Influence of the Quenching and Partitioning Process on the Transformation Kinetics and Hardness in a Lean Medium Manganese TRIP Steel

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Abstract: The quenching and partitioning (Q&P) process of lean medium Mn steels is a novel approach for producing ultra-high strength and good formable steels. First, the steel is fully austenitized, followed by quenching to a specific quenching temperature \((T_Q)\) in order to adjust an appropriate amount of initial martensite \((\alpha'_\text{initial})\). Subsequently, the steel is reheated to a partitioning temperature \((T_P)\) in order to ensure C-partitioning from \(\alpha'_\text{initial}\) to remaining austenite \((\gamma\text{remain})\) and thus retained austenite (RA) stabilization. After isothermal holding, the steel is quenched to room temperature (RT), in order to achieve a martensitic-austenitic microstructure, where the meta-stable RA undergoes the strain-induced martensitic transformation by the so-called transformation induced plasticity (TRIP) effect. This paper systematically investigates the influence of the Q&P process on the isothermal bainitic transformation (IBT) kinetics in a 0.2C-4.5Mn-1.3Al lean medium Mn steel by means of dilatometry. Therefore, the Q&P annealing approach was precisely compared to the TRIP-aided bainitic ferrite (TBF) process, where the samples were directly quenched to the temperature of the IBT after full austenitization. The results indicated an accelerated IBT for the Q&P samples, caused by the formation of \(\alpha'_\text{initial}\) during quenching below the martensite start \((M_S)\) temperature. Furthermore, a significant influence of the annealing parameters, such as \(T_Q\) and \(T_P\), was observed with regard to the transformation behavior. For further characterization, light optical microscopy (LOM) and scanning electron microscopy (SEM) were applied, showing a microstructure consisting of a martensitic-bainitic matrix with finely distributed RA islands. Saturation magnetization method (SMM) was used to determine the amount of RA, which was primarily depending on \(T_Q\). Furthermore, the hardness according to Vickers revealed a remarkable impact of the annealing parameters, such as \(T_Q\) and \(T_P\), on the predicted mechanical properties.

Keywords: Q&P; TRIP; lean medium Mn steel; transformation kinetics

1. Introduction

Both environmental and safety regulations force the automotive industry to the application of new steel grades [1,2]. In order to meet the challenging requirements concerning crashworthiness, formability, and reduction of CO\(_2\) emissions, “Advanced High Strength Steels” (AHSS) are currently under development [3–5]. By the increase of the material strength without a deterioration of ductility, the sheet thickness can be downgauged, resulting in a significant weight reduction of the body structures [1,6]. Furthermore, the combination of both high strength and ductility is of vital importance...
in order to manufacture complex automotive parts. Therefore, a high research effort has been put into
the further development of AHSS.

The 1st generation AHSS, being already in industrial application, is represented by Dual Phase
(DP), TRIP, and Complex Phase (CP) steels [6]. These steels mainly have a multiphase structure,
resulting in tensile strengths up to 1200 MPa and total elongations up to 40% [7–9]. Typically, they are
characterized by \( R_m \times A_{80} \) up to 20,000 MPa% [6]. The 2nd generation AHSS includes Twinning Induced
Plasticity (TWIP), Nano-TWIP, Duplex, and Triplex steels, which are characterized by highly promising
mechanical properties, especially an excellent combination of strength and ductility with \( R_m \times A_{80} \) between
40,000 and 60,000 MPa% [10]. However, due to the high alloying costs and challenging
processing they are hardly used in industrial applications [11].

Currently, recent research activities focus on the development of steels belonging to the 3rd
generation AHSS, including the concepts of Q&P and medium Mn steels, to fill the property gap
between the 1st and 2nd generation AHSS [9,12–14].

The Q&P process has been proposed as a promising approach for steel grades having the
microstructure consisting of a carbon-depleted martensitic matrix and retained austenitic islands [15].
In this process, the steel is fully austenitized followed by quenching to a specific temperature below the
\( M_S \) temperature, where the optimal amount of RA can be adjusted. Subsequent reheating and holding
in the over-ageing region triggers the \( \alpha' \) tempering, whereby the cementite formation during this stage
will be significantly postponed by the Si and/or Al additions. Therefore, C can partition to remaining
\( \gamma \) resulting in its appropriate stabilization upon final cooling to room temperature (RT) [16,17].

