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Microstructure model for the heat-affected zone of X80 linepipe steel

T. Garcin¹, M. Militzer*, ¹, W. J. Poole⁴ and L. Collins²

An important aspect of the integrity of oil and gas pipelines is the heat-affected zone (HAZ) of girth welds where the microstructure of the as-hot rolled steel is altered with potentially adverse effects on the HAZ properties. Therefore, it is critical to evaluate the HAZ microstructure for different welding scenarios. Here, an integrated microstructure evolution model is proposed and applied to the HAZ of an X80 linepipe steel. The model considers dissolution of Nb-rich precipitates, austenite grain growth and austenite decomposition into ferrite and bainite. Microstructure maps showing the fraction of transformation products as a function of distance from the fusion line are obtained and used to compare the effect of different welding procedures on the HAZ microstructure.

**Keywords:** Linepipe steel, Heat-affected zone, Microstructure modelling, Precipitate dissolution, Austenite grain growth, Austenite decomposition, Ferrite, Bainite, Martensite/austenite

**Nomenclature**

- \( a \): thermal diffusivity
- \( b_i \): parameters for bainite start model
- \( C_v \): volumetric heat capacity for iron
- \( D_0 \): pre-exponential factor for the carbon diffusivity in austenite
- \( D_{Nb}^0 \): pre-exponential factor for the niobium diffusivity in austenite
- \( D_C \): carbon diffusivity in austenite
- \( D_{Nb} \): niobium diffusivity in austenite
- \( d_y \): mean austenite grain diameter
- \( f_{N(CN)(L)} \): volume fraction of large Nb(CN) precipitates
- \( f_{N(CN)(S)} \): volume fraction of small Nb(CN) precipitates
- \( f_0 \): initial volume fraction of Nb(CN) precipitates in the as-received material (small and large precipitates)
- \( f_{Ti,N} \): initial volume fraction of TiN precipitates
- \( F^F,B \): true fraction of ferrite (F) and bainite (B)
- \( F^M,A \): fraction of martensite/austenite constituent
- \( F^U,B \): fraction of upper bainite
- \( F^L,B \): fraction of lower bainite
- \( F^U,L,B \): normalised fraction of upper bainite (UB) and lower bainite (LB)
- \( k_B \): Boltzmann constant
- \( K_{N(CN)} \): temperature-dependent solubility product for Nb(CN) precipitates
- \( M^A \): mobility of the austenite grain boundary
- \( M_{Nb}^A \): pre-exponential factor for the mobility of austenite grain boundary
- \( n \): JM AK exponent
- \( q \): heat input
- \( Q^* \): activation energy for the austenite grain boundary mobility

\( Q_C^0 \): activation energy for the carbon diffusivity in austenite
\( Q_{Nb}^0 \): activation energy for the niobium diffusivity in austenite
\( r \): distance from centre line
\( f_{TiCN(L)} \): mean radius of large (L) Nb(CN) precipitates
\( f_{TiCN(S)} \): mean radius of small (S) Nb(CN) precipitates
\( r_{TiN} \): mean radius of TiN precipitates
\( R \): gas constant
\( R_f \): radius of the growing ferrite nuclei
\( T \): absolute temperature
\( T_{pre} \): pre-heat temperature
\( T_{nuc} \): nucleation temperature for bainite
\( T_{ferrite} \): nucleation temperature for ferrite
\( T_{peak} \): peak temperature
\( T_{start} \): bainite start transformation temperature
\( T_{start} \): ferrite start transformation temperature
\( v \): weld speed
\( V \): mean atomic volume for iron
\( V_{N(CN)} \): mean atomic volume for Nb(CN) precipitates

\( X^C \): mean atomic fraction of carbon in the matrix
\( X_{Nb}^C \): nominal atomic fraction of niobium in the steel
\( X_{Nb}^M \): equilibrium atomic fraction of Nb in ferrite
\( X_{M,Nb} \): solute atomic fraction of carbon and nitrogen in the matrix

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Greek symbols

\( \alpha \)  

geometrical constant for the grain curvature

\( \beta \)  

geometrical constant for particle pinning of the grain boundary

\( P_{\text{JM}} \)  

JMVK rate parameters for austenite decomposition

\( \varepsilon \)  

intensity of the interaction of solute Nb with the austenite/ferrite interface

\( \phi \)  

thermal conductivity for iron

\( \phi_{\text{Ni}} \)  

parameters for ferrite start model

\( \sigma_{\text{gb}} \)  

austenite grain boundary energy

\( \sigma_{\text{m}} \)  

matrix precipitate interfacial energy

Introduction

There is currently a need to build new or replace existing natural gas and oil pipelines in North America in order to distribute these resources in a safe and cost-effective manner from northern locations to market. The trend in the industry is to use pipelines of larger diameter and/or increase the operating pressure of the pipeline. This leads to using higher strength linepipe steel grades to avoid thicker wall dimensions, such as X80 or even X100 linepipe grades with a minimum yield stress of 550 and 690 MPa, respectively, for new pipeline construction. In addition to the higher operating pressures, minimisation of construction costs is a prime concern. Using higher strength steels translates into less material being required thereby lowering material and transportation cost. Lighter wall thicknesses may also require fewer weld passes thereby improving construction efficiency. Further, pipeline integrity hinges to a significant degree on having high-efficiency girth welds, that is, the in-field welds of pipeline segments. Thus, advanced welding procedures are being considered such as tandem-wire, dual-torch and laser-hybrid welding. An important concern here is the heat-affected zone (HAZ) in which mechanical properties are modified compared to the base metal. In particular, the austenite (\( \gamma \)) – ferrite (\( \alpha \)) transformation assumes a critical role in determining HAZ microstructures and resulting mechanical properties. It is critical for future pipeline projects, which incorporate new design approaches and welding procedures that fundamental microstructure-property knowledge is advanced for the HAZ.

