Modeling Solid-State Phase Transformations of 13Cr-4Ni Steels in Welding Heat-Affected Zone

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Fatigue damage is commonly encountered by operators of Francis type hydraulic turbine runners made of 13Cr-4Ni soft martensitic stainless steel. These large and complex welded casting assemblies are subjected to fatigue crack initiation and growth in the vicinity of their welded regions. It is well known that fatigue behavior is influenced by residual stresses and the microstructure. By including solid-state phase transformation models in welding simulations, phase distribution can be evaluated along with their respective volumetric change and their effect on residual stresses. Thus, it enables the assessment of welding process on fatigue crack behavior by numerical methods. This paper focuses on modeling solid-state phase transformations of 13Cr-4Ni soft martensitic stainless steel, used for manufacturing hydraulic turbine runners, occurring upon welding. It proposes to determine the material parameters of the models for both the austenitic and the martensitic transformation by nonisothermal dilatometry tests. The experiments are conducted in a quenching dilatometer with applied thermal conditions as experienced in the heat-affected zone of homogeneous welds. The activation energy and the kinetic parameters of the austenitic transformation from fully martensitic state are measured from the experimental results. The martensitic transformation modeling from a fully austenitic domain is also presented.

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I. INTRODUCTION

SINCE its elaboration in the late fifties, 13Cr-4Ni soft martensitic stainless steel has been extensively preferred to low carbon steel for the manufacturing of hydraulic turbines as it offers better mechanical properties, toughness, and corrosion resistance.^[1,2] Moreover, this steel is weldable, has a higher yield strength than common austenitic stainless steels, and it is less expensive, owing to a low alloying nickel content. These benefits made this material a good candidate for the manufacturing of Francis type hydraulic turbines, which are large and complex structures generally made of multiple cast components assembled with homogeneous Flux Cored Arc Welding (FCAW) deposits. Despite the advantages of 13Cr-4Ni steels, fatigue damage initiation and growth in the welded regions continue to be a recurrent problem that must be addressed by hydraulic powerplant operators.^[3,4] Numerous factors promote

fatigue damage such as geometry, microstructures, casting defects, weld porosity, excessive cyclic loading, and residual stresses that are inherent to welding operations. Several studies have addressed the interaction between welding residual stresses and fatigue crack initiation and growth.^[5–8] Nevertheless, a better understanding of the welding residual stresses build up and of their interaction with the hydraulic load applied on the turbine requires the development of appropriate numerical tools. Once developed, these tools can be used to simulate the welding operations and to model the associated solid-state phase transformations the material undergoes. This will allow modeling the microstructure distribution and use of appropriate constitutive law for residual stresses calculation. Moreover, it enables to assess the effect of the welding process on fatigue crack behavior by numerical methods.

During welding, 13Cr-4Ni soft martensitic stainless steels experience different solid-state phase transformations.^[9] Figure 1 shows an example of a dilatometric curve of subsequent heating and cooling. Upon heating from room temperature, the martensite to austenite transformation is characterized by a volumetric contraction. It starts at temperature (A_{c1}), which depends on the heating rate. This transformation is diffusive and its kinetics is driven by numerous factors such as temperature, soaking time, chemical composition, and microstructural morphology. In some circumstances

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Fig. 1—Dilatation as a function of temperature during heating and cooling of 13Cr-4Ni soft martensitic stainless steel.

when facing very high heating rates, the austenitic transformation can exhibit massive transformation behavior. However, to the knowledge of the authors, this phenomenon has not yet been observed in the martensitic stainless steel under study and will not be distinctly considered.

The Leblond's model^[10] and the Johnson–Mehl–Avrami–Kolmogorov (JMAK)^[11–14] models can be used to simulate diffusive phase transformation kinetics. The latter is preferred in this study since it is based on reaction parameters that can be measured from experiments. Numerous studies^[15–18] use the JMAK model in the context of welding simulation. For example, Zhang *et al.*^[16–18] applied this model to estimate the austenite fraction formed during gas tungsten arc welding of 1005 mild carbon steel with good agreement with the experiments.

Upon cooling from an austenitic state, 13Cr-4Ni soft martensitic stainless steels undergo a martensitic transformation at the martensitic start temperature (M_s), regardless of the cooling rate. This transformation takes place within a temperature range at which the mobility of atoms in the iron matrix is negligible. Thus, the transformation occurs briskly when a large amount of atoms of the iron crystal lattice move from the austenitic face-centered cubic f.c.c. structure to a more stable b.c.t. structure by a shearing process.^[19] This transformation is athermal and is driven by several factors excluding time such as temperature, chemical composition, external stresses, *etc.* A volumetric expansion is associated with this transformation which induces compressive residual stresses.^[20–23] Most studies on residual stress calculation in materials subjected to martensitic transformation during welding found in the literature^[24–27] use the Koistinen–Marburger^[28] (K–M) relation to address the austenite to martensite phase transformation problem. Since then, subsequently derived models have been proposed within the same framework.^[29,30]

