

# Effect of the Interpass Temperature on Simulated Heat-Affected Zone of Gas Metal Arc Welded API 5L X70 Pipe Joint

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## Research Article

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# Abstract

Welding costs associated with the laying of pipes for deepwater oil and gas extraction can be reduced using high interpass temperatures (ITs). However, a high IT can decrease the mechanical properties of the heat-affected zone (HAZ) of welded joints. With the use of high strength-toughness steels, this decrease may be an acceptable trade-off. Therefore, it is necessary to evaluate the influence of high ITs on the HAZ. The influence of the IT on the coarse grain HAZ (CGHAZ) and intercritically reheated coarse-grain HAZ (ICCGHAZ) of an API 5L X70 pipe joint welded by gas metal arc welding was investigated. The welding was numerically simulated using finite element method software. The microstructure of the HAZ was predicted using thermodynamic simulation software. The CGHAZ and ICCGHAZ were also physically simulated and evaluated via optical microscopy and scanning electron microscopy, dilatometry, and Vickers microhardness and Charpy V-notch (CVN) impact tests. The increase in IT led to a decrease in CGHAZ microhardness, but did not affect the ICCGHAZ. The CVN energies obtained for all ITs (CGHAZ and ICCGHAZ) were higher than that set by the DNVGL-ST-F101 standard (50 J). These results show that increasing the IT is an interesting and effective method to reduce welding costs. In addition, thermodynamic simulation proved to be a valid method for predicting the phases in the HAZ of API 5L X70 pipe welded joints.

## 1. Introduction

To strengthen their position in a highly competitive market, oil and gas companies have funded research projects aimed at developing solutions to reduce their process costs. The exploration of oil and gas in high-deep-water requires the use of longer rigid risers. These pipes are initially welded on support ships (~ US \$650,000/day rent) and then laid by the J-lay method. One way to reduce the costs of installing pipes in deep water is to reduce the welding time. The reduction of the welding time can be achieved by developing welding parameters, such as high heat input [1] and welding current [2]. However, this can decrease the hardness and impact toughness [3–4] of the heat-affected zone (HAZ) of high-strength low-alloy (HSLA) steels because of the formation of brittle regions, such as a coarse grain HAZ (CGHAZ) [5] and an intercritically reheated coarse-grain HAZ (ICCGHAZ) [6], which increase the risk of welded joint failure [7].

Another method that has been studied is the use of a high welding interpass temperature (IT), which reduces the time required for joint cooling, thus increasing productivity. Sirin et al. [8] observed that intermittent welding (with interruptions for joint cooling) increased the fracture toughness of the welded joint by approximately 18% compared with a continuous process. They concluded that the IT directly influenced the toughness of the HAZ of an API 5L X65 pipe. However, Neves and Loureiro [9] reported that a high IT reduces the HAZ fracture toughness, compromising the performance of the welded joint in service. The increase in IT reduces the cooling rate of the welded joint owing to the lower temperature gradient in the HAZ [10]. Shi and Han [11] reported a drop in the HAZ toughness owing to the amount of martensite-austenite (M-A) constituents formed at lower cooling rates. Dornelas et al. [12] observed that the increase in the IT significantly changed in the microstructure, and consequently decreased the CGHAZ

hardness and fracture toughness of a Cr-Mo low-alloy steel. On the other hand, Wang et al. [13] reported an improvement in the HAZ toughness due to an increase in IT between 80°C and 130°C; however, a drop was observed when the IT increased beyond 130°C. Therefore, the use of a higher IT seems to be critical in alloys sensitive to the cooling rate, justifying its control to ensure good fracture toughness. Steels with high fracture toughness may lose some of their properties without compromising their performance during service, thus reducing production costs. Thus, it is necessary to study the influence of an increase in IT on the CGHAZ and ICCHAZ of these alloys.

In welding with different parameters from those specified by the welding procedure (e.g., higher IT), it is necessary to perform a series of tests to evaluate the mechanical behavior of the welded joint, followed by a requalification of the new parameters, which is expensive. In this case, computer simulations of the welding process and thermodynamic simulations can be used to predict the HAZ thermal cycles and microstructures under different welding conditions (e.g., IT); this speeds up the analysis and makes it less expensive. Computer simulations of welding (finite element method, FEM) [14–18] and thermodynamics (Calphad method) [12, 19–20] have developed in recent years.

To reduce welding costs, this paper examines the effects of welding at  $IT \geq 300^\circ\text{C}$  on the microstructures and impact toughness of the simulated CGHAZ and ICCGHAZ of a high strength-toughness API 5L X70 pipe. The welding thermal cycles of the CGHAZ and ICCGHAZ were simulated using a commercial FEM software. Then, the CGHAZ and ICCGHAZ were evaluated by optical microscopy (OM), scanning electron microscopy (SEM), and Vickers microhardness and Charpy V-notch (CVN) impact tests. The simulated welding thermal cycles were also reproduced in dilatometry tests. A commercial thermodynamic simulation software (Calphad method) was used to predict the CGHAZ and ICCGHAZ microstructures. A comparative study was performed to evaluate the accuracy of the predictions.

## 2. Materials And Methods

### 2.1. Materials

A hot-rolled (seamless, Mannesmann process) HSLA API 5L X70 pipe with a ferritic microstructure (Fig. 1) was welded in this work. Table 1 lists the chemical compositions (wt. %) of the pipe and weld metal. Table 2 lists the mechanical properties of the pipe. The carbon equivalent (CEIIW) was calculated from Eq. 1.

Table 1  
 Chemical compositions (wt. %) of the API 5L X70 pipe and weld metal analyzed by optical emission spectroscopy, and the carbon equivalent ( $CE_{IIW}$ ) of the API 5L X70 pipe.

Chemical element	Material	
	API 5L X70 pipe	Weld metal
C	0.019	0.071
Mn	1.397	1.430
Si	0.254	0.710
P	0.010	0.015
S	0.002	0.011
Cr	0.023	0.048
Mo	0.057	0.017
Ni	0.148	0.027
Nb + V + Ti	0.080	< 0.012
Al	0.028	0.004
Cu	0.475	0.106
$CE_{IIW}$	0.32	□

$$CE_{IIW} = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Cu + Ni)}{15}$$

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Table 2  
 Mechanical properties of the API 5L X70 pipe and weld metal.

Mechanical property	Pipe	Weld metal (specified)
Ultimate tensile strength, UTS [MPa]	575	583
Yield strength, YS [MPa]	540	483
Elongation, El. [%]	61	26
Charpy V-notch impact energy (at - 30 °C), CVN [J]	293	□
Vickers microhardness, HV1	193	□

## 2.2. Gas metal arc welding

The API 5L X70 pipe was gas metal arc girth welded to obtain the data to be used as input in the numerical simulation and to perform its validation. Table 3 presents the parameters adopted for gas metal arc welding. Figure 2 shows the groove configuration and a schematic of the weld passes performed.

Table 3  
Parameters adopted in the gas metal arc welding of the API 5L X70 pipe.