In the HAZ, austenite formation and austenite grain growth occur during rapid heating as the welding torch passes. Experimental measurements indicate that the heating rate is \( >1000 \text{°C s}^{-1} \) under typical welding conditions. An extensive review by Mishra and DebRoy summarises non-isothermal austenite formation and grain growth behaviour with significant spatial and temporal variations that are relevant for the HAZ.\(^5\) Ashby and Easterling considered a complete welding cycle to produce grain growth diagrams for the prediction of the mean grain size in real or simulated welds,\(^2\) whereas Andersen and Grong developed an analytical approach for grain growth in the presence of coarsening and dissolution.\(^3\) More recently, Banerjee et al. investigated austenite formation and grain growth during rapid heating for the same X80 steel considered in this work and developed a grain growth model that accounts for the presence of precipitates and their potential dissolution.\(^4\) The physically based model used in the work of Banerjee et al. incorporates the work of Moon et al. who quantified the effect of particle pinning on grain boundary migration during isothermal austenite grain growth associated with particle coarsening and dissolution.\(^3\)

The subsequent decomposition of austenite and its relation to the prior austenite grain structure has been studied for the HAZ. For example, Henwood et al.\(^6\) used a finite element heat transfer analysis in combination with the Ashby–Easterling grain growth model and the solid-state Kirkaldy reaction approach for the austenite decomposition kinetics to compute the microstructure as a function of space and time. Bhadeshia et al.\(^7\) calculated the ferrite growth kinetics under paraequilibrium conditions using the principle of additivity. A review of these approaches can be found in the classic textbook by Grong.\(^8\) More recently, Zhang et al. described the austenite decomposition in the HAZ of a low carbon steel using a numerical model based on the Johnson–Mehl–Avrami–Kolmogorov (JMVK) analysis and Monte-Carlo simulations.\(^9\) A critical aspect of these various approaches is the need to explicitly link the various metallurgical processes during a welding cycle, for example, grain growth, dissolution of microalloying precipitates and decomposition of austenite to predict the graded microstructure present in the HAZ and its associated mechanical properties.

The concept of microstructure engineering applied to the development of linepipe steels has recently gained increasing attention to link the operational parameters of a welding operation to the microstructure and resulting properties of the material.\(^10\) In this approach, the evolution of the microstructure is described by a set of linked differential equations, which are integrated over the relevant thermal history aided, as required, by empirical relationships. For the HAZ in microalloyed steels, the modelling approach carefully considers austenite grain growth and dissolution of precipitates,\(^4,11\) since the austenite grain size and the solute content of Nb and Ti strongly affects the subsequent decomposition of austenite.\(^12,13\) In particular, a large austenite grain size and/or a high level of Nb in solid solution favours the formation of transformation products such as bainite and the associated martensite/austenite (M/A) constituents, which may have detrimental effects on weld toughness, hydrogen resistance, cold and reheat cracking.\(^2,14,19\) A recent study by Gaudet\(^25\) has shown the variation of tensile properties and tear resistance over a range of temperatures (20 to –60°C) for relevant microstructures found in the HAZ of the current X80 steel (with the exception of the intercritical region, which is currently under investigation).

As part of a larger project, a significant body of experimental and theoretical work has recently been conducted for a Ti–Nb microalloyed X80 linepipe steel.\(^4,11,20–25\) The aim of this paper is to integrate the previous work on austenite grain growth, dissolution of microalloyed precipitates and decomposition of austenite into a single model to allow for the prediction of microstructure evolution in the weld HAZ for this steel with an emphasis on relevant conditions of girth welds. Additional experiments have been conducted to validate and, where necessary, to refine the model parameters previously reported in the literature and examine the predictive capability of the model. A simplified temperature model (calibrated by experimental measurements in welds) has been
implemented with the microstructure evolution model in order to construct maps of the microstructure in the HAZ for different welding conditions.

### Background and experiments

The material used for the construction of the model is a hot-rolled X80 linepipe steel supplied by EVRAZ, Inc. NA (Regina, SK) with a nominal composition reported in Table 1. The microstructure of the as-received steel consists mainly of irregular ferrite in which MA particles are randomly distributed. The experimental studies previously reported for this steel are briefly summarised in the following. Based on the review of these experimental results, a number of additional experiments are proposed to validate and/or refine parameters required for the integrated model.

Gaudet conducted a series of weld trials on plates instrumented with thermocouples spot welded in the bottom of small holes drilled at various positions from the weld fusion line. The temperature cycles, recorded in situ during the weld trials, provide critical information on the relevant heating rates, peak temperatures and cooling rates at different positions in the HAZ for gas metal arc welding (GMAW) in single- and dual-torch configurations. More recently, similar experiments were also extended to submerged arc welding (SAW) conditions in order to compare the two welding techniques and their influence on the resulting microstructure.

The state of precipitates present in the as-received hot rolled steel was quantified by transmission electron microscopy (TEM) investigations coupled with energy dispersive X-ray (EDX) and selected area diffraction pattern analysis. It was concluded that the initial structure after hot rolling and coiling showed three families of precipitates homogeneously distributed within the structure, that is, Ti-rich, Nb-rich and Mo-rich particles. Their sizes range from 1 to 200 nm. A detailed analysis of the Nb-rich precipitates revealed two populations, that is, large precipitates that presumably had formed in austenite during hot rolling with a radius >10 nm and small precipitates formed in ferrite during coiling with a radius <5 nm. Thermodynamic calculations with Thermocalc (TCFE7 database) indicate that the dissolution temperature of Mo-rich precipitates is 670°C, whereas the dissolution temperature for the Nb and Ti-rich precipitates are 1080 and 1445°C, respectively. Thus, it is assumed that the Mo-rich precipitates are fully dissolved when the steel reaches the austenite region during weld heat treatment cycles, that is, all Mo is in solid solution. Second, according to the high dissolution temperature for the TiN precipitates, it is assumed that the equilibrium fraction of TiN precipitates is present in the as-received material and will remain stable throughout the HAZ. Thus, only the Nb(CN) precipitates are considered to evolve (in size and volume fraction) during the HAZ heat treatment cycles.

<table>
<thead>
<tr>
<th>Elements</th>
<th>C</th>
<th>Mn</th>
<th>Mo</th>
<th>Nb</th>
<th>Ti</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt-%</td>
<td>0.060</td>
<td>1.65</td>
<td>0.24</td>
<td>0.0340</td>
<td>0.0120</td>
<td>0.005</td>
</tr>
<tr>
<td>at.-%</td>
<td>0.278</td>
<td>1.67</td>
<td>0.14</td>
<td>0.0204</td>
<td>0.0139</td>
<td>0.019</td>
</tr>
</tbody>
</table>

The kinetics of austenite grain growth were investigated during two main experimental test series. First, the austenite grain size was examined during continuous heating and the influence of heating rate on the austenite grain size was correlated to the peak temperature. The second test series aimed at identifying in more detail the relative fraction of small and large precipitates. To this aim, the limiting austenite grain size was measured during isothermal holding at various temperatures by conducting a series of in situ grain size measurements with a laser ultrasonics for metallurgy (LUMet) system. The relationship between the limiting austenite grain size and the limiting pinning force associated with the presence of Nb(CN) precipitates and their dissolution was thoroughly examined using detailed analysis of the precipitates solubility and considering a size distribution for the precipitates using MatCalc. This detailed analysis led to a slightly modified model compared with that proposed by Banerjee et al. with both models leading to satisfactory descriptions of grain growth during continuous heating scenarios relevant for HAZ thermal cycles. Here, we will adopt the model proposed by Banerjee et al. because of its simplicity and ease to incorporate into an integrated model, for example, no coupling to MatCalc software is required.