A welding process involves a high thermal energy input source bounded within a small volume that moves along with the arc during welding. This creates a thermal transient phenomenon characterized by a large temperature gradient. It therefore generates different thermal histories specific to the locations, giving rise to a

variety of metallurgical states, depending on whether the location is within the fusion zone (FZ), the heat-affected zone (HAZ), or elsewhere in the base metal. The thermal history of a particular region will therefore dictate its microstructural evolution over time. Material thermo-mechanical governing constitutive relations in a particular region are conditioned by its phases content. Hence, it becomes necessary to model the kinetics of phase transformations in order to adequately predict the formation of residual stresses. In the context of welding simulation, the objective of this paper is to establish the parameters of the austenite and martensite solid-state phase transformation models in 13Cr-4Ni soft martensitic stainless steel. For this purpose, homogeneous welding of UNS S41500 hot rolled martensitic stainless steel using E410NiMo weld metal is studied. Although hydraulic turbines are made of CA6NM cast martensitic stainless steel, all three steels have the same chemical composition and have very similar thermophysical properties. They also exhibit very similar dilatometric behavior. We then assume that these three steels behave thermally the same and also exhibit the same phase transformations.

II. MODELING SOLID-STATE PHASE TRANSFORMATION

A. Austenitic Transformation During Welding

The parameters governing austenitic transformation kinetics need to be determined for the aforementioned stainless steel. The development of the model and how the parameters are obtained from experimental measurements is presented next.

1. Modeling diffusive phase transformations

The JMAK theory^[11–14] is intended to model diffusive transformation kinetics. It consists of an equation describing the formed phase f , over time usually expressed as

$$f = 1 - e^{-k\tau^n}, \quad [1]$$

where k parameter characterizes the contribution of both nucleation and growth to the reaction rate and is dependent of the temperature T . τ is the time spent from the beginning of the transformation; *i.e.*, $\tau = t - t_0$. The JMAK exponent n , which is related to phase rate of nucleation and growth mechanisms, is a material constant. Since welding involves a thermal transient phenomenon, Eq. [1] must be further developed to integrate over the time the temperature changes. It is proposed for continuous heating rate to use the following form for the JMAK equation.^[16–18,31–34]

$$f = 1 - e^{-(k\tau)^n} \quad [2]$$

The Arrhenius function often used to model chemical reactions is used to take into account the evolution of the temperature during the thermal history. As such, Eq. [2] is modified as suggested by Zhang *et al.*^[16]:

$$f = 1 - e^{-\left[\sum_{i=1}^m k_0 e^{\frac{-E_a}{RT_i}} \Delta t_i\right]^n}, \quad [3]$$

where Δt_i and T_i are the i^{th} interval of time and temperature, respectively, m the total number of increments, k_0 is the chemical reaction pre-exponential factor of the Arrhenius function, E_a the phase transformation apparent activation energy, and R the universal gas constant. Variables E_a , k_0 and n must be determined from appropriate experiments to describe the transformation kinetics. It is then proposed^[16] to define s as follows:

$$s = \sum_{i=1}^m e^{\frac{-E_a}{RT_i}} \Delta t_i \quad [4]$$

which further simplifies Eq. [3] to

$$f = 1 - e^{-(k_0 s)^n} \quad [5]$$

2. Calibration of JMAK model parameters

The amount of formed phase can be evaluated by measuring thermophysical properties such as heat capacity, thermal expansion, or magnetization among others. In this study, the austenitic phase volume content is calculated from dilatometric measurements. A dilatometer is an equipment that allows recording the change of length of a sample over controlled temperature and time exposure. Since free thermal strain can generally be expressed by either a linear or a quadratic function, a disparity with this trend in the recorded strains is an indication that a phase transformation is ongoing. A lever rule can then be used to evaluate the phase transition. Figure 2 is an example of a dilatometer experiment showing a complete austenitization from a fully martensitic state from which the austenitic phase volume content, f_γ , can be extracted at each time during the experiment. The austenitic transformation on Figure 2 leads to the following equation:

$$f_\gamma = \frac{\varepsilon_{\alpha'} - \varepsilon}{\varepsilon_{\alpha'} - \varepsilon_\gamma}, \quad [6]$$

Fig. 2—Dilatation vs temperature during the continuous heating of a sample. The total strains needed for the evaluation of the austenitic phase volume content at a given temperature using the Eq. [6] are identified.

where ε is the total strain of the sample at time t . $\varepsilon_{\alpha'}$ and ε_γ are the thermal strains at the correspondent temperature T of the starting martensitic and ending austenitic phases, respectively.

The Eq. [5] can be rearranged in the following form:

$$\ln(-\ln(1-f_\gamma)) = n \ln s + n \ln k_0 \quad [7]$$

Therefore, from experimental data in which the evolution of the austenitic phase volume content f_γ over the temperature T and the time t is recorded, a plot of $\ln(-\ln(1-f_\gamma))$ against $\ln(s)$ would allow determining the values of k_0 and n .

