Influences of Thermomechanical Treatment and Nb Micro-alloying on the Hardenability of Ultra-High Strength Steels

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For the construction of mobile crane booms, ultra-high strength steels produced *via* thermomechanical processing (TMP) have widely substituted steels fabricated through the conventional quenching and tempering (Q+T) route. A strong deformation of the austenite grain during hot rolling followed by direct quenching (DQ) offers benefits in terms of strength and toughness. To guarantee an optimal through-hardening, alloying elements retarding the γ to α transformation are used. To explore the influence of the processing route on the critical cooling rate and the hardenability, hot deformation tests were performed on a deformation dilatometer. Different cooling rates were applied after deformation corresponding to two different rolling cycles with varying finish rolling temperatures (FRTs). The obtained hardness values were compared to those received through conventional quenching after austenitization. These investigations conducted on three steels with varying micro-alloying contents showed that Nb in combination with TMP raises strength significantly, and promotes a bainitic and ferritic transformation in solid solution. When applying low FRTs and in combination with other micro-alloying elements, NbC coarsens and reduces the effect of precipitation hardening.

<https://doi.org/10.1007/s11661-019-05235-8>
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I. INTRODUCTION

IN the recent decades, thermomechanical processing with subsequent direct quenching (TM-DQ) has generated an attractive way to produce ultra-high-strength steels.^[1–6] Their high demand arises from an improved payload-to-weight ratio for mobile crane applications. The requirements of the user and manufacturer were met through both economic and efficient production routes. The highest strength and proficient toughness as well as good weldability are realized with these steel grades.^[7] This is accomplished by the systematic combination of an alloying and a rolling schedule in which the recrystallization processes are tuned to achieve a fine-grained austenitic microstructure prior to martensitic transformation. The influence of the condition of the prior austenite grain (PAG) on the properties of the emerging steel product has been investigated in depth.^[8,9] The dimension of the emerging substructure

such as packets and blocks^[10] is proportional to the size of the PAGs.^[11] A decrease in the PAG size results in smaller martensitic constituents.^[12] Consequently, the Hall–Petch correlation can also be extended on martensitic microstructures. A decrease of the size of the prior γ -grain promotes a strength improvement through the smaller dimensions of martensite. The number of barriers for the movement of dislocations increases, which is followed by a rise in strength. Furthermore, rolling in the nonrecrystallization γ -regime (T_{NR}) promotes an elevated number of nucleation sites for the following γ to α transformation enhancing the grain refining.^[4,8,13] It has been well established that decreasing the finish rolling temperature improves both strength and toughness.^[6,14–16] For a further improvement in the strength of these steels, a martensitic transformation is performed by instant quenching.^[4,8,17] Therefore, alloying elements such as Mn, Cr, Mo, Si, and B come to application, which, in dissolved condition, decelerate the diffusion-controlled γ to α transformation and enable full martensitic strength.^[18] During cooling, Mo retards the pro-eutectoid ferrite and perlite transformation^[19] and prevents the Nb-precipitation to Nb(CN).^[20] Si, Mn, and Cr decelerate the perlite and bainite transformations through different mechanisms. The role of Si is based on its inhibition to form carbides due to its low solubility in cementite.^[18] The influence of Cr on the hardenability is less pronounced than the effect of Mn, as Cr requires higher additions to decrease the critical

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Manuscript submitted December 14, 2018.