The subsequent austenite decomposition occurring upon cooling was investigated by conducting a number of continuous cooling transformation (CCT) experiments. Test samples were heat treated to bring all Nb in solid solution and quenched before conducting the austenitisation treatment for the CCT studies. The austenitisation treatments were designed to obtain a predetermined austenite grain size and then the samples were either directly continuously cooled such that all Nb remain in solution or subjected to a holding of 20 minutes at 900°C (where Nb precipitates and only a small fraction remains in solid solution) before continuous cooling. The fraction of Nb remaining in solution as a function of holding time at 900°C was estimated from ageing tests at 570°C based on the magnitude of the age-hardening peak.

The austenite decomposition kinetics were measured with a dilatometer for different initial austenite conditions (grain size, Nb in solution) and a range of cooling rates (1–40°C s⁻¹) that are of relevance for the HAZ and different welding scenarios. Conventional metallography with Nital and Le Pera etches was employed to quantify the as-transformed microstructures in terms of the area fraction of each microstructure constituent, that is, ferrite/bainite and MA, respectively. Based on the results of these CCT tests, a phenomenological model for the austenite decomposition into ferrite and bainite was developed. The model consists of submodules for ferrite start, ferrite growth, bainite start and bainite growth. Furthermore, the fraction of MA constituents in the different samples generated from CCT tests was related to the transformation start temperature measured upon cooling. The proposed overall austenite decomposition model replicates the CCT observations with sufficient accuracy.

The initial fraction of Nb present in solid solution in the base metal is an important parameter in the model but is difficult to quantify accurately. Initially, it was considered in the work of Banerjee et al. that the equilibrium fraction of Nb is reached after hot rolling and coiling of
Modelling strategy

Initial state of precipitates

The initial volume fraction of TiN precipitates \( f_{\text{TIN}}^0 \) is estimated to be \( 2.3 \times 10^{-4} \) assuming that all Ti is precipitated. The initial volume fraction of Nb(CN) precipitates \( f_{\text{NbCN}}^0 \) is assumed to be lower than the equilibrium fraction and can be related to the atomic fraction of niobium \( X_{\text{Nb}}^0 \) in the matrix of the as-hot rolled steel by

\[
\frac{f_{\text{NbCN}}^0}{v_{\text{Fe}}} = (X_{\text{Nb}}^0 - X_{\text{Nb}}^0) \frac{v_{\text{Al}}}{v_{\text{Fe}}} \tag{1}
\]

where \( X_{\text{Nb}}^0 \) is the nominal atomic fraction of niobium in the steel, \( v_{\text{Fe}} \) is the atomic volume for iron and \( v_{\text{Al}} \) is the atomic volume for Nb(CN). As mentioned earlier, the initial atomic fraction of niobium in the matrix is not known accurately and is thus considered as an adjustable parameter in the present approach.

In addition to the initial volume fractions precipitated, the initial mean particle radii are required in the model. Although well aware of the uncertainties associated with the measurement of precipitate sizes using TEM, Banerjee et al. had estimated the mean radii for the various precipitate families. In particular, they found two families of Nb(CN) precipitates, that is, larger precipitates (mean radius of \( r_{\text{NbCN(L)}}^0 = 69 \text{ nm} \)) and smaller precipitates (mean radius of \( r_{\text{NbCN(S)}}^0 = 2 \text{ nm} \)). A mean radius \( r_{\text{TIN}}^0 = 61 \text{ nm} \) for TiN precipitates was adopted for the present approach.

As proposed by Banerjee et al., the initial volume fraction of large Nb(CN) precipitates is estimated from the ratio of the overall precipitate volumes in the as-received material for large Nb(CN) (23 pct.) and TiN (61 pct.) such that

\[
\frac{f_{\text{NbCN(L)}}^0}{f_{\text{TIN}}} = \frac{23}{61} \tag{2}
\]

The initial volume fraction of small Nb(CN) precipitates is then deduced from the total initial volume fraction of Nb(CN) precipitates \( f_{\text{NbCN}}^0 \) by

\[
f_{\text{NbCN(S)}}^0 = f_{\text{NbCN}}^0 - f_{\text{NbCN(L)}}^0 \tag{3}
\]
This initial precipitation state is only applicable for the as-hot rolled state of the investigated steel and would have to be reconsidered for a different steel chemistry or even a significantly different rolling schedule for the present steel.

**Dissolution of precipitates and austenite grain growth**

During welding, the initial population of precipitates evolves with respect to the temperature history at a given distance from the fusion line. For the rapid heat treatment conditions in the HAZ, only dissolution of precipitates will be considered, whereas potential growth and coarsening that are slower processes will not be taken into account. Further, the austenite grain size is described as the mean value of the size distribution, and potential changes in the width of the size distribution are not considered in this approach. The evolution of the mean austenite grain diameter $d_i$ is affected by the evolution of the precipitate population via a Zener-type pinning pressure. In such a situation, the evolution of the mean austenite grain diameter can be expressed as:

$$\frac{d}{dt}d_i = M^\gamma \sigma^\theta\left(\frac{\alpha}{d_i} - \beta \sum_i \frac{d_i^*}{d_i}\right)$$  \hspace{1cm} (3)

where the summation index $i$ represents the types of precipitates present, TiN, small Nb(CN) and large Nb(CN) precipitates. The parameters $\alpha$ and $\beta$ are geometrical constants, $\sigma^\theta$ is the grain boundary mobility, $M^\gamma$ can be expressed by an Arrhenius relationship, that is, $M^\gamma = M^\gamma_0 \exp\left(-\frac{Q}{RT}\right)$, where $M^\gamma_0$ is the pre-exponential factor and $Q^\gamma$ is the activation energy, $R$ is the gas constant and $T$ is the absolute temperature.

