Physically based modeling, characterization and design of an induction hardening process for a new slurry pipeline steel

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HIGHLIGHTS
• An empirical model has been developed for prediction of phase fraction of a new steel after different cooling rates.
• The model has been coupled with the simulated thermal history to predict the microstructure of each pipe thickness layers.
• The effect of heating rate on the austenitization critical temperatures has been considered to the model.
• The simulation shows slightly faster austenitization kinetics probably due to neglecting the effect of grain size.
• The model can be used to optimize the hardness or microstructure gradient based on the industrial considerations.
• The designed thermal cycles produced a gradient of material properties giving ~3 mm hardening depth in a 10 mm thick pipe.

ABSTRACT
Numerical and Gleeble experimental data are combined to predict potential microstructure and hardness profiles through the wall thickness of an induction hardened slurry transportation pipe made of a recently developed 0.4 wt% C, Nb-microalloyed steel. The calculated thermal history of various positions through the wall thickness of an industrial pipe (400 mm diameter, 10 mm thick) were combined with a model that predicts the phase transformations, microstructures and final hardness values on heating and cooling along arbitrary thermal cycles. The accuracy of the hardness profile predictions was verified by experimental data, i.e. reproducing the thermal cycles on a Gleeble thermomechanical simulator. The results indicated that the approach should be a feasible way to optimize induction heating and cooling parameters to obtain desired hardness profiles through the wall thickness.

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1. Introduction

In the long-distance transportation of cement and mining slurry, steel pipeline transportation dominates because of its high efficiency, low cost and environmental advantages. The severe erosive wear conditions in such pipelines set high demands with respect to the steel properties: wear resistance, toughness and hardness. It has been pointed out by Gupta [1] that, depending on the pipeline operating pressure, one way to reduce the steel wear rate is to use a pipe with a gradient of material properties such that there is extremely good wear resistance on the inner surface, and adequate toughness and hardness on the outer surface, able to withstand the impacts and external loads that happen during conveying, installation and maintenance. Kyriakides and Corona [2] mentioned that the properties of the pipe material can be reached by the induction hardening and thermomechanical schedule during the production process. Hence, to be able to manufacture a pipe with these properties the following seven steps would be a cost-effective production route: i) continuous casting, ii) controlled hot strip rolling and coiling, iii) slitting into skelp, iv) cold forming and v) high-frequency induction welding into pipe followed by vi) internal induction heat treatment to give a hard-inner pipe surface together with a tough pipe body.

The mechanical properties of the pipe will be controlled by the choice of chemical composition, the pipe production process, and the final heat treatment. One suitable steel chemistry for the production process under consideration has been developed by the authors [3] considering the main parameters involved in slurry erosion and thermomechanical processing as reported in references [4-6]. A key factor in the design of the gradient of material properties through the pipe body, i.e. microstructural and hardness profiles, and full hardening depth, is the understanding of the material behavior during the induction hardening, especially the phase transformations occurring at the different heating and cooling rates. Scanning induction hardening is a process involving various coupled physical phenomena, i.e. electromagnetic effects, heat transfer and phase transformations. These need to be simulated numerically due to their nonlinearity. Many studies have been carried out using numerical methods to understand and optimize the induction hardening of the pipes and tubes. For example, Lu et al. [7] used a numerical model to design and optimize the induction heating process of a large diameter pipe for production of elbows. Han et al. [8] also simulated a medium-frequency induction heat treatment of a welded pipe in order to optimize the welding quality and energy consumption on the basis of the predicted temperature. They mainly studied the effect of different parameters on the temperature of inner and outer walls, but there are only a few works dealing with the effect of induction hardening parameters on microstructure and material properties. Material properties obtained by induction hardening can vary through the thickness from the hardened surface toward the bulk as the changes in the peak temperature during the induction hardening cause a wide variety of phase transformation conditions. Peak temperature during the induction hardening cause a wide variety of phase transformation conditions. Recently, Montalvo-Urquizo et al. [9] numerically investigated the induction hardening and phase transformations occurring through the thickness of a gearing component made of 42CrMo4. However, they reported that due to the high cooling rate in their designed system, most of the austenite was transformed to martensite and there was no gradient of material properties through the hardened depth. In contrast, a gradient of properties is required in the case of pipe and tubes for slurry transportation, hence a reliable computational method is required to be able to design a suitable induction hardening process.

The aim of this work is to demonstrate a methodology and develop a model that can be used to design the induction heating parameters needed to obtain a desired gradient of material properties through the thickness of an internally hardened pipe. As an example a 10 mm-thick pipe with an inner diameter of 400 mm and the chemical composition described in [3] is used. An empirically-based phase transformation model is combined with the calculated thermal history across the pipe body to predict the profile of prior austenite grain size, as-quenched microstructure and hardness through the pipe wall thickness.

2. Methodology

2.1. Experimental

2.1.1. Material and heat treatments

A Gleeble 3800 thermomechanical simulator has been employed in order to physically simulate different thermal cycles, i.e. different heating and cooling rates and also to provide the dilatometric phase transformation data required for developing a continuous cooling transformation (CCT) model based on the approach of Pohjonen [10]. The sample dimensions and the applied thermal cycles used to determine the CCT diagrams are presented in Fig. 1. As shown, to observe the microstructure and to measure the hardness, Gleeble samples were cut
in half perpendicular to their axis at the position of thermocouple used for monitoring and controlling the temperature during the thermal cycles. The starting material was laboratory hot rolled and directly water quenched to the quench-stop temperature of 420 °C following by slow furnace cooling to produce an essentially isothermal lower bainite starting microstructure. The composition of the steel studied together with the initial bulk mean hardness value and mean prior austenite grain size (PAGS) on the RD-ND cross-section are given in Table 1. More details about the material composition, as-rolled properties and initial microstructure can be found elsewhere, in references [3, 6].

