Materials Science and Engineering A 452–453 (2007) 244–253

MATERIALS SCIENCE & ENGINEERING A

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Analysis of flow behaviour of an aluminium containing austenitic steel

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Received 14 September 2005; received in revised form 17 October 2006; accepted 18 October 2006

Abstract

Experimental true stress–true strain data, at both ambient and elevated temperatures, of an Fe–Ni–Cr–Al alloy in solution treated as well as aged conditions have been analysed using different flow relationships. Ludwigson relationship provides the best fit of the data for all the conditions investigated, though all the relationships, except Voce, describe the flow behaviour well at a high temperature of 673 K. The transition in macroscopic flow behaviour of the alloy with strain, in solution treated condition, can be correlated with the transition in dislocation mechanism from planar slip in low strain regime to slip plus deformation twinning in the high strain regime. Although ageing does not appear to alter the macroscopic flow behaviour, it causes considerable change in flow parameters of the Ludwigson relationship and substructural evolution. The flow data of the aged alloys fitted according to Ludwigson model not only yield a unique set of flow parameters for each ageing condition but also exhibit a systematic trend with ageing time. The transition in macroscopic flow behaviour of the alloy with strain, in aged conditions, can be correlated with a change in dislocation mechanism from dislocation–precipitate interaction at lower strains to dislocation–dislocation interaction at higher strains leading to formation of a dense dislocation tangled networks in the matrix regions surrounding the precipitates.

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Keywords: Flow curves; Deformation twins; Flow relationships; Dislocation-precipitate interaction

1. Introduction

Knowledge of strain hardening behaviour of alloys is a key to the understanding of their forming characteristics. Analysis of strain hardening beaviour of metallic materials is facilitated by the flow curve obtained from a tensile/compression test. The flow curve signifies macroscopic plastic deformation behaviour under a definite structure state and deformation condition. Several flow relationships such as Hollomon [1], Ludwik [2], Voce [3], Swift [4] and Ludwigson [5] have been proposed (Table 1) in the past in order to establish an analytical relation between true stress (flow stress, σ) and true strain (ε). Yet, description of plastic flow behaviour of several metals and alloys is frequently carried out using either the Holloman ($\sigma = K_{\mathrm{H}}\varepsilon^{n_{\mathrm{H}}}$) [1] or the Ludwick ($\sigma = \sigma_0 + K_{\mathrm{L}}\varepsilon^{n_{\mathrm{L}}}$) [2] relation. Ludwigson [5] has demonstrated that the flow behaviour of face centered cubic (fcc) metals and alloys having low stacking fault energy, could not, however, be adequately described either by the Holloman or Ludwick equation and hence an alternative relation has been suggested, which is given by

$$\sigma = K_1\varepsilon_{\mathrm{p}}^{n_1} + \exp(K_2 + n_2\varepsilon_{\mathrm{p}}) \tag{1}$$

where σ is the true stress, ε_{p} the true plastic strain, K_1 and n_1 are the strength coefficient and strain hardening exponent, respectively, and K_2 and n_2 have been introduced as additional constants. It was suggested [5] that the additional term exp ($K_2 + n_2\varepsilon_{\mathrm{p}}$) in Eq. (1) accounts for the deviation from the Holloman equation at lower strain levels.

Attempts have been made earlier to describe the true stress–true plastic strain data of HSLA steels [6] and stable austenitic stainless steels [7,8] using various flow relationships (Table 1). Sivaprasad et al. [7] and Choudary et al. [8] have reported that while both the Ludwigson and the Voce models describe flow data of austenitic stainless steels reasonably well, the latter model is more accurate in fitting the data at higher temperatures.

The Fe–Ni–Cr–Al (FNCA) alloy considered for the present investigation is a stable austenitic alloy and is distinct from the 300 series austenitic stainless steels in two aspects, i.e., the former contains lower chromium (10–14 wt.%) and higher aluminium (to the tune of 5 wt.%) as compared to the latter. 300 series stainless steels contain chromium in the range 17–25 wt.%

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0921-5093/$ – see front matter © 2006 Published by Elsevier B.V.
doi:10.1016/j.msea.2006.10.141

Table 1
Flow relationships relating true stress, σ, and true plastic strain, ε

S. no.	Flow relationship
(1)	Hollomon: $\sigma = K_H \varepsilon^{n_H}$
(2)	Ludwik: $\sigma = \sigma_0 + K_L \varepsilon^{n_L}$
(3)	Swift: $\varepsilon = \varepsilon_0 + K_S \sigma^{n_S}$
(4)	Voce: $\sigma = \sigma_s - K_V \exp(n_v \varepsilon)$
(5)	Ludwigson: $\sigma = K_1 \varepsilon^{n_1} + \exp(K_2) \exp(n_2 \varepsilon)$

and aluminium <0.5 wt.%. The presence of higher aluminium renders FNCA alloys precipitation hardenable. The present study is aimed at: (i) identifying an appropriate flow equation (Table 1, [1–5]) that provides the best fit of true stress–true plastic strain data of the alloy in different microstructural conditions (i.e., solution treated as well as aged conditions) and (ii) investigating the effect of ageing time (precipitate size) on various parameters in the flow equation (that provides the best fit) and substructure evolution with accumulation of strain.

