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Conference Paper · November 2015

DOI: 10.1002/9781119223399.ch44

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DEVELOPMENT OF PROCESSING MAPS FOR INTERCRITICAL ANNEALING USING THE PHASE FIELD APPROACH

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Keywords: Intercritical annealing; Phase field model; Dual-phase steel

Abstract

A phase field model has been developed to simulate microstructure evolution during intercritical annealing. The model describes ferrite recrystallization, intercritical austenite formation and decomposition. In particular, the overlap of ferrite recrystallization and austenite formation for sufficiently fast line speeds has been taken into account in the model. The model has been benchmarked and validated with experimental data for a DP600 steel. Further, simulations have been performed with a systematic variation of processing parameters, i.e. line speed and intercritical holding temperature. Processing maps have been constructed based on these simulations providing martensite fraction and ferrite grain size as a function of line speed and intercritical temperature for the investigated DP600 steel. This study demonstrates the phase field approach as a promising tool to develop through-process models for advanced high strength steels.

Introduction

Intercritical annealing has been widely used in the steel industry to manufacture advanced high strength steels (AHSS), e.g. dual-phase (DP) steels and transformation-induced-plasticity (TRIP) steels which provide a better combination of strength and formability than conventional high strength steels. Continuous annealing or hot dip galvanizing lines are mainly used for intercritical annealing to manufacture sheets of DP and TRIP steels. Rapid heating and cooling techniques are commonly used in modern continuous annealing lines with a maximum line speed of about 1000 m/min [1]. For sufficiently fast line speeds, ferrite recrystallization of the cold-rolled steels may not be completed before austenite formation such that both microstructure phenomena proceed concurrently and interact with each other. Experimental results have demonstrated that the overlap of austenite formation and ferrite recrystallization can lead to different microstructural morphologies than in cases without overlap [2], [3]. Thus, it is of great significance to develop a through-process model for intercritical annealing which is able to take the potential overlap of ferrite recrystallization and austenite formation into account.

In the past decades, phase field models have been used to describe microstructure evolution in steels, e.g. the austenite-to-ferrite transformation during cooling. As a meso-scale approach, the phase field method has a number of advantages over the other meso-scale models, e.g. cellular automata (CA) and Monte Carlo (MC) methods. One merit of phase field models is that there is no need to track interfaces explicitly, which enables it to model microstructures of complex morphology, e.g. dendrite formation. Moreover, phase field models do not require scaling between numerical physical time scales that is an issue for CA and MC methods.

In this paper, a through-process model for a intercritical annealing based on a phase field approach [4] is used to systematically investigate the effects of line speed and intercritical holding temperature on the volume fraction of hard phases, i.e. martensite and/or

martensite/bainite constituents, for a DP600 steel (Fe-0.11wt%C-1.86wt%Mn-0.34wt%Cr-0.16wt%Si).

Methodology

Basics of the multi-phase field model

In this work, a multi-phase field model originally developed by Steinbach [5] is employed where each grain *i* is defined by a unique phase-field parameter ϕ_i. The value of ϕ_i is equal to 1 inside grain *i* and 0 outside grain *i*. Within the interface of width η, ϕ_i changes continuously from 0 to 1. The phase field parameter ϕ_i represents the local fraction of each grain such that the interface consists of a mixture of grains with the constraint of $\Sigma\phi_i = 1$. The temporal evolution of each field variable is described by the superposition of the pair-wise interaction with its neighboring grains:

$$\frac{d\phi_i}{dt} = \sum_{i \neq j} m_{ij} \left\{ \sigma_{ij} \left[\left(\phi_j \nabla^2 \phi_i - \phi_i \nabla^2 \phi_j \right) + \frac{\pi^2}{2\eta^2} \left(\phi_i - \phi_j \right) \right] + \frac{\pi}{\eta} \sqrt{\phi_i \phi_j} \Delta G_{ij} \right\} \tag{1}$$

where m_{ij} is the interfacial mobility that is assumed to obey an Arrhenius relationship, σ_{ij} is the interfacial energy and ΔG_{ij} is the driving force for interface migration which can be either stored energy for recrystallization or the difference of chemical potentials for phase transformation.
Further, the phase field model is coupled with the diffusion equation for carbon to account for long-range diffusion during phase transformation:

$$\frac{\partial C}{\partial t} = \nabla \left[\sum_i D_i \phi_i \nabla C_i \right] \tag{2}$$

Here D_i is the carbon diffusivity in grain *i* and *C* is the local carbon concentration which, in the interface region, is the sum of carbon concentrations C_i weighted with the phase field parameter:

$$C = \sum_i \phi_i C_i \tag{3}$$

The long-range diffusion of substitutional elements is neglected in this work.