The medium Mn steels typically contain between 3 and 10 wt.% Mn and their microstructure
consists of an ultrafine-grained ferritic matrix (with a typical grain size below 1 \( \mu m \)) and approximately
30 vol.% of RA. These steel grades are characterized by an excellent combination of tensile strength
and total elongation achieving the product of \( R_m \times A_{80} \) exceeding 30,000 MPa% [12,14,18].

In the present case, the concept of lean medium Mn Q&P steels combines both the approach
of medium Mn steels and the Q&P process, leading to a microstructure consisting of a tempered
martensitic matrix with an increased amount of RA islands. The RA can be stabilized by the
combination of the C and Mn enrichment, ensuring an optimum strain-induced \( \gamma \) to \( \alpha' \) transformation
(TRIP-effect). Moreover, the presence of the hard C-depleted martensitic matrix ensures the superior
performance of these steels in the forming operations such as sheet cutting, bending and hole expansion,
since the ability to resist high local strains is linked to the hardness gradients in the microstructure [19].

In order to predict the amount of RA depending on \( T_Q \), the constrained carbon equilibrium (CCE)
model proposed by Speer et al. [20] can be applied. This simplified model allows the calculation of the
C-content in \( \gamma \) and therefore the prediction of the amount of RA under three main conditions: (1) All
of the C partitions to \( \gamma \) and the partitioning kinetics are ignored, meaning the partitioning is already
complete; (2) no phase boundary movement during the partitioning process; and (3) no competing
reactions like carbide or \( \alpha_B \) formation take place during the Q&P process [20].

However, the decomposition from \( \gamma \) to \( \alpha_B \) during the C-partitioning step has been observed
in some instances [21–23]. Clarke et al. [24] showed that the formation of carbide-free bainite (\( \alpha_B \))
during the partitioning process led to a reduced RA fraction in the Q&P samples compared to the
amounts predicted by the CCE-model. However, in some cases, the formation of \( \alpha_B \) can contribute to
the stabilization of \( \gamma \) by its C-enrichment.

Therefore, in the present paper, the transformation behavior of a 0.2-4.5Mn-1.3Al steel was
studied for both Q&P and TBF processes in order to evaluate the influence of the presence of \( \alpha_{\text{initial}} \)
on IBT. Several annealing parameters were varied (\( T_Q, T_P \)), in order to examine their influence on the
transformation kinetics and thus the volume fraction of the individual constituents, i.e., \( \alpha' \), \( \alpha_B \), and
RA, in the final microstructure.
2. Materials and Methods

Table 1 shows the chemical composition of the investigated steel grade in wt.%. The steel was melted in a medium frequency induction furnace and cast under laboratory conditions in an ingot of 80 kg. First the material was hot rolled to a thickness of 4 mm, followed by tempering in a batch-annealing-like furnace at a temperature of 550 °C for 16 h. Finally, the material was cold rolled to a thickness of 1 mm.

Table 1. Chemical composition of the investigated steel Fe-C-Mn-Al in wt.%.  

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>Al</th>
<th>Si</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt.%</td>
<td>0.20</td>
<td>4.49</td>
<td>1.30</td>
<td>0.04</td>
</tr>
</tbody>
</table>

The transformation behavior and the influence of the different annealing parameters were investigated by means of dilatometry using a Bähr 805 A/D dilatometer. Therefore, specimens with dimensions of $10 \times 4 \times 1 \text{ mm}^3$ were produced by wire-electrical discharge machining. Figure 1 shows the applied time-temperature schedules for the Q&P (a) and the TBF (b) heat-treatments, adapted to suit an industrially feasible continuous annealing line. All samples were fully austenitized at a $T_{an}$ of 900 °C for 120 s ($t_{an}$) with a heating rate $HR_1$ of 10 K/s. For the Q&P process, the austenitization was followed by quenching to various $T_Q$ in the range of 130 °C–330 °C with a 20 °C step using a cooling rate $CR_1$ of 50 K/s. The $T_Q$ was held for 10 s and subsequently the samples were reheated with 20 K/s ($HR_2$) to a certain $T_P$, which was varied between 350, 400, and 450 °C and held for 600 s ($t_p$), respectively. In contrast, the TBF samples were directly cooled to a $T_B$ of 350, 400, or 450 °C and held for 600 s ($t_B$). Finally, both Q&P and TBF samples were cooled to RT with a cooling rate $CR_2$ of 50 K/s.

