The construction of mobile crane booms requires the usage of ultra-high strength steels. Micro-alloying elements promote grain refinement during hot rolling and result in increased toughness. The relevant strength is given through a martensitic microstructure, which is accomplished by elements retarding the γ to α transformation. Direct quenching (DQ) from the rolling heat and quenching after a preceding re-austenitization (RQ) are two different production routes. They differ regarding their productivity, their achievable strength levels, and their resulting microstructures. In order to explore the influence of the production route in combination with prominent micro-alloying elements, which come to application during hot-rolling, six steels with varying content of V and Nb are investigated concerning their different properties after DQ and RQ as well as their behavior after tempering. It is found, that Nb strongly improves the strength after thermomechanical processing in the as-rolled condition. Furthermore, Nb compensates the loss of strength during tempering. This effect is not thoroughly discussed in literature so far. Although Nb leads to grain refining during re-austenitization, the effects on the strength of RQ steels are minimal. The effect after tempering is also weaker than after direct quenching. It is also shown, that V offers a high strength potential after tempering, however weakens the impact toughness significantly.

1. Introduction

Steels with a yield strength above 1100 MPa are necessary to optimize the payload-to-weight ratio for the design of advanced mobile crane booms. Ultra-high strength steels (UHSS) meet these extreme strength requirements and simultaneously offer good ductility and thus excellent notch impact values. For the production of UHSS, two processing routes are facing each other: the conventional route consists of quenching (Q) after re-austenitization followed by a tempering (T) process to relax the residual stresses of the martensite, and the thermomechanical (TM) route, in which quenching is performed directly after hot-rolling (DQ). The globularization of the prior austenite grain (PAG) accomplished by normalization and reheating to the γ-region offers the advantage of an isotropic microstructure. TM processing of UHSS represents the alternative route, offering an economic benefit by avoiding a supplementary reheating step. Furthermore, the production via TM can provide an additional strength contribution through direct quenching from the rolling heat. This additional asset in strength, accomplished through a strongly deformed non-recrystallizing austenite, offers the benefit over Q + T steels, that for a given strength, alloying elements can be reduced. In turn, the waiver of mainly carbon has a favorable effect on the weldability. Prominent micro-alloying elements (MAE) which come to application are Ti, Nb, and V, which increase the non-recrystallization temperature (\(T_{NR}\)) in dissolved condition and by formation of strain-induced precipitates. Consequently, they promote an elevated number of nucleation sites for the subsequent γ to α transformation through a highly deformed austenite grain. For Q + T steels, they also serve as grain refiners during hot rolling and promote additional refinement during re-austenitization. Furthermore, the emerging precipitates possess a significant strength contribution, which compensates the loss of strength during tempering.

The influence of several MAE during hot-rolling has been broadly investigated. Prominent micro-alloying elements (MAE) which come to application are Ti, Nb, and V, which increase the non-recrystallization temperature (\(T_{NR}\)) in dissolved condition and by formation of strain-induced precipitates. Consequently, they promote an elevated number of nucleation sites for the subsequent γ to α transformation through a highly deformed austenite grain. For Q + T steels, they also serve as grain refiners during hot rolling and promote additional refinement during re-austenitization. Furthermore, the emerging precipitates possess a significant strength contribution, which compensates the loss of strength during tempering.

The influence of several MAE during hot-rolling has been broadly investigated. Furthermore, the role of processing conditions such as the finish-rolling temperature (FRT) is considered to be also well explored. However, there is a lack of information on the combined effect of the processing route and MAE on the final mechanical properties. Particularly, the influence of MAE on a tempering treatment, in which it is distinguished between DQ and re-austenitization and quenching is an open scientific and technological field. The present work is therefore intended to cover the microstructural development, including the resulting grain sizes, depending on the process conditions, and the type and content of MAE. These investigations are supported by the mathematical description of the underlying microstructural processes through
2. Materials and Processing