3. Transformation activation energy from continuous heating rate experiments

In this work, the Kissinger method^[35] is used to determine the transformation apparent activation energy, E_a . This method has been proven effective to characterize the austenitization activation energy of steels from various continuous heating dilatometry experiments.^[33,36] The proposed method consists in determining the temperature at which the maximum reaction rate occurs for dilatometry experiments conducted with different continuous heating rates. Mittemeijer *et al.*,^[33,34] made the assumption that the phase volume content is determined by a state variable β . For the nonisothermal case, β relies on the integration over time of the transformation

$$\beta = \int_0^t k dt, \quad [8]$$

where k is expressed with the Arrhenius function:

$$k = k_0 e^{\frac{-E_a}{RT}} \quad [9]$$

The heating rate ϕ is given as

$$\phi = \frac{dT}{dt} \quad [10]$$

Using Eqs. [9] and [10], the Eq. [8] becomes

$$\beta = \frac{k_0}{\phi} \int_{T_0}^T e^{\frac{-E_a}{RT}} dT \quad [11]$$

An approximate solution for this exponential integral is given by Mittemeijer *et al.*^[34]:

$$\begin{aligned} \beta &= \frac{T^2}{\phi} \frac{R}{E_a} k \left\{ 1 - 2 \frac{RT}{E_a} + \dots \right\} \\ &\simeq \frac{T^2}{\phi} \frac{R}{E_a} k_0 e^{\frac{-E_a}{RT}} \end{aligned} \quad [12]$$

Considering that RT/E_a is small, only the first term of a developed series of the exponential function is considered. Since the maximum reaction rate is expected to occur after the same amount of transformation^[35] regardless of the heating rate, β is constant at the maximum transformation rate temperature T_{inf} . Eq. [12] is rearranged to give Eq. [13]:

$$\ln \frac{T_{\text{infl}}^2}{\phi} = \ln \frac{E_a}{Rk_0} + \frac{E_a}{RT_{\text{infl}}} + \ln \beta_{\text{infl}}, \quad [13]$$

$$= \frac{E_a}{RT_{\text{infl}}} + C$$

where T_{infl} is the sample inflection temperature recorded in Kelvin at which the maximum transformation rate occurs with the correspondent β_{infl} and C is a constant. The determination of the maximum transformation rate temperature, used in the Kissinger method, relies on the determination of the temperature at which the first derivative, df/dt is maximum, *i.e.*, when $d^2f/dt^2 = 0$. This is by definition the inflection point of the transformed fraction curve deduced from the continuous heating dilatometric curve.

This relation, firstly developed for homogeneous reactions, can also be applied to heterogeneous transformations with preferential nucleation sites. When applicable, a plot of $\ln \frac{T_{\text{infl}}^2}{\phi}$ against $\frac{1}{T_{\text{infl}}}$ leads to a linear relation between these two quantities. The slope of this curve is equal to the ratio E_a/R .

B. Martensitic Transformation During Welding

The modeling of the martensitic transformation occurring upon cooling of welded 13Cr-4Ni steel is now addressed. The following section proposes a review of few models and their experimental calibration steps.

1. Modeling diffusionless phase transformations

The martensitic transformation behavior has been extensively modeled through the Koistinen–Marburger (K–M) relation.^[28] This empirical relation is written as follows:

$$f_{\alpha'} = 1 - e^{-0.011(M_s - T)} \quad [14]$$

Equation [14] gives the martensitic phase volume content $f_{\alpha'}$ as a function of the temperature T and the martensitic transformation start temperature M_s . The constant -0.011 was obtained by the authors through the fitting of a series of experiments in which carbon steel samples were quenched at temperatures ranging from -79°C to room temperature. The remaining austenite fraction after the quench was measured by X-ray diffraction.

Van Bohemen (VB) *et al.*^[29] studied the influence of the chemical composition on the martensitic transformation. Adding more experimental data to the plain carbon steels experiments, the authors proposed to review published data on martensitic transformation and to develop an empirical expression for the K–M relation's parameters in terms of the chemical composition of steels. Following their work, Eq. [14] can be rewritten as

$$f_{\alpha'} = 1 - e^{-a_m(T_{\text{KM}} - T)}, \quad [15]$$

where T_{KM} is the theoretical martensitic start temperature and a_m is a transformation rate parameter which is calculated from a fit of martensitic transformation

data. The transformation rate is affected by the change in transformation driving force with temperature and the deformation energy related to the volume and shear strain accommodation.^[29] The determination of M_s generally relies on the measurement of the first measurable indications of transformation that occur when the martensitic transformation starts. On the other hand, the evaluation of T_{KM} has for purpose to model the transformation and is evaluated from an extension of the thermal expansion curve. It therefore misses the very first steps when the martensitic transformation gradually starts and its value is therefore below M_s (cf. Figure 4). However, for very small austenitic grain size, M_s can approximate T_{KM} . After the determination of coefficients for 19 different chemical compositions of steel, the following expressions were obtained:

$$T_{\text{KM}} = 462 - 273x_{\text{C}} - 26x_{\text{Mn}} - 16x_{\text{Ni}} - 13x_{\text{Cr}} - 30x_{\text{Mo}} \quad [16]$$

$$a_m = 0.0224 - 0.0107x_{\text{C}} - 0.0007x_{\text{Mn}} - 0.00005x_{\text{Ni}} - 0.00012x_{\text{Cr}} - 0.0001x_{\text{Mo}} \quad [17]$$

with T_{KM} and a_m expressed in $^\circ\text{C}$ and $^\circ\text{C}^{-1}$, respectively. The x_i variables refer to the weight percentages of elements i . The authors^[29] reported that the largest discrepancies between calculated values of a_m with Eq. [17] and values determined from transformation data fitting were found for steels with high concentrations of both nickel and chromium, which is the case for the 13Cr-4Ni steel studied in this paper. Moreover, these authors suggest, for their model to be applicable, to limit the carbon content to the range 0.3 to 1 pct. This range is however one order of magnitude beyond the acceptable level for the fore-mentioned chromium alloy used in this study, which chemical composition is shown in Table I.