Article published online April 29, 2019

cooling rate,^[21] yet, like Mo, it counteracts the softening during tempering.^[18,22] The strongest impact on the hardenability possesses B, which already influences the γ to α transformation with very low alloying contents.^[23–25]

It is generally recognized that the critical cooling rate and M_S are a function of alloying contents and elements.^[20,26] The influences of several alloying elements retarding the γ to α transformation have been broadly investigated. Moreover, the influences of the rolling conditions and FRTs on the mechanical properties of the martensitic steel product are largely scientifically explored.^[8,16,17] Nevertheless, there is a lack of research, if the rolling conditions and especially the FRTs influence both M_S and hardening characteristics. To investigate the effect of the FRT on the hardenability, a rolling scenario with two different FRTs was performed on a Deformation Dilatometer Bähr 805 A/D. In order to examine, if micro-alloying possesses an effect on the hardenability in combination with different rolling parameters, three different steels were analyzed. In detail, two hardenable steels with different Niobium contents were compared to a hardenable and temper-resistant steel in order to investigate the hardenability depending on different production routes. Steels processed *via* TMP still are in competition to steels produced *via* the classical quenching and tempering (Q+T) route, as manufacturers promote the improved impact strength and better isotropy concerning their mechanical properties of Q+T steels.^[14,27–30] Therefore, we further intend to compare the hardenability of TM steels to that of the steels produced *via* a traditional Q+T route. In addition, CCT diagrams of the investigated steels will demonstrate the influences of the processing route and the Nb content on the γ to α phase transformation and the resulting hardness values.

II. MATERIALS AND EXPERIMENTAL PROCEDURE

A. Materials Investigated

The investigations on the influence of the TM processing on the hardenability have been conducted on three different steels. The corresponding melts were produced in an induction furnace according to the composition listed in Table I. The steels represent hardenable steels with a carbon content of 0.17 pct. Steels 1 and 2 are alloyed with Mn, Si, and B retarding the γ to α transformation to ensure a martensitic microstructure under corresponding cooling conditions. These steels only differ in their Nb content in order to examine the influence of Nb on the hardenability after TMP. In Steel 3, the Mn content is reduced and is

yet alloyed with the micro-alloying element (MAE) V, optimized for a TM-processing route. In addition, steel 3 is dispensed with B, however, modified with elevated contents of Cr, Ni, Mo, and Cu compensating for the softening during tempering.^[31]

B. Experimental Procedure and Sample Preparation

To prepare the material for the investigations, the raw material was prerolled, and dilatometer samples were extracted *via* wire-eroding with dimensions of 10 mm length and a diameter of 5 mm. In the following, tungsten platelets were utilized in order to reduce the thermal conduction between the sample and the deformation stamp, and a “type S” thermocouple was mounted on the sample to determine the sample temperature during deformation. To reproduce a TM-rolling scheme, the deformation sequence according to Table II was performed using a deformation Dilatometer Bähr 805 A/D. Two different FRT were studied, the deformation program contained a nominal overall compression of $\varphi = 1.0$. After a solution annealing of 5 minutes at 1250 °C, the subsequent deformations were split into 5 passes, starting at a temperature of 1000 °C and finishing at 875 °C (rolling scenario FRT 1), and starting at 1075 °C, respectively, aiming for an FRT of 950 °C (rolling scenario FRT 2). The samples were quenched after deformation and an additional holding time of 3 seconds in order to portray a realistic representation of an industrial TM-rolling process, as quenching after the finish rolling steps occurs after several seconds. Five different cooling rates were performed: $\lambda = 1, 3, 10, 30$, and 100 K/s.

To compare the hardenability of the steel samples to those quenched after re-austenitization (Q+T route), prerolled samples were subjected to an austenitization of 930 °C for 5 minutes. Subsequently, the samples were quenched with the same cooling rates implemented on the deformed samples. From the dilatation data, the transformation temperatures were analyzed through the length expansion $\Delta L/L$ occurring at the phase transformations.^[32,33] The *three-tangent method* was used to determine the 5 pct—start and 95 pct—end of the phase transformation. Only the beginning and the end of the first observed transformation during cooling were taken into account.