2.2.1. Phase transformation modeling

Fig. 2 shows the carbon isopleth phase diagram for the studied steel (wt%) has been taken as the carbon content at the eutectoid point. Between the specific volumes of different phases or the possible carbon enrichment of them [14]. This gives:

\[ W_{\alpha, \text{max}} = \frac{C_{\text{max}}(T) - C_0}{\frac{C_{\text{max}}(T) - C_{\alpha}(T)}{K^3}} \]  

where \( C_0 \) is the initial carbon concentration of the steel (wt%), \( C_{\text{max}}(T) \) is the equilibrium carbon concentration of austenite at temperature \( T \), and \( C_{\alpha}(T) \) is the equilibrium carbon concentration of ferrite. The maximum volume fraction of ferrite between \( A_{\alpha 1} \) and \( A_{\alpha 3} \) is then obtained from Eq. (2):

\[ \chi_{\alpha, \text{max}} = \left( 1 + \frac{\rho_\alpha}{\rho_\gamma W_{\alpha, \text{max}}} \right)^{-1} \]  

where \( \rho_\alpha \) is the density of ferrite and \( \rho_\gamma \) is the density of austenite at the given temperature. Below \( A_{\alpha 1} \), the maximum amount of ferrite that can form is taken from the eutectoid composition. Here the value of 0.70 (wt %) has been taken as the carbon content at the eutectoid point. Between \( A_{\alpha 1} \) and \( A_{\alpha 3} \) temperatures the maximum fraction of austenite is \( 1 - \chi_{\alpha, \text{max}} \) and above the \( A_{\alpha 3} \) temperature the material can become fully austenitic. During cooling, in the temperature range where bainite and martensite can form, austenite can fully transform into bainite and martensite, i.e. there is no limiting maximum fraction for bainite or martensite based on the equilibrium conditions.

Calculation of the initial delay of the transformation, i.e. the incubation time, was not included for polygonal ferrite formation, since within the experimental accuracy, it seemed that the incubation time was zero, and that this transformation could be modelled by simply fitting the temperature dependent rate parameter \( k(T) \) and the \( n \) of the equation developed by Avrami [15], as described below. The incubation time was included for austenite to bainite transformation by applying the rule of Scheil, in the way described in ref. [10]. The transformation is assumed to start (1% transformed fraction) when \( \sum 1/x = 1 \), where \( x = K(1 + 273.15/T) \) is calculated when \( T < A \).

Leblond et al. [16] proposed that, Once the transformation has started, the rate of formation of austenite on heating and polygonal

\[ (A-T)^m \exp \left[ \frac{Q}{K(1 + 273.15)} \right] \]  

is calculated when \( T < A \).
ferrite and bainite on cooling \( \frac{d\chi}{dt} \) was calculated with the differential form of the Avrami equation as expressed in Eq. (3).

\[
\frac{d\chi}{dt} = (\chi_{\text{max}} - \chi) \left[ \ln \left( \frac{\chi_{\text{max}}}{\chi_{\text{max}} - \chi} \right) \right]^{\frac{n-1}{n}} \frac{1}{k(T)} \tag{3}
\]

where \( k(T) \) is the temperature-dependent rate parameter, and \( n \) is the Avrami exponent, which is assumed to be a constant in the current model. The following functional forms were used in the fitting for \( k(T) \).

\[
k_\alpha(T) = \exp \left( -a_\alpha(T-b_\alpha)^2 - c_\alpha \right) \text{ for polygonal ferrite, } (\gamma \rightarrow \alpha) \tag{4}
\]

\[
k_b(T) = \frac{C}{\gamma} \exp \left( -a_b(T-b_b)^2 - c_b \right) \text{ for bainite, } (\gamma \rightarrow \alpha) \tag{5}
\]

\[
k_\gamma(T) = A_\gamma(T-Ae1)^m_\gamma \exp \left( -\frac{Q_\gamma}{RT+273.15} \right) \text{ for austenite, } (\alpha \rightarrow \gamma) \tag{6}
\]

where \( a, b, \) and \( c \) are fitting constants for the formation of polygonal ferrite (subscript \( \alpha \)) and bainite (subscript \( b \)), and \( A_\gamma, m_\gamma, \) and \( Q_\gamma \) are fitting constants for austenite formation. For transformation to bainite, the carbon enrichment of austenite due to prior transformation to polygonal ferrite is taken into account by calculating the factor \( C/C_\gamma \) as function of ferrite fraction \( \chi_\alpha \) [14]:

\[
C_\gamma = \frac{C_\gamma - \chi_\gamma}{1 - \chi_\alpha} \tag{7}
\]

The parameters \( k(T) \) and \( n \) were numerically fitted to the experimentally determined transformation data by minimizing the difference between the computational and experimental results using Matlab function \textit{fminsearch} [17].