Figure 1. Time-temperature regimes used for (a) the quenching and partitioning (Q&P) and (b) the transformation induced plasticity (TRIP) aided bainitic ferrite (TBF) heat-treatments.

The microstructure was characterized by means of light optical microscopy (LOM) using LePera etching [25]. Furthermore, the samples were electrochemically polished for the SEM investigations. Vickers hardness testing (HV1) was performed on polished samples using an Emco Test DuraScan 20 device. The volume fraction of RA was determined by the use of SMM [26]. Furthermore, the RA content depending on the $T_Q$ was calculated using the CCE-model proposed by Speer et al. [20]. Therefore, first the volume fraction of $\alpha'$ formed during quenching to $T_Q$ was determined by the Koistinen-Marburger (KM) equation, which can be used to predict the $\gamma$ to $\alpha'$ transformation rate and is given by [27]:

$$f_M = 1 - e^{-0.011(M_S - T_Q)}$$
Here, \( f_M \) is the fraction of \( \alpha' \) transformed during quenching from the \( \gamma \)-region to the \( T_Q \). \( M_S \) is the martensite start temperature, which was calculated according to Mahieu et al. [28]:

\[
M_S = 539 - 423C - 30.4Mn - 7.5Si + 30Al
\]

Figure 2a displays the phase fractions of \( \alpha' \) and \( \gamma \) determined using the CCE-model and Figure 2b represents the related Q&P heat-treatment. Both, KM- and \( M_S \)-formula are applied twice in the CCE-model. First, the \( M_S \) temperature of the initial \( \gamma \) is calculated, followed by the determination of the \( \alpha' \) fraction formed during quenching to a certain \( T_Q \) using the KM equation (blue line). The orange curve represents the remaining initial \( \gamma \) after quenching. Subsequently, the \( M_S \) temperature of the C-enriched \( \gamma \) after the partitioning step is calculated assuming full C-partitioning from \( \alpha' \). Given that, using the KM equation once again, the amount of \( \alpha' \) formed upon final cooling to RT is calculated (red line). The resulting RA fraction as a function of the \( T_Q \) is indicated in green. It is evident that the largest fraction of \( \gamma \) can be retained at that \( T_Q \) where no fresh \( \alpha' \) is formed upon final cooling. However, it must be taken into account that this applied CCE-model does not consider a phase transformation from \( \gamma \) to \( \alpha_B \) during isothermal holding at \( T_P \). Since this contribution focusses on the investigation of the bainitic transformation kinetics during Q&P processing, it can be expected that the amount of retained austenite stabilized to RT will be smaller, compared to the calculated one.

![Phase fractions](image)

**Figure 2.** (a) Calculated phase fractions of martensite (\( \alpha' \)) and austenite (\( \gamma \)) depending on the \( T_Q \) and (b) the related Q&P heat-treatment with the schematic microstructural evolution; the phase fractions are indicated for the initial quench to \( T_Q \) and for the final quench to room temperature (RT).

