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Microscopic mechanism of plastic deformation in a polycrystalline Co–Cr–Mo alloy with a single hcp phase

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Abstract

A Co–Cr–Mo alloy with a single ε (hexagonal close-packed, hcp) phase exhibits excellent tensile properties with a 0.2% proof stress of 630 MPa, an ultimate tensile stress of 1072 MPa and an elongation to fracture of 38.3%. The dominant deformation modes are basal ⟨*a*⟩ slip and prismatic ⟨*a*⟩ slip, and the apparent respective critical resolved shear stresses at room temperature are calculated to be 184 and 211 MPa. This simultaneous activation of both ⟨*a*⟩ slips can be explained in terms of the lattice constant ratio c/a of 1.610. There is a tendency for the geometrically necessary dislocations (GNDs) to accumulate at grain boundaries, and the magnitude of this GND accumulation at a particular boundary is dependent on its character. Numerical analysis using a dislocation-model-based strain gradient crystal plasticity calculation makes it possible to characterize the distributions of dislocation density, local stress and local strain in the polycrystalline ε Co–Cr–Mo alloy, and the calculation is largely consistent with the experimental results. This simulation reveals that the activity of the prismatic ⟨*a*⟩ slip in addition to the basal ⟨*a*⟩ slip contributes to the stress relaxation at the boundary. For this reason, excellent tensile ductility is obtained in the polycrystalline ε Co–Cr–Mo alloy.

Keywords: Cobalt–chromium–molybdenum alloy; Deformation mode; Dislocation slip; Crystal plasticity analysis; Geometrically necessary dislocations

1. Introduction

Cobalt–chromium–molybdenum (Co–Cr–Mo, hereinafter CCM) alloys have been used in various medical applications because of their excellent biocompatibility and mechanical properties [1,2]. The ternary phase diagram of CCM alloy in the compositional region generally used in practice reveals that the equilibrium phases are γ (face-centered cubic, fcc), ε (hexagonal close-packed, hcp) and σ (CoCr, *P*42/*mnm*). A massive transformation from the supersaturated γ phase is found to result in the formation of a single ε phase in Co–(27–29 mass%) Cr–(5–6 mass%)Mo alloy [3]. Therefore, a single ε phase can be obtained without σ precipitation using an optimal thermomechanical process inducing this massive transformation in Co–27%Cr–5%Mo alloy. Quite recently, the authors have reported that a Co–27%Cr–5%Mo alloy consisting of a single hcp structure exhibits both high strength and excellent tensile ductility along with a plastic elongation of >30%. In this investigation, the simultaneous activation of both the basal ⟨*a*⟩ slip and prismatic ⟨*a*⟩ slip was also noted [4]. The ε phase in the CCM alloy is reported to improve its wear properties [1]. Therefore, there is a possibility that a CCM alloy with this ε microstructure can be further functionalized for use in biomedical metallic implants. This work aims at examining the basic properties of the deformation behavior of CCM alloys consisting of a

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single ε phase based on experimental and computationally simulated results.

2. Experimental procedures

An ingot of Co–27Cr–5Mo (in mass%) alloy with chemical composition 27.1 mass% Cr, 5.03 mass% Mo, <0.01 mass% N and O, and balance Co was produced using a vacuum-induction melting furnace. After a homogenization treatment at 1523 K for 21.6 ks, the homogenized ingot was isothermally hot-forged at 1373 K to achieve a reduction in height of 87%, after which it was furnace-cooled to room temperature to yield an alloy consisting of a single ε phase. A tensile test was carried out at an initial strain rate of $1.5 \times 10^{-4}\ s^{-1}$ at room temperature. The deformed microstructure was identified by conducting electron backscatter diffraction (EBSD) analysis and transmission electron microscopy (TEM) observation. The EBSD analysis was conducted using a Philips XL30 FEG SEM equipped with an orientation imaging microscope (OIM) produced by TexSEM Laboratories. The density of geometrically necessary dislocations (GNDs) was calculated from the EBSD data following the procedure reported by Calcagnotto et al. [5] (see details in Section 3.3.1). Furthermore, in order to characterize the dislocation accumulation at the grain boundaries, numerical analysis was carried out using a dislocation-model-based strain gradient crystal plasticity calculation (see details in Sections 3.3.2.1 and 3.3.2.2) [6–10].