Last but not the least, the martensite transformation was modelled by applying an exponential correlation between fraction of martensite \( \chi_m \) and temperature developed by Koistinen-Marburger [18] as given in Eq. (8):

\[
\chi_m = \left[ 1 - \exp(-\eta(M_3-T))/\left(1 - \chi_\alpha - \chi_b \right) \right] \tag{8}
\]

For the purpose of fitting the model, the transformed fraction is determined from the measured dilatometer length change both for austenitization during heating and for phase transformation during cooling. The amounts of polygonal ferrite, bainite and martensite were compared to experimental micrographs in order to verify that the model correctly reproduces the volume fractions of polygonal ferrite, bainite and martensite.

### 2.2.2. Induction hardening simulation

This section briefly explains the 2D geometrical setup for simulating the induction hardening of a thick-walled pipe that can provide a reliable theoretical temperature distribution close to that which might be expected through a real pipe wall thickness. Fig. 3 shows a layout of the setup of the coil, quench and pipe assembly. The pipe is considered to move at a constant velocity of 10 mm/s past a stationary induction heating inductor and quenching system.

#### Table 2

The temperature-dependent material properties considered for the simulation.

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Resistivity ( \rho ) ( \times 10^{-6} ) ( \Omega \cdot m )</th>
<th>Specific heat ( C_p ) ( \text{J/kg} \cdot \text{°C} )</th>
<th>Thermal conductivity ( K ) ( \text{W/m} \cdot \text{°C} )</th>
<th>Density ( \rho ) ( \text{kg/m}^3 )</th>
<th>Emissivity ( C_s ) ( \text{J/m}^2 \cdot \text{s} \cdot \text{K}^4 )</th>
<th>Quenching shower heat transfer coefficient ( \text{W/m}^2 \cdot \text{°C} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>0.15</td>
<td>481.06</td>
<td>51.70</td>
<td>7858</td>
<td>1</td>
<td>1100</td>
</tr>
<tr>
<td>25</td>
<td>0.16</td>
<td>483.09</td>
<td>51.50</td>
<td>7850</td>
<td>0.9</td>
<td>3100</td>
</tr>
<tr>
<td>100</td>
<td>0.22</td>
<td>494.04</td>
<td>50.60</td>
<td>7832</td>
<td>0.8</td>
<td>8000</td>
</tr>
<tr>
<td>200</td>
<td>0.29</td>
<td>522.93</td>
<td>48.10</td>
<td>7801</td>
<td>0.8</td>
<td>11,300</td>
</tr>
<tr>
<td>300</td>
<td>0.39</td>
<td>561.03</td>
<td>45.60</td>
<td>7766</td>
<td>0.8</td>
<td>13,000</td>
</tr>
<tr>
<td>400</td>
<td>0.49</td>
<td>599.13</td>
<td>41.90</td>
<td>7730</td>
<td>0.8</td>
<td>17,200</td>
</tr>
<tr>
<td>500</td>
<td>0.62</td>
<td>669.89</td>
<td>38.10</td>
<td>7692</td>
<td>0.8</td>
<td>18,400</td>
</tr>
<tr>
<td>600</td>
<td>0.77</td>
<td>720.13</td>
<td>33.50</td>
<td>7652</td>
<td>0.8</td>
<td>17,100</td>
</tr>
<tr>
<td>700</td>
<td>0.93</td>
<td>808.89</td>
<td>28.00</td>
<td>7613</td>
<td>0.8</td>
<td>13,800</td>
</tr>
<tr>
<td>800</td>
<td>1.11</td>
<td>561.03</td>
<td>24.80</td>
<td>7635</td>
<td>0.8</td>
<td>10,100</td>
</tr>
<tr>
<td>900</td>
<td>1.15</td>
<td>586.15</td>
<td>25.70</td>
<td>7590</td>
<td>0.8</td>
<td>5200</td>
</tr>
<tr>
<td>1000</td>
<td>1.18</td>
<td>628.02</td>
<td>26.90</td>
<td>7538</td>
<td>0.8</td>
<td>3000</td>
</tr>
</tbody>
</table>
In ferromagnetic materials including the studied steel, several physical properties have a non-linear dependency on temperature. For example, magnetic permeability drops to 1 above the Curie temperature while the specific heat capacity increases with increasing temperature and peaks at around the Curie temperature [19]. The convection coefficient during quenching is also a non-linear function of the surface temperature of the pipe. Some of the main temperature-dependent properties which have been used in the computations are listed in Table 2 and they were shown with a verification experiment to be reliable. Details are presented in Appendices A and B, please see pages 26–27.

Fig. 4. Different type of micrograph of an exactly same area using a) FESEM-EBSD (image quality), b) FESEM-EBSD (inverse pole figures), c) FESEM (secondary electron image 1), d) FESEM (secondary electron image 2), and e) light optical and laser scanning confocal along with f) surface height profiles along the lines identified in (e).
In order to simulate the whole induction hardening process, a coupled electromagnetic - thermal calculation model has been applied using the commercially available finite element software Flux 2D. The electromagnetic model solves a non-linearly coupled system of Maxwell’s equations where the magnetic properties have a non-linear dependency on both magnetic field strength and temperature, and the electric resistivity is dependent on temperature. The resulting power distribution is used to calculate the temperature distribution in the pipe by solving the governing heat transfer nonlinear differential Fourier equations considering convection and radiation during both heating and cooling. It should also be noted that the pipe and coil have a fixed fine mesh during the entire computation with a combination of rectangular and triangular elements while the air domain was re-meshed for every time-step. In total, when the coil was inside the pipe, there was around 80 K mesh elements.

