Effect of the Heating Rate on Austenite Formation

Austenite formation during continuous heating on a grade X80 pipeline steel was studied

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ABSTRACT

The extent of the heat-affected zone (HAZ) in welding is typically estimated from thermodynamic considerations of austenization; however, thermodynamics are a poor predictor of the HAZ location in microalloyed steels. This work addresses the problem through the study of austenite formation during continuous heating on a grade X80 pipeline steel with an initial ferritic and bainitic microstructure. The methodology involved dilatometry, electron microscopy, and thermodynamic calculations. A continuous heating transformation diagram was developed for heating rates varying from 1° to 500°C/s. For the slower heating rates, austenite start-transformation temperature was higher than the one dictated by the equilibrium, while for the faster heating rates, start-transformation temperature gradually approached the theoretically calculated temperature at which the ferrite can transform (possibly through a massive transformation) without a long-range diffusion into austenite. Partial-transformation experiments suggested that austenite formation occurs in the following two stages: 1) the transformation of bainitic zones into austenite, and later, 2) the transformation of polygonal ferritic grains.

KEYWORDS

- Phase Transformation • Austenite Formation
- Continuous Heating Transformation • X80 • Dilatometry
- Continuous Heating

Introduction

The mechanical properties of welded materials are often limited by those of the heat-affected zone (HAZ). In the case of ferritic steels, the outer edge of the HAZ is closely associated with the beginning of transforming to austenite during heating (A_{c1}). While the effect of cooling on the HAZ is relatively well understood, and much literature has been devoted to it, the effect of the heating rate has received scant attention until recently. This lack of attention was justified, because common arc welding operations in traditional steels generate heating rates in which the material transforms qualitatively as expected from thermodynamics, that is, the transformation temperatures are only slightly affected by the heating rate, and

the nature of the transformations during heating or cooling remains similar. The widespread use of microalloyed steels and processes, such as laser beam welding, laser cladding, and laser heat treatment, has brought attention to phenomena that cannot be explained based on the traditional understanding of pure diffusive transformations.

The case of medium-carbon and eutectoid steels with a ferritic-pearlitic initial microstructure has been studied in detail. Studies of austenite formation on slightly or fully pearlitic initial microstructures concluded that austenite preferably nucleates at cementite/ferrite interfaces of the pearlite colonies, and at conventional heating rates, the growth is controlled by carbon diffusion (Refs. 1–9).

Less research has been performed for ultra-low-carbon and interstitial-free steels (Refs. 10–13), medium-Mn steels (Ref. 14), ferrite and carbide aggregates (Ref. 15), and bainitic (Refs. 4, 16–18) or martensitic structures (Refs. 5, 19–21). Even fewer studies are available in literature concerning thermomechanically controlled processed (TMCP) modern low-carbon (< 0.1 wt-%) microalloyed steels. Previous research addressing initial microstructures, such as ferritic/pearlitic (Ref. 5) and ferritic/bainitic or ferritic/martensitic structures (Refs. 22, 23), are helpful to begin an understanding of these systems but not yet comprehensive.

The effect of the heating rate has been studied at low and conventional heating rates (< 50°C/s), and the general agreement (though not unanimous) is that the transformation into austenite is a thermally activated process in which the transformation temperatures increase with the heating rate. This behavior is generally expected to extend to faster heating rates; however, for fast (50° to 100°C/s) and ultra-fast (>100°C/s) heating rates, recent work on low- and medium-carbon steels indicated a possible transition from a diffusion-controlled to an interface-controlled transformation mechanism in austenite formation (Ref. 24). Additional supporting evidence of austenite formation via an interface-controlled reaction, even for low heating rates (1°C/s), has also been provided for medium-carbon steels (Ref. 25), interstitial-free steel (Refs. 10, 15), and low-carbon, medium-Mn steels (Ref. 14). A complete modeling of austenite formation kinetics at high heating rates considering both transformation mechanisms can be found in Refs. 26 and 27.

The characteristic times and wide range of heating rates

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associated to different welding thermal cycles (depending on the welding process and parameters) demand a comprehensive study of austenite formation during continuous heating. It is well known that critical transformation temperatures, kinetics, and mechanisms involved during the austenization are susceptible to the heating rate. Thus, to have a better understanding of all of the microstructural changes that will define the HAZ properties (on cooling transformations), it is important to comprehend how the initial conditions prior to cooling are achieved. The present work becomes more relevant especially for the intercritical and the outer region of the HAZ, where the peak temperatures are slightly above the transformation-finish temperatures and there is no time for austenite grain growth.