Parameter	Welding pass			
	Root	Hot	Fill	Cap
Arc voltage, U [V]	13–19	25–31	25–31	14–22
Welding current, I [A]	132–180	239–337	239–298	126–220
Welding speed, WS [ $\text{cm} \cdot \text{min}^{-1}$ ]	22–30	34–42	32–47	35–44
Heat input, HI [ $\text{kJ} \cdot \text{mm}^{-1}$ ]	0.6–1.0	1.0–1.6	0.9–1.4	0.3–0.7
Welding current type and polarity	DC +			
Welding position	2G			
Filler metal	AWS A5.18 ER70S-6			
Shielding gas type	80% Ar + 20% CO <sub>2</sub>			
Shielding gas flow rate [ $\text{L} \cdot \text{min}^{-1}$ ]	20–33			
Preheating temperature, T <sub>o</sub> [°C]	90			
Interpass temperature, IT [°C]	245			

## 2.3. Welding modeling

To describe the HAZ thermal cycles, the physical welding thermal cycle simulators used simplified analytical models such as Rosenthal [21] and Rykalin 2D [22]. These models simplified some factors; for example, the physical properties are temperature-independent, the metals are solid during welding, no solid phase transformation occurs, there is no heat loss by radiation and convection, and the heat source has zero volume (point or line) [23]. However, these simplifications lead to errors in the HAZ thermal cycle simulation. Numerical simulation methods (e.g., FEM) have been used to overcome these limitations and make the physical simulation more realistic [24–25]. The commercial FEM software Sysweld® was used to accurately describe the CGHAZ and ICCGHAZ thermal cycles. The weld metal elements were simulated using the birth and death technique. The procedures for executing, documenting, and validating the welding simulations followed the ISO 18166 standard.

The welding was modeled using the principle of energy conservation [26]. The fundamental equation of heat conduction in multiphase solids is expressed as Eq. 2, where the temperature distribution ( $T$ ) is a function of the time ( $t$ ) and position ( $x, y, z$ ).  $k$  is the thermal conductivity,  $c_p$  is the specific heat,  $\rho$  is the density,  $T$  is the temperature distribution,  $P$  is the phase proportion, and  $\dot{Q}$  is the heat generated per volume.  $i$  and  $j$  are phases indexes,  $A_{ij}$  is the proportion of phase transformation per unit time, and  $L_{ij}(T)$  is the latent heat of the phase transformation.  $c_p$ ,  $k$ , and  $\rho$  (Table 4) are obtained from the chemical composition-based thermodynamic simulation of the pipe using the JMatPro<sup>®</sup> software. The same properties were used for the base and weld metals during the simulation because their chemical compositions were similar.

$$\frac{\partial}{\partial x} \left( \left( \sum_i P_i(T) \right) k_i(T) \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( \left( \sum_i P_i(T) \right) k_i(T) \frac{\partial T}{\partial y} \right) + \frac{\partial}{\partial z} \left( \left( \sum_i P_i(T) \right) k_i(T) \frac{\partial T}{\partial z} \right) + \dot{Q} - \sum_{i < j} A_{ij} L_{ij}(T) = \rho \sum_i P_i(T) c_{p,i} \frac{\partial T}{\partial t}$$

Table 4  
Density ( $\rho$ ), thermal conductivity ( $k$ ), and specific heat ( $c_p$ ) of the HSLA API 5L X70 pipe at a constant pressure and various temperatures (T).

Temperature, T [°C]	Density, $\rho$ [g·cm <sup>-3</sup> ]	Thermal conductivity, $k$ [W·(m·K) <sup>-1</sup> ]	Specific heat, $c_p$ [J·(g·K) <sup>-1</sup> ]
2000	6.58	42.92	0.82
1900	6.67	41.07	0.82
1800	6.75	39.23	0.82
1700	6.84	37.38	0.82
1600	6.92	35.54	0.82
1500	7.25	35.45	0.73
1400	7.33	34.33	0.69
1300	7.38	33.11	0.67
1200	7.44	31.89	0.66
1100	7.49	30.66	0.64
1000	7.55	29.45	0.63
900	7.60	28.23	0.61
800	7.59	30.38	0.96
700	7.61	32.81	0.98
600	7.64	35.01	0.82
500	7.68	37.54	0.69
400	7.72	40.03	0.62
300	7.75	42.03	0.56
200	7.78	43.15	0.52
100	7.82	43.10	0.48
25	7.84	42.21	0.45

The initial condition of the model assumed an IT of 245°C at 25 mm away from the groove (a common measurement location in welding). The IT is heterogeneous in the joint owing to the localized heating induced by the electric arc [27]; hence, the model considers the temperature distribution as the initial condition. The boundary condition considers the heat flow due to convection and radiation in the welded joint according to Eq. 3, where  $q$  is the energy supplied by the heat source,  $\varepsilon$  is the emissivity coefficient

(0.8),  $\sigma$  is the Stefan–Boltzmann constant ( $5.6704 \cdot 10^{-8} \text{ Wm}^{-2}\text{K}^{-4}$ ),  $T_0$  is the room temperature ( $20^\circ\text{C}$ ), and  $h$  is the convective heat transfer coefficient ( $25 \text{ Wm}^{-2}\text{K}^{-4}$ ). The latent heat of the phase transformation is also considered.

$$-k \left( \frac{\partial T}{\partial x} + \frac{\partial T}{\partial y} + \frac{\partial T}{\partial z} \right) = q - \epsilon \sigma (T^4 - T_0^4) - h (T - T_0)$$

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The moving Goldak double-ellipsoid model [28] was adopted to represent the welding heat source. This model is a combination of two ellipsoids, one in front and the other at the rear, as described by Eq. 4.  $a$ ,  $b$ , and  $c_i$  are the semi-axes of the double ellipsoid (Table 5);  $v$ ,  $q_i$  and  $t$  are the welding speed (Table 3), power density distribution in the double ellipsoid, and time, respectively.  $U$  is the electric arc voltage,  $\eta$  is the thermal efficiency of welding (0.86), and  $I$  is the welding current (Table 3).  $r$  and  $f$  are the rear and front quadrants of the Goldak double ellipsoid;  $f_r$  and  $f_f$  are the fractions of the heat at the rear ( $f_r$ , Table 5) and front ( $f_f = 2 - f_r$ ), respectively. Table 5 presents the dimensions of Goldak's heat source (I and II) used to simulate the weld passes 7–8 (I) and 5–6 (II). Figure 2 shows the weld pass numbers 5–8.

$$q_i(x, y, z, t) = \frac{6\sqrt{3}f_i\eta UI}{abc_i\pi\sqrt{\pi}} \exp \left( -3\frac{x^2}{a^2} - 3\frac{y^2}{b^2} - 3\frac{(z - vt)^2}{c_i^2} \right), i = f \text{ and } r$$

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Table 5  
Goldak's heat source parameters.

Heat source	$a$	$b$	$c_f$	$c_r$	$f_f$
I	3.50	2.50	6.00	6.00	1.09
II	3.00	1.00	1.15	3.85	1.20

The accuracy of an FEM model is influenced by the mesh density used in the physical analysis. Thus, it is necessary to perform mesh refinement in the study regions (CGHAZ and ICCGHAZ) because they experienced severe temperature gradients during welding. The adequate meshing for the problem was based on the literature [29] and the authors' simulation experience. Figure. 3 presents the FEM model mesh and a macrograph of the welded joint. Figure. 4 shows the thermal profile during the welding simulation of the sixth pass (forming a CGHAZ – left) and the eighth pass (forming a CGHAZ and ICCGHAZ – right). The ICCGHAZ was formed because of the intercritical reheating ( $900-773^\circ\text{C}$ ) of the previous CGHAZ.

## 2.4. Physical simulation of the CGHAZ and ICCGHAZ

The simulated CGHAZ and ICCGHAZ welding thermal cycles (Sect. 3.1) were reproduced in the specimens using a Gleeble® 3800 physical simulator. The specimens (11 mm × 11 mm × 80 mm) were removed along the longitudinal direction of the pipe. After the physical simulation, these were subjected to microscopy, Vickers microhardness measurements, and Charpy V-notch impact testing.

