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Cryogenic treatment of an AISI D2 steel: the role of isothermal martensite formation and “martensite conditioning”

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ABSTRACT
Vibrating sample magnetometry, VSM, was applied to follow the evolution of the austenite-to-martensite transformation in AISI D2 tool steel at cryogenic temperatures and to determine the associated changes in magnetic hardness. Data shows that the transformation can progress during cooling from room temperature (RT) to -193°C and, more surprising, during re-heating from -193°C to RT, indicating the formation of, so called, isothermal or thermally activated martensite. The thermally activated conversion of austenite into martensite occurred in the temperature range -33°C ≤ T ≤ -173°C and was fastest at -113°C. Additionally, VSM revealed that the magnetic hardness decreases in correspondence with the formation of martensite. Data is interpreted in terms of continuous growth of the (ferromagnetic) martensite units at cryogenic temperatures that reduces the number of (paramagnetic) austenite regions retained in the material. Eventually, the work is put in perspective to arrive at an understanding of the effect of cryogenic treatment on the microstructure and performance of martensitic steels.

KEYWORDS: martensitic transformation; cryogenic treatment; tool steel; retained austenite; magnetometry

1. INTRODUCTION
Conventional heat treatment of martensitic steels consists in three steps: austenitization, quenching (i.e. fast cooling) to room temperature and tempering.

In the forties of the previous century, Cohen and co-workers [1] in the USA and Gulyaev [2] in the USSR, independently conceived the idea of subjecting steel to a cooling cycle to cryogenic temperatures between quenching and tempering to improve its properties. These activities followed

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up on two observations by Mathews [3]: 1) retained austenite is always present in medium- and high-carbon steels after quenching to room temperature; 2) immersion of steel in cryogenic liquids may reduce the fraction of retained austenite and increase the hardness.

In the mid seventies, Barron and co-workers gave the proof of concept that the properties of steel, in particular the resistance to wear, can be improved importantly by cooling the material to cryogenic temperatures [4,5], so-called deep cryogenic treatment, DCT. Barron investigated the effect of various DCT parameters on wear resistance, among which, the cooling rate to cryogenic temperature, the lowest temperature reached during the treatment and the holding time at cryogenic temperature, and showed that the holding time (at -196°C [4]) plays an important role in DCT.

For half a century, scientists have debated the mechanisms taking place in steel at cryogenic temperatures and discussed how these mechanisms influence the response of steel to wear. Despite these efforts, the metallurgical understanding of the microstructural changes involved in DCT of steel is still incomplete. The effect of cryogenic treatment, and in particular of the holding time at cryogenic temperature, on the microstructure, properties and performance of AISI D2 tool steel, which in Barron’s work [6] is the steel that appeared most responsive to cryogenic treatment, is representative for the debate.

In 1994, Meng et al. [7] claimed that DCT modifies the crystallography of the transition carbides that form during low temperature tempering (180°C, 0.6 to 1.8 ks) and that this modification is crucial to understand the enhanced wear resistance of DCT samples. In this work, carbides were < 40 nm in their largest dimension and their crystallographic structure was hexagonal close-packed (ε carbides) in conventionally treated samples and orthorhombic (η carbides) in samples subjected to DCT. Unfortunately, the mechanism responsible for a modification in the precipitation behaviour by DCT remained unaddressed.

In 1996/1997, Collins [8,9] suggested that two processes take place during DCT: i) transformation of austenite into martensite, which is presumed independent of time and completed during cooling to -100°C, and ii) low temperature “conditioning of the martensite”, an unidentified time-dependent process that would be controlled by carbon diffusion at cryogenic temperatures and would promote the precipitation of finer and more numerous carbides during tempering. Experimental evidence in support of “martensite conditioning” was the observation that the holding time at -196°C during DCT has a significant influence on various materials aspects evaluated after one-stage tempering at 200°C, including the hardness, the number of “fine” carbides revealed by optical microscopy (i.e. approximatively 1-5 μm large) and the resistance to wear.

Approximately a decade later, Das et al. presented additional experimental evidence in support of martensite conditioning [10,11]. They focused on the carbides larger than 0.1 μm and performed all observations after tempering twice at 210°C for 2 hr. The work showed that by holding the material
at -196°C it is possible to significantly modify the carbide population and the resistance of the material to wear; the fraction of retained austenite is negligible after DCT to -150°C and below, independently of the holding time. Das et al. also reported that DCT does not modify the type of the carbides, identified as M₇C₃ and M₂₃C₆, present in the material. Farina et al. were unable to reproduce the results in Ref. [10] and suggested that all the carbides observed by Das et al. had actually formed at high temperature, prior to quenching, implying that the conclusions by Das et al., stating that the carbide distribution changes upon DCT, were unjustified [12].