The aim of this paper is to study the effect of the heating rate on austenite formation in TMCP steels, represented by a modern X80 pipeline grade with a ferritic bainitic microstructure. Dilatometry, microscopy, and thermodynamic calculations were used to determine and interpret a continuous heating transformation (CHT) diagram to explore the features of the austenitization process.

Experimental Procedure

Material System

Cylindrical solid and hollow samples (10 mm in length and 4 mm in diameter with a 3 mm hole) were extracted from a 12-mm-thick skelp of commercial X80 steel. Dilatometry specimens were carefully machined from a fixed depth on the skelp surface (along the longitudinal direction) to avoid the extraction of samples with traces of the typical centerline microsegregation zone usually presented in these thermomechanically processed steels (Refs. 28, 29). The chemical composition of the steel expressed in wt-% is shown in Table 1.

Dilatometry

Continuous heating experiments were performed with a high-resolution Linseis L-78 RITA quenching dilatometer. To study the influence of the heating rate on the austenite formation, full austenitization cycles followed by an immediate rapid cooling (600°C/s) were performed at different heating rates of 0.1°, 1°, 10°, 30°, 70°, 100°, 200°, 300°, and 500°C/s up to a peak temperature between 950° to 1100°C, depending on the heating rate. For low heating rates (< 10°C/s), the peak temperature was initially selected as 950°C, and for high heating rates (> 10°C/s), peak temperature was raised up to 1100°C to ensure that transformation was complete prior to quenching. Additionally, heating cycles with different peak temperatures and different holding times were performed for selected heating rates to study the

Fig. 1 — Dilatation curve (solid circles) and its first derivative (diamonds) done by the forward finite difference technique and later smoothed using the moving average of 20 points of the samples fully austenitized at 100°C/s.

different stages of austenite formation. For a fixed heating rate of 100°C/s, samples quenched from 835° and 900°C were used to study different stages of austenite formation. Additional annealing experiments, with a heating rate of 1°C/s for 0, 20, 300, and 600 s at 750°C, were performed to study the austenitization process below the experimental A_{c1} temperature. All tests were performed at a constant pressure of inert gas helium, and the temperature was controlled by a K-type thermocouple spot welded to the midsection of each specimen.

Ideally, before and after the ferrite-to-austenite transformation, the specimen will expand at a constant rate determined by the coefficient of thermal expansion of each phase; hence, the derivative of the elongation change with respect to the temperature will present a constant value. However, the effects of the magnetic transition, recrystallization, and carbide dissolution can deviate this linear behavior before and after the structural transformation. Therefore, in the present work, the critical transformation temperatures, A_{c1} and A_{c3} , were determined using the first derivative methodology (Refs. 30, 31), as it is shown in Fig. 1.

To obtain the transformed fraction as a function of the temperature, an improved methodology from the lever rule proposed by Kop et al. (Ref. 32) was employed. This correction, contrary to the lever rule, accounts for the difference in densities and thermal coefficient expansion between austenite and ferrite.

The Curie temperature, T_c , was also determined from the dilatometric experiments. To identify the onset temperature of the magnetic transformation, the power delivered by the

Table 1 — Chemical Composition of the X80 Alloy Studied (wt-%)

C	Mn	Si	Cr + Cu + Ni	Mo	Nb	V	Ti	Al	N	Fe
0.035	1.7	0.28	0.66	0.305	0.094	0.003	0.017	0.044	0.0058	bal.

Fig. 2 — Optical (A) and SEM (B) micrographs of granular bainitic zones of the initial microstructure.

Fig. 3 — SEM micrograph of a carbon-rich zone located between two bainitic ferrite grains.

high-frequency module was evaluated. Below the Curie temperature, two different phenomena, Joule heating and hysteresis loss, contributed to the heating of the sample. When the Curie temperature is reached, magnetic properties change, and only the Joule effect is responsible for the heating. This is manifested in a significant increase in power delivered to keep the heating rate constant. This methodology to evaluate T_c from dilatometry experiments has been successfully employed by several authors (Refs. 32, 33).