### 3. Results

#### 3.1. Transformation Behavior

The annealing parameters, especially \( T_Q \) and \( T_P \), remarkably influenced the transformation behavior of the investigated steel. Figure 3 shows the influence of the \( T_P \)-variation in a range of 350–450 °C on the dilatometric curves using the Q&P heat-treatment with a \( T_Q \) of 310 °C (Figure 3a) compared to the TBF cycles (Figure 3b). After full austenitization at 900 °C, the dilatometric samples were cooled to \( T_Q \) or \( T_B \), depending on the applied heat-treatment. In the case of the Q&P regime, \( \alpha'_\text{initial} \) was formed during quenching to \( T_Q \), since the \( M_S \) temperature was approximately 325 °C (Figure 3a). Although the CCE-model, describing the microstructural development during Q&P processing, assumes full C-partitioning from \( \gamma \) to \( \alpha' \) without the occurrence of any phase transformations during isothermal holding at \( T_P \), in the present case after reheating to \( T_P \), \( \gamma \) partially transformed to \( \alpha_B \) during isothermal holding, accompanied by a linear expansion visible in the dilatometric curves. Further, it is obvious that the amount of \( \alpha_B \) described by the volume expansion during isothermal holding decreased with increasing \( T_P \). As a result, \( T_P \) directly influenced the
formation of $\alpha'_{\text{final}}$ during cooling to RT. For the Q&P sample heat-treated at a $T_P$ of 350 °C, no $\alpha'_{\text{final}}$ was formed, whereas at an increased $T_P$ of 400 and 450 °C, a $\gamma \rightarrow \alpha'_{\text{final}}$ transformation was observed by a deviation from the linearity of the dilatometric curves during final cooling. In contrast, for the TBF samples, the largest amount of $\alpha_B$ was formed at a $T_B$ of 400 °C, whereas its smallest amount was contemplated at 450 °C. The correlation between the amount of $\alpha_B$ and $\alpha'_{\text{final}}$ was proven for the TBF regime as well. Thus, the largest amount of $\alpha'_{\text{final}}$ was transformed at a $T_B$ of 450 °C, whereas the lowest $\alpha'$ formation was obtained at 400 °C.

![Dilatometric curves at different $T_P$ (TB)](image)

**Figure 3.** Dilatometric curves at different $T_P$ ($T_B$) of (a) Q&P heat-treatment ($T_Q = 310 \degree \text{C}$) and (b) TBF heat-treatment.

The influence of $T_P$ or $T_B$ on the $\alpha_B$ formation during isothermal holding for the Q&P ($T_Q = 310 \degree \text{C}$) and TBF process is displayed in detail in Figure 4. For both the Q&P and TBF regimes, an increasing $T_P$ and $T_B$ led to accelerated IBT kinetics. However, with an increase of $T_P$ and $T_B$, the $\alpha_B$ formation saturated already after shorter times. In the case of the Q&P samples, this behavior led to the largest amount of $\alpha_B$ at a $T_P$ of 350 °C, whereas with increasing $T_P$, lower $\alpha_B$ fractions were formed during isothermal holding. On the one hand, the IBT in the TBF regime was remarkably slower compared to that of the Q&P process. On the other hand, the IBT did not saturate at a $T_B$ of 350 °C and 400 °C, even after a $t_B$ of 600 s.

![Dilatation due to $\alpha_B$ formation as a function of isothermal holding time at different $T_P$ ($T_B$)](image)

**Figure 4.** Dilatation due to $\alpha_B$ formation as a function of isothermal holding time at different $T_P$ ($T_B$) for (a) Q&P steel ($T_Q = 310 \degree \text{C}$) and (b) TBF steel.

Furthermore, an influence of $T_Q$ on the transformation behavior could be determined by means of dilatometry (Figure 5). With increasing $T_Q$, the amount of $\alpha'_{\text{initial}}$ decreased, whereas the amount of
\( \alpha_B \) formed during isothermal holding at \( T_P \) increased. However, too high \( T_Q \) led to the formation of \( \alpha_{\text{final}} \) during final cooling to RT, especially at a \( T_P \) of 450°C.

\[ \text{Figure 5. (a) Dilatometric curves at different } T_Q \text{ (} T_P = 400 \degree C \text{) and (b) dilatation due to } \alpha_B \text{ formation as a function of isothermal holding time at different } T_Q \text{ (} T_P = 400 \degree C \text{).} \]

3.2. Microstructure

The annealing parameters, especially \( T_Q \), significantly influenced the microstructural constitution of the investigated samples. Figure 6a–c shows the LOM images of the Q&P samples, quenched to a temperature of 190, 250, and 310 °C, respectively. In comparison, the microstructure of the sample annealed in the TBF regime is shown in Figure 6d. All four conditions were heat-treated at a \( T_P \) \((T_B) \) of 400 °C. For the Q&P steels, the matrix consisted of a mixture of \( \alpha'_{\text{initial}} \) and \( \alpha_B \), which appeared as bluish and brownish areas in the micrographs. Furthermore, RA and/or \( \alpha'_{\text{final}} \) islands (white and brownish areas) could be observed as finely distributed in the matrix. At the lowest \( T_Q \) of 190 °C, the secondary phase only consisted of RA, whereas the increase of \( T_Q \) led to the pronounced formation of \( \alpha'_{\text{final}} \)