## 2.5. Thermodynamic simulation

Thermodynamic simulation was performed using the JMatPro® software to obtain the physical properties of the API 5L X70 pipe (Table 4) and the continuous cooling transformation (CCT) diagram (Fig. 5). To predict the microstructures, simulations of the phase contents during cooling for all ITs (CGHAZ and ICCGHAZ) using the average cooling rate as the input data were performed. The average cooling rate was calculated as 1350–300°C (CGHAZ) for IT = 300, 350, 400, and 425°C (9, 4.6, 2.9, and 2.6°C/s, respectively) from the simulated thermal welding cycles (Sect. 3.1). For the ICCGHAZ, the average cooling rate was similar for all conditions (1°C/s, owing to the lower peak temperature, 850°C); hence, only a single simulation was performed for all ITs. All the previously mentioned simulations were based on the chemical composition of the base metal (Table 1).

## 2.6. Dilatometry

Dilatometry tests were performed using the CGHAZ and ICCGHAZ simulated welding thermal cycles (Sect. 3.1) in a Gleeble 3800 physical simulator. The well-known tangent method [30] was applied to determine the phase formation starting temperatures during cooling.

## 2.7. Microstructural characterization

The physically simulated specimens were microstructurally characterized by OM (Zeiss Axio Imager M2m) and SEM (Tescan VEJA 3). The OM and SEM specimens were ground (sandpaper from 100 to 1200 grade), polished (diamond paste from 3 to 1 µm), and etched with 2% nital followed by 2% picral solutions.

## 2.8. Mechanical testing

Full-size CVN impact testing was performed using a JBW-300 pendulum impact tester at -30°C, in accordance with ASTM A370. The microhardness test was performed using an HV-1000 Digimet hardness tester with a load of 1 kgf, in accordance with ASTM E92.

# 3. Results And Discussion

## 3.1. FEM model and its validation

The welding simulation was validated by comparing the simulated welding thermal cycles (Sysweld®) in the last four weld passes with the actual thermal cycles measured with thermocouples; both were

obtained from the base metal 25 mm away from the joint groove. A similar validation procedure is specified in ISO 18166, which has been used by other authors in the literature [4, 12, 31]. Figure 6 shows the excellent correlation between the simulated and measured welding thermal cycles. After validation of the FEM model, welding simulations were performed considering the ITs of 300, 350, 400, and 425°C. Figure 7 shows the thermal cycles obtained for the CGHAZ and ICCGHAZ.

## 3.2. Microstructural characterization

Figure 8 shows the macrographs of the simulated CGHAZ. For IT = 300°C, the CGHAZ was formed by needle ferrite and massive and slender aligned martensite-austenite-carbide (M-A-C) morphologies, which were also observed by Di et al. [32] and Zhu et al. [33]. For IT  $\geq$  350°C, the ferritic grains were larger and only massive M-A-C formed compared with IT = 300°C. Needle ferrite was formed for IT = 300°C because of the shorter time available (i.e., with thermodynamically favorable conditions) for the diffusion of alloying elements, and consequently grain growth. This behavior occurred because of the higher cooling rate of the alloy during its phase transformation at 850–400°C (CCT diagram in Fig. 6) for IT = 300°C compared with the other ITs (Fig. 9). According to the literature [34–36], the increase in heat input changes the M-A-C morphology from slender to massive owing to the decrease in the cooling rate. Figures 8 and 9 show that the IT increase decreased the cooling rate, creating the necessary thermodynamic conditions for the formation of a massive instead of slender M-A-C. The microstructures observed in the CGHAZ were similar to those observed by other researchers [32–33, 37–38], although some have also identified the formation of granular bainite [37, 39]. The phase content during cooling, obtained from the thermodynamic simulation (Fig. 10), indicate the presence of ferrite for all ITs. For IT  $\geq$  350°C, the minor formation of perlite and bainite was predicted (< 1%). The formation of ferrite for all ITs was confirmed by dilatometry tests (Fig. 11).

Figure 12 shows micrographs of the simulated ICCGHAZ. For IT = 300°C, ferrite and massive M-A-C were formed. However, owing to intercritical reheating, the ferritic grains were equiaxial and larger than those in the CGHAZ. For IT  $\geq$  350°C, M-A-C formation was observed at the grain boundaries of the prior austenite, which was also observed by other authors [40–41]. The grain boundary M-A-C observed for IT  $\geq$  350°C may be related to the lower cooling rate after intercritical reheating (Fig. 9), which can produce the thermodynamically favorable conditions for the diffusion of alloying elements over a longer period. The formation of ferrite for all ITs was confirmed by dilatometry tests (Fig. 13). The thermodynamic simulation (JMatPro<sup>®</sup>) indicated the major presence of ferrite (Fig. 14) and minor perlite (< 1%) for all ITs.

From the microstructural analysis and dilatometry, it was possible to obtain good agreement between the observed and predicted microstructures in the CGHAZ and ICCGHAZ for all ITs. These results show that thermodynamic simulation for phase prediction is a valid and powerful method for analyzing the welding of API 5L X70 pipes.

## 3.3. Mechanical testing

Figure 15 shows the Vickers microhardness values of the CGHAZ and ICCGHAZ. For the CGHAZ, the increase in IT led to a decrease in microhardness owing to the reduction of the cooling rate (Fig. 9). Such

thermodynamically favorable conditions promoted greater diffusion of the alloying elements, thus decreasing the hardening by solid solution in the resulting ferritic microstructure. This difference in diffusion explains the change in the needle to equiaxial ferrite, as previously reported (Sect. 3.2). The decrease in microhardness with the reduction in cooling rate was also observed in other studies [4, 37, 42]. For the ICCGHAZ, the increase in IT did not strongly influence the Vickers microhardness because the cooling rates arising from the different ITs are similar (Fig. 9b).

The CVN energies for all ITs, for both the CGHAZ and ICCGHAZ, reached the maximum value of the equipment (300 J), i.e., the specimens did not fracture. The high CVN energies are attributed to the maintenance of the ferritic microstructure in all cases. Figures 8 and 12 show that in both the CGHAZ and ICCGHAZ, M-A-C was formed with slender and massive morphologies [34, 43]. Several studies have investigated the influence of the M-A-C morphology on the toughness. Despite the controversy over whether the slender [34, 44] or massive [45–46] morphology is more detrimental to toughness, all studies agree that both morphologies can cause the HAZ of the API 5L X70 pipe to become brittle [32, 40, 47]. Because of the high CVN energies obtained, it was not possible to assess the influence of M-A-C formation. However, even if this influence is negative, the obtained CVN energies were considerably higher than the limit set by the DNVGL-ST-F101 standard (50 J). Thus, all the investigated ITs do not jeopardize the fracture toughness of the HAZ.

## 4. Conclusions

In this study, the influence of the IT on the numerically and physically simulated CGHAZ and ICCGHAZ of an API 5L X70 pipe was studied. The increase in IT decreased the Vickers microhardness of the CGHAZ owing to lower cooling rates. For the ICCGHAZ, the influence of the IT was weaker. Despite the formation of M-A-C in the CGHAZ and ICCGHAZ, the CVN impact energies in these regions were considerably higher than the limit set by the DNVGL-ST-F101 standard (50 J). These results show that a high IT can reduce the welding time and increase productivity. Moreover, the thermodynamic simulation proved to be an efficient method for predicting the HAZ microstructure of API 5L X70 pipe, reducing costs through tests and analyses.

