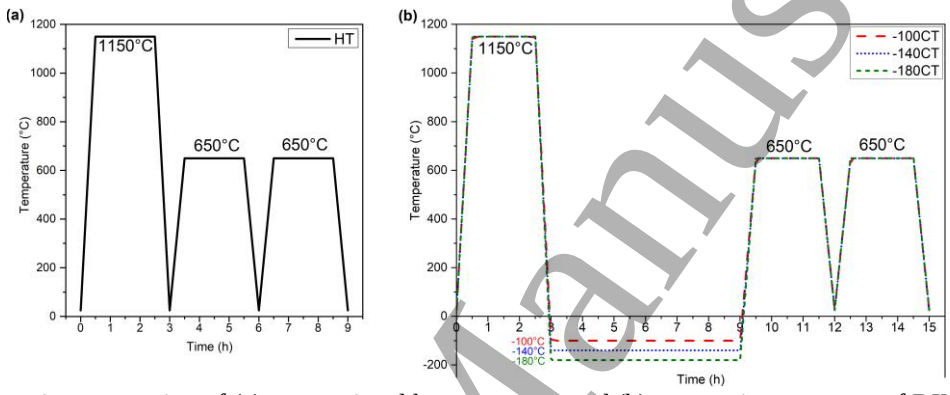


Fig. 1 Schematic presentation of (a) conventional heat treatment and (b) cryogenic treatments of DIN 1.2888 hot work tool steels

For the microstructural investigations, the sample surfaces were ground, polished, and etched with Nital 3 (3mL nitric acid and 97mL ethanol). Microstructure images of the samples were examined with a scanning electron microscope (SEM, Philips XL 30 SFEG). Carbide structures in steels were determined by the SEM-EDS analysis. Additionally, the carbide distribution of the samples was investigated by ImageJ software v.1.54d employing threshold techniques on the SEM images of the samples with 5000X magnification as mentioned in Figure S1 and S2, supplementary material. After determining the region, the photos were converted to 8-bit format, and a threshold procedure was used. The samples were subjected to the same threshold criteria, and the carbides were automatically chosen based on the variations in color observed in the SEM pictures. The software then gave each carbide's characteristics as well as a summary that included the carbide counts, average diameters, and percentage of the carbide area.

The composition of phases was determined using Panalytical Empyrean XRD using a Co Kα source (λ=1.78901 A⁰) operated at 40 kV and 40 mA using a step size of 0.01 °/sec. Rietveld refinement was applied using HighScore Plus software and PDF-4 (ICDD) mineralogical database to determine retained austenite volume fraction.

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Microhardness measurements were made with a Hardway MHVD 1000IS model microhardness device with a load of 200 g. Each sample was subjected to five measurements, and the average results were provided.

Wear tests were performed by a pin-on-disc wear tester at room temperature and 350 °C (copper forging temperature) using varying loads (15 N, 20 N, and 25 N) with a sliding speed of 0.9 m/s for a 300 m distance, according to the ASTM G99-17 standard. Also, a wear trace diameter of 20 mm was applied. The pin was a 3 mm diameter Al_2O_3 ball. Wear volume losses were calculated by tracking the wear depths with a 3D measuring approach in a Keyence optical microscope. Volume losses (V) after the wear tests were calculated by Eq. 1, where R, r, and d were wear scar radius, Al_2O_3 ball radius, and wear scar width, respectively. Obtained V values were used in Eq. 2 to determine specific wear rate (W_R) of the samples by utilizing L as the sliding distance and P as the applied load. After the wear test, SEM was used to examine the surface, and the wear behavior of the samples was identified [7,41,42].

$$V = 2\pi R\left[r^2 \sin^{-1}\left(\frac{d}{2r}\right) - \left(\frac{d}{4}\right)(4r^2 - d^2)^{\frac{1}{2}}\right] \quad (1)$$

$$W_R = \frac{V}{LP} \quad (2)$$

Notch impact tests were performed in 10 x 10 x 55 mm (V-notch) dimensions according to ASTM E23 standard. Each sample was tested three times at room temperature and 350 °C, and the average results were reported.