The microstructure, especially at a \( T_Q \) of 250 and 310 °C exhibited a banded structure, resulting from the segregation of alloying elements such as Mn during casting and subsequent solidification. In contrast, the microstructure of the TBF samples consisted of a ferritic-bainitic matrix with small amounts of finely distributed RA islands, surrounded by large fractions of \( \alpha'_{\text{final}} \).

\[ \text{Figure 6. Light optical microscopy (LOM) images of the (a–c) Q&P samples } (T_Q = 190, 250, \text{ and } 310 \degree C) \text{ and (d) TBF sample at a } T_P \text{ (} T_B \text{) of } 400 \degree C \text{ (magnification } 1000 \times \).} \]

For higher resolution investigations, the microstructure was further characterized by SEM. The SEM images of the Q&P samples quenched to a \( T_Q \) of 190, 250, and 310 °C are displayed in Figure 7a–c, whereas in Figure 7d the microstructure after the TBF heat-treatment is shown for comparison. In the case of the Q&P samples, the microstructure contained cementite precipitates with
an obvious triaxial alignment, which means that the precipitations are arranged in an angle of 120°. This indicates the presence of tempered α’init. Furthermore, the fraction of αB in the matrix rose along with the amount of α’fin with increasing TQ. In comparison, the microstructure of the TBF steel consisted of a αB-matrix with large amounts of α’fin and low fractions of RA.

Figure 7. SEM images of the (a–c) Q&P samples (TQ = 190, 250, and 310 °C) and (d) TBF sample at a Tp (TB) of 400 °C (magnification 5000 ×).

Figure 8 depicts the amount of RA, determined by SMM, for the TBF samples (dotted lines) and the Q&P samples at a Tp of 350, 400, and 450 °C, respectively. In addition, the RA fraction calculated according to the CCE-model proposed by Speer et al. [20] is presented in the diagram. The CCE-model proposed a RA maximum (RAmax) of 19.7 vol.% at a TQ of 210 °C. In accordance with the model calculations, the experiments also confirmed the shape of the RA evolution as a function of TQ. At lower TQ the formation of high amounts of α’init led to lower RA fractions. By a further increase of the TQ, the RAmax was achieved followed by a decrease of the RA. This was due to its lower stabilization and therefore the formation of α’fin upon cooling to RT. Furthermore, the experiments showed an influence of Tp on the amount of RA: with increasing Tp, RAmax rose from 14.8 to 19.0 vol.%. In this context, at the Tp of 450 °C, the amount of RA correlated well with the model calculations. On the contrary, at the Tp of 350 °C and 400 °C, a lower amount of RA was achieved compared to the model predictions. Moreover, the TQ at which the RAmax occurred, was shifted to higher temperatures by up to 80 °C compared to the CCE-model. It is evident that the RA contents of the TBF samples were lower compared to the Q&P samples: 7.1 vol.% RA at a TB of 350 °C, 9.7 vol.% RA at TB = 400 °C, and 3.6 vol.% RA at TB = 450 °C.

Figure 8. Retained austenite (RA) content as a function of TQ for a Tp (TB) of 350, 400, and 450 °C.
Figure 9 represents microstructure charts, where the phase fractions are plotted as a function of $T_Q$ for different $T_P$. In addition, the bar charts, situated close to the right rim of each diagram, depict the microstructural constitution for the TBF steel. In case of the Q&P samples, the decrease of $T_Q$ led to a considerable increase of $\alpha'_\text{initial}$. As a consequence, a lower amount of $\alpha_B$ was formed during IBT, regardless of $T_P$. However, particularly at higher $T_Q$, a vivid influence of $T_P$ is obvious: the increase of $T_P$ led to a decreasing fraction of $\alpha_B$, resulting in the formation of $\alpha'_\text{final}$ upon cooling to RT. Therefore, the RA content steadily rose with increasing $T_Q$ until the onset of the formation of $\alpha'_\text{final}$ occurred. For the TBF samples, it is evident that the largest amount of $\alpha_B$ was formed at a $T_B$ of 400 °C as already shown by dilatometry (Figure 4b). Due to the lowest $\alpha_B$ fraction at a $T_B$ of 450 °C, the largest amount of $\gamma$ transformed to $\alpha'_\text{final}$ during final cooling, resulting in the lowest amount of RA, followed by the samples heat-treated at a $T_B$ of 350 °C and 400 °C.