2.2.3. Calculation of As-quenched hardness

An estimation of the hardness value after a continuous cooling can be made using the available models [10,20,21] depending on steel composition, mixture of steel microconstituents, cooling rate and heating conditions. In this work, two regression-type models and one best-fit model are considered and evaluated against the experimental measurements. The first model, which was developed at the Creusot Laboratory by Maynier et al. [21], is given by the following equations:

\[
HV_m = 127 + 949C + 27Si + 11Mn + 16Cr + 8Ni + 21 \log_{10}v_R
\]  
(9)

\[
HV_b = -323 + 185C + 330Si + 153Mn + 144Cr + 191Mo + 65N + \log_{10}v_R(89 + 53C - 55Si - 22Mn - 20Cr - 33Mo - 10Ni)
\]  
(10)

\[
HV_{f-p} = 42 + 223C + 53Si + 30Mn + 7Cr + 19Mo + 12.6Ni + \log_{10}v_R(10 - 19Si + 8Cr + 4Ni + 130V)
\]  
(11)

where \(HV_m\), \(HV_b\) and \(HV_{f-p}\) are, respectively, the Vickers hardness of martensite, bainite and ferrite-pearlite mixtures, \(v_R\) is the cooling rate from austenitized state expressed in °C/h. The hardness of the steel is then given by a linear rule of mixtures:

\[
HV = f_{f-p}(HV_{f-p}) + f_b(HV_b) + f_m(HV_m)
\]  
(12)

where \(f_{f-p}\), \(f_b\), \(f_m\) are the fractions of the constituents ferrite-pearlite, bainite and martensite, respectively. As observed, Maynier et al., the model does not take austenitizing temperature and heating time into account, therefore the effects related to austenite grain size and homogeneity of the austenite are neglected.

The second model was developed by Trzaska [20] considering again a main role for composition and cooling rate as expressed by the following equations:

\[
HV = -86 + 492C + 92Mn + 69Cr + 25Ni + 102Mo + 267V + 0.0647x_{Fe} - 38F - 70P - 32B + 72M
\]  
(13)

where \(T_a\) is austenitizing temperature; \(v_R\) is the cooling rate expressed in °C/h; \(F, P, B, M\) and \(M\) are step functions of the following forms:

\[
F = \begin{cases} 1 & \text{if } f_{f-p} \geq 0.5 \\ 0 & \text{if } f_{f-p} < 0.5 \end{cases} \quad P = \begin{cases} 1 & \text{if } f_b \geq 0.5 \\ 0 & \text{if } f_b < 0.5 \end{cases} \quad B = \begin{cases} 1 & \text{if } f_m \geq 0.5 \\ 0 & \text{if } f_m < 0.5 \end{cases} \quad M = \begin{cases} 1 & \text{if } f_{f-p} \geq 0.5 \\ 0 & \text{if } f_{f-p} < 0.5 \end{cases}
\]  
(14)

\(f_f\) is the fraction of ferrite and \(f_p\) is the fraction of pearlite, which, in contrast to the model of Maynier et al., are considered separately. The model developed by Trzaska involves austenitizing temperature \(T_a\) as a parameter. Still, as the time of austenitizing is omitted, the model does not take the effects connected to austenite grain size entirely into account.

As the last model, the recently developed (best-fit) model for a 0.2C bearing steel by Pohjonen et al. [10], as expressed in Eq. (15), has been examined against the experimental data. To include the contribution of ferrite, the model has been extended at this research using a regression-type equation for the ferrite in the same manner as Eq. (12).

\[
HV = f_b(C + A \log_{10}v_R) + f_mHV_m
\]  
(15)

In Eq. (15), \(A\) is the the material dependent coefficient defined as \(A = 89 + 53C - 55Si - 22Mn - 10Ni - 20Cr - 33Mo - C\) and \(HV_m\) are fitting parameters. \(f_m\) and \(f_p\) are the fractions of martensite and bainite, respectively.

3. Results and discussion

3.1. Evaluation of microstructure

The continuously cooled microstructures of the Gleeble samples comprised polygonal ferrite, various bainitic morphologies and martensite, the fractions of which depended on the cooling rate. In order to characterize the microstructure and to overcome the challenges of microconstituent identification, various imaging methods (LSCM,
FESEM, and EBSD) have been employed. An example of microstructure identification using different images taken from exactly the same area is presented in Fig. 4. Fig. 4(a) shows the image quality (IQ) which is one of the common maps constructed from the EBSD data using the quality of diffraction patterns. The contrast observed in the IQ can arise from microstructure, surface topology, phase, composition and/or other features of the microstructure like residual plastic deformation left from the sample preparation [22]. By comparing the acquired IQ image with the secondary electron In-Lens (SE1) image (Fig. 4(c)) and secondary electron (SE2) main image (Fig. 4(d)), it could be assumed that, here, IQ mainly depends on the quality (sharpness) of the diffraction patterns, which depends on the strains in the lattice, i.e. dislocation density. Dislocation density increases as the transformation temperature drops through the sequence polygonal ferrite to bainite to martensite, indicating that IQ provides useful visualization of microstructural information. Therefore, IQ has been used for the microstructure characterization as identification of microconstituents by means of FESEM-SEI alone was not always possible and also EBSD is incapable of distinguishing ferrite, bainite, and martensite as they are all essentially body-centered cubic with indistinguishable diffraction patterns. Moreover, the laser

![Graph](image1)

**Fig. 6.** Comparison of simulated vs. experimental amount of austenite formed as function of temperature for different heating rates. The experimental results were obtained by calculation from dilatometry data.

