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Deformation-induced internal stresses in multiphase titanium aluminide alloys

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Abstract

Long-range internal back stresses developed during room temperature tensile deformation of titanium aluminide alloys have been determined. Dip tests were implemented during the strain-controlled tensile deformation and the resulting sample deformation was monitored in the relaxation regime of the machine. The internal stress was determined as the critical stress at which the inelastic sample relaxation is reversed, i.e. going from the tensile direction to the compression direction. The investigation involves a wide range of alloy compositions with a corresponding variation in the strength properties. For the alloys investigated, the internal stress is about 80% of the yield stress. The mechanical tests were coupled with electron microscopy examination of the defect structure in order to assess the strain accommodation occurring during deformation. Possible sources for the built up internal stresses are discussed.

Keywords: Titanium aluminides; Internal stresses; Strain accommodation; Stress relaxation experiments; High-resolution electron microscopy

1. Introduction

A commonly held concept in crystal plasticity is the subdivision of the yield stress σ of a material into internal and effective stress parts according to [1]

$$\sigma = \sigma_\mu + \sigma^*(\dot{\varepsilon}, T) \tag{1}$$

The internal (or athermal) stress σ_μ results from glide obstacles with long-range stress fields, which can only be overcome mechanically. Thus, σ_μ shields part of the total applied stress σ and is almost independent of the deformation temperature, apart from a small variation in the shear modulus with temperature [1]. Prime examples of long-range (or athermal) glide obstacles are the grain and phase boundaries occurring in multiphase assemblies. The effective (or thermal) stress part σ^* arises from glide obstacles with short-range stress fields, which can be overcome with the aid of thermal activation. Thus, σ^* depends on strain rate $\dot{\varepsilon}$ and temperature T. Typical examples of short-range obstacles are impurity-related defects or monoatomic jogs dragged behind moving screw dislocations. In terms of the thermally activated dislocation glide model Eq. (1) can be extended to [2]

$$\sigma = \sigma_\mu + \frac{M_T}{V}[\Delta F^* + kT\ln(\dot{\varepsilon}/\dot{\varepsilon}_0)] \tag{2}$$

The quantity M_T (=3.06) is the Taylor factor and k is the Boltzmann constant. The quantity $\dot{\varepsilon}_0$ is a constant reference strain rate, involving the Burgers vector, the mobile dislocation density, the attempt frequency of the dislocations and the slip path of the dislocations after a successful activation [2]. $V = ldb$ is the activation volume of the thermally activated process for overcoming the short-range obstacles. V is the product of the obstacle distance l, the obstacle diameter d and the Burgers vector b. Hence, V can be considered as the number of atoms that have to be coherently thermally activated for overcoming the short-range obstacles by the dislocations [2]. The quantity ΔF^* is the free energy of the thermally activated process and describes the total energy for overcoming the

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short-range obstacles at a given stress and temperature. The activation parameters V and ΔF^* depend mainly on the alloy's composition and deformation temperature [3–5]. The values measured for two-phase ($\alpha_2 + \gamma$) alloys at room temperature are [3]

$$V = (1.6 - 3) \times 10^{-18}\ \text{mm}^{-3} = (70 - 130)b^3 \text{ and}$$
$$\Delta F^* = 1.3\ \text{eV}$$

Within this model, the net stress σ^* to produce slip is the difference between the applied stress σ and σ_μ. In order to separate the internal and effective stresses, several experimental approaches have been proposed, which were recently reviewed by Kruml et al. [6]. In the present study, deformation-induced internal stresses within TiAl alloys were determined by the incremental unloading technique, which was implemented during constant strain rate tensile tests. For TiAl alloys, knowledge about the relative magnitude of σ_μ and σ^* is important because these materials exhibit strong plastic anisotropies and are prone to cleavage fracture. Upon deformation, high long-range internal stresses may develop, which could exceed the cohesion strength of the material. Internal stresses are also important for the evolution of the microstructure during thermo-mechanical treatments, as recovery and recrystallization are triggered by heterogeneities in the deformed state.