3. Results and Discussion

3.1 The effect of cryogenic treatment on microstructure

The SEM images of heat treated and cryogenic treated with a magnification of 5000X and 25000X are given in Fig. 2.

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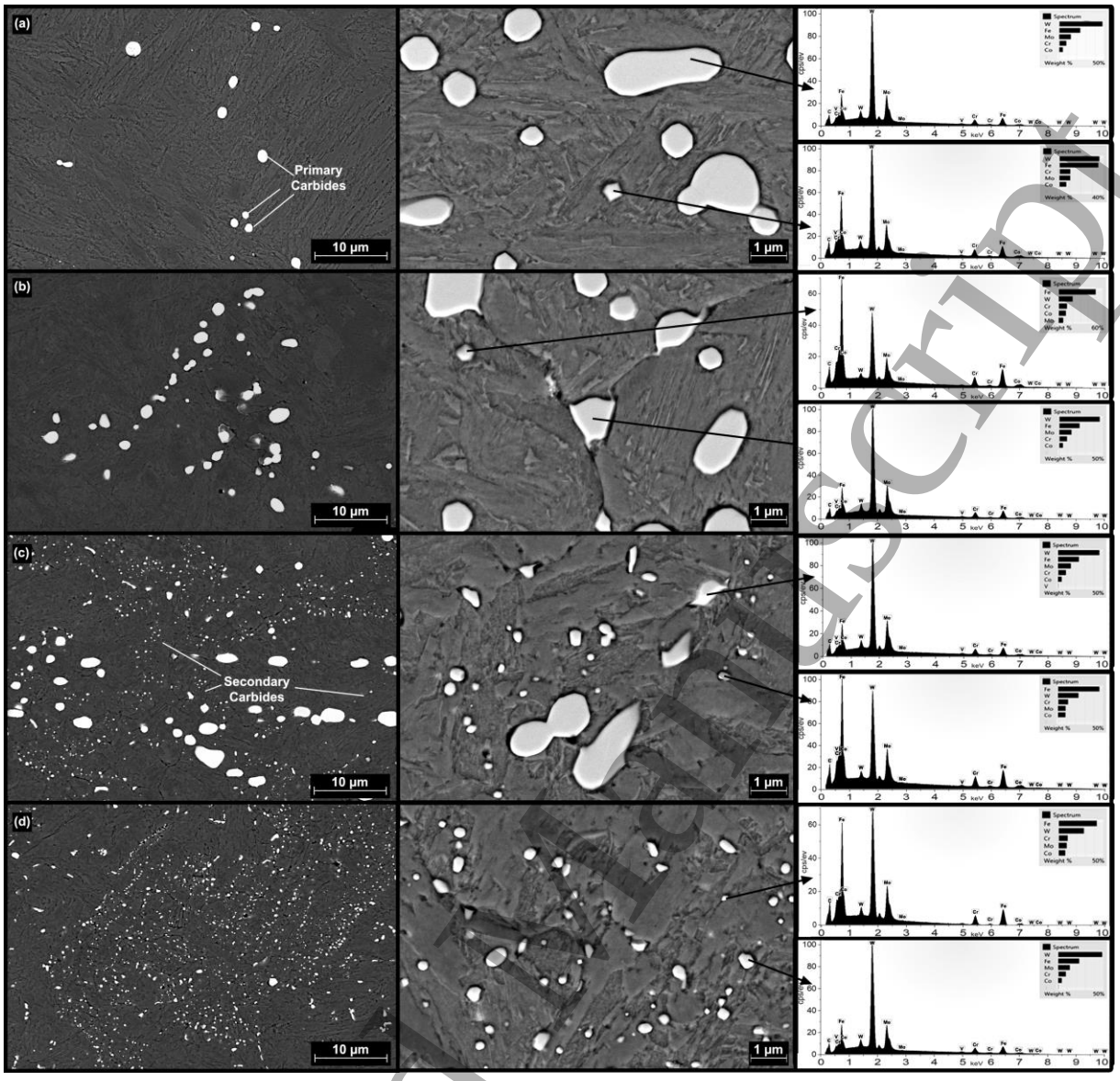


Fig. 2 SEM images and EDX analysis of the carbides obtained from **(a)** HT, **(b)** -100CT, **(c)** -140CT, and **(d)** -180CT

Fig. 3 and **Table 2** represent the retained austenite (RA) ratios calculated by Rietveld Refinement conducted by Highscore Plus software and the carbide distribution obtained by ImageJ software.