Characterization of Microstructure

After the thermal cycles, samples were cross sectioned and polished prior to the etchant process. Samples etched with 2% Nital were examined by optical microscopy and field emission scanning electron microscopy (FESEM), Zeiss Sigma 300 VP-FESEM, with an operating voltage of 5 to 15 KeV.

The ferrite grain size was measured on several scanning electron microscopy (SEM) micrographs with different magnifications. The average ferrite grain diameter (d_a) was estimated using the mean linear intercept method, and the effect of a nonequiaxial structure was eliminated by using lines in different orientations with an approximately equal weight.

Results

As-Received Material

Optical and SEM micrographs of the as-received material are shown in Fig. 2A and B, respectively. The initial microstructure, typical of low-carbon thermomechanically processed steels, consisted of a complex structure formed of polygonal ferrite (PF) and granular bainite (GB). The larger ferrite grains, transformed at high temperatures, were surrounded by the ferritic bainitic zones, which were composed of much finer ferrite grains and second phases mainly along the grain boundaries — Fig. 2B.

The average grain size, d_w , and the standard deviation were 2.06 and 0.70 μm , respectively.

The second phases are identified as bright zones in the high-magnification SEM micrograph — Fig. 3. Although carbon content was significantly low, it exceeded its solubility in ferrite at low temperatures, thus the excess was located in these second phases. The low-carbon content available made the formation of pearlite difficult during austenite decomposition, resulting in a microstructure practically free of pearlite. Instead, the residual austenite located between the bainitic grains will transform into block-like M/A constituents, carbides, or even retained austenite at room temperature in absence of pearlite formation.

From a previous extensive characterization work, in the same studied alloy and processing conditions, the following precipitates with a NaCl-face-centered-cubic structure were identified: (I) relatively large (230–170 nm) Ti-rich nitrides, (II) medium-sized (30–80 nm) Nb-rich (with variable amounts of Ti) carbonitrides, and finally (III), nanosized (less than 10 nm) Nb-Mo carbides. The Ti- and Nb-rich nitrocarbides (I and II) usually precipitate at high temperatures in the field while the nanosized precipitates (containing Mo) formed at lower temperatures close to the coiling temperature (CT). More detailed information about the precipitates in the as-received condition can be found in work performed by Lu (Ref. 34).

From the study of the as-received microstructure, it can be concluded that austenite forms from a nearly complete ferritic structure composed by PF, GB with minimal amount of carbon-rich second phases, and a variety of MX precipitates where M = (Ti, Nb, Mo) and X = (N, C).

Fig. 4 — CHT diagram of the studied alloy.

Fig. 5 — Time required for the transformation to complete as a function of the heating rate.

Dilatometry: Fully Transformed Samples

The values of A_{c1} , A_{c3} , and T_c for all dilatometry tests were determined using the procedure outlined above. The results, shown in Table 2, were used to construct the CHT diagram shown in Fig. 4. The experiments revealed systematic trends for A_{c1} and A_{c3} .

The start of the austenization temperature, A_{c1} , showed a slightly gradual increase with heating rates approaching 810°C at the highest rates. In contrast, the end of austenization, A_{c3} , exhibited higher susceptibility to the heating rate. The magnetic transition, T_c , was unaffected by the heating rate for all the experiments.

Although there was a significant increase in the transformation-range temperatures ($A_{c3}-A_{c1}$) with the heating rate (hr), the time to complete austenization decreased significantly with the heating rate, calculated as $(A_{c3}-A_{c1})/hr$ as shown in Fig. 5.

At high heating rates (above 10°C/s), during the structural transformation, the elongation curve against the temperature presented two main different stages, which were much more noticeable on the derivative curves. As it is shown in Fig. 6, each derivative curve was composed by a strong peak at lower temperatures close to A_{c1} , followed by a weaker peak at higher temperatures, determining the end of the structural change, A_{c3} . From Fig. 6, it can be observed that the first peak moved slightly toward high temperatures as the heating rate increased, while the second peak presented much higher susceptibility to the heating rate.

Partial Transformation Experiments

To explore the sequence of transformations, experiments with intermediate peak temperatures were performed. By having partial austenization, the austenite developed was distinguished from the untransformed structure by the presence of a much finer microstructure characteristic of fast-cooling transformations.

Three different peak temperatures were explored by heating samples at 100°C/s up to 835°, 900°, and 1110°C, followed by a rapid cooling of 600°C/s to room temperature. Figure 7 shows the resulting microstructure of these experiments.