A few years afterward, Gavriljuk et al. studied the tempering reactions in detail and reported that DCT can modify the precipitation sequence during tempering, i.e. the sequence of carbide crystal structure [13]. In conventionally treated material, carbide precipitation starts in the temperature interval 100-200°C with the formation of nm-sized transition ε carbides, which are converted into cementite (10-500 nm) at approx. 300°C. Tempering at 500°C yields partial dissolution of the cementite with simultaneous formation of very fine, < 10 nm in dimension, alloying-element carbides. DCT consisted in instantaneous cooling to -196°C followed by 24 hr at either -196°C or -150°C. It was observed that DCT suppresses the formation of alloying-element carbides during tempering at 500°C. Moreover, depending on the holding temperature during DCT, the treatment either promotes the formation of transition η carbides in addition to ε (-196°C), or suppresses the formation of both (-150°C). According to Gavriljuk et al., the key factor to understand DCT is the formation of a special type of martensite at cryogenic temperature [14]. This martensite forms with time dependent kinetics and is labelled isothermal. The formation of isothermal martensite would be associated with the formation of carbon clusters during DCT and these clusters would influence the precipitation reactions during tempering. Furthermore, Gavriljuk et al. also indicated that isothermal martensite formation is not possible at -196°C, which appears in conflict with the observation that DCT, consisting of instantaneous cooling to this temperature followed by isothermal holding for 24 hr (and subsequent heating to room temperature) has a measurable effect on the tempering reactions.

In the last 15 years, the activities in our research team have aimed at resolving the fragmentary and partly conflicting interpretations by focusing on the time-dependent nature of martensitic transformation in steel at cryogenic temperatures [15-18]. The present contribution addresses time-dependent martensite formation in AISI D2 steel. The aim of the work is twofold. Firstly, the work aims at identifying the temperature range where isothermal martensite occurs in D2. Secondly, it aims at extracting information on the evolution of the material structure during DCT by analyzing changes in the magnetic hardness. Stojko reported variations in the magnetic hardness of high carbon steels during DCT [18] and speculated that these changes could be related to the movement of carbon atoms within the martensite, which would eventually reveal the formation of carbon clusters, if any.
2. MATERIALS AND METHODS

The material under investigation was a rod, Ø 10 mm, AISI D2 cold work tool steel (1.55%C, 0.3%Si, 0.4%Mn, 11.3%Cr, 0.8%Mo, 0.8%V) supplied by Uddeholm in soft annealed condition. The material was austenitized, cooled to room temperature, stored, and finally subjected to investigation. Two types of investigations were conducted: 1) in situ vibrating sample magnetometry, VSM, to reveal the evolution of martensite formation at cryogenic temperatures (cryo-VSM); 2) ex situ room temperature VSM to study the evolution of the magnetic properties of the material vs. variations in its internal structure (RT-VSM).

For the two types of investigations, two material conditions were applied, labelled as Q1000 and Q1030 after the austenitization temperature. These conditions correspond to:

1) Q1000: austenitization in Ar at 1000°C for 30 minutes, followed by cooling in Ar (average cooling rate 25°C/min, and 120°C/min for T>550°C), storage at room temperature for 1-3 months prior to investigation;

2) Q1030: austenitization in vacuum at 1030°C for 30 minutes, followed by gas quenching in N₂ (average cooling rate 120°C/min, 4 bars), tempering and subsequent investigation.

Magnetometry is a well-established technique to qualitatively determine the fraction of retained austenite in steel [15-19]. In steel, austenite is paramagnetic, while b.c.c. / b.c.t. phases (i.e. ferrite, and martensite) are ferromagnetic. Carbides, which are abundant in AISI D2, can be ferromagnetic or paramagnetic. As a rule of thumb, only iron-based carbides are ferromagnetic. Accordingly, only ferromagnetic phases contribute to the signal that is measured with magnetometry. Under the assumption that the fraction of carbides does not change during the thermal cycle and that diffusion-controlled transformation of austenite into a b.c.c. phase is not possible, an increase in magnetization, ΔM, can be directly attributed to the transformation of austenite into martensite.