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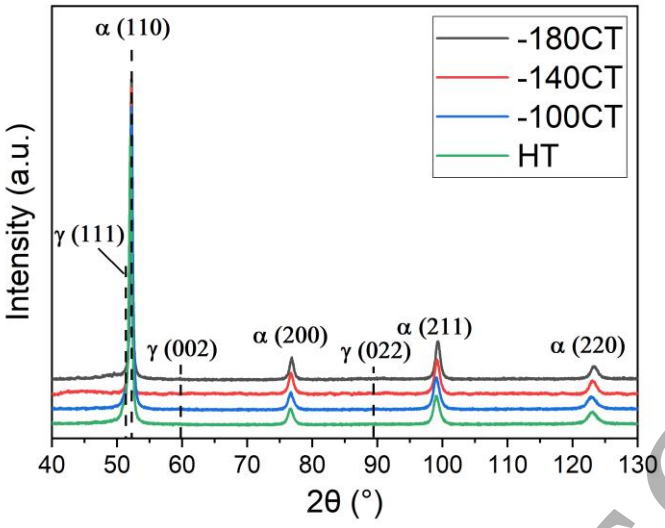


Fig. 3 XRD peaks of the steels obtained after tempering process

Table 2 RA and carbide distribution of 1.2888 samples with varying treatments

Samples	Retained Austenite (vol.%)	Average Carbide Size (µm)	Carbide Amount (area %)
HT	1.8	0.813	0.751
-100CT	0.8	0.760	3.006
-140CT	0.2	0.152	3.312
-180CT	< 0.1	0.057	3.609

According to Fig. 2 and Table 2, retained austenite amounts in the samples showed a decrease with the lowering of the cryogenic treatment temperature. After conventional heat treatment and tempering processes, it is a well-known issue that the austenite phase cannot transform to martensite totally, and therefore a certain amount of austenite phase retained due to the martensite finish temperature which is lower than the room temperature [4,43]. The retained austenite is under significant compressive stress after conventional quenching. This occurs because the retained austenite regions are surrounded by martensitic areas and are subjected to compressive forces due to the volume expansion that happens during the transformation from retained austenite to martensite [25,26]. Lowering cryogenic temperature accelerates the martensitic transformation by providing a sufficient driving force to overcome the energy barriers for nucleation and growth of martensite. Also, CT refines the martensitic structure, with martensite laths becoming much thinner than conventional treatment[25]. In the present study, the RA was transformed down to a non-detectable value after cryogenic treatment. Varying studies claim that a total transformation is not possible due to the volume expansion of transformed martensite and its compressive pressure on the RA [4,44–46].

Moreover, a significant change in the carbide precipitation was observed. For the heat-treated sample, without cryogenic treatment) (1.2888 HT) the carbides have precipitated non-homogenously with an average carbide size of 0.813 µm, with a volumetric area of 0.751 %. On the other hand, after shallow cryogenic treatment, the carbide formation increased significantly while a slight reduction in the average carbide size was encountered. Lowering the cryogenic temperature to -140 °C resulted in both bigger and

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smaller carbides distributed throughout the structure. The average carbide size was nearly five times lower than the shallow cryogenic treated sample. On the other hand, deep cryogenic treated sample illustrated a very fine carbide formation and homogeneous carbide distribution. The supersaturated carbon and alloying element atoms in Martensite were forced out of the lattice due to the high lattice contraction of Martensite at cryogenic temperatures. Also, new nucleation sites for complicated carbides were produced. With the tempering process, fine secondary carbides formed [19,39,47–49]. Additionally, cobalt played a significant role in the secondary carbide precipitation. As Arici et al. [50] explained, Co is not an element that forms carbides. However, it assists indirectly in the formation of metal carbides. A synergetic increment in the density of dislocations is observed with the addition of Co [51]. The nucleation of precipitates and grains occurs at areas where the dislocation density increases. Therefore, the presence of cobalt raised the chemical potential of alloying elements and enhanced their precipitation [37–40,50,51].