In the sample heated up to 835°C (approximately 25% of transformation, according to the dilatometry analysis and after the first peak of the derivative [Fig. 7A]), only the bainitic zones in the surroundings of the larger ferrite grains presented evidence of transformation into austenite, γ' . The PF grains, still untransformed, resulted in a similar heterogeneous microstructure to the as-received sample.

For the sample heated up to 900°C, right after the second peak where the transformation was completed (Fig. 7B), the larger ferrite grains were absent, evidencing a growth of austenite into them, which later transformed into ferrite and bainite during cooling. The result was a more homogeneous microstructure composed of ferritic and bainitic zones.

Figure 7C corresponds to heating up to 1100°C, and it shows the resulting microstructure of two phenomena: complete transformation into austenite and significant austen-

Table 2 — Critical Transformation Temperatures Determined from Dilatometry

Heat Rate	°C/s	1	10	30	70	100	200	300	500
T_c	°C	729	728	722	725	730	732	728	728
A_{c1}	°C	780	797	798	795	798	810	805	812
A_{c3}	°C	888	875	879	885	892	922	934	957

Fig. 6 — Derivative curves for heating rates: A — 10°C/s; B — 30°C/s; C — 100°C/s; D — 500°C/s.

ite grain growth, expected from the high peak temperature reached. During the rapid cooling, the austenite decomposed into a classic upper bainitic structure.

Annealing Experiments

For the continuous heating experiments, there was no evidence of transformation below 780°C, even at the lowest heating rate (1°C/s); however, the A_{e1} temperature expected from thermodynamics was 640°C.

To investigate the discrepancy between experimental and thermodynamic austenization temperatures, isothermal annealing experiments were carried out in a sample at 750°C for 0, 20, 300, and 600 s.

In these annealing experiments, the as-received material experienced transformation into austenite after 20 s of the isothermal annealing and, after 300 s, reached a volume fraction close to the one predicted by thermodynamics, approximately 15%.

Figure 8 indicates that austenite grows preferably in the bainitic zones, where most of the carbon-rich phases were located. In the SEM micrograph, no carbon-rich phase zones are seen, suggesting that they were consumed by the austenite.

From the carbide/martensitic structure formed during cooling at 200°C/s, it can be derived that during the annealing at high temperatures, there was enough time for a sig-

nificant alloying element partition into the austenite and, hence, an increase on the hardenability compared to the samples quenched immediately after reaching the peak temperature.

Mass Balance

Thermodynamic calculations were performed using the Thermo-Calc Software TCFE10 database to obtain transformation temperatures, A_{e1} , A_{e3} , paraequilibrium boundaries, and the T_0 curves for the studied steel. The calculated A_{e1} and A_{e3} were 640° and 840°C, respectively.

During the thermomechanical-controlled processing, titanium, niobium, and vanadium nitrocarbides were intentionally precipitated, depleting the austenite (that later transformed into ferrite) from carbon. These precipitates are known to be thermally stable at high temperatures, and they will trap the consumed carbon even at high temperatures above 1100°C, especially for short time periods (Refs. 23, 35).

In the as-received microstructure, the carbon available to take part in the austenite formation during heating will be distributed between the ferrite grains (with very low-carbon solubility) and a small volume fraction of second phases (such as M/A or retained austenite), which are richer in carbon and prone to transform at high temperatures.

Neglecting the Ti-rich precipitates (which are usually

Fig. 7 — SEM micrographs of the samples transformed up to 835°C (A), 900°C (B), and 1100°C (C), followed by a rapid cooling of 600°C/s (D).

combined with N at high temperatures during the first stages of the thermomechanical processing and do not have further reactions) and considering only Nb-rich precipitates (which may contain other alloying elements in solution such as Mo, V, and Ti) from now on referred to as MX, the following mass balance can be used to estimate the carbon content in austenite after precipitation has ceased:

$$x_c = x_c^{\gamma} m_{\gamma} + x_c^{\text{MX}} m_{\text{MX}} \quad (1)$$

where all fractions are in wt-%, x_c^{γ} is the carbon content of austenite, x_c^{MX} is the carbon content in MX, m_{γ} is the mass fraction of austenite, and m_{MX} is the mass fraction of MX, such that $m_{\gamma} + m_{\text{MX}} = 1$.

