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Acta Materialia 55 (2007) 4359–4368

Acta MATERIALIA

www.elsevier.com/locate/actamat

The austenite to ferrite transformation of Fe–Ni under the influence of a uniaxially applied tensile stress

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Received 22 November 2006; received in revised form 29 March 2007; accepted 31 March 2007
Available online 25 May 2007

Abstract

Differential dilatometry has been employed to study the austenite (γ)-ferrite (α) massive phase transformation of Fe-3.1 at.% Ni upon cooling under the influence of an applied constant uniaxial tensile stress. The applied stress level was chosen to be below the yield stress of the alloy. It was found that an extra length change in the specimens occurs during the transformation as a result of transformation-induced plasticity. The local plastic deformation induced by transformation plasticity results in an anisotropic volume change of the specimen in the longitudinal direction. A phase transformation model, involving site saturation, interface-controlled continuous growth and an appropriate impingement correction, has been employed to extract the velocity of the γ/α interface. The observed scatter in the interface velocity decreases with increasing applied stress due to relaxation of misfit deformation energy by transformation-induced inhomogeneous plastic deformation. The interface velocity remains almost constant during the whole transformation.

Keywords: Austenite–ferrite phase transformation; Dilatometry; Tensile stress; Interface migration; Thermodynamics

1. Introduction

The austenite (γ) to ferrite (α) phase transformation in iron-based alloys (steels) is of great importance from both a fundamental and a technological point of view [1]. Understanding the $\gamma \rightarrow \alpha$ transformation behaviour is essential to control the final microstructure of the α phase. The microstructure of a material is decisive for its mechanical properties. Research has been carried out on the influence of uniaxial stress on (the kinetics of) the austenite to pearlite transformation (e.g. [2]), the austenite to bainite transformation (e.g. [3]) and the austenite to martensite transformation (e.g. [4]). However, according to the present authors' knowledge, no publication exists on the influence of applied uniaxial tensile stress, in the elastic range, on the austenite to ferrite massive transformation.

This paper reports on the quantitative influence of an applied uniaxial tensile stress, below the yield stress of the alloy, on the start temperature, length (and volume) change evolution, interface velocity and driving force of the massive austenite to ferrite transformation in Fe-3.1 at.% Ni alloy.

2. Phase transformation kinetics

The Fe–Ni partial equilibrium phase diagram relevant to the present study is shown in Fig. 1 [5]. This phase diagram was determined employing very long annealing times (days) in order to assure that equilibrium was reached (e.g. [6]). The investigated Fe-3.1 at.% Ni alloy reaches the single phase region upon cooling (at $20\ \mathrm{K\ min^{-1}}$) at about 960 K (see the dashed line in Fig. 1). The T_0 curve, indicating the temperature where the Gibbs energies of the product (α) and parent (γ) phases of the same composition are equal, has been shown as a function of Ni content in Fig. 1 as well.

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1359-6454/$30.00
doi:10.1016/j.actamat.2007.03.032

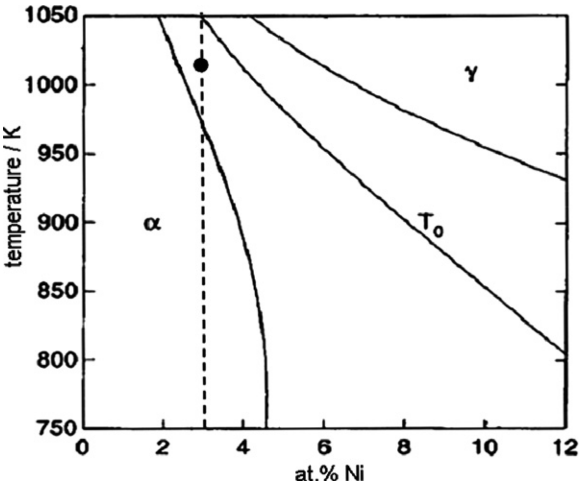

Fig. 1. Partial equilibrium phase diagram for Fe–Ni showing the T_0 line (temperature corresponding to equal Gibbs energies of the metastable product and parent phases (α and γ) of the same composition). The dashed line represents the alloy composition of Fe-3.1 at.% Ni. The measured start temperature of transformation of $\gamma \rightarrow \alpha$ transformation under an applied stress of 0.005 MPa is given by the circular point.