3.2 The effect of cryogenic treatment on Hardness

The hardness measurements were conducted on the samples after the double tempering process are given in **Fig. 4**. A hardness value of 507 HV in was obtained for the heat-treated sample and a slight increase was observed with the decreasing the cryogenic temperature. The measured hardness values of HT, -100CT, -140CT and -180CT were 521, 528 HV, and 529 HV, respectively. In the study of Koneshlou et al., it was observed that the hardness of H13 hot work tool steel increased at a similar rate in the shallow cryogenic treatment and deep cryogenic treatment. Lowering the temperature results in a lower change in hardness after most of the austenite is transformed [52]. In a study investigating the cryogenic treatment on H11 steel, the increase in the microhardness in the shallow and deep cryogenic treatment samples was reported to be 7% and 18%, respectively, where the increase in the percentage of retained austenite transformation to martensite was the cause of the increase in hardness [24]. Akbarizadeh et al. indicated that this hardness was associated with the conversion of retained austenite into martensite, as well as with secondary carbide deposition in deep cryogenic processing in D6 tool steels [8].

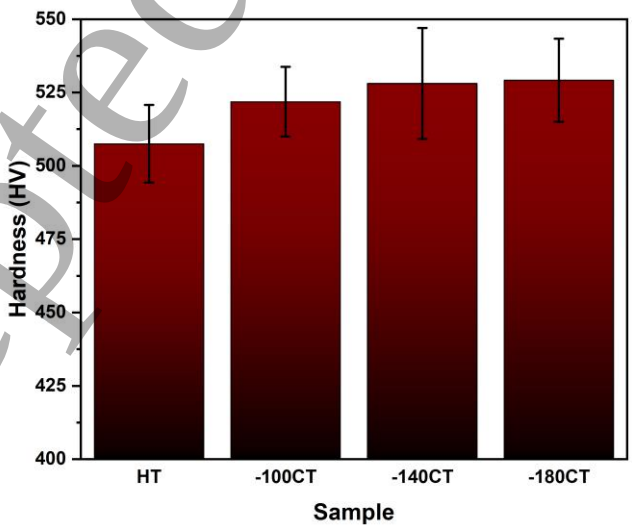


Fig. 4 Vickers Hardness results for heat treated and cryogenic treated 1.2888 steel samples

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In the present study, it was observed that very low amounts of austenite retained in the structure especially with the lowering cryogenic temperature and accordingly a significant hardness change was not observed.

3.3 Effect of Heat Treatment on Toughness

As shown in **Fig. 5 (a)** that there is a small increase in impact energy with the decrease of the cryogenic processing temperature in the notch impact tests performed at RT. In 1.2888 steel, there was an increase of less than 1 J (joule) by cryogenic treatment. While the impact energy of the HT sample was 3 joules, the impact energy was measured as 3.10, 3.44, and 3.63 J in the -100CT, -140CT and -180CT samples, respectively.

Senthilkumar et al. showed the impact energies of SCT and DCT applied 4140 steel samples at room temperature. A slight increase in impact energy was observed in the SCT and DCT samples. As a result of the cryogenic process, a large amount of austenite was converted to martensite, resulting in increases in volume causing the martensite lattice to deteriorate. This was interpreted as possible evidence of low-temperature conditioning of martensite, leading to a finer transition carbides precipitation during tempering [15]. Idayan et al. applied shallow and deep cryogenic processes to AISI 400C bearing steel. They showed that the conventional cryogenic treated material had the highest impact energy, and the impact energy was reduced by cryogenic treatment. The lowest impact energy was seen in DCT material [20].