## Declarations

**Authors' contributions:** Formal analysis, P.H.G.D., V.H.P.M.O., and J.d.C.P.F.; funding acquisition, D.O.M., P.Z.J., and J.d.C.P.F.; investigation: P.H.G.D.; method: P.H.G.D.; and V.H.P.M.O.; project administration: D.O.M., P.Z.J., and J.d.C.P.F.; supervision, D.O.M., P.Z.J., and J.d.C.P.F.; original draft, P.H.G.D. and V.H.P.M.O.; writing-review and editing, P.H.G.D., J.d.C.P.F., V.H.P.M.O., D.O.M., and P.Z.J.

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**Data availability:** The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

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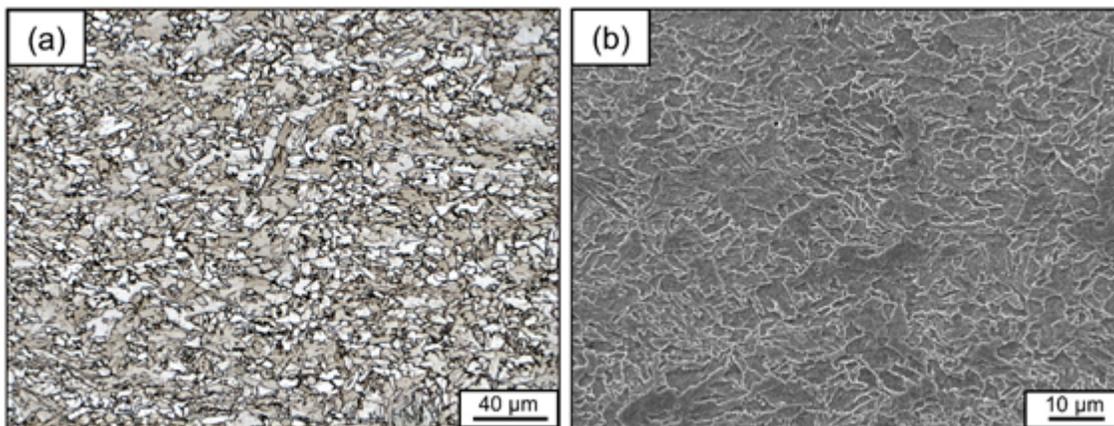
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## Figures



**Figure 1**

Optical (a) and scanning electron (b) micrographs of the fine-grained ferritic steel API 5L X70 pipe.

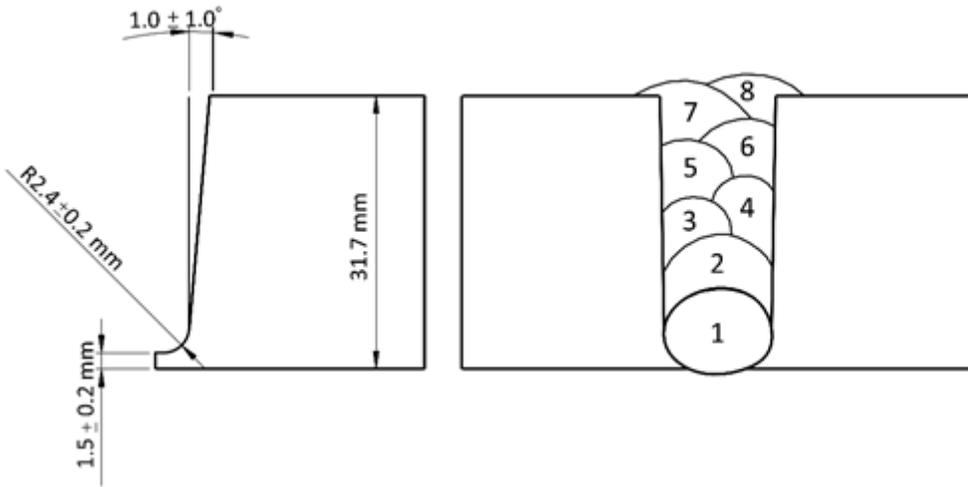


Figure 2

Groove configuration (left) and schematic of the weld pass sequence (right).

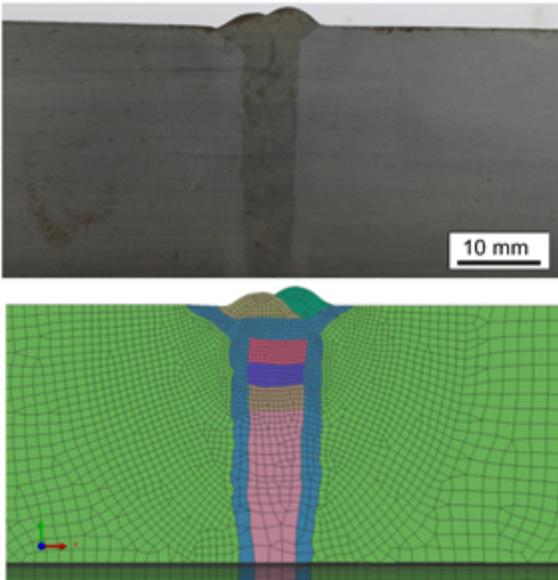


Figure 3

Macrograph of the welded joint (top) and the finite element method model mesh (bottom).

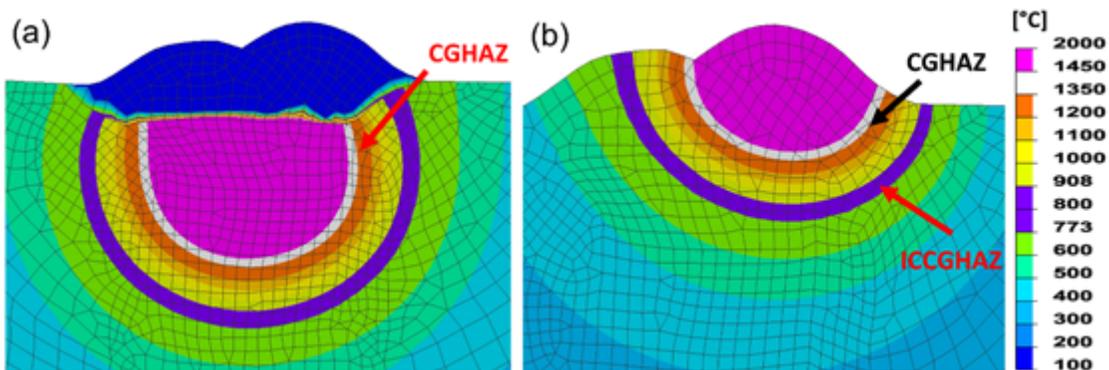
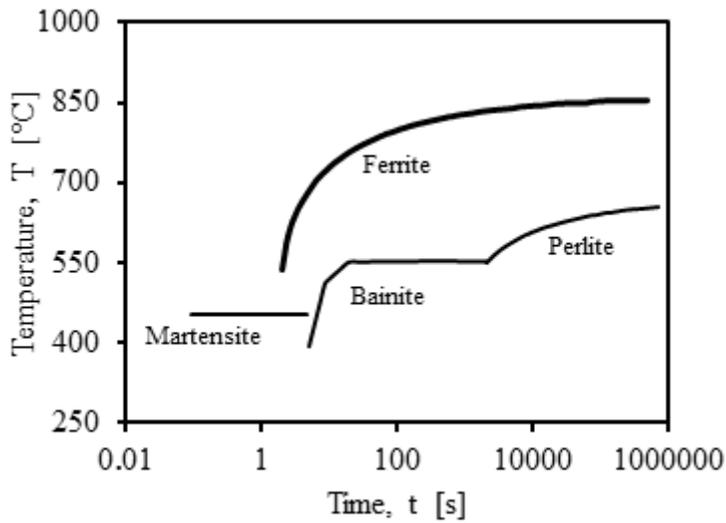