In this approach, the dissolution of only two families of precipitates is considered, that is, large and small Nb (CN). Due to their high dissolution temperature, the variation in the fraction of TiN precipitates is neglected in the present model. The diffusion-controlled dissolution rate of spherical Nb(CN) precipitates embedded in a solid solution is described by:

$$\frac{d}{dt}X_{NbCN(L,S)} = \frac{D_{NbCN}}{v_{Fe}} X_{NbCN(L,S)}^M - X_{NbCN(L,S)}^I$$  \hspace{1cm} (4)

here, $D_{NbCN}$ is the diffusivity of niobium in the matrix of austenite and is given by an Arrhenius relationship, $D_{NbCN} = D_{NbCN}^0 \exp\left(Q_{NbCN}/RT\right)$. The variable $X_{NbCN}^M$ is the atomic fraction of niobium in the precipitate, that is, 0.5 for Nb(CN). $X_{NbCN}^I$ is the solute atomic fraction of niobium in the matrix and $X_{NbCN}^s$ is its equilibrium atomic fraction at the precipitate/matrix interface. Due to the small sizes of the precipitates, the interface curvature plays an important role in the equilibrium atomic fraction at the interface, that is, the so-called Gibbs–Thomson effect, such that the equilibrium solubility depends on the precipitates radius and the matrix precipitate interfacial energy $\sigma$ as:

$$X_{NbCN}^s = K_{NbCN} \exp\left(\frac{4\sigma_{NbCN}^s}{RT}\right)$$  \hspace{1cm} (5)

where $X_{NbCN}^s$ is the solute atomic fraction of carbon and nitrogen in the matrix, $K_{NbCN}$ is the temperature-dependent solubility product for Nb(CN) precipitates and $k_B$ is Boltzmann’s constant.

In this approach, the number density for each precipitate family is considered constant such that the volume fraction for each class can be updated according to the evolution of their mean radii. The volume fraction is then used to evaluate the mean solute atomic fraction of niobium as well as carbon and nitrogen in the matrix according to mass balance, that is,

$$X_{Nb}^M = X_{Nb}^i - \frac{1}{2} \frac{v_{Fe}}{v_{Fe}} (f_{NbCN(L)} + f_{NbCN(S)})$$  \hspace{1cm} (6)

$$X_{C+N}^M = X_{C+N}^i - \frac{1}{2} \frac{v_{Fe}}{v_{Fe}} (f_{NbCN(L)} + f_{NbCN(S)})$$  \hspace{1cm} (7)

where $X_{C+N}^i = X_C^i + (X_N^i - X_N^L)$ is the nominal carbon concentration plus the excess nitrogen that is not precipitated in TiN. Following this approach, the evolution of the mean austenite grain diameter as well as mean radius and volume fraction for each precipitate family is calculated along the thermal path.

Table 2 summarises the parameters used in the combined precipitate dissolution and austenite grain coarsening model. This model provides two important parameters for the subsequent austenite decomposition model, that is, (1) the prior austenite grain diameter and (2) the mean atomic fraction of niobium in solution in the matrix. In addition to cooling rate, these two parameters affect the onset of the austenite decomposition during subsequent cooling.

**Austenite decomposition**

In this approach, two separate models are utilised for the prediction of the transformation start temperature since austenite decomposition may start with either ferrite or bainite in the present steel depending on the HAZ cooling conditions.

The model constructed to predict the onset of ferrite formation was originally developed for plain carbon steels.\(^{27}\) It considers the carbon diffusion-controlled early growth of ferrite nuclei formed at austenite grain boundary corners at a temperature $T_f$. The model was extended to include the solute drag effect of niobium on the moving austenite/ferrite interface.\(^{22,40}\) The evolution of the mean radius of the corner-nucleated ferrite grain can be written such that:

Table 2 Parameters for the austenite grain growth and precipitate dissolution model

<table>
<thead>
<tr>
<th>Parameters</th>
<th>Value</th>
<th>References</th>
</tr>
</thead>
<tbody>
<tr>
<td>$D_{NbCN}^{0} \text{m}^2 \text{s}^{-1}$</td>
<td>$5.3 \times 10^{-2}$</td>
<td>33</td>
</tr>
<tr>
<td>$Q_{NbCN}^{0} \text{kJ mol}^{-1}$</td>
<td>344</td>
<td>33</td>
</tr>
<tr>
<td>$Q_{C+N}^{0} \text{kJ mol}^{-1}$</td>
<td>350</td>
<td>34</td>
</tr>
<tr>
<td>$\alpha$</td>
<td>4</td>
<td>35</td>
</tr>
<tr>
<td>$\beta$</td>
<td>12</td>
<td>36</td>
</tr>
<tr>
<td>$\sigma_{NbCN}^{s} \text{m}^{-2}$</td>
<td>0.5</td>
<td>37</td>
</tr>
<tr>
<td>$\sigma_{C+N}^{s} \text{m}^{-2}$</td>
<td>0.5</td>
<td>33</td>
</tr>
<tr>
<td>$\log 10(K_{NbCN})$</td>
<td>$-1.32 - 6670/T$</td>
<td>38</td>
</tr>
<tr>
<td>$D_{C}^{0} \text{m}^2 \text{s}^{-1}$</td>
<td>$1.5 \times 10^{-5}$</td>
<td>39</td>
</tr>
<tr>
<td>$D_{N}^{0} \text{kJ mol}^{-1}$</td>
<td>142.1</td>
<td>39</td>
</tr>
</tbody>
</table>
where \( R_i \) is the radius of the growing ferrite grain, \( D_C \) is the carbon diffusivity in austenite described by an Arrhenius relationship \( D_C = D_C^0 \exp(-Q_C/R_T) \) and \( \varphi \) is the instantaneous cooling rate. \( X_C^a \) and \( X_C^b \) are the equilibrium atomic fraction of carbon in austenite and ferrite, respectively. The coefficient \( \epsilon \) is an adjustable parameter describing the intensity of the interaction of solute Nb with the austenite/ferrite interface following the approach described by Fazeli and Militzer.\(^{25}\) The temperature dependence of \( X_C^a \) and \( X_C^b \) are calculated with Thermocalc using the TCFe7 database and assuming full equilibrium for all alloying elements, that is, orthoequilibrium. The condition for measurable start temperature of ferrite formation \( T_S^F \) is given by attaining a critical carbon level \( X_C^c \) at the austenite grain boundary such that

\[
R_i > \frac{1}{\sqrt{2}} \left[ X_C^c - X_C^N \right] d_f
\]

According to the analysis by Militzer et al.,\(^{28}\) the parameter \( X_C^c \) is related to \( d_f \) with an empirical expression of the form \( X_C^c/X_C^N = b_1 + b_2/d_f \). Further, it is considered that the transformation start condition coincides with reaching a measurable fraction of ferrite, that is, 5% transformed.