![Diagram](image2)

**Fig. 7.** The calculated continuous cooling transformation (CCT) diagram together with the experimental, dilatometer, results and microstructure for each experimental cooling rate. a) 1 °C/s, b) 5 °C/s, c) 10 °C/s, d) 20 °C/s, e) 40 °C/s, and f) 60 °C/s.
scanning confocal microscopy image has been used for the surface roughness measurement (Fig. 4(f)) to check the depth of each phase after etching. According to all mentioned, following will discuss the rationale for identification of observed various microconstituents which are labeled in the micrographs:

3.2. Microconstituents identification

3.2.1. (Quasi) polygonal ferrite (PF)

The formation (Q)PF which is the first phase to form during the continuous cooling is accompanied by the rejection of carbon to the austenite. Hence, it would be the softest phase with a lower resistance to etching in comparison to phases containing more carbon. Thus, it appears as flat and deeply etched grains in the microstructure, see Fig. 4(d). Moreover, it can also be seen that the ferrite has a lower hardness than the surrounding microconstituents from the width of the scratch lines that can be seen crossing some of the PF grains. Etching depths are shown by the surface height profiles shown in Fig. 4(e) and Fig. 4(f), Line 1 is between (lath-like) bainite and PF, line 2 is between (granular) bainite and PF, and line 3 is between (lath-like) bainite and martensite. It is clear that the most etched phase is PF, then bainite followed by martensite which has the highest surface height. According to this rationale, all of the identified PF in the selected region have been labeled in the SE1 image Fig. 4(c). Apart from the above points, PF is...
easily identified from the fact that it appears very dark in the SE1 image and light grey in the IQ image. Also, no internal features or precipitates are observed inside PF grains.

3.2.2. Bainite (B)
As the temperature drops further during continuous cooling, the condition becomes more favorable for the formation of other components. For the studied composition, during slow cooling, bainite forms after PF. The different types of bainitic microconstituents found in the present steel have been described elsewhere by the authors [6]. However, in this paper, all kinds of bainitic morphologies have been considered under the generic term of bainite (B). Bainite has been distinguished by its internal structure in the IQ and SE1 images and also by the fact that it is more strongly etched compared with surrounding martensite but less etched compared to PF in SE2 images. In IQ images, bainite appears with light contrast with internal substructure, see Fig. 4(a) and (d).

3.2.3. Martensite (M)
In the IQ image high values (lighter area) are associated with little lattice distortion, i.e., low dislocation densities, while low IQ values (darker area) are considered to be the result of intense lattice distortion. Accordingly, the black area in the IQ image indicates the presence of martensitic. However, as can be seen in Fig. 4(a), there are a few small regions with a full-black color which are near the PF grains (for instance the grains pointed by red arrows in Fig. 4). These do not look like the main martensitic areas, which display various shades of dark grey, and

![Graph showing comparison of simulated vs. experimental volume fractions of ferrite, bainite and martensite at different cooling rates.](image-url)

**Fig. 9.** Comparison of simulated vs. experimental volume fractions of ferrite, bainite and martensite at the different cooling rates.

![Graph showing simulated vs. experimental results for ferrite, bainite, and martensite.](image-url)

**Fig. 10.** An example of the output of the graphical user interface, which can be used to experiment with different cooling paths in order to find an appropriate path resulting in the desired amounts of ferrite, bainite and martensite.
they have been identified as high carbon martensite formed from austenite enriched in carbon rejected by the surrounding phases.

3.3. Phase transformation model results

The thermal and metallurgical calculations were coupled by fitting the phase transformation model to the transformed austenite fraction vs temperature data obtained from the experimental dilatometry measurements by using the Nelder-Mead minimization algorithm within Matlab, applying the `fminsearch` function. The result of the fitting is shown in Fig. 5.

Moreover, to check the reliability of the model regarding the effect of heating rate on the austenitization critical temperatures, the austenitization model was fitted to experimental data obtained from dilatometry measurements during heating. Fig. 6 shows that there is good agreement between the experimental results and the model predictions. However, it should be noted that despite the fact that austenite decomposition kinetics are known to be affected by the austenite grain size [23–25], here only the effect of heating rate on the austenite decomposition temperature has been considered. The model could then be used to generate the CCT diagrams like that given in Fig. 7 for example.

Fig. 8(a–f) shows IQ images of the microstructures obtained after continuous cooling at different constant cooling rates from 1 °C/s

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Fig. 11. The temperature profile through the pipe wall thickness at sections A and B (defined in Fig. 3).

Fig. 12. Induction heating rate vs temperature and through-thickness position a) at the beginning of the process (section A) and b) at section B as representative of the rest of process (steady state condition).

Fig. 13. The peak temperature for the different thicknesses after heating and at the beginning of cooling.
to 60 °C/s. Microconstituents are labeled in the micrographs on the basis of the rationale given in Section 3.2. It can be seen that the sample cooled at 1 °C/s mainly consists of large PF/QPF grains along with blocky bainite islands (Fig. 8(a)). Increasing the cooling rate to 5 °C/s, causes an obvious decrease in the size and amount of PF and the appearance of a significant amount of martensite. On a further increase of the cooling rate to 10 °C/s, the microstructure is dominated by bainite and martensite, with just a few PF grains. Martensite was the predominant microconstituent in the microstructures acquired at the cooling rate of 20, 40, and 60 °C/s. Fig. 9 shows the fractions of all phases as determined using the point counting method on 5 fields of view. The figure also shows the fractions predicted by the simulation model. It can be seen that the agreement is good.