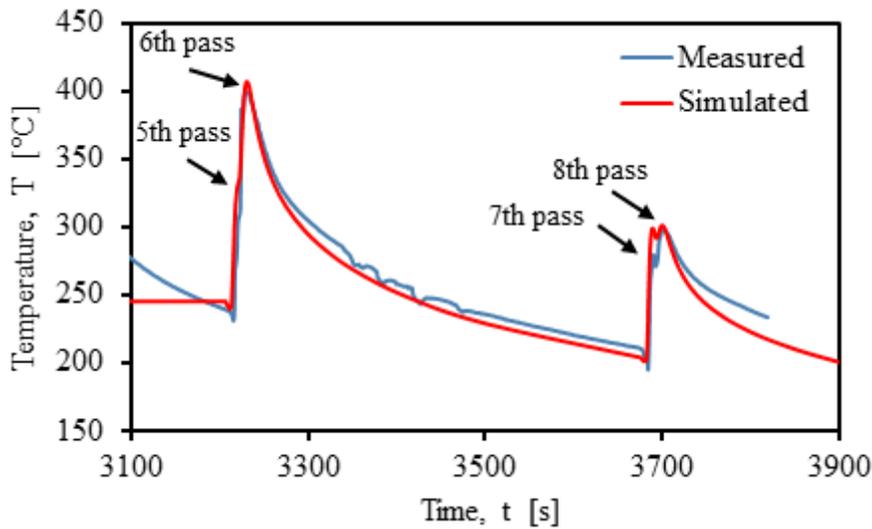
Figure 4

Thermal profile of the simulated welded joint. Simulation of the fifth and sixth weld pass forming the (a) coarse grain heat-affected zone (CGHAZ); and the seventh and eighth weld pass forming the (b) CGHAZ and intercritically reheated coarse-grain HAZ (ICCGHAZ).



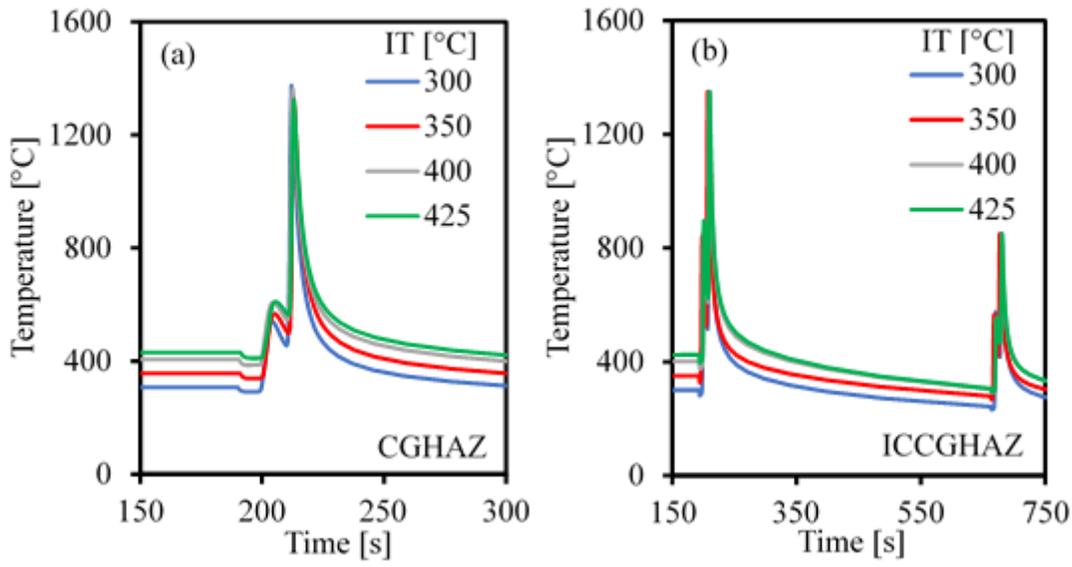
**Figure 5**

Simulated API 5L X70 pipe continuous cooling transformation diagram.



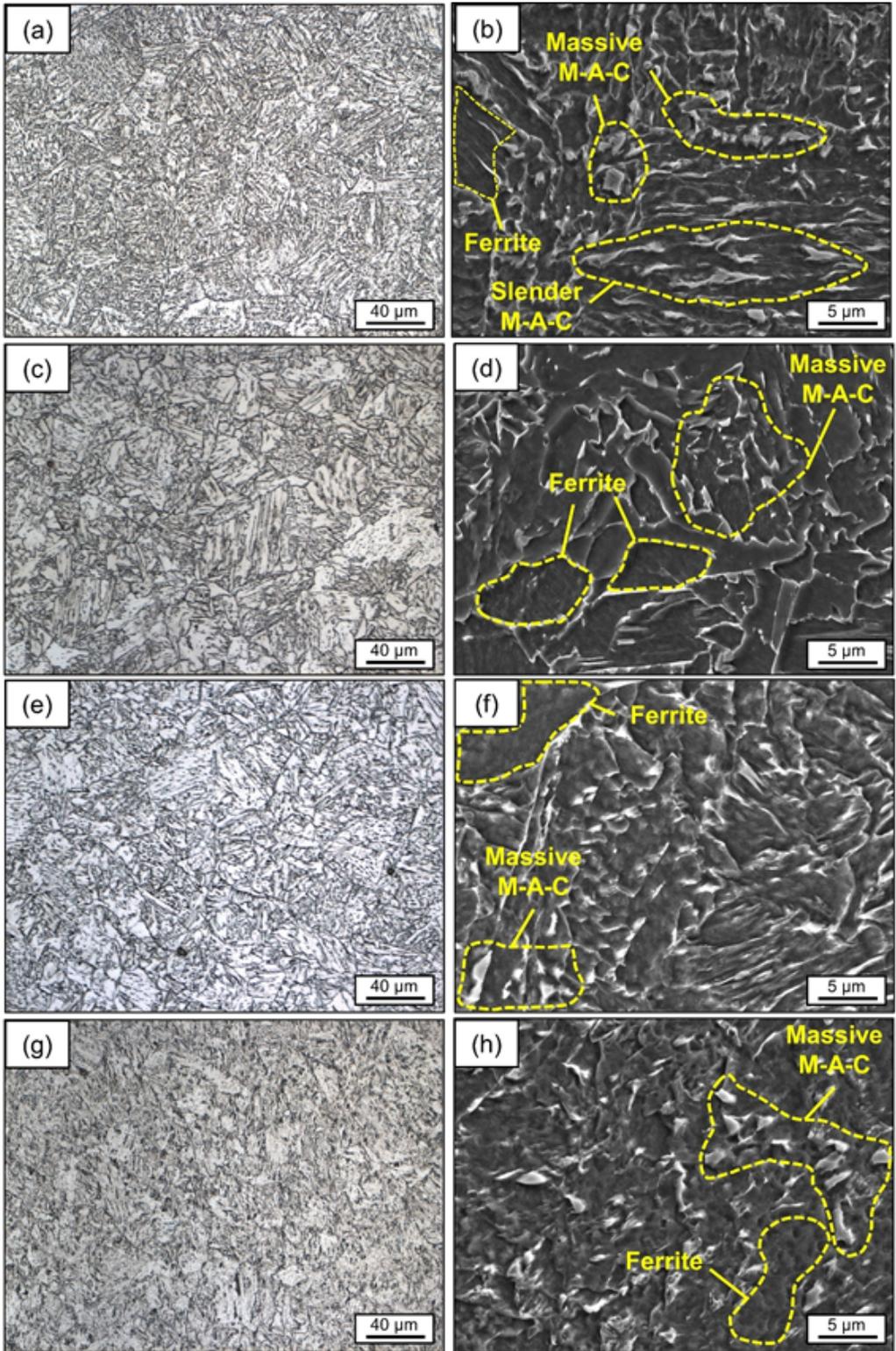
**Figure 6**

Welding thermal cycles measured with thermocouples and simulated with the commercial FEM software Sysweld®, both in the base metal 25 mm away from the joint groove.