The six steels tested were melted in an induction furnace in quantities of 300 kg each and their chemical composition is listed in Table 1. Based on the basic composition of steel 1, which represents a UHS steel, successive additions of Nb and V were applied. Besides the carbon content \( \approx 0.17% \), the base alloy contains \( \approx 0.2\% \) Si, \( \approx 2.3\% \) Mn, and \( \approx 0.002\% \) B to guarantee a through-hardening during quenching. Further surcharges of Cr and Cu control temper resistance,\cite{32,33} Al and Ti are responsible for grain refinement and additional strength increase.\cite{12,13,34,35} The variants of steel 1 possess additions of Nb of \( 0.02\% \) and \( 0.2\% \) V to compare the microstructural and mechanical anisotropy. It will be established, whether a correlation between an elongated austenitic grain and the anisotropy in the mechanical properties between the longitudinal and transversal direction exists. For this matter, six steels with varying contents of Nb and V, processed via TM are investigated and compared to the conventional RQ-route. Mechanical properties were determined for the longitudinal and transversal direction in the as-rolled condition and after tempering. In addition, two different FRT were performed in the DQ route. Thereby it should be established, whether an increasing grain elongation, which is accompanied with a decreasing rolling temperature influences the mechanical anisotropy.

### Table 1. Chemical composition of the six steels investigated.

| Base alloy | Variation | C  | Si  | Mn  | Al  | Cr  | Cu  | V    | Nb   | Ti  | B  |
|------------|-----------|----|-----|-----|-----|-----|-----|------|------|-----|----|
| 1          | Nb+       | 0.173 | 0.20 | 2.38 | 0.049 | 0.280 | 0.089 | 0.005 | 0.002 | 0.021 | 0.0025 |
|            | Nb ++     | 0.175 | 0.20 | 2.37 | 0.043 | 0.280 | 0.079 | 0.005 | 0.021 | 0.020 | 0.0024 |
| 2          | V+        | 0.170 | 0.21 | 2.30 | 0.047 | 0.710 | 0.078 | 0.005 | 0.002 | 0.021 | 0.0026 |
|            | V ++      | 0.179 | 0.22 | 2.29 | 0.049 | 0.700 | 0.081 | 0.194 | 0.002 | 0.021 | 0.0026 |

### 2.2. Determination of the Activation Energy and Non-Recrystallization Temperature

To derive the activation energy for dynamic recrystallization and describe the yield stress in terms of temperature and strain rate, the Zener-Hollomon parameter was determined for the selected steels. For the experiments, which were conducted on a deformation Dilatometer Bähr 805 A/D, cylindrical samples with a diameter of 6.0 mm and a length of 10 mm were machined from the milled plates with a wire-electro discharge machine. The deformation procedure of the six investigated steels involved the calculation of the activation energy \( E_A \) for dynamic recrystallization (DRX) via the Zener-Hollomon parameter.\cite{30,31} Finally, the present work should stretch the bow between the microstructural and mechanical anisotropy. It will be established, whether a correlation between an elongated austenitic grain and the anisotropy in the mechanical properties between the longitudinal and transversal direction exists. For this matter, six steels with varying contents of Nb and V, processed via TM are investigated and compared to the conventional RQ-route. Mechanical properties were determined for the longitudinal and transversal direction in the as-rolled condition and after tempering. In addition, two different FRT were performed in the DQ route. Thereby it should be established, whether an increasing grain elongation, which is accompanied with a decreasing rolling temperature influences the mechanical anisotropy.

### Table 2. Overview of the processing cycles and conditions and the used nomenclature.

| Processing cycles | Direct quenched | Re-austenitized and quenched |
|-------------------|-----------------|-----------------------------|
| As-rolled         | DQ850           | DQ930                       |
| Tempered          | DQ850 + T       | DQ930 + T                   |

FRT = finish rolling temperature.
solution annealing at 1250 °C and subsequent forming at 850, 900, 950, 1000, and 1050 °C, respectively, at four different strain rates $\varphi = 0.01, 0.1, 1.0$, and $10 \, s^{-1}$. The peak strain of the flow stress curves was used for the calculation of the activation energy and the Zener-Hollomon parameter according to refs.\cite{30,31}. For the calculation of $T_{NR}$, which is dependent on the chemical composition, the mathematical model described by Boratto\cite{24} was applied.