The impact energy of the HT sample was measured as 12.35 J. It was measured as 20.57, 20.97 and 23.44 in the -100CT, -140CT and -180CT samples, respectively, with a decrease in the cryogenic processing temperature (**Fig. 5 (b)**). While the -180CT sample has an impact energy of 3.63 J in the notch impact tests performed at room temperature, the same sample has an impact energy of 23.44 J in the test performed at 350°C.

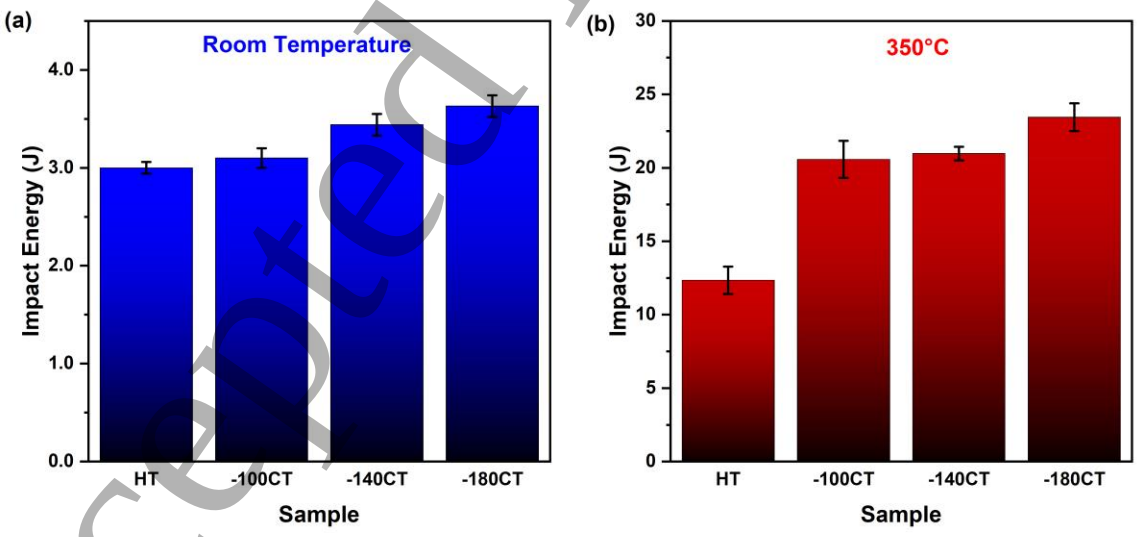


Fig. 5 Impact energy values of samples at **(a)** room temperature and **(b)** 350°C

Cryogenic treatment (CT) induces elevated compressive macro-stresses that suppress additional martensitic transformation, thereby effectively stabilizing retained austenite. During tempering, these stresses are partially relieved, allowing the retained austenite to transform. This process requires a reduction in the

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tetragonality of surrounding martensite, accompanied by carbide precipitation. Therefore, it can be concluded that, carbides precipitated in tool steels can reduce the internal stresses of martensite, and the homogeneous dispersion of fine carbides increases wear resistance. In many studies, it was implied that there were two main reasons for the improvements in the wear resistance and hardness of cryogenically treated tool steels. First, the conversion of the retained austenite structure to martensite, and second the fine carbide precipitation [23]. With the conversion of retained austenite to martensite, a decrease in toughness is expected. However, the increase in toughness was due to finer precipitation of carbides. In line with these results, an increase in impact energies of -100CT, -140CT and -180CT samples can be expected [15]. Dhokey et al. demonstrated that cryogenically treated samples of M35 cobalt-containing tool steels exhibited thinner and more uniformly dispersed carbide structures. As the steel absorbs thermal energy from the surroundings during the subsequent heating period in the cryogenic process, it facilitates atomic movement, thereby promoting the precipitation of fine tertiary carbides [38].