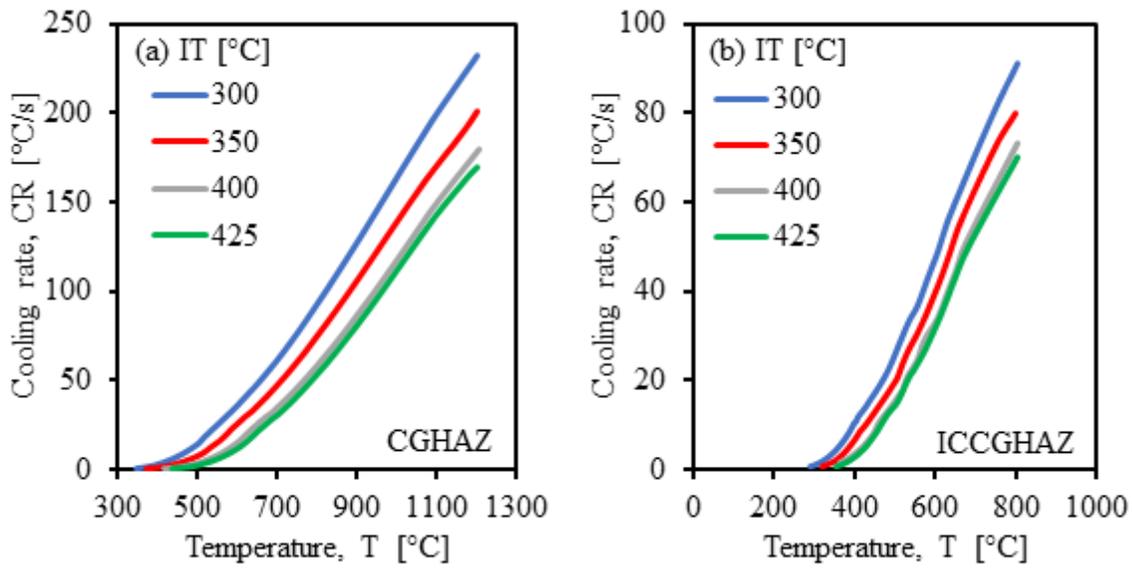
**Figure 7**

Simulated welding thermal cycles in the (a) coarse grain heat-affected zone (CGHAZ) and (b) intercritically reheated coarse-grain HAZ (ICCGHAZ) for different interpass temperatures (IT)



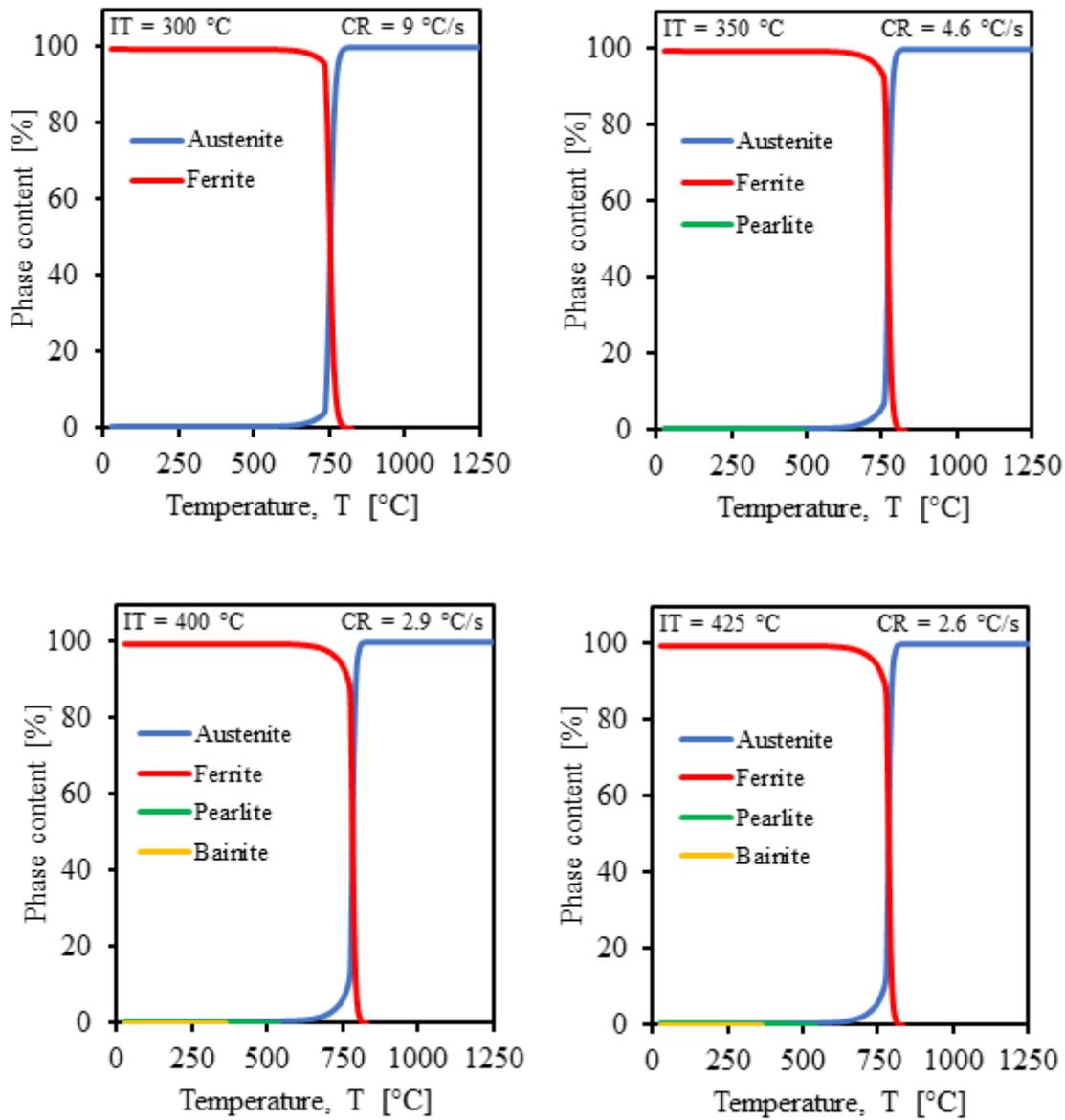
**Figure 8**

Optical (left) and scanning electron (right) micrographs of the simulated coarse grain heat-affected zone for the following interpass temperatures: (a) 300 °C, (b) 350 °C, (c) 400 °C, and (d) 425 °C.