Prior to the hardness testing, the specimens were hot embedded and then ground with 320 to 4000 grit SiC paper for at least 30 seconds. Subsequently, the samples were polished with 3- μ m diamond paste for at least 3 minutes and with 1- μ m paste for 30 seconds. Five HV10 hardness measurements were performed on each sample. Care was taken to ensure that the hardness was extracted from the center of the sample. To reveal the

Table I. Chemical Compositions of the Steels Investigated [in Percent]

Steel/Composition	C	Si	Mn	Al	Cr	Ni	Mo	Cu	V	Nb	Ti	B
Steel 1	0.17	0.2	2.3	0.05	0.25			0.08			0.02	0.002
Steel 2 (+ Nb)	0.17	0.2	2.3	0.05	0.25			0.08		0.04	0.02	0.002
Steel 3 (+ MAE)	0.17	0.2	1.4	0.05	0.70	1.0	0.4	0.5	0.03	0.04	0.02	< 0.001

Table II. Deformation Sequence Performed at a Deformation Dilatometer Bähr 805 A/D to Investigate the Influences of Two Different FRTs on Hardenability

Deformation Step/Parameter	1	2	3	4	5
ϕ (–)	0.25	0.25	0.15	0.15	0.1
$\dot{\phi}$ (s ^{–1})	10	10	10	10	10
FRT 1: T (°C)	1000	950	925	900	875
FRT 2: T (°C)	1075	1025	1000	975	950
t (s)	2.5	2	1.5	1.5	(3)

PAGs, the samples were etched with a picric acid etchant. Their compositions and procedures are described in Reference 34. For the metallographic analysis of the PAGs, the equivalent grain diameters and the aspect ratio of the PAGs were determined using the image analysis software MIPAR™.^[35] For the microstructural analysis of the transformed microstructure, the samples were polished with a silicate polishing (*Struers OPS*) for 10 minutes prior to a finishing electrolytic ablation of 5 seconds with 35 V. Subsequently, the samples were dipped into a Nital etchant. The images of the microstructures were recorded using an optical microscope and a *FIB Versa FEI 3D DualBeam* scanning electron microscope (SEM). The characteristics of the emerging precipitates of NbC were evaluated by means of EDS analysis using an EDAX Octane Plus detector and the software package TEAM 4.3.

III. RESULTS

The effect of the process route on the hardenability is shown in Figure 1. Steel 1 exhibits no significant influence of the FRT on the hardening behavior, the hardness progressions dependent on the cooling rates are congruent. Nevertheless, the hardness is noticeably increased with lower FRT (448 HV10) compared to the FRT of 950 °C (433 HV10) and the sample quenched without deformation (435 HV10) at the highest cooling rate of 100 K/s. Steel 3 possesses the same phenomenon. The hardness at high cooling rates > 3 K/s is clearly elevated for the lower FRT of 875 °C and amounts 472 HV10 at a cooling rate of 100 K/s compared to 460 HV10 for the sample deformed with an FRT of 950 °C and 451 HV10 for the undeformed, plain quenched sample. However, at a cooling rate of 1 K/s, both the undeformed quenched sample and the compressed sample with an elevated FRT possess higher hardness values: 413 HV10 and 417 HV10, respectively, compared to 372 HV10 measured for the sample deformed with an FRT of 875 °C.

Figure 1 also contrasts the hardening behaviors of steel 1 and steel 2, and therefore the influence of Nb on the hardening behavior depending on the process route. The upper diagram compares the hardness values of steels 1 to 3 in a plain hardened condition dependent on the cooling rate. At low cooling rates < 10 K/s, the Nb-alloyed steel (2) only reaches low hardness levels below 350 HV10. A full martensitic hardness only is