While the simulation model has been calibrated using linear heating and cooling rates, it is nevertheless possible to use the model to predict the final microstructure and hardness for any arbitrary heating and cooling cycle. This is done with the aid of a graphical user interface (GUI), an example of which is presented in Fig. 10. Using this GUI, any continuous cooling path can be easily drawn by the user and then the model estimates the phase transformation products and related hardness within 1 s. Similarly, a tool for calculating austenitization for any thermal cycle was created. The fitted parameters for the both on-heating and on-cooling phase transformation model have been listed in Appendix B.

3.4. Induction hardening simulation results

3.4.1. On-heating temperature profile

The simulated temperature profile of every millimeter thickness of the pipe wall as a function of time is presented in Fig. 11 for the two positions A and B shown in Fig. 3. The positions are such that when section B meets the beginning the coil, section A is crossing the end of the coil. The aim was to reach a peak temperature of about 950 °C in 10 s at the inner surface with a gradual decrease in peak temperature toward the outer surface. The through-thickness temperature profile of an induced heated pipe is affected by the temperature dependent physical properties of the pipe material and the induction heating parameters i.e. penetration depth and air gap. As Fig. 12(a) shows, in the very beginning of the heating process (section A) up to the Curie temperature, i.e. 760 °C, the heating rates near the inner surface are significantly higher than those further from the surface. This is due to the high relative magnetic permeability and low penetration depth. At this stage the inner surface is heated directly by the induced current while regions further inside the pipe are heated by thermal conduction from the inner surface. Above the Curie temperature, the relative permeability drops to unity which increases the penetration depth and leads to a moderate heating rate [19]. Then, by continuing the process, as clearly seen in Fig. 12(b), the heating rate in section B is not as high as in the beginning the process and it converges also to about 50 °C/s in the important temperature window i.e. between 400 °C to the peak temperature. Section B is representative of the thermal cycles seen by the hardened pipe in general, i.e. all cross-sections along the pipe length.

When the induction heating begins, eddy currents are most intense just below the surface due to “skin effect”. Therefore, beside the highest heating rate, the highest peak temperature will also be experienced at or just below the inner surface. In the other words, both the induced joule heating and temperature are much greater on the inner surface than on the outer surface, which leads to the clear initial temperature difference and heating rate near the inner and outer pipe surfaces. Later, thermal conductivity causes the heating rate to converge to the almost same level, here around 50 °C/s, through the wall thickness. Recently, the present authors [26] have shown that austenitization at a heating rate of 50 °C/s produces the finest prior austenite grain size in the steel under consideration.

The peak temperature of every millimeter of pipe thickness obtained from the simulations of the designed heating process is given in Fig. 13.

It is also well known that the heating rate influences the phase transformation temperature i.e. austenite start temperature (Ac1) and finish formation temperature (Ac3) [27,28]. Fig. 14 shows the effect of heating rate on the change of austenite formation temperatures obtained from the dilatometry results for the different constant heating rates. It can be seen that both Ac1 and Ac3 temperature increase with increasing the heating rate and that the finish temperature increases faster than the start temperature. Comparing Figs. 13 and 14, it can be seen that almost 30% of wall thickness should be fully austenitized i.e. see temperatures above the austenitization finish temperature (red line), 25% is predicted to be heated into the two-phase region (between the blue and red lines) and 45% should remain below the lower critical temperature (the blue line), i.e. remain as simply tempered as-rolled microstructure. In this way, a smooth gradient of material properties is expected with hard inner layers and a softer, tougher outer body.

3.4.2. On-cooling temperature profile

As shown in Fig. 11, there are three steps during the cooling, i) natural cooling between the end of heating and start of water quenching (distance between coil and quenching zone) during which heat conduction leads to an increase in temperature in the outer layers and a decrease in the temperature of the inner layers, ii) forced water cooling that makes the inner layers of the pipe experience earlier and

![Fig. 14. The effect of heating rate on the transition of austenite start and finish formation temperature.](image)

![Fig. 15. Measured and predicted hardness values of the Gleeble samples vs. cooling rate.](image)
faster cooling than deeper layers, and iii) again natural cooling at the point where the pipe section leaves the cooling zone.

3.5. Hardness prediction

It is obvious that different cooling rate lead to different microstructures and thereby different final hardness levels. As described earlier, the final hardness can be estimated by considering the material chemistry, cooling path and consequent final microstructure while neglecting the effect of austenite grain size. Fig. 15 shows a comparison of the measured hardness values of the samples used to determine the CCT diagram with the predictions of the models of Maynner et al. (Eqs. (9)–(12)), Trzaska (Eqs. (13)–(14)) and Pohjonen et al. (Eq. (15)). Although for the cooling rates 5 and 20 °C/s, Trzaska’s model shows slightly better results than the other models, in some cases it deviates significantly from the experimental measurements. The overall root means square error (RMSE) shows that, for the present steel and conditions, the models of Maynner et al. and Pohjonen et al. are the best, and equally acceptable for practical use. The RMSE values were 17.2 HV and 17.8 HV for Mayner’s and Pohjonen’s models, respectively, whereas for Trzaska’s model it was 46.4 HV. Hence, Mayner’s and Pohjonen’s model were coupled to the electromagnetic, thermal and phase transformation models and employed for the calculations of hardness through the pipe thickness as shown in the next section. It should be remarked