2.3. Sample Preparation for Metallographic Analysis

To study the microstructure dependent on the processing route and MAE, specimens of the hot rolled sheets were hot embedded and then ground from 320 grit to 4.000 grit SiC paper for at least 30 s. Subsequently, the samples were polished with 3 μm diamond paste for at least 3 min and with 1 μm for 30 s. To reveal the PAGs, the samples were etched with a picric-acid etchant. The composition and procedure is described in ref.\cite{38}. To reveal the transformed martensitic microstructure, the samples were dipped in a Nital etchant. The images were recorded with an optical microscope and a FIB Versa FEI 3D DualBeam scanning electron microscope (SEM). For the metallographic analysis of the PAGs, the equivalent grain diameters and the aspect ratio of the PAGs were determined with the image analysis software MIPARTM\cite{39}. Since the temper treatment has no influence on the PAGs, only those samples were examined due to a favorable impact of the etchant on the tempered specimens.\cite{38}

3. Results

3.1. Microstructural Analysis

Light-optical microscopy and SEM investigations of the transformed microstructure revealed a mainly martensitic microstructure with small bainitic sections as depicted in Figure 1. The samples, which were directly quenched from the rolling heat (DQ850) showed martensitic blocks, which are orientated in 45° toward the rolling direction (Figure 1a). These laths and blocks emerge from the highly pancaked γ grain, as Figure 2a reveals on a tempered sample of Base alloy 1 Nb ++ . After re-austenitization, the PAGs develop a globular shape (Figure 1c), the blocks are not arranged in a preferential orientation and rather randomly organized. After T, carbides emerge from the supersaturated a-grains.

The analysis of the PAGs, which were revealed with a picric-acid etchant, illustrates the influence of the processing route and the content of MAE. As expected, the lower FRT of 850 °C leads to an increased aspect ratio compared to the FRT of 930 °C as demonstrated by Figure 2. Already small additions of Nb strongly promote the pancaking of the PAGs, however, this effect seems to be effective only at lower final rolling temperatures. Further additions of Nb lead to grain refinement. This can be observed as well after re-austenitization. The influence of the Cr surcharge to base alloy 2 seems negligible concerning the microstructural appearance. Cr has no influence in retarding recrystallization, the aspect ratio and mean grain size of base alloy coincides with base alloy 1, however with increasing V, the aspect ratio increases slightly but not in the same extent as through Nb. The light-optical images of the steels investigated,
are displayed in Figure 3, and compare the TM route, processed with a FRT of 850°C (a–c) to the condition after re-austenitization (d–f). The PAGs of base-alloy 1 are almost equiaxed in the DQ850 condition (Figure 3a). Niobium promotes immensely the pancaking of the γ-grains (Figure 3b), the effect of Vanadium however is less pronounced (Figure 3c). Re-austenitization delivers equiaxed grains as the light optical images (Figure 3d–f) reveal. RQ of base alloy 1 leads to the same mean grain size (Figure 3d), yet, globularized PAGs. Nb (Figure 3e) and V (Figure 3f) both act as grain refiner after re-austenitization. Nevertheless, Nb refines the grains in every processing route significantly. The influence of V general, the influence of the FRT on the mean prior austenite grain size cannot be determined with certainty, yet, re-austenitization clearly refines the γ-grains (Figure 3d–f).