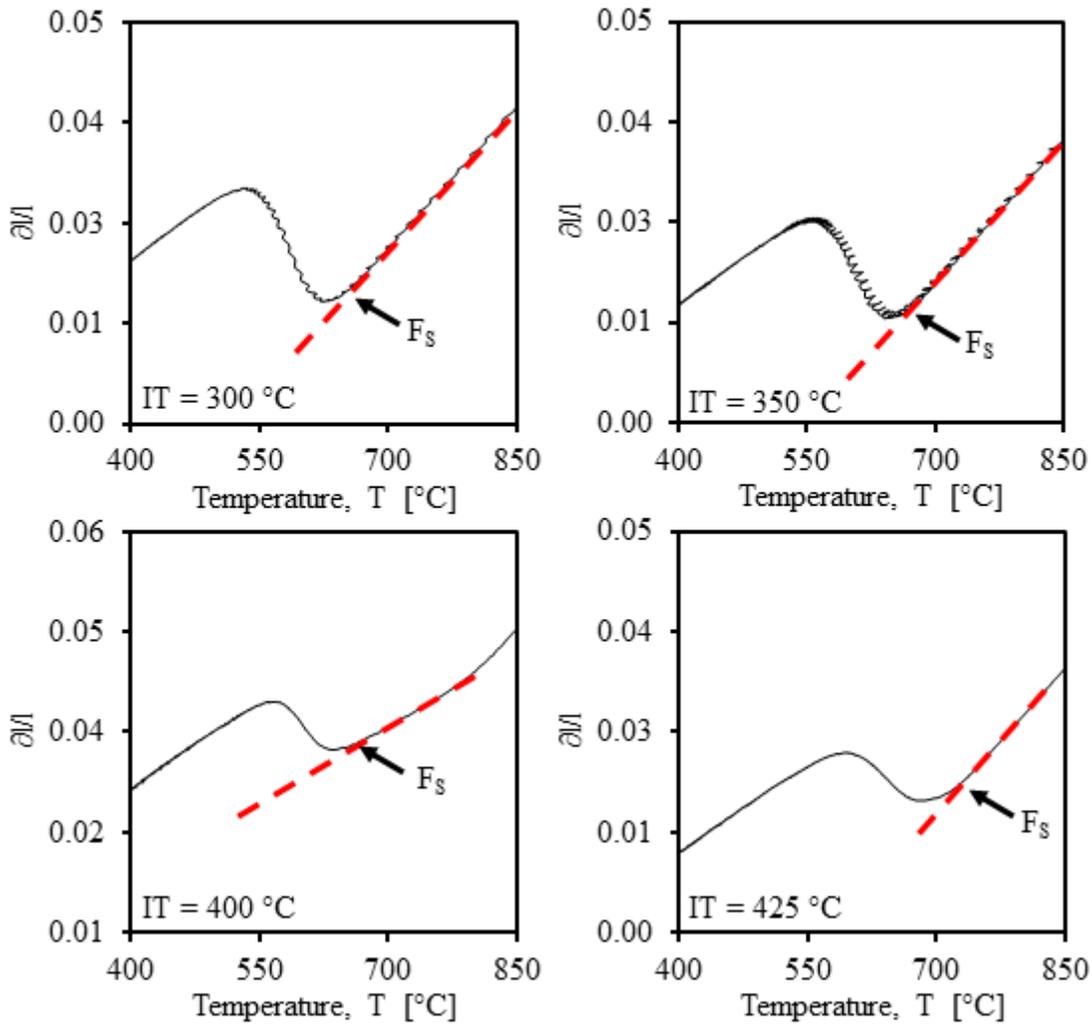
**Figure 9**

The coarse grain heat-affected zone (CGHAZ) (a) and intercritically reheated coarse-grain HAZ (ICCGHAZ) (b) cooling rates for interpass temperatures (IT) of 300, 350, 400, and 425 °C.



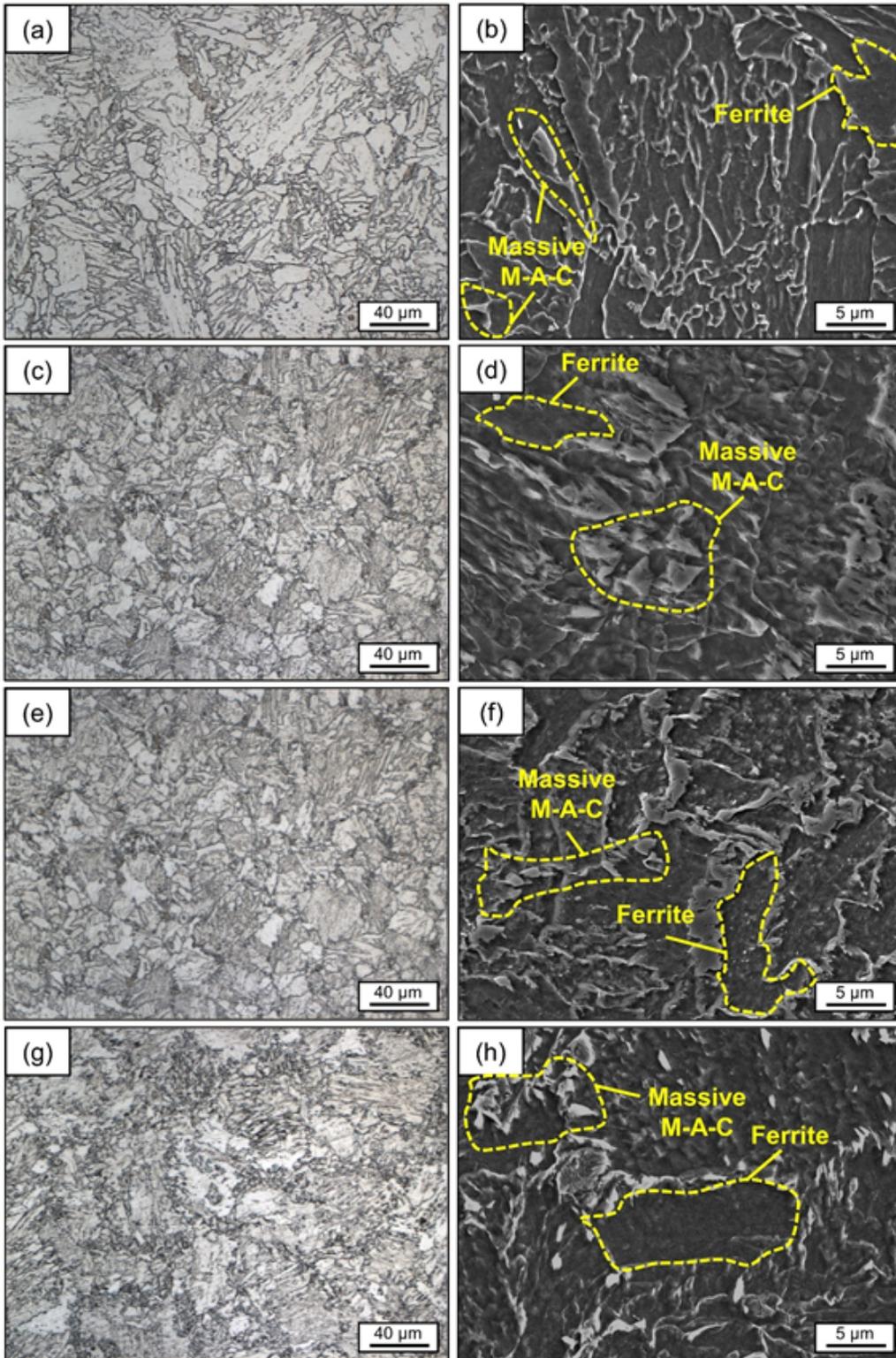
**Figure 10**

Simulated phase contents of the coarse grain heat-affected zone during cooling for interpass temperatures (IT) of 300, 350, 400, and 425 °C. CR means cooling rate.