<table>
<thead>
<tr>
<th>Constituent (%)</th>
<th>Inner surface</th>
<th>+1 mm</th>
<th>+2 mm</th>
<th>+3 mm</th>
<th>+4 mm</th>
<th>+5 mm</th>
<th>+6 mm</th>
<th>+7 mm</th>
<th>&gt;7 mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ferrite</td>
<td>0.00</td>
<td>0.00</td>
<td>0.00</td>
<td>0.00</td>
<td>0.05</td>
<td>0.30</td>
<td>0.50</td>
<td>0.50</td>
<td>0.00</td>
</tr>
<tr>
<td>Quenched Bainite</td>
<td>0.10</td>
<td>0.10</td>
<td>0.40</td>
<td>0.80</td>
<td>5.10</td>
<td>10.3</td>
<td>0.50</td>
<td>0.50</td>
<td>0.00</td>
</tr>
<tr>
<td>Martensite</td>
<td>99.0</td>
<td>98.0</td>
<td>97.6</td>
<td>97.2</td>
<td>53.0</td>
<td>30.8</td>
<td>0.10</td>
<td>0.00</td>
<td>0.00</td>
</tr>
<tr>
<td>Initial Bainite</td>
<td>0.00</td>
<td>0.00</td>
<td>0.00</td>
<td>0.00</td>
<td>39.9</td>
<td>58.1</td>
<td>98.7</td>
<td>99.0</td>
<td>100</td>
</tr>
<tr>
<td>Retained Austenite</td>
<td>0.90</td>
<td>1.90</td>
<td>2.00</td>
<td>2.00</td>
<td>1.95</td>
<td>0.50</td>
<td>0.20</td>
<td>0.00</td>
<td>0.00</td>
</tr>
</tbody>
</table>

Table 3: Microstructures predicted through the pipe wall thickness after induction hardening.
that in adopting the model of Pohjonen, the fitting parameters $HV_m = 645.7$ HV and $C = 12.4$ have been employed.

### 3.6. Prediction of through-thickness microstructure and hardness profiles

In order to verify the ability of the model to predict the final microstructure for each layer based on its thermal history, all of the calculated thermal cycles for section B in Fig. 11 were applied to the studied material using the Gleeble simulator. For each layer, dilatometer traces were recorded and the hardness values of the simulated microstructures were measured.

Fig. 16 shows the dilatometric heating and cooling curves as a function of temperature for the studied samples. Three types of behavior can be identified: i) samples which transformed fully, or almost fully to austenite during heating (inner, inner + 1 mm, and inner + 2 mm), ii) samples which were partially austenitized (inner + 3 mm, inner + 4 mm, and inner + 5 mm), and ii) samples seeing no transformation to austenite on heating (inner + 6 mm - outer). Fig. 16(d–f) shows examples of typical microstructures for each group. It can be clearly seen that if austenitization is incomplete, for example 5 mm below the inner surface, some original bainite is retained after heating and cooling. In such layers, there are complex microstructures consisting of as-quenched martensite and as-quenched bainite along with the original and untransformed but tempered bainite (Fig. 16(e)). This leads to a gradual transition of microhardness from the 630 HV of the fully martensitic region in the inner 3 mm of the wall thickness to the 350 HV of the tempered bainite within four millimeters of the outer surface. Figs. 16(c & f) are representative of the outer layers, >6 mm from the inner surface. The dilatation curve in Fig. 16(c) shows no signs of transformation to austenite during heating, i.e. the bainitic base material is just tempered.

Taking into account the above mentioned effects regarding partial transformation of bainite to austenite and the role of heating rate on the austenitization start and finish temperatures, the empirical model [10,13] was used to predict the fractions of the various microstructural components that would be expected on the basis of the exact heating and cooling cycles obtained from the induction heating and cooling model (Fig. 11). The results are shown in Table 3. There is a gradual transition from the hard-martensitic layers to the ductile, mainly bainitic, layers, which ought to result in excellent performance during slurry transportation. Fig. 17 presents a summary of the phase transformation behavior and consequential microstructure including prior austenite and as-quenched structure for the different positions through the wall thickness layers. It can be seen that the higher heating rate at the inner surface generated a finer prior austenite grain structure (6.5 μm) than a few millimeters below the surface (7.5 μm). In the central part
of the wall thickness, the maximum temperatures only reached intercritical values in the $A_1$–$A_3$ interval. Here, the prior austenite grain structure consisted of fine new austenite grains formed during heating interspersed among coarse pancaked-shaped untransformed as-rolled grains giving an overall average prior austenite grain size of 15 μm. The prior austenite of the outer layers, where the peak temperature remained below $A_1$, was that of the original thermomechanically rolled strip. Here, the average ECD size was 29 μm.

The hardness profile through the pipe wall was also predicted by the phase transformation model on the basis of the calculated mean value of the cooling rate for each thickness layer. It is compared with the experimental hardness values obtained from the Gleeble simulations of the various thermal profiles, section B in Fig. 18. It can be seen that the designed heating and cooling parameters should provide a smooth sigmoidal hardness profile with a 3 mm thick, uniformly hard inner surface and a 4 mm thick uniformly hard outer surface, where the upper and lower hardness levels are 636 and 350 HV. Except the middle layers, which have a complex microstructure, the predictions of the phase transformation model are in good agreement with the physically simulated hardness values, giving confidence that such a model can be used to evaluate the outcome of other induction heating and cooling parameters.