3.2. Mechanical Properties

The ultimate tensile strength (UTS) values are displayed in Figure 4 for the different compositions investigated. Regarding the strength values of the steels in the as-rolled condition, the DQ variants with a FRT of 850°C are clearly above those of the DQ variant with a FRT of 930°C (60–70 MPa) and as well above the RQ samples (30–100 MPa). Base alloy 2, which is alloyed with 0.7% Cr, possesses compared to the low-Cr variant (base alloy 1) in the as-rolled condition less strength. The surcharge of Cr however leads to a slight increase in strength when tempered. The effect of the alloying elements Nb and V becomes evident in TM processing route, whereas no influence is visible in the RQ condition. Processed with a FRT of 850°C, 0.02% Nb increase the UTS by an amount of 49 MPa and at a FRT of 930°C by 42 MPa. The strength increase however saturates, which is visible in the weaker strength increase with double the amount of Nb. The same effect can be observed when alloying with V. However, the strength contribution of 0.1% V is less pronounced than the one of Nb, although V is alloyed with fivefold the
The impact toughness values of the as-rolled specimens at 0 and the TM processed variants with a FRT of 930°C are pronounced for the DQ cycles. Nb results in strength increase after tempering. This effect is even more pronounced for the DQ cycles. In as-rolled condition, micro-alloying elements have no strength contribution of tempering. V counteracts even stronger with a flow stress increase of 0.02% Nb achieve an additional strength of ∼100 MPa of the TM rolled steels, and ∼50 MPa after tempering the RQ steel. V counteracts even stronger with a flow stress increase of 0.1% V and with ∼200 MPa by the addition of 0.2% V. Nevertheless, the effect of both, Nb and V alloying, seems to saturate after tempering too.

Figure 5 exhibits the impact toughness for the as-rolled conditions (left) and the tempered state (right) for the 6 steels investigated at 0°C (upper graphs), −20°C (middle), and −40°C (bottom part). The DQ variants with a FRT of 850°C and the RQ variants possess the best toughness values followed by the TM processed variants with a FRT of 930°C. The micro-alloying with Nb and V has no significant influence on the impact toughness values of the as-rolled specimens at 0 and −20°C. Yet, at −40°C, V seems to deteriorate the notch impact values considerably. As expected, tempering generally improves the toughness, but when comparing the variants clear differences appear. The RQ variants show slight advantages over the DQ variants. Nb and Cr exhibit no significant improvement, however, Vanadium decreases the toughness after tempering noticeably.

In order to evaluate the influence of the processing route (DQ and RQ) on the mechanical properties in longitudinal and transversal direction, the comparison of the notch impact test results is useful. Microstructural variations affect Charpy impact tests (Figure 5) more than tensile tests (Figure 4). The differences of the impact test results of the longitudinal and transversal direction at 0, −20, and −40°C can be found in Figure 6. Comparing both TM routes, the higher FRT of 930°C (gray bars) possesses an improved isotropy as the FRT of 850°C (black bars). This can be identified at all three temperatures. However, RQ does not show an optimization of the mechanical isotropy (red bars), especially at 0 and −20°C, despite having globular PAGs. For all process routes an improvement is accomplished through tempering (dashed bars). For the Vanadium alloyed variants, the anisotropy even seems to be eliminated completely. However, as can be seen in Figure 5, the impact toughness decreases considerably after tempering for variant 4 and 5. Also the decrease in anisotropy with decreasing test temperatures can be related to the fundamental deterioration of the notch impact values at lower temperatures. Nb exhibits no comprehensive influence in modifying the longitudinal–transversal isotropy of the TM rolled variants. Nevertheless, all surcharges to base alloy 1, Nb, V, and Cr reveal benefits concerning the isotropy of the RQ samples.

3.3. Flow Stress Behavior and Activation Energy for Dynamic Recrystallization

For the evaluation of the recrystallization behavior and the influence of V and Nb on the activation energy for the dynamic recrystallization (DRX), dilatometer experiments were performed with deformation temperatures between 850 and 1050°C, at strain rates of 0.01, 0.1, 1.0, and 10 s⁻¹. Figure 7 shows the corresponding true stress–strain curves at a strain rate of φ = 1.0 for the five temperatures. For the sake of a better clarity, only the base alloys (black and gray curves) and their corresponding “high” alloy variants (Nb ++: blue dashed line and V ++: red dashed line) are plotted. The flow stress increases significantly with decreasing deformation temperatures, the kf max increases from ∼130 MPa at 1050°C to 300 MPa at 850°C. At both 850 and 1050°C, only a small effect of V and Nb can be observed with a slightly elevated flow stress compared to base alloy 1 and 2, which behave almost congruent regarding their flow stress behavior. However, at the temperatures between 900 and 1000°C, the effect of the MAE becomes evident. At 1000°C, the Nb ++ variant is clearly elevated above the others and maintains the level of flow stress, while the other curves show a decrease at high strains. The decline of the curve of the V variant is also decelerated. This effect is repeated at 950°C, where the curve of Nb clearly stands out and shows no significant decrease. Also at 900°C the Nb ++ steel has elevated flow stress behavior,
whereas the V−+ variant enqueuers in the behavior of base alloy 1 and 2.