For sufficiently fast cooling, ferrite does not form and austenite decomposition starts with the formation of bainite. The bainite start temperature can be attributed to a critical driving pressure,\(^{41}\) but measurable bainite start temperature is similar to the ferrite start temperature associated with 5% transformed. Based on the CCT studies, the following expression had been proposed to find the bainite start temperature \( T_S^B \):\(^{42}\)

\[
0.05 = \int_{T_S^B}^{T_S^B} \left[ b_1 + b_2(T - 273.15) \right] dT
\]

Here, the temperature \( T_S^B \) is the temperature where the critical driving pressure is reached for the given steel composition and the \( b_i \) are empirical parameters. Similar to the ferrite start temperature, it was found that an Nb solute drag effect is operational in the early stages of bainite formation as described by the parameter \( b_3 \).

In the intermediate cases where both ferrite and bainite form, the temperature \( T_S^B \) is taken as the onset of bainite formation. Since the critical driving pressure is now dependent on the carbon enrichment of austenite during ferrite formation, this temperature decreases with ferrite fraction from its value for the nominal steel composition.

Subsequent growth of ferrite and bainite is described using the JMAK model and adopting additivity such that the normalised fraction transformed \( F_N^{FB} \) along a cooling path is given by

\[
F_N^{FB} = 1 - \exp \left( -\int_{T_S^F}^{T_S^B} \frac{\beta(T, X_{Nb}, d_f)}{\varphi} dT \right)^n
\]

Here, \( n \) is the JMAK exponent and the subscript \( N \) is introduced to refer to the normalised fraction of ferrite (F) or bainite (B). The rate parameter \( \beta \) can be written as

\[
\beta = \exp \left( \frac{\beta_1 T + \beta_2}{d_f} \right)
\]

where the \( \beta_i \) are taken to be linear functions of the amount of niobium in solution, that is,

\[
\beta_i = \beta_{i1} X_{Nb}^N + \beta_{i2}
\]

A linear dependence is adopted for simplicity as a relatively narrow range of Nb levels is considered in the present work. Further investigations would be necessary to obtain a more accurate description of the phenomenon.

When no bainite forms, the true ferrite fraction after completion of the austenite decomposition \( F_F \) is given by 1 - \( F_{M/A} \) where \( F_{M/A} \) is the true M/A fraction that can be related to the transformation start temperature \( T_S^B \) by\(^{23}\)

\[
F_{M/A} = 0.02 + 0.107 \exp \left( \frac{- (T_S^B - 859)^2}{648} \right) + \frac{0.03}{1 - \exp \left( -0.014 T_S^B + 124 \right)}
\]

Here, the M/A fraction is about 0.05 for a mostly ferritic microstructure leading to a true ferrite fraction of 0.95 which is consistent with the paraequilibrium ferrite fraction at temperatures of 650°C and below where ferrite formation would typically cease. When both ferrite and bainite form, the true ferrite fraction is obtained by dividing the normalised ferrite fraction, \( F_F^0 \), with the paraequilibrium ferrite fraction at the bainite start temperature \( T_S^B \). The paraequilibrium ferrite fraction is obtained with Thermocalc (TCFe7 database). The total bainite fraction at the end of the transformation, \( F_B \), is then deduced from the true fractions of ferrite and M/A as 1 - \( F_F - F_{M/A} \). Further, the relative fraction of upper and lower bainite can be estimated from the total fraction of bainite based on the transformation temperature. The distinction between upper bainite and lower bainite has been clarified in the work of Takayama et al.\(^{49}\) and Reichert et al.\(^{25}\) by detailed microstructure characterisation using electron backscatter diffraction pattern (EBSD) mapping. Each austenite grain transforms to primarily one Bain group variant in the case of upper bainite, while lower bainite is characterised by the occurrence of all three Bain groups in each parent austenite grain. Further, the determination of the Kernel Average Misorientation (KAM) from the EBSD maps indicates increasing KAM values and, thus, increasing dislocation density, with lowering the transformation temperatures. In addition, the M/A fraction study reported by Reichert et al.\(^{23}\) for this steel shows that 12.8% of M/A is present in the upper bainite microstructure with a transformation start temperature of 586°C, whereas only 2% of M/A is present in the lower bainite microstructure with a transformation start temperature of 540°C. Thus, to rationalise the gradual change from upper-to-lower bainite between these two transformation start temperatures, the relative normalised volume fractions of upper and lower bainite \( (F_{UB}, F_{LB}) \) can be deduced from
\[ F_{N}^{LB} = (1 - F_{N}^{B}) = \frac{F_{MA}^{M/A} - 0.02}{0.108} \] (15)

The true fraction of upper bainite \( F_{N}^{UB} \) and lower bainite \( F_{N}^{LB} \) are then evaluated proportionally to the total fraction of bainite \( F_{N}^{B} \). Table 3 lists the parameters used in the austenite decomposition model for ferrite and bainite formation, respectively.

### Simplified temperature model

Based on the data from the weld trial, a simplified temperature model (in °C) is proposed by employing the Rosenthal equation for thick plates,\(^{16}\) that is,

\[ T(t, r) = T_0 + \frac{q}{2\pi k v} \exp \left( \frac{-r^2}{4at} \right) \] (16)

Here \( t \) is the time, \( r \) is the distance from the centre line, \( v \) is the weld speed, \( q \) is the heat input, \( k \) is the thermal conductivity, \( a \) is the thermal diffusivity and \( T_0 \) is the pre-heat temperature. The peak temperature is given by

\[ T_{\text{peak}}(r) = T_0 + \frac{2q/v}{\pi \nu C_v r^2} \] (17)

where \( C_v \) is the volumetric heat capacity for iron and \( \nu \) is Euler’s number. According to equation (16), the cooling time from 800 to 500°C is given by

\[ t_{8-5} = \frac{q/v}{2\pi k} \left( \frac{1}{500 - T_0} - \frac{1}{800 - T_0} \right) \] (18)

To describe the measured time–temperature curves, \( q/v \) is employed as an effective heat input parameter while for simplicity assuming \( k = 41 \text{J m}^{-1} \text{s}^{-1} \text{K}^{-1} \) and \( C_v = 600 \text{J kg}^{-1} \text{K}^{-1} \) as representative values for low-carbon steel. The HAZ is defined with respect to the position of the fusion line where the peak temperature is the melting temperature of 1500°C. The width of the HAZ can then be determined from the locations of the peak temperatures of 900°C for the zone fully austenitised where the proposed microstructure evolution model is applied and the width of the overall HAZ including the intercritical region is determined from the location of the peak temperature of 700°C where the austenite/ferrite two-phase field is reached.