Lee *et al.*^[30] proposed a new model for martensitic transformation. This model reads

$$f_{\alpha'} = 1 - e^{-K_{\text{LV}}(M_s - T)^{n_{\text{LV}}}}, \quad [18]$$

where K_{LV} and n_{LV} are material parameters. The authors used the equation developed in the study of Capdevila *et al.*^[37] for the determination of M_s . They also suggest the following relations for the determination of the model's parameters based on the chemical composition.

$$K_{\text{LV}} = 0.0231 - 0.0105x_{\text{C}} - 0.0017x_{\text{Ni}} + 0.0074x_{\text{Cr}} - 0.0193x_{\text{Mo}} \quad [19]$$

$$n_{\text{LV}} = 1.4304 - 1.1836x_{\text{C}} + 0.7527x_{\text{C}}^2 - 0.0258x_{\text{Ni}} - 0.0739x_{\text{Cr}} + 0.3108x_{\text{Mo}} \quad [20]$$

Table I. Chemical Composition of UNS S41500 Stainless Steel

Weight Percent	C	Mn	P	S	Si	Cr	Ni	Mo
ASTM ^[38]	< 0.05	0.50 to 1.00	< 0.030	< 0.030	< 0.60	11.5 to 14.0	3.5 to 5.5	0.50 to 1.00
S41500	0.034	0.68	0.018	0.001	0.44	12.7	4.0	0.57

Though, the usage of this model should be limited to plain carbon or low alloyed steels.

2. Calibration of martensitic transformation model parameters

The martensitic transformation models presented in this section are often used in the literature to conduct simulation on materials and processes for which the parameters are unknown. Insufficient attention to these particularities inevitably leads to misvaluation of the phases content and erroneous residual stress calculation. In this work, the martensitic phase volume content is experimentally evaluated by the analysis of dilatometric curves upon cooling. Figure 3 shows an example, where at a given temperature, the total strain ϵ is measured. Again, a lever rule is used, with the extrapolated values of the fully martensitic $\epsilon_{\alpha'}$ and fully austenitic ϵ_{γ} thermal strains. Equation [21] expresses the relation used for the evaluation of the sample martensite content :

$$f_{\alpha'} = \frac{\epsilon - \epsilon_{\gamma}}{\epsilon_{\alpha'} - \epsilon_{\gamma}} \quad [21]$$

Some discussed models use the theoretical martensitic transformation start temperature T_{KM} , which is determined with the onset method. Figure 4 shows an example of the application of the method that consists to find the intersection between the free thermal strain of the dilatometric curve portions before and after the start of the martensitic transformation. For this purpose, second-order polynomials are fitted on each described portion of the dilatometric curve. The intersection point of the two polynomials is the theoretical martensitic transformation temperature start.

III. MATERIAL AND EQUIPMENT

The material under study is a 13Cr-4Ni soft martensitic hot rolled stainless steel denominated as UNS S41500 in accordance with ASTM-A240/A240M-18^[38] delivered in 60-mm-thick plate. The carbon and phosphorus contents have been measured with a combustion infra-red detection technique. The other chemical constituent proportions have been measured by glow-discharge atomic emission spectrometry. The measured values are presented, together with standard requirements, in Table I.

Prior to the tests, the material has been fully austenitized at 1050 °C for 1 hour to allow microstructure homogenization and then furnace-cooled to room temperature. The microstructure obtained following the

Fig. 3—Dilatation vs temperature during the continuous cooling of a sample. The total strains needed for the evaluation of the martensitic phase volume content at a given temperature using the Eq. [21] are identified.

Fig. 4—Determination of the theoretical martensitic start temperature T_{KM} by the onset method.

heat treatment is fully martensitic with less than 1 pct of retained austenite as observed by Godin *et al.*^[39] Tubular samples of 10 mm length, 4 mm external, and 3.4 mm internal diameters are then prepared by wire electro-discharge machining. The sample longitudinal axis is aligned with the rolling direction. A small specimen thickness is intended to achieve high heating rates. These samples are meant to be used in a TA Instruments DIL805 A/D/TTM quenching dilatometer to characterize the austenite and the martensite transformation behaviors. The quenching dilatometer is set in alpha mode, which allows a more sensitive thermal expansion measurement through a linear variable differential transformer (LVDT). Sample heating is performed with an induction coil to allow fast heating capability, whereas

the cooling is controlled by the admission of argon around the sample.

IV. DETERMINATION OF EXPERIMENTAL HEATING AND COOLING RATES BY FE SIMULATION OF WELDING

At a particular location of a welded assembly, the maximum reached temperature and the rate at which it is attained are closely related to the weld procedure and the distance from the welding heating source path. Numerical simulation allows to determine the range of heating rates in these regions. This range will provide an experimental domain of heating rates that are meant to reproduce those encountered in the heat-affected zone (HAZ) of 13Cr-4Ni during the FCAW welding process. This diversity of thermal histories during heating generates various volume fractions of austenite that eventually, upon cooling, give rise to different amounts of martensite transformation, ultimately acting on the welding residual stress development. The dilatometric experiments are meant to be performed at least within these two heating rate limits to ensure that the austenitic transformation mechanisms involved during FCAW are correctly modeled by the JMAK model developed in this work.