reached at high cooling rates of 30 and 100 K/s. Regarding the latter, the hardness of the Nb steel (2) even exceeds that of the Nb-free variant (435 HV10) with a value of 448 HV10. The additional strength contribution through Nb is even more apparent in the deformed samples at both finish deformation temperatures of 875 °C and 950 °C. At high cooling rates of 100 K/s, the hardness gain through an addition of 0.04 pct Nb amounts + 41 HV10 (from 446 HV10 to 487 HV10) at an FRT of 875 °C (middle graph). This effect is repeated at the higher FRT of 950 °C (bottom diagram). An addition of 0.04 pct Nb results in a hardness plus of 43 HV10. At low cooling rates of 3 K/s, the higher Nb content seems to retard the γ to α transformation both at the FRTs of 875 °C and 950 °C. At the lowest FRT, martensitic hardness is even reached with a Nb content of 0.04 pct at a cooling rate of 1 K/s. The lower-temperature-deformed samples again attain higher hardness levels. At a cooling rate of 100 K/s, the samples deformed with an FRT of 950 °C reach hardness values of 433 HV10 (Nb-free) and 477 HV10 (Nb+), respectively, whereas the corresponding hardnesses at an FRT of 875 °C reach values of 446 HV10 (Nb-free) and 487 HV10 (Nb+).

Figure 2 shows the transformation start and finish temperatures of steels 1 and 2 and the resulting microstructures related to the adopted processing route and the implemented cooling rate. Steel 1 exhibits a balanced behavior concerning its phase-transformation temperatures with a slight upward trend toward the decreasing cooling rates. A martensitic microstructure is attained at cooling rates between 10 and 100 K/s, which is confirmed both through hardness measurements and metallographic analysis. At low cooling rates between 1 and 10 K/s, the microstructure contains both bainitic and martensitic segments in variable proportions. The Nb variant, steel 2, behaves differently. The martensite transformation temperatures are slightly reduced for steel 2 for all three process routes. At high cooling rates, a full martensitic microstructure is attained. However, the CCT diagrams (Figure 2) illustrate that, at low cooling rates, the γ to α transformation occurs at elevated temperatures. These values are significantly above the theoretical M_s (calculated from Reference 26) and suggests, considering the reduced hardness values as illustrated in Figure 1 that the ferritic and bainitic phases are forming. Figure 3 presents the metallographic analyses of steels 1 and 2 quenched at 10 K/s. Despite undergoing the same cooling conditions, steel 2 forms ferritic components after re-austenitization

Fig. 1—Effect of FRTs on the hardenability of steels 1 to 3 compared to quenching after previous austenitization for 5 min at 930 °C. A decreased FRT delivers higher hardness values for all steels. However, steel 3 exhibits a significant hardness decrease at low cooling rates compared to the higher FRT and the Q-route. Top graph: Nb retards the hardenability on the undeformed and simple quenched samples. Middle graph: with the increasing Nb content, higher hardness values are achieved at an FRT of 875 °C. Bottom graph: Nb influencing the final hardness and retarding the γ to α transformation at low cooling rates at an FRT of 950 °C.

(Figure 3(c)). After TMP with an FRT 875 °C, it becomes fully martensitic (Figure 3(b)).

Figure 4 displays the corresponding phase transformations and the resulting hardness values and microstructures of steel 3, dependent on the processing route. At medium cooling rates, the phase transformations occur at analogical temperatures for all three different processing routes. Starting with a M_s of ~ 400 °C at a cooling rate of 3 K/s with a descending trend to values of M_s ~ 370 °C at 30 K/s. However, this trend is disrupted at low cooling rates. The phase transformations are shifted toward the elevated temperatures to values of M_s ~ 484 °C (FRT 875 °C) and ~ 486 °C (FRT 950 °C) for a cooling rate of 1 K/s and to M_s ~ 417 °C (FRT 875 °C) and ~ 403 °C (FRT 950 °C) for a cooling rate of 100 K/s, respectively. Nevertheless, compared to steel 2, a fully martensitic microstructure is reached down to cooling rates of 10 K/s. It is shown that an increase in cooling rate above 10 K/s not necessarily increases the hardness of the respective production route.

Fig. 2—Phase-transformation temperatures of steels 1 and 2 depending on cooling rates and processing route. Whereas M_s and M_F are dependent on the chemical composition, the cooling rates necessary for a complete martensite formation differ between the processing routes. The M_s was calculated according to Ref. [26].