Assuming that the number of MX Nb-rich precipitates is equal to the maximum amount given by thermodynamic calculations (at 700°C), the maximum carbon content of austenite prior to its decomposition, x_c^{γ} , was estimated at 0.019 wt-%. This mass balance compares well with a similar material analyzed in Ref. 34, in which the carbon content in austenite was assessed as $x_c^{\gamma} = 0.0207$ wt-%.

Analysis of Paraequilibrium and Diffusion Time Scales

Figure 9 shows the paraequilibrium and T_0 curves calculated with Thermo-Calc for the studied alloy. The A'_{e1} and A'_{e3} curves represent the ferrite/austenite phase boundaries under paraequilibrium, and the T_0 curve defines the conditions at which the Gibbs-free energy of the ferrite and austenite

(with the same chemical composition) are equal. For a set amount of carbon content, the temperature given by the T_0 line determines the temperature above which austenite can form with the same chemical composition as the parent phase (ferrite), and hence no long-range diffusion is required. The T_0'' curve was calculated with the same basis as T_0 but adding 400 J/mol to the ferrite phase in the thermodynamic calculations to account for the strain energy associated to the displacive character of the bainite formation upon cooling (Ref. 36).

During last stages of the TMCP, austenite decomposed into two different products: at high temperatures, PF; and as the temperature decreased (usually around the CT), mixed displacive/reconstructive transformations, such as acicular ferrite or bainite, occurred. The precipitation of Ti- and Nb-rich carbides occurred at high temperatures, and the Nb-Mo nanosized carbides and the bainitic transformation took place at comparable temperatures. However, the remaining carbon, after all precipitation ceased, exceeded the solubility in ferrite at low temperatures, $x_c^{\gamma} = 0.0207$ (Refs. 37, 38).

The transformation of austenite into GB at the CT, which originally forms from an austenite with a carbon content x_c^{γ} , will not obey the lever rule and will likely exhibit classical incomplete reaction phenomena (Ref. 36) in which the reaction ceases when the carbon content of the remaining austenite reaches the T_0'' curve, x_c^{un} . The untransformed austenite, referred to in this work as carbon-rich second phases, will likely transform partially into M/A constituents in absence of carbide precipitation.

The carbon content in ferrite was assumed to be between 0.007 and 0.003 wt-%, which is consistent with the solubility curves of carbon in ferrite for low temperatures. The approximated value was similar to the reported values of 0.005 and 0.008 wt-% for the same material (Ref. 34) and a similar alloy (Ref. 39), respectively. Since the CT was unknown, the carbon content of the untransformed austenite (product of the incomplete reaction phenomena), x_c^{un} , was estimated, assuming an intermediate CT of 600°C, and its intersection with the T_0'' curve was calculated. The value given by $T_0''(\text{CT})$, pointed in Fig. 9, represents the carbon composition of the untransformed austenite.

Fig. 8 — Optical (A) and SEM (B) micrographs of the annealed sample at 750°C during 600 s and then rapidly cooled down at 200°C/s.

Fig. 9 — T_0 and paraequilibrium lines calculated with Thermo-Calc.

The diffusion coefficients of substitutional alloying elements are orders of magnitude lower than those of interstitial elements, and the difference increases as the temperature decreases. With an approximated ferrite grain size of 2 μm , the diffusion coefficient of substitutional elements in the CT range was between 10^{-17} to 10^{-20} m^2/s (Ref. 38). For a characteristic diffusion distance $x = \sqrt{Dt}$, it took days for the substitutional alloying elements to partition and reach orthoequilibrium conditions. Thus, the ferrite inherited the bulk composition of substitutional alloying elements during rapid thermal cycles, and paraequilibrium conditions were used to study the ferrite/austenite boundaries upon continuous heating.

To consider a partitionless transformation of ferrite to austenite, it was necessary to use the composition of carbon in the ferrite to estimate the temperature A_M , at which this process was thermodynamically possible. The temperature A_M is the intersection of the T_0 curve and the carbon content of ferrite. For x_c^α between 0.007 and 0.003 wt-%, Fig. 9 yields a temperature A_M in the range of 812° to 808°C.