Sub-model development

The schematic of a thermal path for intercritical annealing is shown in Figure 1. The through-process model [4] only considers the microstructure evolution until cooling to 475 °C which is the temperature for galvanization in a hot-dip galvanizing line. Sub-models for ferrite recrystallization, austenite formation and austenite to ferrite transformation were developed by assuming the formation of hard phases, either martensite or martensite/bainite aggregates, is negligible above 475 °C [4]. Further, it is assumed that the remaining austenite at 475 °C will transform fully to either martensite or martensite/bainite aggregates during subsequent cooling or galvanizing. Thus, the austenite fraction at 475 °C is taken as the fraction of hard phases in the final microstructure. The physical parameters in the sub-model of ferrite recrystallization are a critical value of stored energy above which nucleation can take place, as well as the energy and

mobility of ferrite-grain boundaries. The parameters in the sub-models of austenite formation and austenite to ferrite transformation are austenite nuclei density in pearlite, austenite nucleation rate at ferrite-grain boundaries, the energy and mobility of austenite-pearlite and austenite-ferrite interfacial boundaries. The solute drag effect of the main substitutional element, i.e. Mn, on the austenite-ferrite interfaces is taken into account. The details of the model including parameter selection and validation can be found in our previous work [4].

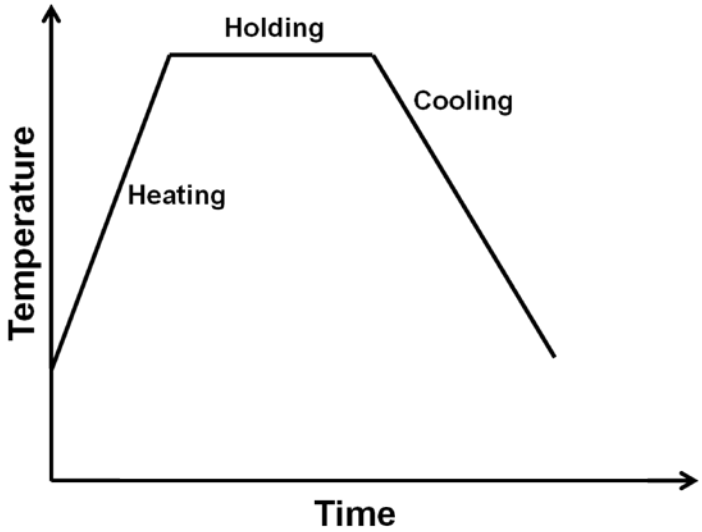

Figure 1. Schematic of a thermal path for intercritical annealing.

Simulation conditions

In this work, simulations of intercritical annealing were performed with various temperature-time paths by adjusting the line speed (80-1000 m/min) and holding temperature (760-800 °C). The initial microstructure of the steel for intercritical annealing consists of pearlite (15 vol.%) and ferrite in a cold-rolled state. The corresponding initial microstructure used in the simulations was constructed using Voronoi tessellation [4]. A simulation domain of 100 μm×100 μm with a grid size of 0.1 μm and periodic boundary conditions was used. The interface thickness η was set to 0.5 μm.

Results and Discussion

Recrystallization kinetics, phase transformation kinetics, ferrite grain sizes, microstructural morphology and carbon distribution, are obtained from the phase field simulations. As an example, Figure 2 shows the fractions of recrystallized ferrite, non-recrystallized ferrite and austenite for a holding temperature of 780 °C and two line speeds, i.e. 80 m/min and 1000 m/min. In the case of 80 m/min where the heating rate is about 2 °C/s, ferrite recrystallization is finished prior to austenite formation and there is no overlap of ferrite recrystallization and austenite formation (Figure 2(a)). Austenite grains nucleate at both pearlite/ferrite interfaces and recrystallized ferrite grain boundaries (Figure 3(a)). In the case of 1000 m/min where the heating rate is about 30 °C/s, ferrite recrystallization is not yet complete when austenite formation starts and proceeds until the austenite fraction reaches about 0.3. In this case, some of the austenite grains originally formed at the boundaries of deformed ferrite grains will remain inside recrystallized ferrite grains after ferrite recrystallizing fronts move across them (Figure 3(b)).