However, at cooling rates < 10 K/s, also sections of bainite or even ferritic phases (quenched sample at cooling rate of 1 K/s) appear. Figure 5 reveals selected PAGs (a to c) and the transformed microstructure (d to f) of steel 1 to 3 at a cooling rate of 100 K/s. Processed with an FRT of 875 °C, steel 1 exhibits still a coarse and globular PAG structure (Figure 5(a)). The corresponding transformed microstructure of steel 1 (Figure 5(d)) is by far coarser than the counterpart of steel 2 (Figure 5(e)). After re-austenitization, the PAGs of steel 2 and 3 very fine, as displayed in Figure 5(b) and (c). The quantification of the dimensions of the PAGs are listed in Table III. The size of the resulting α substructures, such as ferritic grains, bainitic laths, and martensitic blocks, are proportional to their originating γ -grains.^[8,9,11] According to Table III, the finest grains are achieved through TMP at a finish processing at 875 °C, whereas, as aforementioned, steel 2 and 3 show a tendency to produce the finest grains through re-austenitization at 930 °C (Figures 5(b) and (c)).

IV. DISCUSSION

In order to investigate the influence of the production route on the hardenability of UHSS strength, three steels with different alloying contents were subjected to varying cooling rates after applying different

Fig. 3—Microstructures of samples cooled at 10 K/s: (a) steel 1 after re-austenitization and quenching, steel 2 after (b) TMP with an FRT of 875 °C, and (c) quenching after re-austenitization. The microstructure of steel 1 is slightly coarsened compared to the Nb variant. Direct quenching of steel 2 with 10 K/s creates a martensitic microstructure, whereas from the quenching route, ferritic segments arise.

Fig. 4—Phase-transformation temperatures of steel 3. Effects of cooling rates and FRTs compared to plain quenching (Q) on the martensite start (M_s) and finish temperatures (M_F). The corresponding hardness values obtained after applying the different cooling rates are in correspondence with the obtained microstructures.

compressions on a deformation dilatometer. The dilatometer is a simple, yet, unique tool to bring industrial-scale processes down to laboratory dimensions. However, it cannot picture a whole rolling scenario. Further, the given nominal strains are distributed unequally over the deformed sample. Nevertheless, these points were considered through withdrawing both hardness and microstructural investigations consequently only from the middle of the deformed samples.

For all three steels investigated, the highest hardness values were attained at an FRT of 875 °C, compared to the samples compressed at an FRT of 950 °C and the plain-quenched (after an austenitization of 5 minutes at 930 °C) processing route. The influence of the processing route is less pronounced at steel 1. However, the steel grades alloyed with Nb (steel 2 and 3) show a significant increase in hardness when deformation is imposed prior to quenching. This effect can be attributed to the increased reduction in the nonrecrystallization regime,

as Nb shifts the T_{NR} to higher temperatures and promotes pancaking as demonstrated in Figure 6. This is in accordance to literature.^[2,6,8,14,16]

The comparison of the hardening behavior between the three steels emphasizes the significance of micro-alloying in combination with TMP concerning the achievable strengths. While the gain in strength through an addition of 0.04 pct Nb (steel 2) is only marginal for a conventional hardening after re-austenitization (+ 3 pct), TMP with subsequent DQ with 100 K/s gives a hardness plus of 9 pct (FRT 875 °C) and 10 pct (FRT 950 °C) compared to the Nb free variant steel 1. Furthermore, at lower cooling rates of <30 K/s, the hardenability of steel 1 and 2 diverge clearly. Whereas steel 1 possesses at a cooling rate of 1 K/s still martensitic constituents, steel 2 consists mainly of ferritic and bainitic components already at a cooling rate of 30 K/s. The reported effect that Nb micro-alloying and deformation promotes bainite formation by Kajjalainen *et al.*^[16] is in accordance with the present investigations. Moreover, it is shown that the accelerated γ to α transformation is further enhanced through both, lowering the FRT in TMP and quenching after re-austenitization, from which even ferritic components emerge. This can be attributed to the much smaller austenite grains after re-austenitization through the addition of Nb in steel 2, in which the austenite grains are pinned by precipitates prohibiting grain growth. A decrease in the austenite grain size shifts the time-temperature-transformation diagram to shorter times and promotes a ferritic transformation.^[36,37] Steel 3, however, is stronger alloyed with γ -stabilizer and does not exhibit this phenomenon. Whether Nb eliminates the effect of the γ to α transformation retarding elements in a solved or precipitated condition, needs to be clarified by TEM investigations. Nevertheless, the hardening behavior of steel 3 already gives an indication to this hypothesis. A surcharge of further alloying elements in steel 3 with additions of V, Cu, Mo and Ni, does not significantly increase the strength, but influence the hardenability. After the TM processing at an elevated FRT of 950 °C or after quenching from austenitization at 930 °C, the hardness does not considerably decrease after a low cooling rate of 1 K/s. Metallographic investigations reveal both, martensitic and bainitic