3.4 Effect of Heat Treatment on Wear rate

The wear tests were conducted at both room temperature and working temperature of the DIN 1.2888 steel. The volume losses and wear rates, which were calculated by Equation 1 and 2, are provided in **Table 3** and **Fig. 6**. Tests were carried out under 3 loads of 15, 20, and 25 N. In the volume loss values obtained in the abrasion test, the volume loss increases almost proportionally with the applied force. For 1.2888 hot work tool steel, -180CT had the lowest volume loss under 15 N load in tests at room temperature. With the decrease in cryogenic processing temperature, there was a decrease in volume loss and increased wear resistance. The best wear resistance is seen in the -180CT sample. As the cryogenic processing temperature decreases, the wear resistance increases. The highest difference was encountered for the sample that was cryotreated at -180 °C. Cryogenic treatment resulted in an increase in wear resistance in both hot work tool sheets of steel in experiments at room temperature. The reasons for this situation are the transformation of the soft retained austenite phase into a hard phase, martensite, in the steel and the homogenous distribution of the finer and denser carbide structure [16,46].

Table 3 Volume losses and wear rate values of the samples

	Room Temperature				**350 °C**			
	Volume Loss (mm^3)							
Load	**HT**	**-100CT**	**-140CT**	**-180CT**	**HT**	**-100CT**	**-140CT**	**-180CT**
15 N	1.265	0.800	0.749	0.552	0.885	0.739	0.639	0.431
20 N	1.550	1.016	0.837	0.733	1.058	0.851	0.749	0.455
25 N	1.979	1.736	1.640	1.512	1.098	1.084	0.958	0.600
	Wear Rate X $10^4.mm^3.N^{-1}.m^{-1}$							
Load	**HT**	**-100CT**	**-140CT**	**-180CT**	**HT**	**-100CT**	**-140CT**	**-180CT**
15 N	2.812	1.777	1.664	1.159	1.966	1.642	1.419	0.958
20 N	2.584	1.692	1.395	1.222	1.763	1.418	1.248	0.758
25 N	2.638	2.314	2.187	2.016	1.463	1.445	1.227	0.800

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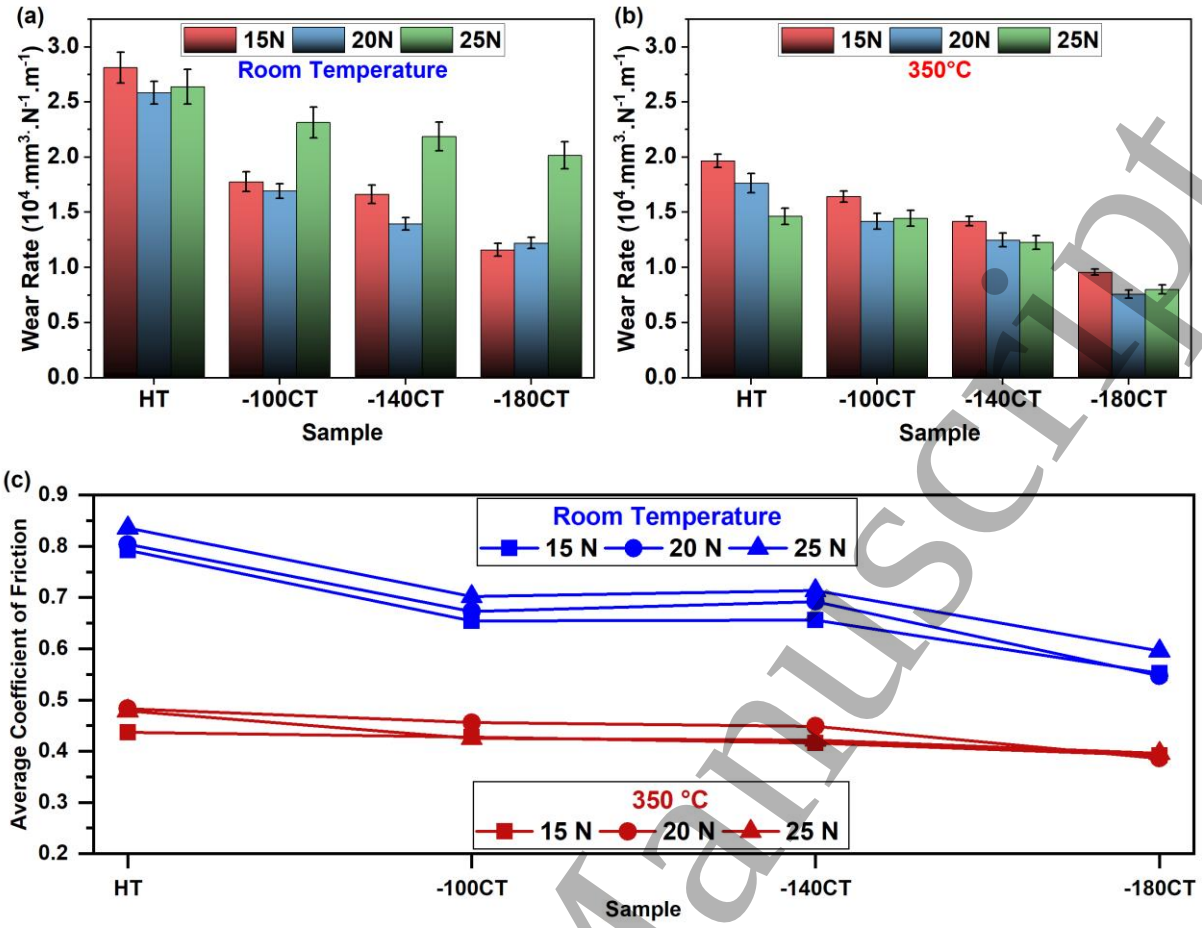