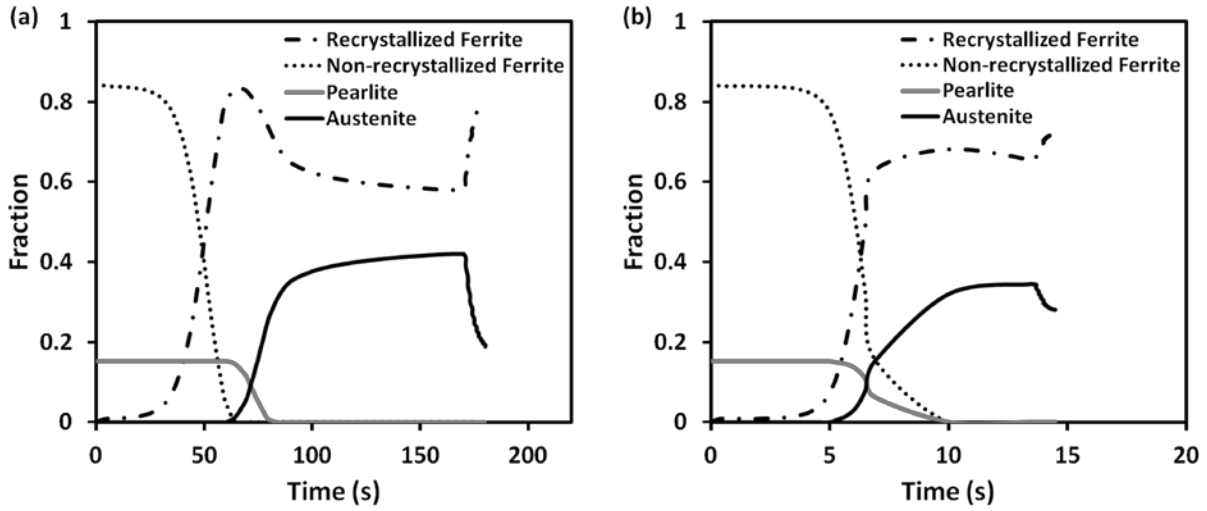

Figure 2. Simulated phase fractions (holding temperature: 780°C; line speed: (a) 80 m/min; (b) 1000m/min).

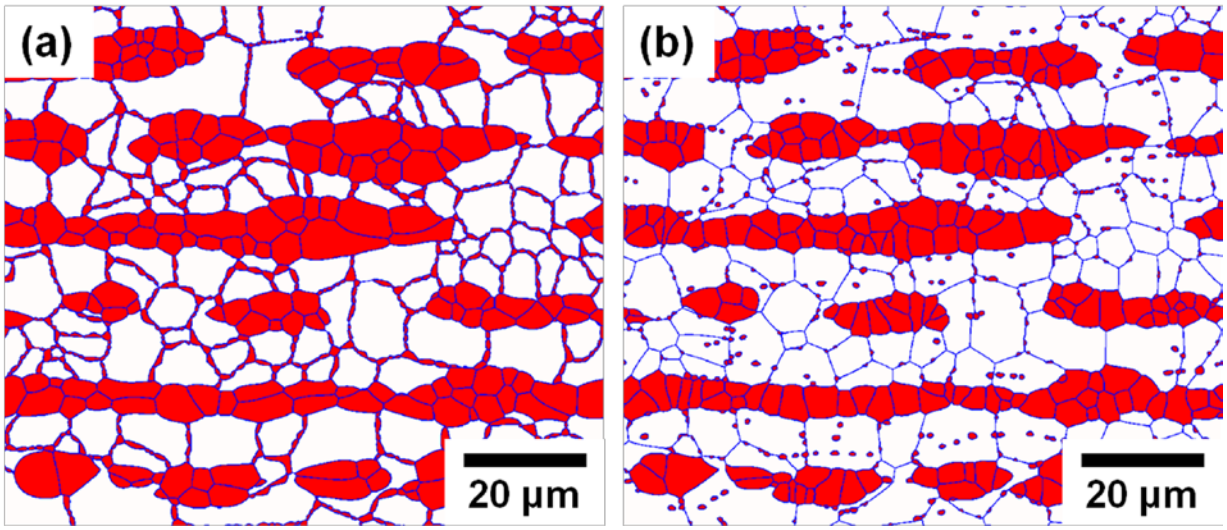

Figure 3. Simulated microstructures (dark: austenite; white: ferrite) after holding at 780°C for two line speeds: (a) 80 m/min; (b) 1000 m/min.