A. Finite Element Model and Simulation Parameters

In order to define the rates controlling solid-state phase transformations during FCAW of 13Cr-4Ni steels, which is the weld procedure used for manufacturing hydraulic turbines, a finite element (FE) simulation of the welding process is performed. *In situ* temperatures resulting from a FCAW on a 13Cr-4Ni steel were measured in a previous study.^[40] A 456-mm-long $\times$ 254-mm-wide $\times$ 36.9-mm-thick UNS S41500 steel plate was prepared with a 60° and 12.7-mm-depth “V”-shaped groove to receive three FCAW beads of homogeneous weld metal E410NiMo. The first bead was performed along the full plate length, the second along the 2/3, and the third along 1/3 of the plate length. Twenty K-type thermocouples were installed on the plate, among which 18 were inserted and spot welded in flat end holes of different depths, to allow collecting temperature measurements in the vicinity of the HAZ. Figure 5 shows the overall dimensions of the plate and the weld groove, along with bead lengths. A FE mesh of this geometry was prepared. The FE model used to calculate the temperature distribution contains 297816 eight-node hexahedral elements. Both the welded region and the HAZ were meshed using 1-mm-side-length elements, whereas in the remaining of the plate, the element size was gradually coarsened as the distance from the weld root increased. The geometry of the beads has been previously determined from metallographic observations.^[40] A close view of the weld region is shown in Figure 6.

Thermal simulation of welding involves solving heat transfer equations for transient phenomena. Temporal discretization was done in several time increments in

Fig. 5—Geometry of the instrumented welded plate. The first bead is colored in blue, the second bead in red, and the third in green. Adapted from Ref. [40].

Fig. 6—Close view of the welded region of the FE mesh.

Fig. 7—Thermal conductivity of S41500 as a function of temperature.

order to integrate adequately the specific and latent heat contributions to the heat transfer, to account for the movement of the heat source and to capture the temperature dependence of material properties. The simulation has been conducted using a 0.01-s time step during welding and 1.0-s time step at the interpasses. The thermal properties of the material were updated at the current temperature at each time step. Figures 7 and 8 show the temperature dependence of thermal conductivity and specific heat of 13Cr-4Ni steel that were used.^[40] It is noteworthy that although the solid-state transformation kinetics was not taken into account in the FE simulation, both material properties,

conductivity and specific heat, did incorporate the effect of the latent heat of the transformations. These two properties may therefore be considered as apparent thermal conductivity and specific heat. The thermal conductivity has been measured up to 900 °C by Laser Flash Analysis (Netzsch LFA457) and the specific heat, up to 1050 °C by Differential Scanning Calorimetry (Netzsch DSC402). These maximal temperatures were limited by the range of applicability of these devices. Databases of steels of comparable chemical compositions, available in the literature^[41] have been used for estimating these two material properties at temperatures exceeding the equipment capabilities and are considered the same for heating and cooling.

The welding simulation also considered radiative and convective heat losses to the environment. A convective heat transfer coefficient of 8 W/(m² °C) was used. As the plate is heated, oxidation occurs on the outer surfaces which modifies the emissivity coefficient ψ . Figure 9, adapted from Bouffard,^[42] shows the dependence between the emissivity coefficient of a 13Cr-4Ni cast steel (CA6NM) as a function of the temperature. It was

Fig. 8—Specific heat of S41500 as a function of temperature.

Fig. 9—Emissivity coefficient of CA6NM as a function of temperature. Adapted from Ref. [42].

observed that emissivity coefficient remained constant during cooling.

The heat input of the welding torch is modeled in the FE simulation with the model proposed by Goldak *et al.*^[43] This model consists of two semi-ellipsoids, at the front and rear of the source, approximating the geometry of the weld pool. Figure 10 shows a schematic of Goldak's model with its geometrical parameters: a and b for the lateral extent of the ellipsoid and its penetration depth, whereas c_f and c_r are the front and rear extents of the semi-ellipsoids, respectively. Equations [22] and [23] give the power density within the double-ellipsoid.^[43]

$$q_f = \frac{6\sqrt{3}f_f Q}{ab c_f \pi \sqrt{\pi}} e^{-3x^2/a^2} e^{-3y^2/b^2} e^{-3(z-vt)^2/c_f^2} \quad [22]$$

$$q_r = \frac{6\sqrt{3}f_r Q}{ab c_r \pi \sqrt{\pi}} e^{-3x^2/a^2} e^{-3y^2/b^2} e^{-3(z-vt)^2/c_r^2} \quad [23]$$

The power density within the double-ellipsoid is assumed to follow a Gaussian distribution that is unevenly shared between the front and the rear ellipsoids. This uneven distribution is triggered with parameters f_f and f_r , representing the fraction of the total heat input distributed in the front and in the rear ellipsoids, respectively. In Eqs. [22] and [23], Q is the total power input, which is equal to

$$Q = \eta IV, \quad [24]$$

where V and I are the source voltage and current, respectively, and η , the source efficiency. Table II presents the list of parameters defining the heat source of the model studied in this work. The source voltage, current, and speed (v) were set during the experiment, whereas the source efficiency was calibrated from temperature measurements. According to the welding procedure for E410NiMo weld metal, preheat and

Fig. 10—Heat source model used for FCAW welding. The power distribution and front/rear ellipsoids are shown. Adapted from Ref. [43].