### General structure of integrated model

The integrated model is constructed from the individual sub-models calibrated using the aforementioned experimental studies. The flow chart of the model is presented in Fig. 2. The temperature–time cycles calculated from the simplified temperature model for each position in the HAZ are used as input for the calculation together with the initial precipitate fractions and radii. The evolution of austenite grain size and precipitate sizes and volume fractions are quantified for all positions along the HAZ thereby providing the starting conditions for the austenite decomposition model during cooling. Both the ferrite and bainite start transformation temperature are evaluated simultaneously. If the ferrite start temperature is higher than that for bainite, the ferrite growth model is called together with the model to determine the transition from ferrite to bainite formation. If the transition temperature to bainite is reached before the completion of the austenite decomposition, the bainite growth model is called. On the other hand, if the bainite start formation temperature is above the ferrite start temperature, only the bainite growth model is called. The fractions of ferrite, upper and lower bainite are renormalised together with the fraction of M/A calculated for a given heating cycle.

### Results

#### Validation tests

Figure 3a compares the measured evolution of the austenite grain size during continuous heating with the predictions of the austenite grain growth model. The initial austenite grain size of 5 μm resulting from the austenite formation from the starting hot-rolled microstructure is independent of heating rate. The onset of measurable grain coarsening scales with heating rate from about 1050°C for 10°C s\(^{-1}\) to 1100°C for 100°C s\(^{-1}\), as predicted by the model. Measurements are extended to about 1250°C above which laser ultrasonic signals become unreliable. As austenite grain sizes measured with LU Met are typically consistent with metallographic data within a 20% margin,\(^{44,45}\) measured and predicted grain sizes are in reasonable agreement in particular for the faster heating rate of 100°C s\(^{-1}\). For the HAZ, an even higher heating rate of at least 1000°C s\(^{-1}\) was measured and the model predictions are included for...
this scenario in Fig. 3a, indicating that negligible grain growth takes place during heating up to 1250°C. Figure 3b illustrates the case for the dual-torch scenario employing a heating rate of 100°C s⁻¹. In the first pass, the measured austenite grain size is 20 µm when reaching the peak temperature of 1225°C and increases during a brief hold of 1 second to about 30 µm before the onset of cooling during which no further grain growth occurs. In the second pass, the austenite grain size obtained is also 30 µm even though there are in detail important differences in the grain growth behaviour in both passes. In the second pass, the pinning of Nb(CN) particles is reduced due to partial dissolution in the first pass and more importantly the initial austenite grain size resulting from austenite formation is 15 µm, that is, significantly larger than that observed in the first pass. Here, one has to consider that the microstructure of the as-hot rolled steel and that obtained after the single pass heat treatment are different. As an illustration one may consider the microstructures that result from different single pass scenarios. For sufficiently low peak temperatures (e.g. 1000°C, Fig. 4a), irregular ferrite forms similar to that found in the as-hot rolled steel. For sufficiently high peak temperatures, for example, 1200°C (Fig. 4d), bainite rather than the irregular ferrite of the base metal forms. These different microstructures have also very different populations of M/A constituents in terms of their fraction, size and morphology such that different austenite grain sizes can be expected to result from austenite formation starting from these varieties of ferritic and/or bainitic microstructures. This observation is similar to that discussed in a recent austenite formation study for a low carbon steel. Assuming an initial austenite grain size of 15 µm for the second pass, the predicted austenite grain size evolution is in reasonable agreement with the measurements. The predicted austenite grain size is in both passes 25 µm, which is within the error of measurements, that is, 20%, consistent with the measured 30

2 Flow chart of integrated microstructure model for the HAZ

3 Laser ultrasonic grain size measurements and comparison with austenite grain growth model predictions: a for continuous heating at different rates, b for dual-pass welding scenario with heating rates of 100°C s⁻¹ and peak temperatures of 1225°C with a holding time of 1 second
µm. Rather than tuning the model parameters in order to obtain better agreement between measured and predicted grain sizes, all model parameters for austenite grain growth and precipitate dissolution are kept identical as those reported by Banerjee et al. The only difference in the implementation of Banerjee’s model is the assumption made with respect to the Nb level in solution, which affects the initial fraction of small Nb precipitates. A sensitivity analysis shows, however, that this is of little consequence for the grain size prediction as substantial austenite grain growth can only occur once all small precipitates are dissolved. Further, for the HAZ portions with larger austenite grain sizes, bainite forms and the bainite transformation kinetics is independent of the prior austenite grain size. Thus, the prediction of a 20% smaller austenite grain size has only a negligible effect on the prediction of the resulting transformation products. Thus, the proposed austenite grain size model is applicable to the HAZ, provided that for dual-torch welding the effect of the HAZ microstructure obtained in the first pass on the initial austenite grain size is taken into account. Here, for simplification 5 µm is assumed for a predominantly ferritic microstructure and 15 µm for a predominantly bainitic microstructure.

Examples of the final microstructures resulting from different single-torch scenarios (see Fig. 1a) are shown in Fig. 4 illustrating the role of peak temperature and cooling conditions. For a peak temperature of 1000°C and a \( t_{85} \) of 15 seconds, 90% of irregular ferrite forms with 10% of M/A constituent that is similar to that in the base metal. Optical metallography gives 10% of irregular ferrite, 82% of bainite and 8% of M/A constituent for the sample cycled with a peak temperature of 1200°C (Fig. 4b). On the other hand, 5% of M/A constituents and 95% of bainite are observed in the sample cycled with a peak temperature of 1350°C (Fig. 4c). For higher peak temperatures (1200 and 1350°C) and a \( t_{85} \) of 15 seconds, the quantitative optical metallography results are consistent with primarily bainitic microstructures containing mainly upper bainite forming for lower cooling rates (e.g. for \( t_{85} = 15 \) second, see Figs. 4b and c). Increasing the cooling rate (i.e. decreasing \( t_{85} \) to 7 seconds) and/or peak temperature promotes the formation of fine bainitic ferrite laths characteristic of lower bainite rather than upper bainite, as illustrated for a peak temperature of 1350°C (see Figs. 4c and d).