The final microstructures and carbon distribution after cooling are shown in Figure 4. In both cases, the austenite grains originally formed at ferrite grain boundaries have shrunk to sizes below 1 μm. For the lower line speed (80 m/min), austenite grains formed in the prior pearlite regions also transforms partially back to ferrite such that the austenite bands tend to break up. The carbon concentration near the austenite-ferrite interfaces is much higher than that in the center of austenite grains due to carbon rejection from growing ferrite. This can lead to the formation of bainite enveloped by a rim of martensite during galvanizing and the subsequent cooling, as experimentally observed [6].

Based on the simulation results for all the investigated holding temperatures and line speeds, processing maps are constructed showing the fraction of austenite after holding and the fraction of hard phases in the final microstructure as a function of line speed and holding temperature (Figure 5 and Figure 6). The fraction of austenite formed after holding increases with holding temperature and decreases with line speed (Figure 5(a)). The fraction of hard phases in the final microstructure increases, however, with both holding temperature and line speed as increasing line speed and associated faster cooling reduce the degree of intercritical austenite

decomposition. The average ferrite grain size in the final microstructure is similar (7.3±0.5 μm) for all simulated processing conditions.

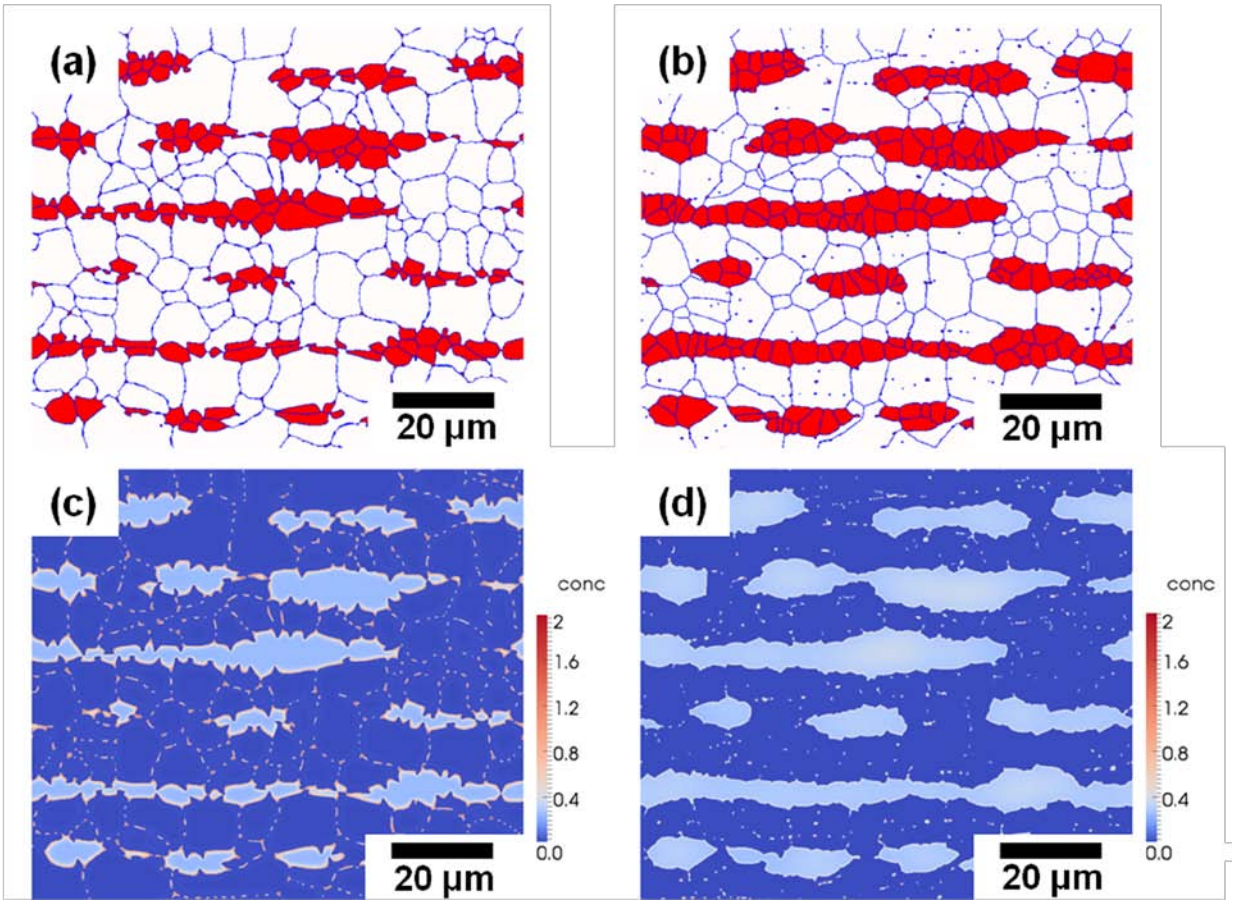

Figure 4. Microstructures (dark: austenite; white: ferrite) and carbon distributions in the simulated final microstructures for a holding temperature of 780°C and line speeds of 80 m/min (a, c) and 1000m/min (b, d).

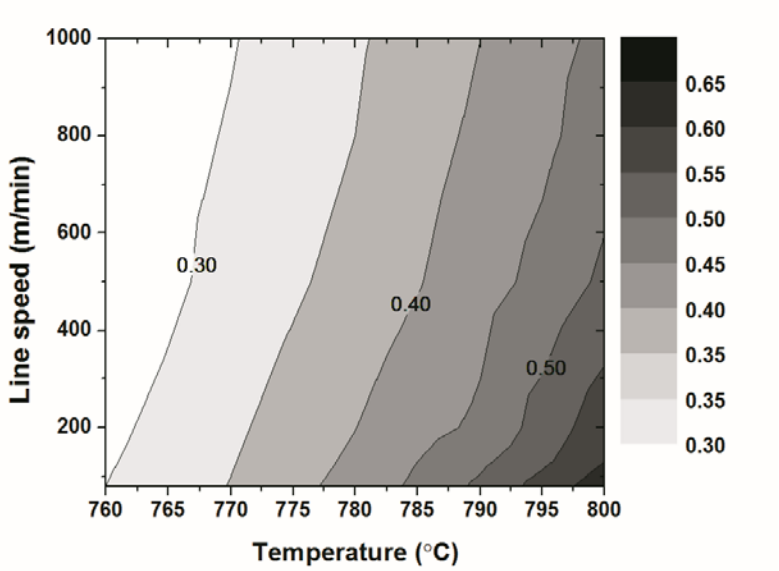

Figure 5. Simulated austenite fraction after holding as a function of line speed and holding temperature.

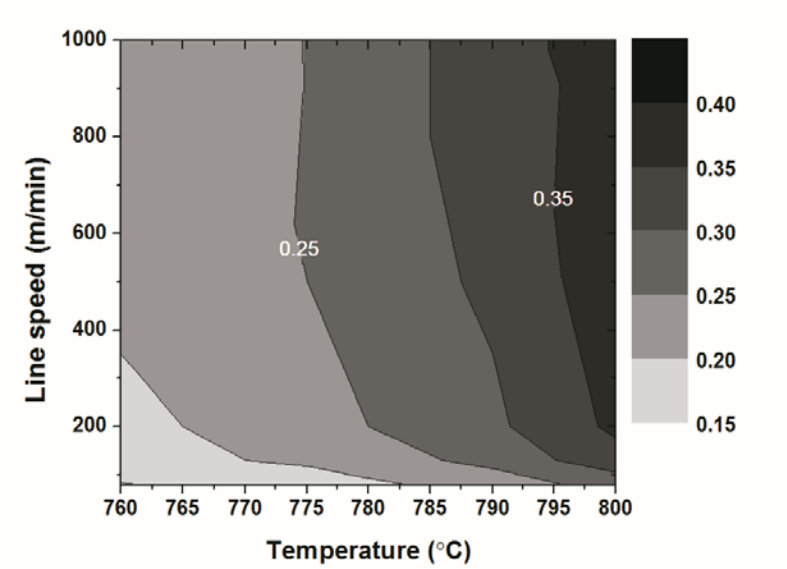

Figure 6. Simulated fraction of hard phases (martensite and/or martensite/bainite) after cooling as a function of line speed and holding temperature.

Conclusions

Processing maps have been constructed based on phase field simulations providing intercritical austenite and martensite/bainite fractions as a function of line speed and intercritical temperature for the investigated DP600 steel. This study demonstrates the phase field approach as a promising tool to develop through-process models for advanced high strength steels.

Acknowledgement

The authors are grateful to the Natural Sciences and Engineering Research Council of Canada (NSERC) and ArcelorMittal Dofasco Inc. for their financial support. They thank M. Kulakov for providing experimental data and W.J. Poole for many stimulating discussions.

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