Table II. Goldak's Welding Source Model Parameters for FCAW of S41500 Steel Plate

a , mm	b , mm	c_f , mm	c_r , mm	f_f	f_r	η	V , V	I , A	v , mm/s
6.6	6.3	12.0	24.0	0.6	1.4	0.80	28.0	292.9	6.5

interpass temperatures were set to 100 °C and 130 °C, respectively.

B. Finite Element Results

Post-processing the welding simulation results described in Section IV–A allows to investigate the temperature and the heating rate distributions in the plate. Austenitic start temperature (A_{c1}) of 13Cr-4Ni stainless steel at 620 °C and 660 °C have been reported after experiment performed at continuous heating rates of 1 °C/min and 5 °C/min, respectively.^[9] It has also been reported that austenitic grain coarsening occurs during expositions to temperatures above 1050 °C.^[39] To be consistent with these observations, the experimental temperature range of interest in which the austenitic transformation from martensitic phase in the weld HAZ will start and finish, encompassing the studied heating rates, should extend from 600 °C to 1000 °C. For this study, it is therefore assumed that if the maximum temperature reached at a location is below 600 °C, the austenitic transformation does not take place. It is also assumed that a complete austenitic transformation was achieved if the maximum temperature is above 1000 °C. The nodal heating rates at each time step of the simulation were calculated using linear interpolation. Figure 11 shows the maximum heating rates reached during welding and the region for which the temperature range described above was achieved. These correspond to the 600 °C and 1000 °C isotherms that are drawn on the figure with solid black and solid white lines. Thus, the minimum heating rate needed for a region to achieve 600 °C is 60 °C/s. Besides, the maximum heating rate that allows reaching a maximum of 1000 °C is 218 °C/s.

Upon cooling, the temperature gradient and the cooling rate in the welded region decrease. As the martensitic start temperature (M_s) is reached, at around 270 °C, the cooling rate is between -6 °C/s and -10 °C/s. These heating and cooling rates are used to guide the selection of the dilatometer experimental test parameters for an adequate determination of the austenitic transformation apparent activation energy and the JMAK parameters. Figure 12 shows a schematic of the proposed dilatometric thermal history imposed to the tubular dilatometric samples previously described in Section III. All specimens were subjected to a continuous heating segment from a starting temperature of 25 °C to an austenitization temperature, $T_\gamma = 1000$ °C, within a vacuum atmosphere. The continuous heating rate ϕ is set to 1, 3, 10, 30, 100, 200, 300 °C/s to assess the kinetics of austenitic transformation. A dwell time, $t_d = 10$ s is then applied at the austenitization

Fig. 11—Simulated maximum heating rate in the welded region. The white and the black solid lines locate the 600 °C and 1000 °C isotherms, respectively.

Fig. 12—Thermal cycle for dilatometric tests. The heating rate ϕ is a variable of this test program.

temperature to ensure the temperature within the sample to homogenize and the dilatometer temperature control device to stabilize. Then, the cooling phase at a constant rate, $v = -10$ °C/s is initiated. Cooling is controlled by the admission of argon in the chamber and around the sample until room temperature is reached within the sample.

V. RESULTS AND DISCUSSION

A. Determination of the Austenitic Transformation Activation Energy in 13Cr-4Ni Steel

The austenitic phase volume content during continuous heating is evaluated from dilatometric experiments using the approach described by Eq. [6] and Figure 2. Figure 13 displays a typical result for each of the tested heating rates. As expected, the increase of the heating

Fig. 13—Kinetics of austenitic transformation for different continuous heating rates. Only 10 pct of data are displayed on the curves.

rate is reflected in an increase in the austenitic transformation start temperature, which leads to a shift of the curves towards higher temperatures. It can be seen in Figure 13 that for a heating rate of 1 °C/s, the transformation rate is significantly different at the end of the transformation. According to the study of Bojack *et al.*^[36] on 13Cr6NiMo steel, this could result from a change of transformation mechanism during the austenitic transformation attributed to a mechanism of nickel and molybdenum partitioning. Thus, enriched regions of an austenite-stabilizing element would have a lower activation energy and would therefore transform first, which would lead to a two-stage complete austenitization transformation. At the last stage, diffusivity of Ni and Mn increases and the remaining amount of martensite transforms while nitrides and carbides dissolve. As the heating rate is increased, the occurrence of the second stage is avoided, which is mostly the case in welding.

Inflection temperatures (T_{infl}) are determined from the experimental austenitic phase volume content curves. A third-degree polynomial fit, $F(T)$, is performed on the central portion for each transformation curves, where maximum transformation is expected. First and second temperature derivatives are computed. The inflection temperature is the temperature for which the second derivative is zero, *i.e.*, $d^2F(T)/dT^2 = 0$. This procedure avoids the error that would result from a numerical derivation of somewhat scattered experimental points. Figure 14 shows a typical test performed at 10 °C/s and the first derivative of the experimental results. The experimental data and the polynomial fit are plotted in Figure 14(a), whereas their derivatives are plotted in (b). The experimental data derivatives are obtained by central derivative approximations. The dashed line indicates the calculated inflection temperature. The average austenitic phase volume content of all dilatometric tests at T_{infl} is of 0.46, with a standard deviation of 0.03, which shows that the transformation is almost halfway completed when its rate reaches a maximum.