VSM was performed with a Lake Shore Cryotronics 7407 vibrating sample magnetometer equipped with a Janis superTran–VP cryostat. Iron 99.99% plates, 0.2 mm thick, 2.2x2.2 mm² were used for baseline calibration of the magnetic moment. The magnetization of iron, M_{Fe}, was taken equal to 217.6 emu/g [20].

For Cryo-VSM of Q1000 0.3-mm-thick 2.2x2.2 mm² plates were investigated. The thermal cycle consisted of isochronal cooling to -193°C at 15°C/min, followed by heating at the same rate to various (one per sample) isothermal holding temperatures, storage at the holding temperature for 24 hours, and finally re-heating to room temperature at 15°C/min. Two samples were directly re-heated to room temperature without interruption for an isothermal storage at an intermediate temperature. The magnetic properties were recorded as follows. During cooling, heating and holding, a constant magnetic field of 1 T was applied to measure the magnetization of the samples at saturation, M. At the beginning of each thermal cycle, the magnetic hysteresis curve was recorded by varying the
magnetic field between +1 T and -1 T (step size 1.33 mT, 0.4 s acquisition time per step). This yielded both the saturation magnetisation, $M$, and the coercivity, $H_c$, i.e. the magnetic hardness. The latter was obtained by subtracting the coercivity measured for a reference iron sample, $H_{cFe}$, from the value measured for the D2 samples, $H_{cm}$:

$$H_c = H_{cm} - H_{cFe}$$  \[1\]

In both cases, the coercivity values were determined by linear interpolation among data in the range from +0.02 T to -0.02 T. Eq. 1 takes into account that the instrumental coercivity is not negligible relative to $H_{cm}$, while the intrinsic magnetization of iron can be neglected.

Samples for RT-VSM on Q1030 were 0.7-mm-thick, Ø 3 mm disks, which were subjected to tempering. Tempering consisted of isochronal heating at 6ºC/min to various temperatures followed by quenching in water. Identical tempering conditions were applied in a parallel study [21], where the microstructure evolution on tempering was investigated in situ with thermal analysis and synchrotron X-Ray Diffraction. Magnetic hysteresis curves were acquired before and after tempering, applying the abovementioned acquisition parameters.

Fig. 1 presents the thermal cycles for the two types of investigation. In Fig. 1, stars indicate measurements of the magnetic hysteresis curve. The continuous line refers to continuous measurement of the magnetization at saturation $M$. Dotted lines are placed were no measurements were performed.

![Fig. 1 – Illustration of the applied thermal cycles: a) example for cryo-VSM where the sample is held 24h at -133ºC; b) tempering cycles for RT-VSM.](image)

In addition to VSM, the Q1000 material was characterized by scanning electron microscopy, SEM, Vickers hardness measurements, and X-Ray Diffraction, XRD. SEM was performed approximately 30 months after quenching and storage at RT. The instrument used was a Phenom Pro X, table top SEM, applying an acceleration voltage of 15 kV; the sample was a Ø10 mm disk that was electropolished in A2 electrolyte from Struers at 25ºC applying a voltage of 20 V for 20 s. Hardness
measurements were performed using a Future Tech model FM-700 micro-hardness tester, equipped with a Vickers indenter applying a load of 0.5 kg and a dwell time of 10 s. Each sample was characterized with 15 independent indentations. Care was taken during hardness measurements to avoid large carbides. XRD was performed in transmission at BESSY II synchrotron facility, EDDI beamline, approx. 3 months after quenching. The primary slits (which define the interaction area for transmission measurements) were 0.5 x 1.5 mm². The scattering angle 2θ was fixed at 14°. The fraction of retained austenite was determined from the relative intensity of the 200 and the 311 reflections of austenite and the 200, 211 and 321 reflection of martensite. For additional details of the quantification procedure and the experimental setup at EDDI, the reader is referred to Ref. [22]. XRD and hardness measurements were performed on the same samples investigated by VSM.