Figure 8 shows the activation energy for DRX and the $T_{NR}$, which were calculated based on the Boratto model.\cite{24} Nb possesses the biggest influence on $E_A$, already 0.02 wt% increase the activation energy for 38 kJ mol$^{-1}$ K$^{-1}$. Cr has no impact on the activation energy, the effect of V is less pronounced compared to Nb, a relatively high amount of 0.2 wt% V rise $E_A$ only for 18 kJ mol$^{-1}$ K$^{-1}$.

4. Discussion

In order to examine the influence of the production route in combination with MAE, six steels with varying contents of V and Nb were rolled in a hot-mill simulation unit with two different FRT of 850 and 930°C, respectively. To evaluate differences between TM rolling with subsequent quenching and quenching after re-austenitization, the six different steels were additionally subjected to a conventional quenching after reheating to 930°C for 5 min. The mechanical properties of the as-rolled plates were compared to tempered specimens to study the role of the production route and MAE on their behavior after tempering. TM processing with a low FRT results in improved UTS (Figure 4). The increasing deformation in the non-recrystallization regime and pancaking of the γ-grains (Figure 3) at lower FRT leads to refinement of lath and block size\cite{50,41} which contributes to the strengthening. DQ with an elevated FRT and RQ possess comparable strength values, despite both processes result in different PAG sizes. The strength increase
resulting from a subsequent quenching from the rolling heat compensates the strength improvement through grain refining after re-austenitization of the RQ probes so that similar strength values emerge. Comparing DQ930 and RQ, the mechanical properties of the TM rolled and direct-quenched route shows benefits, when alloyed with Nb. An elevation of \( T_{\text{NR}} \) promotes an increased deformation in the non-recrystallization regime (Figure 8), so that even the higher FRT profits from a certain TM contribution, as the last rolling steps are already below \( T_{\text{NR}} \).

This profit is expressed by the aforementioned refinement. The effect of V on the \( T_{\text{NR}} \)-which includes the influence on the \( E_A \) for DRX-is very weak (Figure 8). This is reflected in the reduced pancaking (Figure 1, 2) and thus in the quasi non-existent TM-contribution to the DQ route. In Figure 9, the strength contribution of V and Nb is summarized as a function of the alloying content and contrasts the influence of the MAE on the strength in the as-rolled condition and after tempering. The strength potential of V is reached after tempering and is attributable to the precipitation of VC and V\(_4\)C\(_3\) and substitution of cementite carbides.\(^3,14,42\) As in Figure 9 visible, the higher FRT even gains more strength through T, then the counterpart

Figure 7. True stress–strain curves of base alloy 1 and 2 and its variants. At 1050 °C Nb and V have no effect on the flow stress, at 950 and 1000 °C Nb significantly increases flow stress and prohibits DRX, the effect of V is less pronounced. At 850 °C very little DRX appears, the yield stress curves of the individual alloys are almost congruent.

Figure 8. Activation energy \( E_A \) and the non-recrystallization temperature \( T_{\text{NR}} \), calculated based on the Boratto model.\(^24\) Nb increases \( E_A \) significantly, whereas V exhibits only little effect. Same trend can be observed regarding their role on the \( T_{\text{NR}} \).