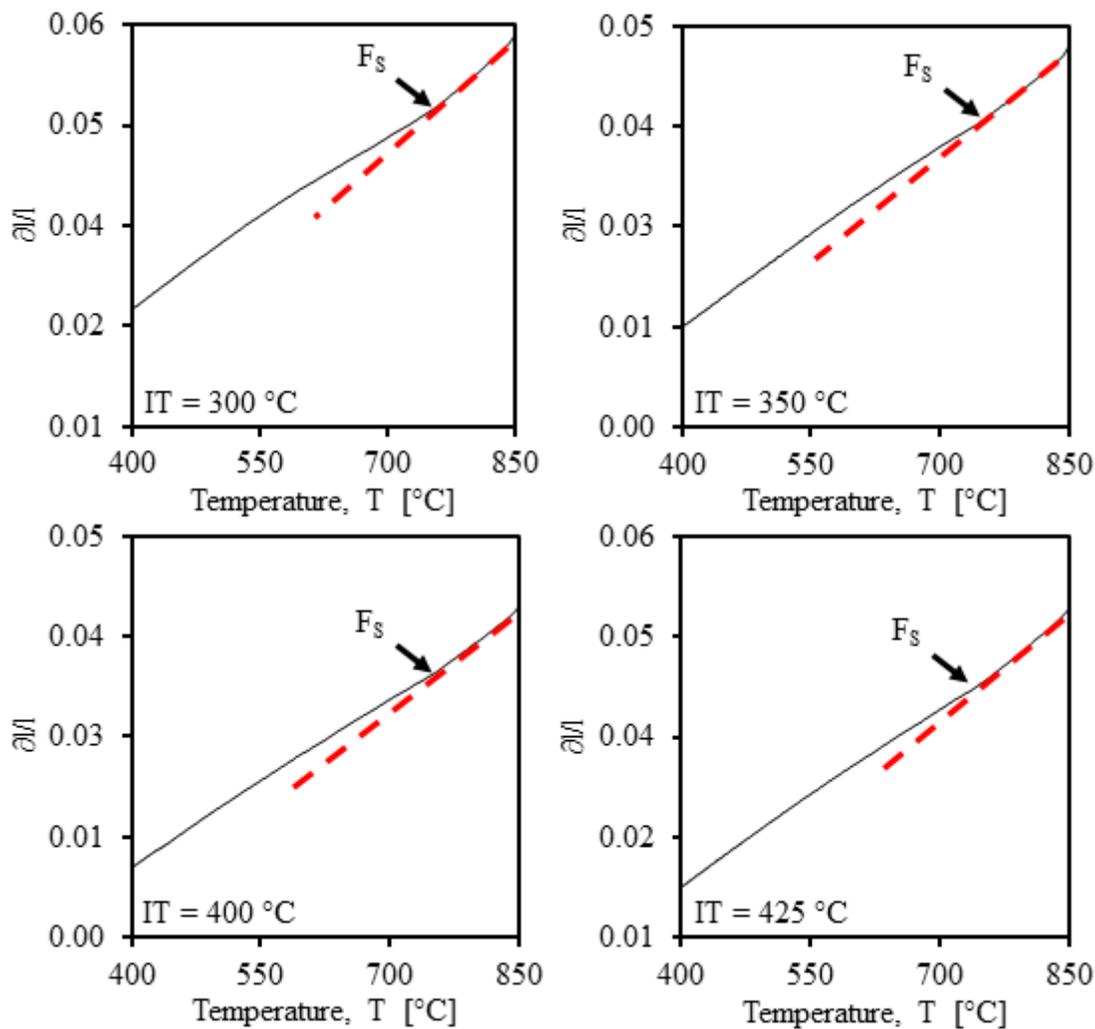
**Figure 11**

Results of the dilatometry test for the coarse grain heat-affected zone indicating the starting temperature of ferrite formation (FS) for all interpass temperatures (IT).



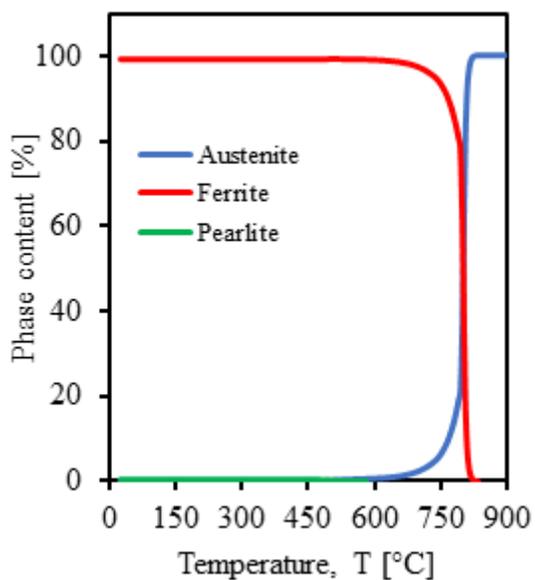
**Figure 12**

Optical (left) and scanning electron (right) micrographs of simulated intercritically reheated coarse-grain HAZ (ICCGHAZ) for interpass temperatures (IT) of (a) 300 °C, (b) 350 °C, (c) 400 °C, and (d) 425 °C.



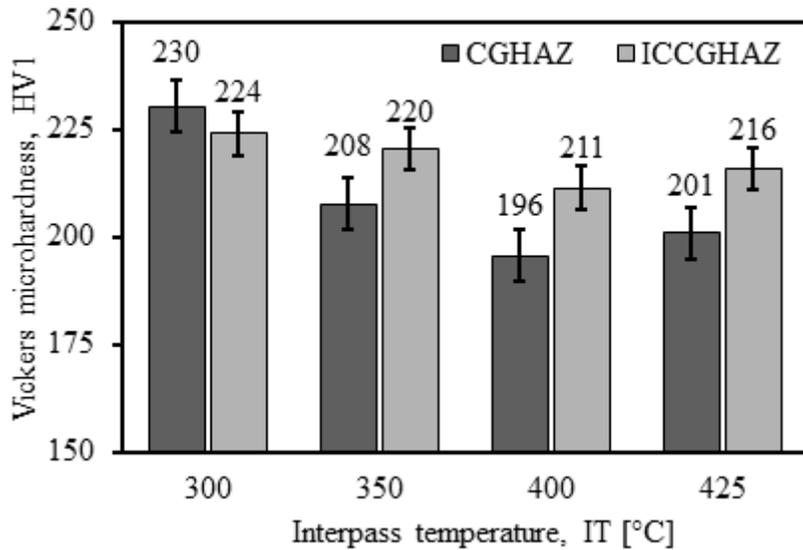
**Figure 13**

Results of the dilatometry test indicating the starting temperature of ferrite formation ( $F_s$ ) for the intercritically reheated coarse-grain heat-affected zone.



**Figure 14**

Simulated phase contents of the intercritically reheated coarse-grain heat-affected zone during cooling for all interpass temperatures (cooling rate = 1 °C/s).



**Figure 15**

Vickers microhardness of the coarse grain heat-affected zone (CGHAZ) and intercritically reheated coarse-grain heat-affected zone (ICCGHAZ) for the interpass temperatures of 300, 350, 400, and 425 °C.