3. RESULTS AND DISCUSSION

3.1. SEM, XRD and hardness

![Image](image.png)

*Fig. 2 – Micrograph showing the microstructure of the material after quenching from 1000°C (Q1000).*

The microstructure of the Q1000 material is presented in Fig. 2. The microstructure consists of carbides dispersed in a martensitic matrix. It cannot be excluded that a minor fraction of bainite is present and that this structure is not resolved at the magnification considered in the present study [23]. The carbides are 0.1 to 30 μm large. The only carbide identified with XRD is M₇C₃. It is remarked that the material was not tempered prior to investigation. Evidently, all the carbides that can be observed in Fig. 2 must have formed during hardening prior to reaching room temperature.
Additionally, XRD revealed the presence of retained austenite, RA. The fraction of RA was quantified at 5.4±0.6% for the material in the as-quenched condition. Conversely, the fraction of RA was 4.7% when averaged over all DCT samples and 4.4±0.5% for the samples isothermally held for 24 h in the temperature interval \(-173^\circ C \leq T \leq -33^\circ C\), where isothermal martensite forms (see below). It follows that XRD indicates a minor, but measurable, reduction in the fraction of RA by DCT. The hardness was 760±20 HV. DCT samples were harder when compared to the as-quenched material and reached maximally 790±20 HV for the samples treated at \(-133^\circ C\).

### 3.2. Cryo-VSM

The results of the cryo-VSM investigations are presented in Fig. 3. The black line in Fig. 3 shows the evolution of the magnetization of the samples during cooling (averaged over all samples) relative to the magnetization of the samples at the beginning of the investigation, \(\Delta M/M_0\). The grey curve guides the reader through the increase in \(M\), as measured after isothermal holding for 24 h at different temperatures. Data points and error bars refer to the average value and standard deviation of 3 independent measurements: i) \(\Delta M/M_0\) evaluated from the last \(M\) value measured on heating from \(-193^\circ C\) to the holding temperature and the first \(M\) value measured during heating from the holding temperature to room temperature; ii) \(\Delta M/M_0\) evaluated from the hysteresis curves recorded at the beginning and at the end of the holding step; iii) \(\Delta M/M_0\) evaluated based on the \(M\) values recorded continuously during the holding step. Variations in the magnetization during the heating steps were below the experimental accuracy of the instrument (<0.2% for this type of experiment).

![Fig.3 – Change of the magnetization of D2 steel for Q1000 during DCT. The black line reports the formation of martensite during continuous cooling to \(-193^\circ C\). The grey data and curve illustrate the additional martensite formation obtained during isothermal holding of the material for 24 h at the indicated temperatures after heating from \(-193^\circ C\).](image-url)

On cooling to \(-193^\circ C\), Fig. 3 shows that the magnetization of the material increases by approx. 1.2%
and increases fastest in the temperature interval -80°C to -120°C (at the bending point at approx. -100 °C). An additional increase in magnetization is measured during isothermal holding of the samples for 24 h at various temperatures after heating from -193°C. This additional increase in \( M \) depends strongly on the isothermal holding temperature and is maximally 0.7% at -113°C.

Direct conversion of \( \Delta M/M_0 \) into the fraction of martensite, \( \Delta f_{\alpha'} \), is not possible, because the fraction of carbide phases present and their magnetic properties are unknown. A rough estimate can be made under the assumption that the effect of the presence of alloying elements in the austenite matrix is negligible. Accordingly:

\[
\Delta f_{\alpha'} = \frac{\Delta M/M_0}{M_{Fe}} \cdot M_0 \quad [2]
\]

where \( M_0 \) equals 148.9 emu/g; it follows: \( \Delta f_{\alpha'} \approx 0.7 \cdot \Delta M/M_0 \).

**Fig. 4** – Evolution of the magnetic hardness of the samples plotted versus the isothermal holding temperature. The two data sets refer to \( H_c \) measured at the end relative to the beginning of the DCT cycle (overall cycle) and during the 24 h isothermal holding (isothermal part), respectively.

VSM also provided the coercivity, \( H_c \), of the material at various stages of DCT. Data is presented in Fig. 4, and indicates that DCT promotes magnetic softening of the samples. Two contributions to magnetic softening can be distinguished: a decrease of \( H_c \) by approx. 1% upon cooling to -193°C and following reheating to RT; this corresponds to the baseline for the black curve in Fig. 3. The isothermal contribution, represented by the grey data, that is 0.5% at the isothermal temperature where the phenomenon is most pronounced: -113°C. Interpretation of data in Fig. 3 is not straightforward. Evidently, an increase in magnetization during DCT (Fig. 2) is associated with a reduction in \( H_c \) (Fig. 3) and the increase in \( M \) and the reduction in \( H_c \) are more pronounced if more martensite forms.

