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ScienceDirect
Acta Materialia 59 (2011) 1716–1724
Acta MATERIALIA
www.elsevier.com/locate/actamat

Solute trapping-free massive transformation at absolute stability

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Received 12 October 2010; received in revised form 8 November 2010; accepted 14 November 2010
Available online 14 December 2010

Abstract

Directional transformation of a hypo-peritectic Fe–17.5 at.% Co alloy was studied. Two consecutive phase transformations—solidification (liquid to delta ferrite) and solid-state transformation (δ ferrite to γ austenite)—were observed and compared with theory. In all experiments, the solidification front was planar and in the steady-state, and therefore produced a homogeneous parent phase for the following δ–γ transformation. Depending on the growth conditions, γ transformed from δ as cells or as a plane front. The cell tip radius decreased with growth rate from $V = 1$–$5\ \mu m\ s^{-1}$. At higher velocities, between 7 and $10\ \mu m\ s^{-1}$, the δ/γ interface morphology became planar. In order to explain this morphological transition, volume diffusion-controlled plane front growth and dendrite growth theory was applied. Good agreement was obtained between theory and experiments. It is concluded that plane front stabilization with increasing velocity is due to absolute stability, with a concentration spike at the transformation front. In the steady-state, this leads to composition invariance, typical for massive transformation. Computed interface velocities for quenching in heat treatment, which can be as high as several centimeters per second, show that, in certain cases, the controlling mechanism of massive transformation is steady-state plane front growth with a narrow concentration spike and not complete solute trapping.

Keywords: Massive transformation; Local equilibrium; Solute trapping; Absolute stability; Directional transformation

1. Introduction

Solid-state transformations are very important in material processing as, in combination with solidification, they largely determine the properties of the final product. Understanding the mechanisms of these transformations is therefore important. One such phase change is the massive transformation characterized by a composition invariant change in the crystal structure with a highly mobile interface. Much work has been published in this area, mostly using isothermal or continuous cooling experiments [1,2]. The high displacement rates of this transformation (typically millimeters per second to centimeters per second) make experiments difficult to interpret, owing to the uncertainty in the interface temperature associated with the latent heat release [2]. Directional growth experiments are more interesting, as the interface velocity V and temperature gradient G can be set, and the interface temperature T_i measured. Borgenstam and Hillert [3] and Zurob et al. [4], used directional growth experiments in a concentration gradient which allowed them to quantify the γ–α transformation in Fe–Ni and Fe–C–Mn.

Vandyoussefi et al. [5] and Kurz and Lima [6,7] used directional growth in a temperature gradient (Bridgman technique) to study the δ−γ transformation in Fe–Ni and Fe–Cr, respectively. In Fig. 1, the differences between isothermal and directional growth experiments are presented, with the aim of showing the advantage of studying solid-state transformations in a temperature gradient. Fig. 1a represents a schematic drawing of a massive transformation product forming from a grain boundary in the center outwards. In this case, the growth direction and the heat flux point in the same direction. At the interface, the rejection of solute and heat sets the interface temperature, which cannot be measured owing to the small boundary layers. On the other side, a directional growth experiment is characterized by a unidirectional temperature field with

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doi:10.1016/j.actamat.2010.11.038