If transformation involves the long-range diffusion of carbon atoms, the carbon-rich second phases will play a role

during the transformation as they are the only available carbon source. If the untransformed austenite does not decompose at a low temperature or during heating, the temperature at which austenite will start growing will be higher than the one indicated by the equilibrium. According to the theory presented by Yang et al. (Ref. 17), the incomplete reaction phenomena during the bainitic transformation will result in a higher volume fraction of austenite than that required by the lever rule. Because of this metastable condition, during heating, the driving force for austenite to transform into ferrite will still be negative until the temperature reaches $T_s = A'_{c3}(x_{u\gamma})$, pointed in Fig. 9. Hence, austenite will start to grow only if the temperature is above T_s . According to the approximations and calculations made above, the T_s temperature for a CT of 600°C is 785°C, similar to the A_{c1} obtained at low heating rate experiments.

Discussion

Two different stages of the transformation were observed on dilatation curves for heating rates above 10°C/s; these two

stages are evident on the first derivative curve (Fig. 6). Although these phenomena can overlap in time/temperature, it was accepted that the dilatation curve during austenite formation was a convolution of two separate transformations (Refs. 8, 40, 41). For the case of classic ferritic and pearlitic structures, the different stages of the dilatation curve were associated with pearlite and ferrite transformation into austenite. However, in the present work, the as-received material was pearlite free; the initial microstructure was composed of considerably larger PF grains embedded in much finer GB grains. The two stages observed were likely to be a first transformation of the granular bainitic zones, which are finer and adjacent to carbon-rich zones, followed by the PF.

Evidence supporting the transformation sequence proposed was in the partial transformation experiments reaching 835°C (immediately after the peak of the first transformation). In these experiments, the microstructure associated with a transformation during heating was observed only in the granular bainitic zones, leaving the large PF grains untransformed (Fig. 7A). A partial transformation experiment reaching 900°C (immediately after the peak of the second transformation) showed that the PF grains were consumed during heating (Fig. 7B).

The first stage of the transformation showed only a small effect on the heating rate, likely because of the faster kinetics involving the higher number of interfaces and possible nucleation sites for austenite. The transformation of PF, with fewer interfaces and carbon-rich zones, is expected to be slower and more influenced by the heating rate.

The finish-transformation temperature (A_{c3}) was much more susceptible to the heating rate than A_{c1} . Experimental noise made the determination of A_{c1} more difficult at lower heating rates, making unreliable any determination at 0.1°C/s. The determination of A_{c3} was always reliable. In both cases, the values determined through dilatometry seemed to follow consistent trends. For the samples transformed at heating rates above 10°C/s, the A_{c3} temperature showed a gradual increase with the heating rate. However, A_{c3} for the sample heated at 1° and 0.1°C/s showed the opposite trend. The same trend was reported by Chang et al. (Ref. 22) for a bainitic microstructure with a similar alloy composition. The trend at fast heating rates was expected based on the superheat necessary to drive a faster transformation. Enomoto et al. (Ref. 27) indicated that at high heating rates, the movement of the austenite interface (i.e., growth of austenite into ferrite grains) can be slower than the actual change in temperature, resulting in a significant increase in the A_{c3} temperature. The trend at slower cooling rates has no obvious explanation.

The experimentally obtained austenite formation temperatures were significantly higher than the ones predicted by the equilibrium; in particular, A_{c1} differed by more than 100°C even at heating rates as low as 1°C/s. In contrast with transformations during cooling that usually start from single-phase austenite, heating transformations involve a much more complex starting microstructure depending on the previous processing and which might include metastable components. The complex starting microstructure will involve a diversity of transformations during heating with different time scales and limited by the time available given by the heating rate.

In the present work, the heating experiments initiated from a nearly complete ferritic/bainitic structure. The minor

fraction of carbon-rich second phases observed between the bainitic ferrite can be associated with the products of an incomplete transformation to bainite. Although the exact nature of these carbon-rich zones has not been identified, according to characterization performed in this work and found in literature (Refs. 39, 42, 43), these carbon-rich zones are believed to be a mixture between M/A constituents and retained austenite containing a significantly lower amount of carbon than predicted by the lever rule at the CT.

If all of the untransformed austenite remains metastable in the initial microstructure, nucleation is not necessary and austenite growth would start when the alloy is heated to the T_s temperature calculated above.