### 3.3. RT-VSM and implications for cryo-data

RT-VSM investigation was implemented to provide a background for the interpretation of the Cryo-
VSM results. The following 4 stages of tempering of D2 steel can be distinguished for isochronal heating from RT to 700°C at 6°C/min [21]: i) precipitation of transition carbides and simultaneous loss of martensite tetragonality (RT-220°C); ii) precipitation of cementite and simultaneous relaxation of phase specific micro-stresses (220-350°C); iii) precipitation of secondary alloying element carbides and simultaneous depletion of austenite in carbon (450-700°C); iv) decomposition of retained austenite and simultaneous relaxation of the martensite structure (570-670°C).†

The evolution of the magnetic properties at the various tempering stage is illustrated in Fig. 5.

Fig. 5 – Evolution of the magnetic properties of D2 steel, Q1030, versus tempering temperature. Tempering consisted in continuous heating of the sample at 6°C/min to the indicated temperatures followed by quenching in water.

Fig. 5 shows that the first 3 tempering stages have no measureable influence on M. This indicates that, within experimental accuracy (±0.7% [18]), the contribution of the Fe atoms to the magnetization of the samples does not change, irrespective of the Fe atoms’ presence in the martensite matrix or in iron-based carbides. Then, upon tempering at a temperature above 550°C the magnetization increases drastically, in accordance with the decomposition of (paramagnetic) retained austenite into (ferromagnetic) ferrite and alloyed carbides in tempering stage iv).

In relation to the coercivity (magnetic hardness), three changes in $H_c$ could be identified. The first upon tempering in stage ii), the second in stage iii) and the third, and most significant, in stage iv). Obviously, the first stage of tempering has no measurable effect on $H_c$. Apparently, the rejection of carbon from martensite and the associated loss of martensite’s tetragonality, has no effect on the magnetic hardness. Similarly, the various phenomena occurring during the second and third stage of tempering have a limited effect. The magnetic hardness decreases significantly during decomposition of retained austenite and simultaneous rearrangement of the martensite internal structure. This is

† Ref. [21] indicated that stage (iv) takes place in the range 600-650°C. More precisely, data indicated this stage starts at 570°C and ends at 670°C. Therefore, in the present work 570-670°C is taken as temperature interval.
interpreted in the following.
Firstly, prior to the application of a magnetic field, a large number of magnetic domains exists within the ferromagnetic regions of the sample (i.e. martensite units and iron carbides). When a magnetic field is applied, the domains align to the external magnetic field to minimize the internal energy. The alignment takes place either by growth of pre-existing magnetic domains, or by nucleation and subsequent growth of new domains. When the field is removed, the force for alignment is lost, and new randomly oriented domains can nucleate and grow. In ferromagnetic materials, some domains remain and a magnetic field in the reverse direction is required to demagnetize the sample. The coercivity is a measure of the intensity of this field. Evidently, the mobility of the magnetic domains is determined by the material’s microstructure. Growth of a magnetic domain can be difficult in a complex microstructure as in hardened D2 steel, where the ferromagnetic areas are confined by non-magnetic / paramagnetic regions. Similarly, demagnetization can be difficult because a large number of magnetic domains needs to nucleate before the magnetization of the sample is lost: regions of retained austenite and large chromium carbides act as (strong) barriers for the movement of the magnetic domains, thus contributing to magnetic hardness.

Returning to DCT, the above interpretation is used to interpret data in Fig. 4. Firstly, the applied method can neither provide information on the evolution of the tetragonality in martensite nor on the accompanying formation of carbon clusters during DCT. Rather, the evolution of $H_c$ during DCT indicates that the mobility of the magnetic domains is enhanced by DCT through the annihilation of (paramagnetic) retained austenite regions during continuous growth of the martensite units at cryogenic temperatures. This interpretation explains the correlation between the increase in magnetization during DCT as illustrated in Fig. 3 and the magnetic softening reported in Fig. 4, as both phenomena are a consequence of the austenite-to-martensite transformation. Following the above interpretation, data in the present work supports our recent research activities in ferrous martensite, which indicated that martensite units do not (necessarily) grow instantaneously to full dimension upon their nucleation. Rather, along with the nucleation of new martensite, existing martensite grows during continuous cooling from RT to cryogenic temperatures as well as during isothermal holding of the material in the cryogenic temperature region [16].