3.3. Hardness

The results of the hardness measurements according to Vickers are plotted as a function of $T_Q$ in Figure 10. When the $T_Q$ was in the range of 130 to 250 °C, both increasing $T_Q$ and $T_P$ led to a slightly decreasing hardness. However, at $T_Q$ exceeding 250 °C, the hardness rose with $T_Q$, especially at a $T_P$ of 450 °C. For this reason, in case of the Q&P samples, the hardness was between approximately 350 and 460 HV1. In contrast, for the TBF samples the lowest hardness of 445 HV1 was measured at a $T_B$ of 400 °C, whereas at a $T_B$ of 350 and 450 °C the hardness reached approximately 470 and 475 HV1, respectively.
4. Discussion

The time-temperature-transformation (TTT) diagrams for isothermal holding at different $T_P$ ($T_B$) depict the comparison between the Q&P heat-treatment (Figure 11a–c) at different $T_Q$ (190, 250, and 310 °C) and the TBF processing route (Figure 11d). It is obvious that the application of the Q&P process resulted in an evident acceleration of the $\alpha_B$ formation. This is in agreement with Wang et al. [29], since the presence of $\alpha'_1$ pronouncedly accelerated the IBT due to the enhanced nucleation rate by the presence of geometrically necessary dislocations. At higher $T_Q$, the amount of $\alpha_B$ significantly increased, in particular at the lower $T_P$ of 350 and 400 °C. According to Smanio and Sourmail [30], this relation can be explained by the fact that at higher $T_Q$ lower amounts of $\alpha'_1$ were formed, which was also confirmed by the LOM and SEM investigations (Figures 6 and 7). As a result, a larger amount of $\gamma$ was present in the microstructure at the onset of isothermal holding, from which a larger amount of $\alpha_B$ could be formed, compared to lower $T_Q$. For this reason, the largest $\alpha_B$ formation was observed in the TBF samples, where especially a lower $T_B$ of 350 °C led to increased $\alpha_B$ fractions up to 80 vol.% after the investigated holding time. Furthermore, a general impact of the $T_p$ ($T_B$) on the transformation kinetics was found: especially at a $T_p$ ($T_B$) of 400 and 450 °C the transformation kinetics was evidently accelerated due to the faster C diffusion into the remaining $\gamma$. In general, the increase of $T_p$ and $T_B$ led to lower fractions of $\alpha_B$, which is in analogy with the $T_0$-concept [31]. Thereby, the difference between the Gibbs free energies of $\gamma$ and $\alpha$ decreases with an increase of temperature. As a consequence, the driving force for $\alpha_B$ formation diminishes with temperature.

![Figure 11. Time-temperature-transformation (TTT) diagrams for (a-c) Q&P steels ($T_Q = 190, 250, \text{and } 310 \degree C$) and (d) TBF steels.](image-url)
In the case of the TBF samples, the swing back phenomenon could be contemplated, which is in accordance with Oka and Okamoto [32]. This effect describes the accelerated formation of $\alpha_B$ due to the formation of thin-plate isothermal $\alpha'$ in a temperature range just above $M_S$. Therefore, at a $T_B$ of 350 °C, the onset of the $\alpha_B$ formation was observed at shorter times compared to a $T_B$ of 400 °C.