2. Experimental procedure

The chemical composition (wt.%) of the alloy investigated was Fe–21Ni–14Cr–4.5Al–0.12Zr–0.02Y–0.025C. The details of alloy melting and ingot processing were given elsewhere [9]. Specimen blanks of dimensions 10 mm × 10 mm × 70 mm cut from the hot rolled plate were solution treated at 1473 K for 1 h and then water quenched. The specimen blanks were then aged at 973 K for 5, 15 and 50 h. Tensile specimens of 25 mm gauge length and 4 mm gauge diameter were machined from these heat-treated specimen blanks. Tensile tests at ambient as well as elevated temperatures were conducted on Instron 1185 universal testing machine at a cross-head speed of 1 mm/min. Microstructural examination of the samples interrupted at different plastic strain levels as well as those strained to fracture was carried out using scanning and transmission electron microscopy. Details pertaining to preparation of TEM specimens and electrolyte used were described elsewhere [9].

3. Results

3.1. Initial microstructure

Fig. 1 is an optical micrograph of the alloy in solution treated (1473 K/1 h) condition depicting equiaxed grain structure and annealing twins with an average grain size of ~120 μm. The alloy in solution treated condition essentially exhibits austenite phase with a few precipitates of zirconium carbide (of size 5 μm) distributed randomly throughout the matrix.

Microstructures of the alloy in peak (973 K/5 h) and overaged (973 K/50 h) conditions are illustrated in Fig. 2. Precipitates with plate-like morphology are seen distributed throughout the austenite matrix following ageing. These precipitates were identified using X-ray and electron diffraction techniques as ordered β-NiAl with CsCl type crystal structure and they follow Kurdjmov–Sachs orientation relationship with the matrix. Microstructural aspects in detail are published elsewhere [9].

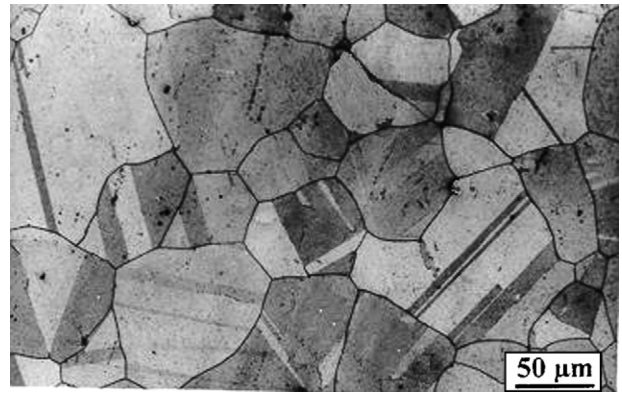

Fig. 1. Optical micrograph of the Fe–Ni–Cr–Al alloy in solution treated condition.

3.2. True stress–true strain data

The engineering stress–strain curves of the alloy for different heat treatment conditions have revealed that UTS and fracture stress are distinct for the alloy in solution treated and overaged (50 h) conditions, while they essentially coincide for the alloy in peak (5 h) and moderately overaged (15 h) conditions. This indicates that fracture in peak (5 h) and moderately overaged (15 h)

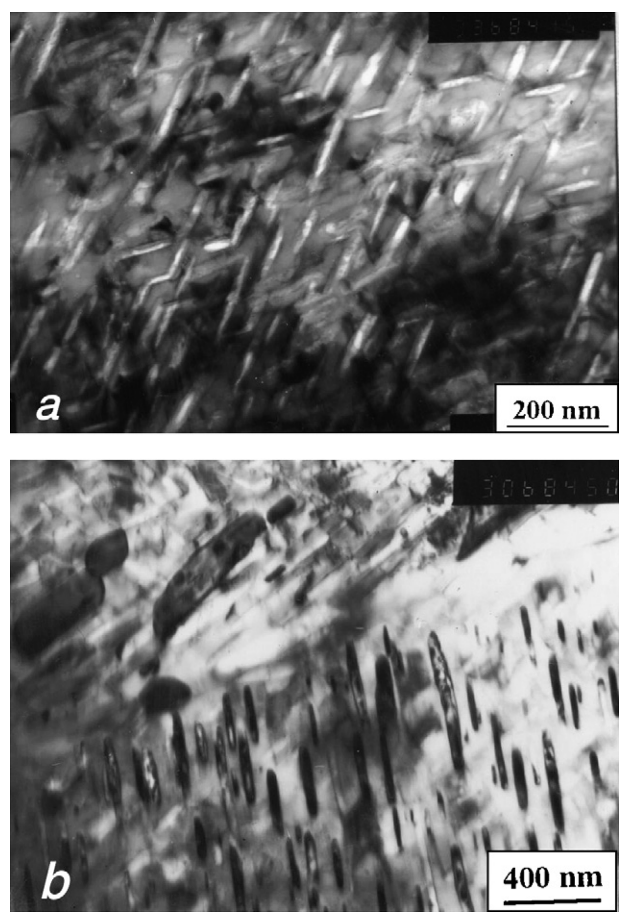

Fig. 2. Transmission electron micrographs revealing morphology of NiAl precipitates in the Fe–Ni–Cr–Al alloy in (a) peak (973 K/5 h) and (b) overaged condition (973 K/50 h).

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