### 3.4. Results in perspective: can martensite formation explain it all?

In relation to the effect of DCT, the present data adds information on the kinetics of martensite formation at cryogenic temperatures in D2 steel and suggests that DCT not only reduces the fraction of retained austenite in the material, but also the total number of retained austenite regions. How this influences the evolution of carbide precipitation on tempering and, subsequently, the performances of the material in service remains to be investigated. Nevertheless, existing theories on the effect of
DCT on the steel microstructure and on the role of time in DCT can be evaluated when the present data are put in context with the literature.

According to VSM data, isothermal martensite formation is possible in D2 during DCT, but the transformation is extremely sluggish at temperatures lower than -170°C. This is in agreement with Gavriljuk et al.’s work. However, an effect of the holding time at -196°C on the microstructure and wear resistance of D2 steel was reported by both Collins and Dormer [9] and Das et al. [11]. Similarly, Barron reported an effect of holding time at -196°C on the wear resistance of another alloyed steel, AISI T4. Farina et al. pointed out that the carbides observed by Das et al. could not have formed during tempering at 210°C and that this conclusion should be considered incorrect and unjustified. The argumentation of Farina et al. is consistent with more recent investigations by Gavriljuk et al. [13] and Pellizzari et al. [21], as well as the data presented in Fig.2 in the present investigation. Following the same line of argumentation, tempering at 200°C cannot modify the carbide population at the scale investigated by Collins and Dormer. Nevertheless, the work of Barron still stands, and points towards the possibility that martensite formation is only a part of the story and a second, unrevealed mechanism (martensite conditioning?) is in place at -196°C.

On the other hand, the austenitization temperature selected in the present investigation was in the lowest range of specifications for D2 and the material was stored a long time at room temperature prior to DCT. The first condition promotes a particularly low fraction of RA, while the second stabilizes the RA against transformation into martensite during DCT. It cannot be excluded that a austenite transforms isothermally into martensite even at -196°C when more favorable conditions for the austenite-to-martensite transformation during DCT apply.

Additionally, at -196°C the driving force for martensite formation is very large; as a rule of thumb, a large driving force is associated with favorable conditions for nucleation. It could be speculated that by holding the material at -196°C it is possible to promote the (isothermal) formation of martensite nuclei; these nuclei are unable to grow at -196°C because the thermal energy is insufficient at this temperature. It follows that the fraction transformed at -196°C is minor and the process remains unrevealed by VSM. Growth occurs only upon heating, at a temperature higher than, say -170°C, and promotes additional martensite formation by autocatalysis. First at this point, the process is detected. Clearly, the longer the holding time at -196°C, the more numerous the martensite nuclei and the larger the fraction of martensite that forms during heating at fixed heating conditions.

Isothermal (i.e. thermally activated) nucleation of martensite is not novel and for many years has been considered as the only process responsible for the isothermal formation of martensite in several materials [24]. Recently, a very comprehensive investigation of the subject was performed by Kakeshita et al. [25-26]. These authors demonstrated that the isothermal nucleation of martensite is possible in ferrous as well as non-ferrous materials at temperatures as low as -269°C [25]. Eventually,
such martensite nucleation could correspond to “martensite conditioning” in the field of DCT.

4. CONCLUSIONS

Vibrating sample magnetometry was applied to follow in situ the evolution of the austenite-to-martensite transformation at cryogenic temperature in D2 steel and to reveal additional information on phenomena possibly associated with this transformation.

The investigations reveal that martensite forms on (moderately fast) cooling from room temperature to -193ºC as well as on re-heating from -193ºC to room temperature during a long isothermal holding step within the temperature interval -173ºC ≤ T ≤ -33ºC.

The formation of martensite at cryogenic temperatures is associated with magnetic softening of the material. Such magnetic softening is interpreted in terms of a reduction of the number of austenite regions retained in the material.

Magnetic hardness does not provide information on the presence of carbon dissolved in the martensite structure. Rather, it appears to reflect (mainly) the presence of retained austenite in the microstructure.

The present work does not provide sufficient information to firmly establish whether all observations in the literature can be ascribed to the austenite-to-martensite transformation, or whether a second mechanism, labelled martensite conditioning, is active at cryogenic temperatures. In this work we speculate that martensite conditioning is nothing else than (isothermal) nucleation of martensite, which lacks the thermal energy to grow.

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