Inflection temperatures thus determined are further used to estimate the activation energy of austenitic transformation, which is related to the decomposition of

Fig. 14—Determination of T_{infl} at a heating rate of $\phi = 10^\circ\text{C/s}$: (a) Polynomial fitting of the central portion of the austenitic phase volume content curve, (b) Derivative of the polynomial determined in (a) plotted against temperature.

Fig. 15—Kissinger's plot used for activation energy evaluation.

a fully martensitic matrix. For this purpose, the effective heating rates recorded during the transformation are used for the application of the Kissinger method. According to Eq. [13], the activation energy can be extracted from the calculation of the slope of a linear regression plot of $\ln(T_{infl}^2/\phi)$ against $1/T_{infl}$, as shown in Figure 15. The inflection temperatures are more scattered for tests performed at heating rates equal to and larger than 100 °C/s. It is noteworthy that these tests also presented more discrepancy between the programmed and the effective heating rates, suggesting that

the experimental device might be used at its limit capacity.

A resulting activation energy of 509.9 kJ/mol is calculated, which agrees with the value previously obtained value by Bojack *et al.*^[36] on a similar fully martensitic steel however tested at lower heating rates. Hereafter, this austenitic transformation activation energy is used to compute the JMAK model parameters.

B. Determination of JMAK Equation Parameters

Once the activation energy is known, the JMAK equation parameters, n and k_0 , can be calculated from the same set of experiments. For each sample, the summation in Eq. [5] is performed as the test progresses. This enables to trace a plot of $\ln(-\ln(1-f_\gamma))$ against $\ln(s)$. Since the data points of the tests are not evenly distributed as the transformation occurs, the data are re-sampled to the nearest 5 pct increment of f_γ to prevent the regression to be biased by high data density regions. Figure 16 displays the selected data points for all tests along with the linear regression equation that was used to calculate the JMAK model parameters. A coefficient of determination of the linear regression $R^2=0.83$ is obtained. Parameters n and k_0 , respectively, 0.66 and 4.04×10^{24} ($\ln(K_0) = 37.19$), are calculated from the slope and the intersect of the linear regression.

Figure 17 presents the variation of the austenitic phase volume content according to the JMAK model with E_a , n , and k_0 previously determined as input to the model. For comparison, experimental data are also shown in the figure. Predicted values of f_γ are in good agreement with the experiments. A maximum difference of 19 pct in the volume fraction is observed between the model prediction and the experiment for the 100 °C/s test sample. Also, a significant difference arises for the 1 °C/s test sample as the volume fraction of formed austenite exceeds 80 pct. As discussed previously in Section V-A, a change of transformation mechanism is the likely cause of the observed reduction in the transformation rate. Nevertheless, the mean error on the austenite phase volume content, all tests being considered, stands at 7.9 pct with a 4.3 pct standard

Fig. 16—Plot of $\ln(-\ln(1-f_\gamma))$ against $\ln(s)$ allowing JMAK parameters calculation.

deviation. Better agreements are observed for the 1 °C/s and 300 °C/s test samples with mean errors of 3.9 and 7.5 pct, respectively. The model is effective in replicating the shift of the curves towards higher temperatures as the heating rate increases, but tends to underestimate the transformation start temperature at higher rates. However, this is thought to be of a limited impact on the overall formed austenite and the welding residual stress calculation since for higher rates, both the maximum reached temperature and the amount of austenite formed are higher.

C. Influence of Time Step on Austenitic Transformation Model

It has been discussed in Section IV-A that the simulation of thermal transient processes involves discretizing the time of simulation in increments for each of which the temperature distribution is computed. This allows updating the material specific heat and thermal conductivity and thus accounting for any temperature change during the time increment. It also allows updating the position of the power source and the calculation of the power density within the double-ellipsoid, to account for the evolution of the model boundary conditions and finally the progress of undergoing temperature and time-dependent phase transformations. The prediction of the austenitic phase volume content when applying the JMAK model to FE thermal computation in welding depends on the duration of the time increments between each calculation step. The influence of the duration of the time increment, Δt (see Eq. [3]), on the prediction of the amount of austenite formed during heating at a rate of 200 °C/s was assessed. The time increments used between calculation steps were ranging from 0.1 to 0.0001 seconds for the case of an explicit integration. The difference is calculated in comparison with a 0.0001-s time length reference. A 0.01-s time step increment appears to be a fair compromise between precision and computation cost as it would generate a maximum difference of 1.19 pct in the resulting austenitic phase volume content, which seems acceptable considering the error of the model against experimental measurements. Using a smaller time step

Fig. 17—Comparison between measured and calculated austenite volume fraction.

would unjustifiably raise the FE welding simulation time while using larger time step would lead to inaccurate predictions.