Figure 5 shows the evolution of the bainite microstructure resulting from a dual-torch scenario where pass 1 has a peak temperature of 1200°C and \( t_{85} = 7 \) second and a pass 2 has a peak temperature of 1200°C and a \( t_{85} \) of 15 seconds. After the first pass, a mixed upper/lower bainite structure forms (see Fig. 5a), which appears after Nital etching as very fine and numerous bainite platelets formed within the prior austenite grain. On the other hand, after the second pass upper bainite forms as coarse bainitic ferrite islands with dispersed carbide precipitates located mainly at the grain boundaries (see Fig. 5b) due to the slower cooling from the peak temperature.

The transformation start temperatures can be taken as a representative transformation temperature for continuous cooling scenarios such as those observed in the HAZ. Figure 6 compares measured and predicted transformation start temperatures for the simplified weld heat treatment simulations. The transformation start temperature is measured with an accuracy of ±5°C by the application of the lever rule on dilatometry data. Reasonably good agreement is obtained provided the initial amount of Nb in solution is assumed to be 60 ppm. The austenite decomposition model is rather sensitive to the amount of Nb in solution. For example, bringing all Nb in solution can, depending on cooling rate and austenite grain size, decrease transformation temperatures by 50–100°C.
compared to the case of having most Nb precipitated.\textsuperscript{22} Having an accurate prediction of the transformation temperature ensures also that the overall fraction of transformation products obtained from the austenite decomposition model is consistent with the microstructure observations shown in Figs. 4 and 5. An initial amount of Nb in solution of 60 ppm in the as-hot rolled steel seems reasonable based on the observed austenite decomposition kinetics and the abovementioned TEM studies of Lu.\textsuperscript{30} A sensitivity analysis conducted by adjusting the initial amount of Nb in solution from 50 to 70 ppm shows that this change modifies the prediction of the transformation start temperature by 10°C.

Application of integrated model

Four cases of welding conditions are considered in the present study: single-torch GMAW with no pre-heat (GMAW-ST), single-torch GMAW with pre-heat (GMAW-ST-P), single-torch SAW with no pre-heat (SAW-ST) and dual-torch GMAW with no pre-heat (GMAW-DT). The three scenarios without pre-heat, that is, GMAW-ST, SAW-ST and GMAW-DT, replicate conditions of the previously conducted weld trials. The dual-torch scenario is a combination of GMAW-ST with a second pass that can effectively be described by starting with a pre-heat. The peak temperatures in both welding passes are assumed to be the same for a given position in the HAZ. Table 4 lists these four cases with the \( t_{85} \) times measured in the welding trials, except for GMAW-ST-P, where this time was obtained using the same effective heat input, ratio of heat input to weld speed, as for GMAW-ST. The Rosenthal fit emphasises to replicate the \( t_{85} \) times by determining the required effective heat input as shown in Table 4 together with the resulting width of the HAZ, which is consistent with the estimates from the metallographic observations made on the weld trial samples. The time–distance temperature profiles in the HAZ for these four cases are illustrated with contour plots in Fig. 7.

Based on these thermal profiles, the integrated model predicts the dissolution of Nb(CN) precipitates, austenite grain growth and the resulting austenite decomposition as a function of time and position in the HAZ. Figure 8 summarises the prediction of three important microstructure parameters, that is, Nb in solution, austenite grain size at the onset of austenite decomposition and the transformation start temperature as a function of distance from the fusion line for the four investigated welding scenarios. For reference, the peak temperatures are shown as well. In all scenarios, Nb is completely in solution near

```
<table>
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<tr>
<th>Case</th>
<th>Effective heat input/kJ. mm(^{-1})</th>
<th>Pre-heat/°C</th>
<th>( t_{85})/s</th>
<th>Position of ( T_{\text{peak}} = 900°C/mm)</th>
<th>Position of ( T_{\text{peak}} = 700°C/mm)</th>
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<td>2.4</td>
<td>3.9</td>
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<td>13.0*</td>
<td>3.1</td>
<td>5.3</td>
</tr>
<tr>
<td>SAW-ST</td>
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<td>27</td>
<td>18.4</td>
<td>4.1</td>
<td>6.5</td>
</tr>
<tr>
<td>GMAW-DT (second pass)</td>
<td>2.02</td>
<td>200**</td>
<td>13.3</td>
<td>3.2</td>
<td>5.3</td>
</tr>
</tbody>
</table>

\*Obtained assuming effective heat input of GMAW-ST.  
\**Effective pre-heat temperature resulting from the first pass.
```
the fusion line within a layer that increases from 0.4 to 0.8 mm with the relative heat input (including pre-heat), that is, the width of this zone is largest for the SAW scenario, which has the highest effective heat input for the considered welding cases. Then, the Nb solute level drops gradually to an intermediate value of 165 ppm that is consistent with the dissolution of the small Nb(CN) precipitates. A further decrease from this intermediate Nb solute level occurs for a sufficiently large distance from the fusion line that coincides with the position where the peak temperatures reaches about 1080°C that is expected as the Nb(CN) dissolution temperature for the present steel composition. The austenite grain size at the fusion line is for all four welding scenarios in the range of 80–90 µm. The magnitude of this grain size is similar to that observed in welding trials by Hamad et al. They, however, measured typically somewhat smaller grain sizes. In single-torch welding most grains were below 70 µm and above 70 µm for dual-torch welding. These apparent discrepancies can be mitigated when considering the predicted rapid decrease of grain size with distance from the fusion line, for example, for the GMAW-ST scenario, the grain size is about 70 µm at a distance of 200 µm from the fusion line and this decrease is less pronounced immediately near the fusion line for higher heat inputs (e.g. for SAW the grain size is still 80 µm at a distance 200 µm away from the fusion line). Further, the present model does not consider minor dissolution and/or coarsening of TiN precipitates which in particular for dual-torch scenarios may lead to a further minor reduction in pinning forces resulting in somewhat larger grain sizes when comparing single- and dual-torch welding. The rapid decrease of austenite grain size with distance from the fusion line is consistent with typical observations made on actual welds. The rate of the decrease in grain sizes mirrors the drop in peak temperatures with distance. In all cases, the position of onset of grain coarsening correlates well with the peak temperature being the Nb(CN) dissolution temperature of 1080°C and can, thus, be taken as the separation of the coarse-grained HAZ (CGHAZ) from the fine-grained HAZ (FGHAZ). The resulting transformation start temperatures reflect the state of austenite in terms of grain size and Nb in solution for each individual welding scenario. In a first approximation transformation, start temperatures correlate well with the amount of Nb in solution, that is, decreasing Nb in solution leads to an increase in the transformation temperature. When comparing different welding scenarios, however, the $t_{8.5}$ times play a significant, if not dominant role. Increasing $t_{8.5}$, that is, decreasing the cooling rate, leads to an increase in transformation start temperatures. GMAW-ST with the lowest $t_{8.5}$ of 6.5 seconds shows the lowest transformation temperatures, GMAW-ST-P and GMAW-DT have both a $t_{8.5}$ of about 13 seconds and thus similar transformation temperatures, whereas SAW-ST with the largest $t_{8.5}$ of about 22 seconds has the highest transformation temperatures.