Considering the RA investigations, a significant difference between the calculated CCE-model and experimental data was determined. This was particularly apparent for the amount of RA and optimal $T_Q$. In analogy to the present results, these observations have already been reported in the literature [24,33,34]. The CCE-model assumes full C-partitioning from $\alpha'_{\text{initial}}$ into the remaining $\gamma$. However, in the present study the microstructural investigations have shown the presence of cementite precipitates in the tempered $\alpha'_{\text{initial}}$ (Figure 7), which contradicts this assumption. The insufficient C-partitioning led to a reduced C-content in the remaining $\gamma$, resulting in a lower amount of RA due to the formation of $\alpha'_{\text{final}}$, caused by the lower chemical RA stability. In particular, for the $T_P$ of 350 °C, the $RA_{\text{max}}$ was remarkably lower than the predicted one. According to Clarke et al. [24], the $\alpha_B$ formation, which predominantly occurred at a $T_P$ of 350 °C in the present study, led to a pronounced decrease in RA as the aftermath of the reduction of untransformed $\gamma$ during isothermal holding.

Furthermore, the microstructural changes were reflected in hardness according to Vickers, as well (Figures 9 and 10). With increasing RA, content hardness steadily decreased, resulting in the lowest hardness for those samples containing the largest amount of RA. Since with increasing $T_P$ the volume fraction of RA stabilized to RT rose, the lowest hardness values could be obtained at $T_P = 450$ °C. However, the increased RA content at higher $T_P$ led to the lower chemical RA stability, and thus to an intensified formation of $\alpha'_{\text{final}}$. Therefore, especially at $T_P = 450$ °C, hardness drastically increased at $T_Q$, exceeding that temperature where the maximum RA content was measured. This is in agreement with [35], where a comparable correlation between the microstructural constituents (i.e., RA and $\alpha'_{\text{final}}$) and resulting hardness was found for lean medium Mn steels.

The evident shift of the $RA_{\text{max}}$ to higher $T_Q$ than predicted could be explained by the fact that $\gamma$ is stabilized by both chemical and mechanical factors, whereas the CCE-model only takes the chemical stabilization into account [36]. Furthermore, the CCE-model includes two empirical formulas, describing the $\alpha'$ kinetics and the $M_S$ temperature. According to Kim et al. [37], slight differences in the chemical composition or the initial microstructure could therefore lead to deviations from the model, as well. In the present contribution, small deviations between the $M_S$ temperature calculated according to Mahieu et al. ($M_S = 350$ °C) and the $M_S$ temperature determined by means of dilatometry ($M_S = 325$ °C, Figure 5a) could be observed, as well. Therefore, an adjustment of the applied $M_S$-formula with an improved validity for increased Mn and C contents is intended for future investigations. Nevertheless, the CCE-model proposed by Speer et al. [20] is a vital tool for the first estimation of applicable annealing parameters in the case of Q&P steels.

5. Conclusions

The results of the present contribution give conclusive evidence that the Q&P process significantly influences the transformation behavior of the investigated lean medium Mn steel. Both $T_Q$ and $T_P$ must be carefully adjusted, since they have a substantial impact on the microstructural evolution during steel processing.

The main findings of the present study are as follows:

- The Q&P heat-treatment accelerated the transformation kinetics during isothermal holding in the partitioning step due to the presence of $\alpha'_{\text{initial}}$, which acted as nucleation sites for the $\alpha_B$ formation.
- For both Q&P and TBF grades, the increase of $T_P$ ($T_B$) resulted in lower amounts of $\alpha_B$, although the transformation kinetics was accelerated.
- With increasing $T_Q$, smaller amounts of $\alpha'_{\text{initial}}$ and larger amounts of $\alpha_B$ were formed, leading to an increased RA content. At too high $T_Q$, $\alpha'_{\text{final}}$ was formed, resulting in a sharp decrease in
RA. Therefore, the microstructure of the Q&P samples consisted of tempered $\alpha'_{\text{initial}}$, RA, and at higher $T_Q$ partially of $\alpha_0$ and $\alpha'_{\text{final}}$.

- The microstructure of the TBF samples with the present lean medium Mn composition consisted of a ferritic-bainitic matrix with large amounts of $\alpha'_{\text{final}}$ and low fractions of RA.
- The comparison of the CCE-model and the experimental data showed a significant deviation of the maximum RA content, in particular at lower $T_P$. Furthermore, the optimum $T_Q$ was shifted to higher temperatures compared to the CCE-model predictions.


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**References**


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