Fig. 5—Prior austenite grains (*a* through *c*) and transformed microstructures (*d* through *f*) of steels 1 to 3, at a cooling rate of 100 K/s. Despite a low FRT, the microstructure of steel 1 (*a* + *d*) is coarse and globular. Steels 2 (*b*) and 3 (*c*) clearly refine through re-austenitization. Steel 2 possesses a very fine lath structure after an FRT of 875 °C (*e*). Due to the described proportionality, the substructure originating from the PAG (*c*) of steel 3 is refined too (*f*).

Table III. Grain Sizes of the Prior Austenite Grains Achieved Through Different Processing Routes

Steel/ Process	Q 930 °C (μm)	FRT 875 °C (μm)	FRT 950 °C (μm)
1	26.7	11.2	21.8
2	8.9	9.0	13.0
3	8.1	21.3	11.7

microstructures. The decreased hardness after processing with an FRT of 875 °C would lead to the assumption that Nb is fully precipitated, whereas regarding the other processing routes, Nb might be still remaining in solution, and thus, according to the hardness values retards in a dissolved state the formation of ferrite. However, SEM investigations showed that the decreased hardness can be attributed to the size of the precipitated NbC as can be seen in Figure 7. As investigated by Maugis *et al.*,^[38] NbC particles become incoherent when they exceed a size of 5 nm. Regarding the dimensions of the particles found in the SEM examinations, which are significantly larger than 5 nm, Orowan's theory of precipitation hardening^[39] can be applied. However, the sizes of the NbC precipitates that emerged at an FRT of 875 °C (Figure 7(b)) are too large to contribute to reach sufficient precipitation hardening.^[40] Similar phenomenon was observed by Klinkenberg *et al.*^[41] In contrast, the small sizes of the NbC particles at an FRT of 950 °C and after re-austenitization at 930 °C possess the ability to achieve a high hardness at a low cooling rate of 1 K/s. It can be assumed that those alloying elements in steel 3 (Cr, Ni, and Mo and foremost the

Fig. 6—PAGs of steel 3 processed at an FRT of 975 °C. The highly pancake-shaped γ grains explain the increased hardness over steel 1.

MAE V which is a very strong carbide former) promote precipitation of Nb and stabilizes it as NbC, under which condition it has no influence on the hardenability itself. On the contrary, Nb, which in steel 2 remains still solved in the matrix, stimulates bainite and ferrite.

Due to the fact that the size of the microstructural components contributes to the strength according to the Hall–Petch relation, the attainable microstructures and substructures were assessed through Nital and picric acid etchants. Regardless of the cooling rate, steel 1 exhibits a very coarse microstructure (Figures 3(a) and 5(a) and (d)) due to the dispensing of MAE which prohibits accelerated grain growth through the

Fig. 7—SEM images of Nb carbides in steel 3 under different processing conditions with a cooling rate of 1 K/s: (a) Undeformed and quenched from 930 °C: NbC particles have a size range from 100 to 120 nm. (b) Processed with an FRT of 875 °C: NbC particles emerge with a size in the range from 2 to 3 μm .