On the other hand, if at low temperatures, all of the austenite transforms into martensite, the reverse transformation upon heating would be energetically favorable to a lower temperature than the previous case. However, the volume fraction of austenite that forms from the martensite, given by the lever rule (considering a system with a carbon content of x_c^y), will be lower than the initial volume fraction of martensite (carbon-rich second phases) until the carbon concentration of the new austenite reaches x_c^{xy} . Hence, the total fraction of austenite will not overcome the initial volume fraction of martensite until temperatures reach T_s .

A more plausible mixed scenario where the carbon-rich second phases are composed of M/A constituents will involve a more complex sequence of austenite formation, but the limiting start-transformation temperature will lie in between the aforementioned scenarios.

The value of T_s was subjected to the CT and T'_0 selected. A change in the CT or strain energy due to the formation of bainite had a direct influence on the T_s value. However, A'_{c3} slightly changes with the carbon content, and a shift of 50°C of CT or a decrease of 200 J/mol in the energy stored would lead to an approximately 20°C difference on the predicted T_s . It is important to remark that the volume fraction of carbon-rich second phases was significantly low, and a partial transformation into austenite is not detectable with the current technique. Therefore, the experimental A_{c1} is likely related to the temperature at which austenite is energetically free to grow, T_s . At heating rates as slow as at 1°C/s, the obtained A_{c1} (780°C) lies between these two scenarios and is in good agreement with the T_s temperature proposed following the Yang et al. (Ref. 17) reaustenitization theory for acicular ferrite. Values reported in literature of ferrite-to-austenite formation temperatures for similar alloy compositions but with lower (Refs. 20, 22, 30, 44, 45) and comparable (Refs. 12, 13) heating rates show A_{c1} temperatures similar to those observed in this work.

Partial transformation experiments reaching 750°C (below T_s but above A_{c1}) indicated no presence of austenite when the quench followed directly from the heating. For the case where there was a 20-s anneal at the peak temperature, the granular bainitic zones showed evidence of transformation starting at the carbon-rich zones. This is consistent with the absence of substitutional alloying elements diffusion during heating, and the presence of enough diffusion during the hold, to drive a transformation through a possible partition mechanism controlled by Mn diffusion to reach an austenite fraction comparable with the orthoequilibrium conditions calculated at the annealing temperature after 300 s.

As the heating rate increased, A_{c1} approached 810°C,

which was similar to the estimated A_M . At this temperature, there was no energy restriction for the ferrite to transform into austenite without composition change.

More experimental and theoretical support is necessary to affirm that the transformation occurs through a partitionless mechanism. Nevertheless, the experimental results and literature evidence suggest that to model the kinetics of austenite formation at high heating rates, it should be considered at least as a coupled mechanism (Refs. 24, 25, 46).

Conclusion

Austenite formation during continuous heating on a grade X80 pipeline steel, its composition listed in Table 1, with an initial ferritic and bainitic microstructure (Fig. 2) was studied by dilatometric analysis and electron microscopy.

A CHT diagram was developed for heating rates varying from 1° to 500°C/s — Fig. 4.

At the slower heating rates tested (below 10°C/s), the austenite start-transformation temperature A_{cl} was higher than the one dictated by the equilibrium (A_{e1}) by approximately 100°C. The A_{cl} determined (780°C at 1°C/s) was in good agreement with the transformation temperatures predicted by the effect of an incomplete transformation to bainite at the CT (785°C). However, specific information about the processing conditions was needed to make more accurate predictions and comparisons with experimental values.

As the heating rate increased, A_{cl} shifted to higher temperatures and gradually approached the A_M temperature, determined thermodynamically (810°C).

On the other hand, the finish-transformation temperatures were also higher than the one dictated by the equilibrium. A_{c3} not only increased significantly with the heating rate but also showed a modest increase trend for low heating rates.

Above A_M , austenite nucleation and growth were thermodynamically feasible from ferrite without a change in composition. The occurrence of transformation near this temperature and the thermodynamic feasibility of an interface-controlled phase transformation suggests that these steels of very low carbon might involve significant amounts of massive transformation during fast heating. Current research is evaluating this possibility.

Interrupted transformation experiments with a maximum temperature of 835°C and above were performed at an intermediate heating rate (100°C/s) followed by fast cooling (600°C/s). In these experiments, the bainitic zones surrounding untransformed, larger PF grains showed evidence of transformation during cooling, indicating that the austenitic transformation starts in the bainitic zones of the starting microstructure, followed by transformation of proeutectoid ferritic grains. And this phenomenon can explain the occurrence of two different stages during the structural transformation.

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