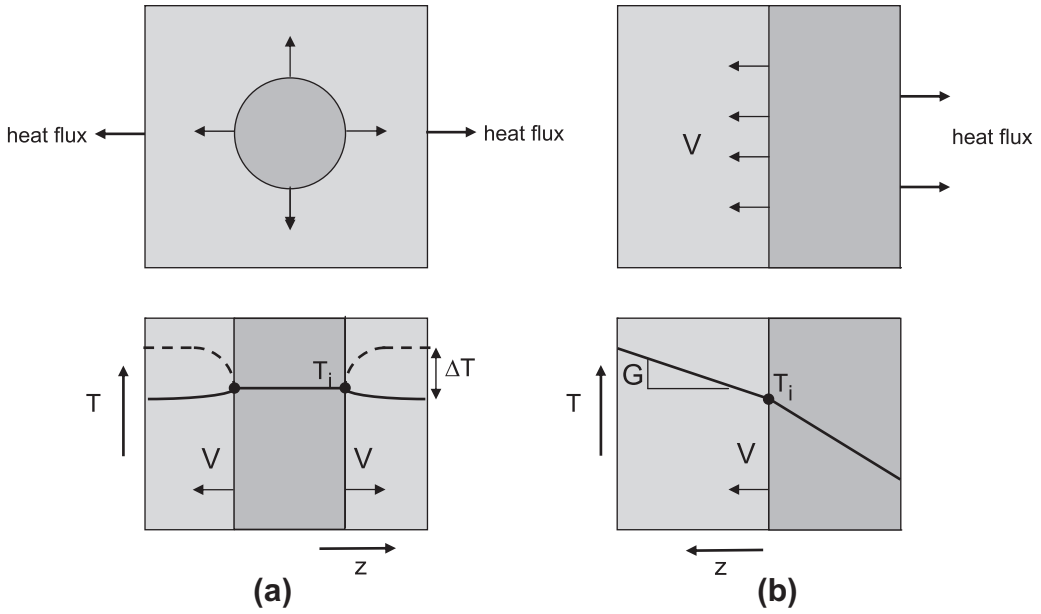

Fig. 1. Solid-state transformation conditions: (a) isothermal transformation with nucleation in the undercooled bulk and outwards growth of the new (dark) phase as a function of the undercooling ΔT; the rejection of solute and heat defines the interface temperature T_i, which cannot be measured; (b) directional transformation with nucleation at the cold outer surface; interface velocity V, temperature gradient G and T_i can be measured.

a positive temperature gradient (Fig. 1b). The interface moves here opposite to the heat flux, and its velocity can be set to specific values which are only limited by the condition of unidirectional heat flux. Owing to this configuration, the interface velocity, the temperature gradient and the interface temperature can be easily measured. From this brief presentation of the two techniques, it should become clear that directional growth experiments have a potential for future solid-state transformation studies.

Fe alloys, which form the basis of the industrially important steels, show a rich variety of solid-state transformations on cooling. This is also the case for peritectic Fe–Co alloys. On the iron-rich side of this system, bcc δ-ferrite ($\alpha_{\text{high T}}$) solidifies on cooling and transforms subsequently to fcc γ-austenite, which in turn transforms back to α-ferrite. The γ−α transformation process is affected by the microstructure of austenite formed during the δ–γ transformation. Very little information is available on the latter transformation. This is attributed to the fact that the growth behavior of γ is not easy to observe, since it occurs close to the solidus temperature of the alloy, and the microstructure of the δ−γ transformation is masked by the consecutive γ−α transformation at lower temperature. Furthermore, owing to the small concentration difference between the phases, metallographic evidence of the transformation product is very difficult to obtain.

Vandyoussefi et al. [5] observed cellular interface morphologies of various shapes and scales in directionally transformed Fe–3%Ni alloys. According to their findings, the cell spacing decreased when the velocity was increased from 5 to 25 μm s^{-1}. At 30 μm s^{-1}, it coarsened and showed the typical form of low-amplitude interface deformations. At velocities >40 μm s^{-1}, signs of a transformation front could no longer be detected. It was argued that the boundary layer at the moving interface was too thin, and the concentration spike too small to become visible in the metallographic etching. Kurz and Lima [6,7] observed similar morphological transitions in Fe–8%Cr. In both cases (Fe–Ni and Fe–Cr), the variation in the dendrite/cell spacing λ was compared with theoretical predictions of the Ivantsov–Marginal stability (I–MS) model [8,9], and it was concluded that the observations could be explained by the well-known growth theory of dendrites, cells and plane front [10]. The disappearance of the transformation microstructure at high velocity was explained by the concept of absolute stability [11].

At absolute stability, the solute diffusion field at the interface becomes localized with respect to the instability wavelength λ. Effects of large solute Péclet number (ratio of diffusion coefficient to interface velocity $D/V \ll \lambda$) and capillarity dominate, leading to a morphologically stable planar transformation front. (Oscillatory instabilities, which do not form in alloys with very small solidus–liquidus intervals, are neglected here [12].) When the transformation produces a plane front in steady-state, the parent phase and the product phase have the same composition.

As the critical velocity required for absolute stability V_{ab} is proportional to the solute diffusion coefficient, the typical velocity for absolute stability in solidification is very high and very difficult to produce [13–15]. An absolutely stable planar growth front has, to the knowledge of the authors, never been shown very clearly, either in solidification or in solid-state transformation experiments. In solid-state transformations, the limit of absolute stability is orders of magnitude smaller and can be easily reached.