2. Alloys and characterization

The compositions of the alloys investigated are listed in Table 1. The alloys were produced by triple vacuum arc remelting, hot extrusion at about 1250 °C and different annealing procedures [7]. The alloys had been characterized by scanning-probe techniques [7]. The thermo-mechanical treatments essentially resulted in a banded duplex microstructure that is composed of equiaxed grains and remnant lamellae. The constitution depends on alloy composition and basically reflects the trends that have been described in several reviews [8–11], i.e. the content of $\alpha_2(Ti_3Al)$ phase increases with decreasing Al concentration. The constitution of the Al-lean alloy nos. 1–6 is more complex, as they contain significant amounts of β/B2 phase, with a body-centred cubic structure, and of an orthorhombic phase (oP4), with a B19 structure. The B19 structure can be described as having a hexagonal superstructure (hP8) of the $D0_{19}$ structure of $\alpha_2(Ti_3Al)$. A subsequent paper [12] describes this alloy constitution in more detail. Blanks cut from the extruded alloys were machined into tensile specimens with the axis parallel to the extrusion direction, and were 24.5 mm in gauge length and 3.5 mm in diameter.

Complementary to the scanning-surface view, deformed samples of selected alloys were examined by transmission electron microscopy (TEM) utilizing conventional and high-resolution techniques. The TEM specimens were prepared by the standard method of jet electropolishing 2.3 mm diameter discs [7]. In situ heating experiments were performed inside the TEM on alloy no. 8. The samples used had been compressed at room temperature to a plastic strain of $\varepsilon = 3\%$.

3. Mechanical testing

For the deformation experiments, a servo-hydraulic closed-loop machine MTS 810 (MTS System Cooperation) was used. All experiments were performed in the tensile mode at room temperature in air with a strain rate of $\dot{\varepsilon} = 2.38 \times 10^{-5}\ \text{s}^{-1}$. The total (elastic plus plastic) sample deformation was measured by an extensometer over the entire gauge length and used as a feedback parameter. The data was digitally recorded utilizing the MTS TESTSTAR software. The internal stress was monitored by the relaxation characteristics observed after incremental unloading. These two elements of the tests will be considered in more detail.

3.1. Stress relaxation

The strain components involved in a constant strain rate test are illustrated in Fig. 1. Stress relaxation is the decay of stress over time when a constant strain rate test is stopped. The specimen continues to deform under the action of thermal activation, thereby expanding the load train. In other words, plastic sample deformation Δl_S^p gradually substitutes for the elastic deformations of the sample Δl_S^e and the machine Δl_M^e. Thus, apart from the thermodynamic factors controlling the sample plasticity, the kinetics of

Table 1
Composition and constitution of the alloys investigated.

No.	Symbol	Composition (at.%)	Main phase composition
1		Ti–42Al–5Nb–1.6 Ga–0.2C–0.2B	$\alpha_2(Ti_3Al)$, γ(TiAl), β/B2, B19
2	●	Ti–44.4Al	$\alpha_2(Ti_3Al)$, γ(TiAl), β/B2,
3	▼	Ti–45Al–5Nb–2Mo	$\alpha_2(Ti_3Al)$, γ(TiAl), β/B2, B19
4	∇	Ti–45Al–1.5Nb–1Mn–1Cr–0.2Si–0.8B	$\alpha_2(Ti_3Al)$, γ(TiAl), β/B2
5	△	Ti–45Al–1.5Nb–1Mn–1Cr–0.2Si–0.2B	$\alpha_2(Ti_3Al)$, γ(TiAl), β/B2
6	*	Ti–45.6Al–7.7Nb–0.2C	$\alpha_2(Ti_3Al)$, γ(TiAl), β/B2, B19
7	□	Ti–46.6Al	$\alpha_2(Ti_3Al)$, γ(TiAl)
8		Ti–48Al–2Cr	$\alpha_2(Ti_3Al)$, γ(TiAl)
9	○	Ti–49.1Al	γ(TiAl)