Fig. 6 Wear Rates of the 1.2888 samples conducted at **(a)** room temperature and **(b)** 350°C, and **(c)** average coefficient of friction values

When the wear results of the room temperature and the working temperature are compared, it was observed that the wear resistance of 1.2888 hot work tool steel improves at 350 °C. Zhou et al. applied a cobalt-based coating to H13 hot work tool steel and their experiments at high temperatures showed that with the increase in temperature, an initial decrement and then increment were observed. Increment in the temperature resulted with the formation of a dense oxide layer that protects the material surface. The higher the temperature increased, the thicker the oxide layer became and accordingly the wear rate were affected values positively [53]. The analysis of the coefficient of friction (CoF) data (Fig. 6 (c)) indicates that cryogenic treatment effectively reduces friction, with deeper cryogenic treatments (e.g., -180CT) providing the most significant improvements. At room temperature, the CoF decreases progressively with more intense cryogenic treatment, likely due to microstructural changes like the transformation of retained austenite to martensite and precipitation of fine carbides, which enhance surface hardness and reduce adhesion [54]. Additionally, a slight decrease in CoF with increasing load at room temperature suggests that higher loads may flatten surface asperities, promoting smoother sliding.

At 350°C, the CoF is notably lower across all samples compared to room temperature, potentially due to the formation of protective oxide layers that act as lubricious barriers. These thermally induced layers, combined with the enhanced hardness from cryogenic treatment, likely contribute to smoother sliding and

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reduced friction [55,56]. The -180CT treatment shows the best overall performance at high temperatures and under high loads, making it a promising choice for high-temperature applications where low friction and high wear resistance are essential.

SEM images of the wear regions of the samples subjected to HT and -180CT after the wear tests conducted at room temperature and 350 °C are given in **Fig. 7**. The higher hardness provided a higher resistance against wear, as expected from the often-valid relation between hardness and wear resistance [22,34]. The increment in the carbide density and finer carbides that are observed in cryogenically treated samples helps to achieve the micro-stress distribution of the material, which results in higher wear resistance [54].