Fig. 8—Dilatation curves of steel 3 cooled at 100 K/s. The M_S temperatures of the TM-processed samples differ clearly from the re-austenitized and quenched samples.

formation of carbides and a low solute drag. In contrast, the micro-alloyed variants, steels 2 and 3, reached the smallest grain sizes, whereby their dimensions differ

depending on the process route. In accordance with literature, the lower the reduction below the FRT is, the finer the microstructure becomes. Steel 3, however, further refines after re-austenitization (Figure 5(c)). This refining can be attributed to the combination of a highly deformed austenite prior to quenching (Figure 6), which recrystallizes during reheating in the γ -region reducing the grain boundary energy. Furthermore, precipitated carbides inhibit an augmented grain growth.

The M_S and M_F temperatures are dependent on the chemical composition^[20,36,42] and as reported by several authors on the PAG size.^[36,43,44] As Figure 2 demonstrates, M_S and M_F are independent of the processing routes for steel 1 and 2. At slow cooling rates, transformation starts at higher temperatures due to the formation of ferritic or bainitic phases. Dilatation data do not allow a precise identification of the originating phase through blurring boundaries. However, microstructural analysis provided the proof with respect to the existence of martensitic, bainitic, and ferritic phases. Nevertheless, for steel 3, the M_S and M_F temperatures differ clearly from each other depending on the different process routes and the cooling rates (Figure 3). At the highest cooling rate of 100 K/s, both TM-processed steels start their martensitic transformation clearly above the calculated M_S of 391 °C,^[26] whereas the quenched steel is delayed in its transformation, yet reaches M_F very close just below M_S (Figure 8). This could lead to the assumption that in a TM-deformed state, the martensite transformation is accelerated, which, however, needs further undercooling for a complete conversion, whereas in a plain austenitized state, the full martensitic transformation is accomplished with less undercooling. Due to the fact that a decrease in the PAG size leads to an elevated M_S ,^[36,43–45] the M_S of the re-austenitized and quenched sample is expected to be above the TM-processed sample, as the determined microstructural dimensions for steel 3 are significantly smaller after re-austenitization. However, it is reported that a high deformation of the austenite as it is performed during TMP triggers the M_S to elevate to higher temperatures.^[16] This argument is supported by the observation that increasing the reduction in the nonrecrystallization regime with lowering of the FRT from 950 °C to 875 °C further elevates the M_S to higher temperatures (Figure 8). On the contrary, re-austenitization releases these deformations by recrystallization. This leads to the assumption that the effect of elevating M_S through a highly deformed austenite dominates over the increased M_S through a finer austenite grain. However, this phenomenon only occurs at very high cooling rates and thus leaves room for a more in-depth investigation concerning this hypothesis.

V. CONCLUSIONS

The aim of this investigation was to study the influences of Nb and different process routes on the achievable hardness and hardenability of ultra-high strength steels. The main findings and conclusions are summarized in the following:

1. TM processing with subsequent direct quenching increases the strength significantly compared to conventional quenching after re-austenitization. The finish rolling temperature plays a major role during TMP: decreasing the FRT leads to a further strength gain.
2. Further hardness increase is observed by additions of Nb. However, investigations showed that the addition of Nb leads to an expedited formation of bainite and ferrite. This is explained by a grain refinement after re-austenitization resulting in an accelerated γ to α transformation, shifting the time transformations curves to shorter times. Finer grains additionally elevate the M_S temperature.
3. Increasing the reduction in the nonrecrystallizing austenite regime elevates the M_S temperature. This effect dominates over an accelerated formation of martensite due to smaller austenitic grain sizes.

ACKNOWLEDGMENTS

Open access funding provided by Montanuniversität Leoben. Funding of the Austrian BMVIT within the framework of the program “Production of the future” and the “BMVIT Professorship for Industry” is gratefully acknowledged.

CONFLICT OF INTEREST

Authors declare no conflict of interest.

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