D. Determination of Martensitic transformation Parameters

The martensitic transformation behavior of 13Cr-4Ni martensitic stainless steel is studied using the same set of dilatometric measurements as described in Section IV-B. For each sample, the deformation upon cooling is used to calibrate the parameters of the martensitic transformation models. At first, the theoretical martensitic transformation start temperature T_{KM} , which is determined using the method described in Section II-B-2 is defined. Averaging the 21 dilatometric tests, a value of 254.2 °C is obtained. This value is significantly different than the value obtained with Eq. [16], proposed by van Bohemen *et al.* (VB),^[29] using the chemical composition of 13Cr-4Ni steel (cf. Table I), which gives 188.8 °C.

The same trend is noticed for the determination of M_s for the Lee *et al.* (L-V) model.^[30] The value of M_s is measured at 263.2 °C while the suggested relation^[37] ended at 548.6 °C. Many empirical equations for the evaluation of M_s from chemical compositions can be found in the literature. In their work, Lippold *et al.*^[44] review some of them which are all nonapplicable at predicting M_s for the steel used in this work. Unfortunately, these relations do not take into account the specific metallurgical interactions that arise in steels. For instance, in 13Cr-4Ni soft martensitic stainless steel, partitioning of austenitic stabilizing elements and carbides precipitation have been reported.^[45–47] These two phenomena, which depend on temperature and time exposure, will locally modify the chemical composition and nucleation sites. Unfortunately, these particularities cannot be taken into account when using empirical relations, especially for high alloy steel. Hence, it is recommended to measure experimentally M_s , based on the specific austenite condition and the cooling rate encountered, when using a martensitic transformation model.

The rate parameter a_m of the VB model Eq. [17] is optimized to fit at best the behavior of the martensitic transformation. Calculation of the rate value according to the relation based on chemical compositions (Eq. [17]) with the chemical composition of Table I results in a value of 0.020 °C⁻¹. To achieve a better agreement with experimental data, a rate parameter is calculated with a least squares method. A transformation rate of 0.0267 C⁻¹ is obtained, which allows minimizing the mean of differences of transformed martensite between Eq. [15] and the experimental results to 1.77 pct.

The parameters of the L-V model in Eq. [18] are firstly determined with the use of Eqs. [19] and [20] according to the chemical composition of Table I. The values obtained for K_{LV} and n_{LV} are 0.0989 and 0.526, respectively. However, a least squares method leads to a better agreement with the experimental data. With the

Fig. 18—Comparison of measured and calculated martensitic phase volume content for different models.

experimentally determined M_s of 263.2 °C and values of 0.0029 for K_{LV} and 1.486 for n_{LV} , a mean of differences of 3.64 pct between Eq. [18] and the experimental results is obtained.

Figure 18 shows, in comparison with experimental values, the reliability of the Koistinen–Marburger (K–M) relation Eq. [14] along with the van Bohemen’s (VB) Eq. [15] and the Lee (L–V) Eq. [18] relations. The first and the last equations use the martensitic start temperature measured in this work. The VB curve uses T_{KM} and the a_m as determined from Eq. [16] and [17]. At last, the calibrated curves for VB and L–V models use the measured T_{KM} , M_s , and fitted parameters determined in the present study. As it can be seen, the VB relation cannot describe properly the expansion associated with the martensitic transformation of 13Cr-4Ni with the parameters determined from chemical compositions. However, when used with the parameters fitted in this work, it has a better agreement with the experimental results.

VI. CONCLUSION

In this study, solid-state phase transformation models parameters for 13Cr-4Ni soft martensitic stainless steel in the context of welding were experimentally determined by nonisothermal dilatometry experiments. The tests were conducted in the temperature rate range experienced in the weld HAZ, where the austenitic transformation is assumed to be only partial. This variation of austenitic phase volume content will influence the amount of martensitic transformation upon cooling and the resulting welding residual stress formation. To determine the heating and cooling rates affecting the transformation kinetics in that particular region of the weld, a FE simulation of the FCAW welding process has been performed. It was found that the region of interest is subject to heating rates varying from 60 °C/s to 218 °C/s and cooling rates from –6 °C/s to –10 °C/s.

Following the methodology in References 33, 35, and 36, the activation energy of austenitic transformation from a fully martensitic state has been estimated to

509.9 kJ/mol. This allowed determining the kinetic parameters of the JMAK model for the austenitic transformation evolution, $\ln(k_0)$ and n , to 37.19 and 0.66, respectively. The capabilities of the model have been evaluated in the context of a simulation with various time increments at a heating rate of 200 °C/s. It showed that a 0.01-s discretization enables sufficient precision for welding computation with a maximum difference of 1.19 pct of austenitic phase volume content.

The martensitic transformation behavior has been studied with the widely used Koistinen–Marburger (K–M)^[28] empirical relation and few derived models. The methodology used to determine $T_{KM} = 254.2^\circ\text{C}$ and $a_m = 0.0267^\circ\text{C}^{-1}$ allows to satisfactorily represent the experimental data with the van Bohemen (VB)^[29] model. The martensitic transformation takes place at a low temperature for which the material mechanical properties, such as the Young's modulus and the yield stress are almost at their maximum value, giving rise to a significant amount of residual stress. These residual stresses are also an influencing parameter of the martensitic transformation expansion behavior. This phenomenon known as transformation induced plasticity (TRIP) will be addressed in a future work.

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