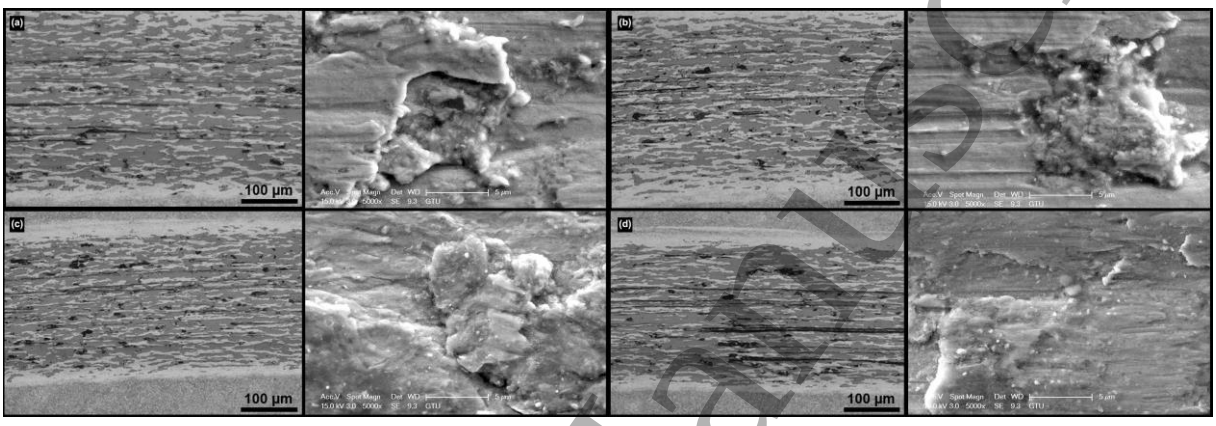


Fig. 7 SEM images of the worn surfaces of **(a)** HT and **(b)** DCT at room temperature, **(c)** HT and **(d)** DCT at 350 °C under 25 N load

The wear mechanism of all the samples exhibited delamination and abrasion under high magnification. Heat treated samples had broken areas and rough surfaces both at room temperature and 350 °C. The sub-surface deformation caused crack initiation and subsequent propagation, resulting in delamination occurring by repetitious plastic flow [57]. Some of the particles disintegrated by delamination formed deposits by ejecting from the worn surface, and the other part formed grooves with the abrasive effect [14]. Delamination was reduced by cryogenic treatment. Most of the spalled particles were thrown out from the tribopair in the DCT due to increased hardness by eliminating the retained austenite. Accordingly, the cryogenically treated samples had shallower grooves than the only heat-treated samples.

4. Conclusions

In the present study, the effect of cryogenic treatment on DIN 1.2888, a hot work tool steel used as die in forging processes, was investigated. To achieve this, DIN 1.2888 steel samples were first heat-treated at 1150 °C, then cooled to room temperature, -100, -140, and -180 °C, followed by double tempering at 650 °C. Microstructural and mechanical properties of the samples at both room temperature and working temperature were examined. According to the results obtained, the following were concluded.

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- When the DIN 1.2888 hot work tool steel was subjected to heat treatment and double tempering, a small amount of retained austenite existed in the structure. On the other hand, after cryogenic treatment, the retained austenite decreased to a non-detectable level.
- Microstructural analysis revealed that cryogenic treatment increased carbide density from 0.751 to 3.609 vol.%. This increase in carbide density depended on the high cobalt content in DIN 1.2888 steel, which promoted carbide formation, even though cobalt was not a carbide former.
- A slight increase in hardness, from 507 HV in the heat-treated sample to 529 HV in the sample treated at -180 °C, was observed with decreasing cryogenic treatment temperature. This increase corresponded with a more homogeneous carbide distribution and a lower amount of retained austenite.
- The impact energy at both room temperature and 350 °C increased as the cryogenic treatment temperature decreased, where finer carbide precipitation played a positive role in the toughness of the steel.
- The cryogenic treatment had a positive effect on the wear rates of the steel at room temperature. At 350 °C, lower wear rates were observed, likely due to the oxidation of metal forming a protective surface layer, and the formation of fine carbides that resisted the abrasive wear.

Overall, DIN 1.2888 steel showed promise as a hot work steel for die applications due to its superior mechanical properties at elevated temperatures. Minimizing retained austenite and optimizing carbide formation with a homogeneous distribution improved the mechanical and microstructural characteristics of the steel.

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Conflict of Interest

On behalf of all authors, the corresponding author states that there is no conflict of interest. The authors have no relevant financial or non-financial interests to disclose.

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