Figure 9. Strength contribution of Nb and V. The effect of V for strength improvement during TM rolling is very small (dashed lines), and even negative if re-austenitized. However, during tempering it significantly increases the UTS by 200 MPa (dotted lines). Nb has no effect as strengthener after re-austenitization and Q, however, delivers a plus of >50 MPa after tempering. During TM processing (DQ850, DQ930), little amounts of Nb already contribute to a strengthening. After tempering, again the effect is even more pronounced.
with the lower FRT or the RQ variant, respectively. It is assumed, that V carbides, which are responsible for the strength improvement are precipitated in a higher amount at lower rolling temperatures of 850 then at 930°C, where a sufficient portion is still available for an improved strength gain during tempering. Despite the improvement of strength, Vanadium shows a negative effect on the impact toughness after tempering. This effect is portrayed by Figure 10, which contrasts the elongation at fracture and the notch impact toughness at 0°C to the ultimate tensile strength. During tempering, the strength decreases in favor of the elongation at fracture. Although, the V variants follow this well-known trend concerning their elongation values, the impact values deteriorate significantly, whereas the other variants are improved concerning their impact toughness. This effect can be deduced to the embrittlement from Vanadium carbides.\[43,44\]

Regarding the Charpy impact values in the as-rolled condition, neither V nor Nb result in a significant improvement or deterioration (Figure 5). However, the TM rolled plates with a FRT of 930°C are generally lower. The DQ850 samples possess surprisingly similar values as the RQ variants, although their strength is remarkably higher. At −40°C, the values of the DQ850 probes are even above those produced via Q. This can be attributed to the aforementioned refinement of lath and blocks through the increased pancaking: the resulting grain refinement offers enhanced strength and toughness and shifts the ductile-to-brittle temperature to lower temperatures.\[14,33,45,46\]

The comparison of the PAG aspect ratio to the Charpy impact toughness in longitudinal and transversal direction did not deliver a clear correlation. Although the reduced pancaking through an elevated FRT in DQ930 exhibits an improved isotropy of the notch impact values compared to DQ850, a globularization of the PAGs through re-austenitization (RQ) deteriorates the anisotropy again. The orientation of the martensitic blocks, which are arranged in 45° toward the RD (Figure 1a, b) and originate from the pancaked γ grain in an angle of 45°, obviously play no significant role on the anisotropy either. The longitudinal–transversal differences of RQ, in which the martensitic components are aligned randomly are even higher compared to DQ. Increasing the pancaking through Nb has no obvious effect on the anisotropy. However, annealing after quenching seems to enhance those differences. As proposed by Kajjalainen et al.\[50,48,49\] rolling textures, which are dependent on the FRT are responsible for unfavorable bending properties.\[50\] These preferred orientations cannot be eliminated by quenching or normalizing and thus can be an explanation for the remaining anisotropy. However, further investigations are necessary to provide information on this matter.

Investigations on the flow stress behavior and the DRX have confirmed the different roles of Nb and V micro-alloying. Nb comes already to affect at relatively high temperatures (>1000°C), and retards the recrystallization with very small additions (0.02 m%) through the formation of carbides.\[14\] These observations are visible in the flow stress curves (Figure 7). Although Vanadium is a strong carbide former, the high solubility product compared to Ti and Nb\[51\] and the low solute drag\[52\] attribute V only a weak effect in retarding recrystallization. As visible in Figure 8, even large amounts of vanadium, which exceed conventional micro-alloying contents, have just little effect on the DRX. These calculations are supported by the similar trend of $T_{NB}$ dependent on DRX (Figure 8).

5. Summary

In order to explore the influence of the production route in combination with micro-alloying elements, six steels with varying content of V and Nb were investigated in their as-rolled and tempered condition. Their mechanical properties after direct quenching and quenching after re-austenitization were examined. The main findings are summarized in the following:

1) Nb is confirmed in its important role in retarding recrystallization. However, its strong impact on the improvement of mechanical properties after tempering is unattended till now. It counteracts in compensating the softening during tempering. This effect is enhanced by a previous TM rolling.

2) Vanadium has a negligible effect on the mechanical properties of both DQ and RQ steels in the as-rolled condition, as it is not capable in retarding recrystallization considerably. However, it manifests its effect after tempering, 0.1 m% of V lead to an additional strength of 130 MPa after tempering. Despite of the positive effect on UTS, Vanadium...
alloyed steels suffer from embrittlement after tempering, followed by a deterioration of the impact toughness. Positive effects of Cr on the temper resistance were not observed.