The transformation start temperatures shown in Fig. 8d provide a critical indicator for the final microstructure that forms as a function of distance in the HAZ.

7 Temperature map resulting from the various simulations conducted: a GMAW-ST, b GMAW-ST-P, c SAW-ST and d GMAW-DT
8 Important temperature and microstructure indicators as a function of distance from the fusion line for the four weld scenarios: a peak temperature, b Nb in solution at onset of austenite decomposition, c austenite grain size at onset of austenite decomposition and d transformation start temperature.

9 HAZ microstructure maps for the simulated welding scenarios: a GMAW-ST, b GMAW-ST-P, c SAW-ST and d GMAW-DT
HAZ microstructure maps obtained with the integrated model are shown for the four welding scenarios in Fig. 9. In the GMAW single-torch scenario with no pre-heat (Fig. 9a), the lowest transformation temperatures are observed and lower bainite forms near the fusion line in the CGHAZ and upper bainite in the remaining portion of the HAZ farther away from the fusion line. For the single-torch GMAW with pre-heat, the HAZ is wider and the $t_{1.5}$ is larger such that upper bainite forms in the CGHAZ and ferrite in the remaining HAZ (Fig. 9b). This observation is similar to that made for the two other cases, that is, upper bainite in the CGHAZ and ferrite otherwise (see Figs. 9c and d). In the SAW scenario, the HAZ is significantly wider (Fig. 9c), whereas the HAZ microstructure and width for the dual-torch GMAW scenario (Fig. 9d) is essentially identical to that obtained for the single-torch GMAW with pre-heat (Fig. 9b) since both cases have the same peak temperatures and $t_{1.5}$ times. When comparing single- and dual-torch GMAW without pre-heat (see Fig. 9a and d), the prediction of lower bainite forming near the fusion line in single-torch welding and upper bainite for dual-torch welding is consistent with the observations of Hamad et al. in weld trials. There are, however, some discrepancies between the calculated and observed M/A fractions for these two welding scenarios. The current model suggests an M/A fraction near the fusion line of 0.02 for single-torch welding but 0.12 for dual-torch welding, whereas Hamad et al. observed an M/A fraction of 0.01–0.02 for both welding scenarios. Further studies are required to clarify the differences in calculated and observed M/A fractions for dual-torch welding.

To further evaluate the model, it is useful to determine the sensitivity of the predictions with respect to a number of fit parameters in particular those that were used to tune the model, that is, the initial level of Nb in solution and the initial austenite grain size resulting when austenite forms from bainite in the second pass during dual-torch welding. Changing the Nb level in solution by ±20 ppm, that is, considering 40 and 80 ppm instead of 60 ppm, has a marginal effect on the calculated microstructure maps. The change in the initial Nb level does primarily affect the transformation temperature in the FGHAZ and the ferrite fraction in this zone away from the fusion line increases with decreasing Nb content in solution when a mixture of ferrite and bainite is present, which here applies only to the GMAW-ST scenario (Fig. 9a). In all other cases, a rather sharp transition from a primarily ferritic to a primarily bainitic microstructure occurs at a position, which is unaffected by the initial amount of Nb in solution. Similarly, changing the initial austenite grain size by ±10 µm, that is, considering 5 and 25 µm instead of 15 µm in the bainitic portion of the HAZ after the first pass in dual-torch welding, has essentially no effect on the resulting microstructure maps as it only applies to the CGHAZ where the austenite decomposition into bainite is independent of austenite grain size.

**Conclusion**

The case study presented here illustrates that the proposed integrated model is a powerful tool to compare HAZ microstructures for different welding scenarios. The model is based on the microstructure engineering concept but employs still a number of empirical relationships, in particular for bainite formation and the M/A fraction. The transition between different transformation products, that is, irregular ferrite, upper and lower bainite, which is of relevance for the HAZ is gradual. The proposed integrated model provides a modelling strategy that could be adopted for different steel chemistries as well which, however, would require to re-adjust some of the model parameters.

EBSD studies and evaluation of mechanical properties indicate that a unified characterisation of these transformation products in terms of an effective grain size (e.g. line length of high angle boundaries per unit area) and effective dislocation density can be considered as a function of transformation temperature. This approach may provide an alternative avenue to formulate an austenite decomposition model with a reduced number of tuning parameters that, moreover, would provide a natural link to mechanical properties, which depend primarily on grain size and dislocation density. Further, a more rigorous model approach is required for the M/A constituents. Here, it is required to also include their size and shape as well as the actual fraction of retained austenite into an improved model.

Connecting the predicted, graded microstructures with a structure–property model will then permit to evaluate the integrity of the HAZ as a function of welding procedure. A direct linkage to the welding parameters (i.e. heat input and weld speed) will, however, require to replace the simplified temperature model with a heat transfer model that accounts for the complexity of the weld geometry as well as the latent heat of both solidification of the weld metal and the austenite decomposition.

The current model is limited to the portion of the HAZ that is completely austenitized. Thus, it will be critical to extend the model to intercritical heat treatments with a primary focus on an intercritical cycling of the CGHAZ in a subsequent weld pass. There is increasing evidence that martensite forming from intercritical austenite in the CGHAZ can lead to a significant deterioration of HAZ properties.

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