4. Summary and conclusions

Physical and numerical simulations have been combined to predict the variation of microstructure and hardness expected through the wall thickness of a pipe induction hardened on the inner surface. One numerical simulation was used to calculate the thermal cycles experienced at various positions through the wall thickness and another to calculate the resultant microstructure and hardness through the wall thickness. Experimental data utilizing a Gleeble simulator provided 1) the CCT data needed to calibrate the phase transformation model, and 2) the actual microstructures and hardness levels expected at different depths below the inner surface on the basis of the numerically simulated thermal cycles. Good agreement was obtained between the hardness levels obtained through both numerical and Gleeble experimental data, giving confidence that the phase transformation modeling approach described in the paper can be used to optimize induction heating and cooling parameters to give desired hardness profiles through a pipe wall. Of course, a separate study might be required to better accuracy verification of the numerically simulated thermal profiles before the approach can be used with confidence to predict an actual pipe hardening process. Specific conclusions related to the bainitic steel used as the present base material are as follows:

- Beside the increasing the re-austenitization critical temperatures with increasing the heating rate, the difference between austenite onset and end temperatures also increases by increasing the rate of heating.
- Cooling rates above 60 °C/s lead to a fully martensitic microstructure with a minimum hardness level of 630 HV.
- The numerical phase transformation and hardness model shows reliable accuracy for the cooling rates above 10 °C/s.
- Based on the regression fitting, the hardness value (HV) of the studied material for a mean prior austenite grain size of about 6–10 μm can be estimated as function of cooling rate ($C_R$) based on the following exponential correlation:

\[
\text{Hardness} = 233.69 + 398.49(1 - \exp(-0.087C_R))
\]

- The designed thermal cycles of the induction hardening produced a gradient of material properties giving a hardening depth of about 3 mm in a 10 mm thick 400 mm diameter pipe.
- The simulation shows slightly faster austenitization kinetics for all of the heating rates most probably due to neglecting the effect of prior austenite grain size.
- The developed models can be used to optimize through-thickness hardness or microstructure gradients depending on the pipe application and practical industrial considerations.

Acknowledgment

The authors are grateful for financial support from the European Commission under grant number 675715 – MIMESIS – H2020-MSCA-ITN-2015, which is a part of the Marie Skłodowska-Curie Innovative Training Networks European Industrial Doctorate programme. The authors would also like to thank EFD induction a.s. for all the helps and support during the industrial secondment for this project, especially Björnar Grande. We would also like to express our sincere appreciation to Juha Uusitalo from University of Oulu for all his efforts and considerations for the Gleeble simulations.

Appendix A. Validation of available material properties in the flux software

To check the reliability of available material properties in the software database, especially the temperature dependent properties, an experiment was performed to verify the simulation parameters and results, by comparing experimental and simulated surface temperature profiles during induction hardening.

For this verification test, a cylindrical bar made from the studied material was subjected to a hardening system that consisted of an inductor and a quenching ring as illustrated in Fig. 19. To provide uniform heating through the whole body, the specimen was made to rotate around its axis both in the experimental set up and the simulation. Quenching was performed with a quenching ring coaxial with the inductor and sample by applying the given coolant shower. The applied frequency and the magnitude of the AC current were 17 kHz and 1950 A, respectively. The quenching ring was set up to spray a mixture of water and 9% commercial AquaQuench 365 at the rate of 0.4 L/s.
The result is presented in Fig. 20, indicating a good agreement between the simulation and experimental results i.e. <5% underestimation of the predicted temperature during the heating stages and natural cooling, and <5% overestimation for the last step of quenching.

Fig. 19. Verification experimental test set up.

Appendix B. Phase transformation modeling parameters
The parameters which have been assigned to the model to fit the phase transformation equations (Eqs. (1)–(11)) to the experimental results are listed in Table 4. These parameters are needed to reproduce the results using the model equations.

Fig. 20. Comparison of surface temperature cycles during induction hardening. Simulation in red and experiment in blue. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)
### Table 4

Fitted parameters for the formation of ferrite, bainite, and martensite ($\gamma \rightarrow \alpha$), as well as the formation of austenite ($\alpha \rightarrow \gamma$).

<table>
<thead>
<tr>
<th></th>
<th>K</th>
<th>A (°C)</th>
<th>m</th>
<th>Q (kJ)</th>
<th>a</th>
<th>b</th>
<th>c</th>
<th>n</th>
<th>Ms</th>
<th>$\eta$</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Ferrite</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>$2.3313 \times 10^{-4}$</td>
<td>720.0</td>
<td>6.240</td>
<td>1.3</td>
<td></td>
<td></td>
</tr>
<tr>
<td><strong>Bainite</strong></td>
<td>0.04915</td>
<td>657.6</td>
<td>0.8681</td>
<td>56.7</td>
<td>$1.2317 \times 10^{-4}$</td>
<td>469.2</td>
<td>3.374</td>
<td>1.0</td>
<td></td>
<td></td>
</tr>
<tr>
<td><strong>Martensite</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>349.7</td>
<td>0.01158</td>
</tr>
<tr>
<td><strong>Austenite</strong></td>
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<td>2.2307</td>
<td>129.11</td>
<td>1.044</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
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</tbody>
</table>

**References**