3) Re-austenitization previous to quenching does not improve the longitudinal–transversal anisotropy despite of a globular PAG. In contradiction, a reduced pancaking through a higher FRT results in an improvement of this anisotropy. Nevertheless, higher FRT have a negative effect on UTS and Charpy impact toughness.

Acknowledgement

Funding of the Austrian BMVIT (846933) in the framework of the program “Production of the future” and the “BMVIT Professorship for Industry” was gratefully acknowledged.

Conflict of Interest

The authors declare no conflict of interest.

Keywords

direct quenching, finish rolling temperature, mechanical anisotropy, Nb micro-alloying, thermomechanical processing, V micro-alloying

Received: December 20, 2018
Revised: February 18, 2019
Published online: March 18, 2019

[1] M. Klein, R. Rauch, H. Spindler, P. Stiaszny, BHM Berg- Hütenmänn. Monatsbl. 2012, 157, 108.
[2] M. Klein, H. Spindler, A. Lugur, R. Rauch, P. Stiaszny, M. Eigelsberger, Mater. Sci. Forum 2005, 500–501, 543.
[3] H. Bhadeshia, R. Honeycombe, Steels: Microstructure and Properties, 4th edn., Butterworth-Heinemann, Oxford, UK 2017, pp. 237.
[4] T. Tanaka, Int. Met. Rev. 1981, 26, 185.
[5] D. T. Llewellyn, R. C. Hudd, Steels: Metallurgy and Applications, 3rd ed., Butterworth-Heinemann, Oxford, [England], Woburn, MA 1998.
[6] C. R. Brooks, Principles of the Heat Treatment of Plain Carbon and Low Alloy Steels, 1st edn., ASM International, Ohio, USA 1996, pp. 3.
[7] C. Ouchi, ISIJ Int. 2001, 41, 542.
[8] A. Kaijalainen, S. Pallaspuro, D. A. Porter, Adv. Mater. Res. 2014, 922, 316.
[9] S. Pallaspuro, A. Kaijalainen, T. Limnell, D. A. Porter, Adv. Mater. Res. 2014, 922, 580.
[10] R. Schnitzer, D. Zügner, P. Haslberger, W. Ernst, E. Kozeschnik, Sci. Technol. Weld. Join. 2017, 22, 536.
[11] P. Haslberger, S. Holly, W. Ernst, R. Schnitzer, J. Mater. Sci. 2018, 53, 6968.
[12] O. Grong, D. K. Matlock, Int. Met. Rev. 1986, 31, 27.
[13] D. J. Abson, R. J. Pargett, Int. Met. Rev. 1986, 31, 141.
[14] T. Gladman, The Physical Metallurgy of Microalloyed Steels, Institute of Materials, London 1997.
[15] I. Tamura, Thermomechanical Processing of High-Strength Low-Alloy Steels, Butterworths, London, Boston 1988.
[16] W. B. Morrison, Mater. Sci. Technol. 2009, 25, 1066.
[17] A. A. Barani, F. Li, P. Romano, D. Ponge, D. Raabe, Mater. Sci. Eng. A 2007, 463, 138.
[18] T. Gladman, Mater. Sci. Technol. 1999, 15, 30.
[19] O. Kwon, A. J. DeArdo, Acta Metall. Mater. 1991, 39, 529.
[20] R. A. Grange, C. R. Hribal, L. F. Porter, Metall. Trans. A 1977, 8, 1755.
[21] H. L. Andrade, M. G. Akben, J. J. Jonas, Metall. Trans. A 1983, 14, 1967.
[22] S. F. Medina, Mater. Sci. Technol. 1998, 14, 217.
[23] L. J. Cuddy, in Proc. 1981 Thermomechanical Processing of Microalloyed Austenite, TMS-AIME, Warrendale, PA 1981, pp. 129–140.