3. Results and discussion

3.1. Microstructure

Fig. 1a shows the EBSD image quality (IQ) map and inverse pole figure showing the crystallographic orientations with respect to the direction perpendicular to the microstructure of the as-hot-forged CCM alloy. This image shows that an equiaxed grained microstructure was formed with an average grain size of 42 μm and random crystallographic orientations. Furthermore, this as-forged microstructure is found to comprise a single ε (hcp) phase according to the EBSD analysis and XRD profile. Hereinafter, this hot-forged CCM alloy is designated the ε-CCM alloy.

3.2. Plastic deformation at room temperature

3.2.1. Stress–strain curve

The nominal stress–nominal strain curve for tensile deformation is shown in Fig. 1b. The 0.2% proof stress is 630 MPa, and augmented work hardening was associated with increasing strain. Failure occurs at a plastic elongation of 38.3%, and occurs catastrophically with very little local necking even though the ε-CCM alloy exhibits excellent tensile ductility. As compared to the strength and ductility of a CCM alloy consisting of a γ (fcc) phase [11], a higher strength and a similar excellent ductility are observed for the present ε-CCM alloy.

3.2.2. Dislocation slip

Fig. 1 shows the EBSD IQ maps of the ε-CCM alloy at tensile strains of (Fig. 1a) 0% and (Fig. 1c) 6%. Striations corresponding to slip lines or twinning traces appeared in all grains after deformation. There were no misorientations observed among these striations in the grain, indicating that the striations appeared only through the formation slip lines. Trace analysis of these striations reveals that almost all striations lie along the (0001) (basal) or $\{1\bar{1}00\}$ (prismatic) planes, indicating that two types of slip systems are dominant: basal slip and prismatic slip. In fact, the simultaneous activations of basal $\langle a \rangle$ slip and prismatic $\langle a \rangle$ slip are already observed at strains of 1.3% and 3% (corresponding to strains just after yielding). Fig. 2a–c show the inverse pole figures, including the counters denoting the Schmid factor (μ) of (Fig. 2a) basal $\langle a \rangle$ slip and (Fig. 2b) prismatic $\langle a \rangle$ slip along with (Fig. 2c) the ratio of these Schmid factors for the basal $\langle a \rangle$ slip and prismatic $\langle a \rangle$ slip. A ratio of 1.0 means that the μ values for basal $\langle a \rangle$ slip and for prismatic $\langle a \rangle$ slip are equal, and a ratio of >1.0

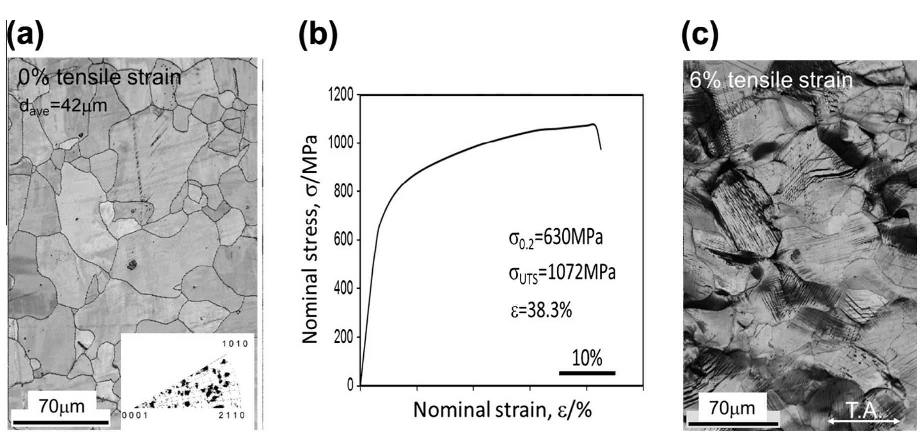

Fig. 1. (a) EBSD IQ maps of ε-CCM alloy at tensile strains of 0%; (b) nominal stress–nominal strain curve of ε-CCM alloy at room temperature; and (c) EBSD IQ maps of ε-CCM alloy at tensile strains of 6%.