2.1. Transformation model

A general procedure for the simulation of phase transformation kinetics on the basis of nucleation, growth and impingement mechanisms has been given in Ref. [7]. The first step of this procedure involves the calculation of the volume of all growing particles, assuming that the grains never stop growing and that new grains hypothetically nucleate also in the transformed material (i.e. at this stage, "hard impingement" is ignored). This volume is called the extended volume (V_e), and the extended volume divided by the total specimen volume (V_0) is called the extended transformed fraction x_e. V_e is given by

$$V_e = g\left(\int v(T,f)\,\mathrm{d}t\right)^3 \tag{1}$$

where g is a geometrical factor (which is 1 for cubical growth and $4\pi/3$ for spherical growth), v is the interface velocity, which depends on the temperature, T, and the transformed fraction, f, which has to be integrated over time, t. The interface velocity is considered to be proportional with the product of the interface mobility (M) and the driving force ($-\Delta G_{\alpha\gamma}$) [8]:

$$v_\alpha = M(T)[-\Delta G_{\alpha\gamma}(T,f_\alpha)] \tag{2a}$$

The interface mobility exhibits an Arrhenius-like temperature dependence

$$M(T) = M_0 \exp(-Q/RT) \tag{2b}$$

where M_0 is the pre-exponential factor and Q is the activation energy. In this work data of an intrinsic, thermally activated mobility involving only the atomic rearrangements in the migrating interface are used deliberately in order to be able to evaluate, knowing the growth velocity and the chemical Gibbs energy difference between ferrite and austenite, the influence of elastic and plastic accommodation energies and of the interface energy. The intrinsic mobility data used have been estimated in Ref. [9] from grain growth and re-crystallization experiments of pure iron yielding $M_0 = 4.9 \times 10^3\ \mathrm{m\,s^{-1}\,mol\,J^{-1}}$, and $Q = 1.47 \times 10^5\ \mathrm{J\,mol^{-1}}$ [9]. The driving force ($-\Delta G_{\alpha\gamma}$) in Eq. (2a) is given by

$$\Delta G_{\alpha\gamma}(T,f_\alpha) = \Delta G^{\mathrm{chem}}_{\alpha\gamma}(T) + [\Delta G^{\mathrm{def}}_{\alpha\gamma}(f_\alpha) + \Delta G^{\mathrm{int}}_{\alpha\gamma}(f_\alpha)] \tag{3}$$

where $\Delta G^{\mathrm{chem}}_{\alpha\gamma}$ is the molar chemical Gibbs energy difference between ferrite and austenite, which depends on the temperature T. $\Delta G^{\mathrm{int}}_{\alpha\gamma}$ is the molar Gibbs energy of the γ/α interface, and $\Delta G^{\mathrm{def}}_{\alpha\gamma}$ is the summation of elastic and plastic molar accommodation energies resulting from the misfit between the ferrite and austenite phases. The driving force, $-\Delta G_{\alpha\gamma}$, consists of a negative term, $\Delta G^{\mathrm{chem}}_{\alpha\gamma}$, which favours the transformation, and two positive terms, $\Delta G^{\mathrm{def}}_{\alpha\gamma}$ and $\Delta G^{\mathrm{int}}_{\alpha\gamma}$, which obstruct the transformation. The chemical driving force depends on temperature and not on the fraction transformed, because the transformation is partitionless. Both $\Delta G^{\mathrm{int}}_{\alpha\gamma}$ and $\Delta G^{\mathrm{def}}_{\alpha\gamma}$ depend primarily on the fraction transformed, f_α (and not directly on temperature).

In the next step, the extended transformed fraction is corrected for "hard impingement" of the growing particles. The impingement correction adopted here corresponds to an intermediate case of ideally periodically and ideally randomly dispersed growing particles [10]:

$$f = \tanh(x_e) \tag{4}$$

with $x_e = V_e/V_0$ as the extended fraction. Thus, the kinetic model is characterized by the specific models adopted for (i) the nucleation process, (ii) the growth process and (iii) the (hard) impingement.

Nucleation of ferrite can take place during cooling in the two phase region before significant decomposition can take place. Thus, for the subsequent transformation occurring upon continuous cooling the nucleation can be supposed to have been completed at the very start of growth, i.e. "site saturation" is adopted. Further, since the transformation is partitionless, interface-controlled growth can be adopted. The interface velocity then can be derived from the rate of transformation, $\mathrm{d}f_\alpha/\mathrm{d}t$, the transformed fraction, f_α, and the grain size, $2\bar{r}_\alpha$, of the fully transformed ferrite phase:

$$\mathrm{d}f_\alpha/\mathrm{d}t = 3(1-f_\alpha^2)\arctan h^{\frac{2}{3}}(f_\alpha)\bar{r}_\alpha^{-1} v_\alpha(T,f) \tag{5}$$

3. Experimental details

3.1. Specimen preparation

Bulk high-purity Fe (99.98 wt.%) and Ni (99.99 wt.%) were used for the preparation of the alloy. The purity of both Fe and Ni has been indicated by the composition data (determined by inductive coupled plasma-optical emission spectrometry (ICP-OES)) in Table 1. Melting of appropriate