[24] R. Barbosa, F. Boratto, S. Yue, J. J. Jonas, in Proc. of the International Conference on Processing Microstructure and Properties of HSLA Steels, The Mineral, Metals and Materials Society, Pittsburgh, PA 1987, pp. 51–56.
[25] C. N. Homsher, PhD thesis, Colorado School of Mines, 2013.
[26] B. Pereda, B. Lopez, J. M. Rodriguez- Ibabe, in Proc. International Conference on Microalloyed Steels: Processing, Microstructure, Properties and Performance, AIST, Pittsburgh 2007, pp. 151.
[27] S. Venynckt, K. Verbeke, B. Lopez, J. J. Jonas, Int. Mater. Rev. 2012, 57, 187.
[28] L. Bracke, W. Xu, T. Waterschoot, Mater. Today Proc. 2015, 2, 659.
[29] A. Kaijalainen, N. Vähäkouropus, M. Somani, S. Mehtonen, D. Porter, J. Kömi, Arch. Metall. Mater. 2017, 62, 619.
[30] J. J. Jonas, C. M. Sellars, W. J. M. Tegart, Metall. Rev. 1969, 14, 1.
[31] H. Mirzadeh, J. M. Cabrera, J. M. Prado, A. Najafizadeh, Mater. Sci. Eng. A 2011, 528, 3876.
[32] G. Krauss, Steel: Processing, Structure, and Performance, 3rd edn., ASM International, Ohio, USA 2005.
[33] D. T. Llewellyn, R. C. Hudd, Steels: Metallurgy and Applications, 3rd ed., Butterworth-Heinemann, Oxford [England], Woburn, MA 1998.
[34] D. K. Matlock, J. G. Speer, Mater. Sci. Technol. 2009, 25, 1118.
[35] P. D. Hodgson, H. Belaldi, M. R. Barnett, Mater. Sci. Forum 2005, 500–501, 39.
[36] G. E. Totten, Steel Heat Treatment: Metallurgy and Technologies, CRC Press, Boca Raton 2006.
[37] R. Esterl, M. Sonnleitner, R. Schnitzer, Steel Res. Int. 2019, DOI: 10.1002/srin.201800500.
[38] R. Esterl, M. Sonnleitner, M. Stadler, G. Wölger, R. Schnitzer, Pract. Metallogr. 2018, 55, 203.
[39] J. M. Sosa, D. E. Huber, B. Welk, H. L. Fraser, Integrating Mater. Manuf. Innov. 2014, 3, 10.
[40] A. J. Kaijalainen, P. P. Suikkkanen, T. J. Limnell, L. P. Karjalainen, J. I. Körni, D. A. Porter, J. Alloys Compd. 2013, 577, 642.
[41] Y. Prawoto, N. Jasmawati, K. Sumeru, J. Mater. Sci. Technol. 2012, 28, 461.
[42] A. I. H. Committee, ASM Handbook: Heat Treating, ASM International, Metals Park, Ohio 1991.
[43] R. Lagneborg, T. Siwecki, S. Zajac, B. Hutchinson, Scand. J. Met. 1999, 28, 186.
[44] A. P. V. Wyk, G. Pienaar, J. Atr. Inst. Min. Met. 1987, 87, 73.
[45] G. Gottstein, Physikalische Grundlagen der Materialkunde, 3. Aufl., Springer, Berlin, 2007.
[46] A. Brownrigg, Scr. Metall. 1973, 7, 1139.
[47] S. Zajac, V. Schwinn, K. H. Tacke, Mater. Sci. Forum 2005, 500–501, 387.
[48] A. J. Kaijalainen, P. P. Suikkkanen, L. P. Karjalainen, D. A. Porter, Mater. Sci. Eng. A 2016, 654, 151.
[49] A. J. Kaijalainen, P. P. Suikkkanen, L. P. Karjalainen, J. J. Jonas, Metall. Mater. Trans. A 2014, 45, 1273.
[50] W. Bleck, R. Grossterlinden, U. Lotter, C.-P. Reip, Steel Res. 1991, 62, 580.
[51] H. Bhadeshia, R. Honeycombe, in Steels: Microstructure and Properties, 4th edn., Butterworth-Heinemann, Oxford, UK 2017, pp. 101.
[52] R. J. Gledowski, in Application Technologies of Vanadium in Flat-rolled Steels -Vanitec Symposium, Suzhou, China, 2005, p. 43.