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Short communication

Microstructure refining of Co-29Cr-6Mo-0.16N alloy in rapid hot-forging process

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ABSTRACT

Explosive dynamic recrystallization (DRX) of Co-29Cr-6Mo-0.16 N (CCMN) alloy with extremely low stacking fault energy (SFE) was observed in rapid hot-forging processes; the DRX was considered to be closely related to the prevailing planar-defects formation in parent grains such as mechanical twin and stacking fault bands as well as their interactions, that finally break the parent grains into fine pieces accompanied by instantaneous reorientation and recovery in the refined grains.

1. Introduction

Dynamic recrystallization (DRX) of metallic materials is closely related to deformation behavior of parent grains, therefore dependent on both the external factors of deformation such as temperature, strain rate, strain, and the internal factor of materials such as stacking fault energy (SFE), composition, respectively [1–3]. Deformations of low SFE materials have been summarized to be achieved by formations of planar defects such as partial dislocation gliding with a result of stacking fault (SF), mechanical twin or martensitic transformations (MT) [4–10], which are more obvious at high strain rate or in shock-deformation [11,12]. As revealed in a research by Rohatgi et cl.[11], microstructure of low SFE Cu-Al alloy after shock compression showed to be dominated by planar twin formation along various directions, with a consequence of twin network partitioning the parent grains into finer parts. In general, mechanical twin and/or SF are activated in limited planes, therefore are quite non-homogenously distributed inside the parent grains, resulting in localized deformation with high orientation gradient across the defects/matrix interfaces and relatively free of deformation in the other areas of parent grains [4,13,14]. Due to the above reasons, as previously reported in a variety of metallic materials, the planar defects such as twin or shear bands from the overlapping of dislocation gliding or SF [4], especially their interactions, are prone to be preferred sites for nucleation of new grains through either mechanical rotational-recrystallization or continuous DRX [15–18]. An explosive DRX in materials with extremely low SFE is expected to be possible if both the extremely low SFE with a consequence of planar defects dominated deformation and appropriate temperature with an aim of providing sufficient driving-force for DRX are satisfied simultaneously in hot forging process. However, SFE generally increases gradually with temperature, and the activation of the above-mentioned planar defects is inhibited greatly at elevated temperature in most of traditional alloy.

2. Experimental method

Co-29Cr-6Mo-0.16 N (hereafter CCMN, wt%) alloy is a special alloy due to its extremely low SFE even at high temperature (approximately 15 mJ m^{-2}) at 1000 °C as calculated by Yamanaka; [19]. The addition of N in this alloy greatly helps us to simplify the analysis of DRX, since MT scarcely takes place in both deformations above 1000°C and the subsequent rapid-cooling process compared to the alloy without N addition. There have been a few researches regarding the DRX behavior of Co-based alloys, in which extremely homogeneous and fine microstructure could be obtained during rapid hot forging process (above 1.0s^{-1}); however, no special research regarding the DRX mechanism at high strain rate was conducted. In the present study, we conducted dynamic hot forging process on CCMN alloys, in which explosive DRX was observed.

The preparation of CCMN alloy has been reported in previous studies [20,21]. We note that the N addition was achieved by adding Cr_2N powder to CCM alloy melts. Cylindrical specimens, 8 mm in diameter and 12 mm in height, were used. In order to reduce the microstructural nonuniformity arising due to the friction between the sample surface

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and the jigs, the flat ends of specimens were machined with concentric grooves with a depth of 0.1 mm so that the molten lubricant could freely flow inside the grooves at high temperature, resulting in a large decrease in the friction coefficient [22]. Compressive tests were carried out in a vacuum at 1000 and 1100 °C using a computer-aided Thermecmaster-Z hot-forging simulator. Induction heating was carried out at a rate of 5 °C/s to the elevated temperature holding for 300 s before compression. As soon as the sample was compressed to the strain levels of 0.2, 0.4 and 1.0, it was quenched with a mixture of N_2 and He at a cooling rate of approximately 50 °C/s to room temperature in order to "freeze" the high-temperature microstructure. Selected true strain rate $\dot{\varepsilon}$ was $10\ s^{-1}$. The thermocouple used was R-type for temperature controlling, and the temperature accuracy was approximately ± 3 K. In order to avoid the heat dissipation from sample end surface to the anvil surface at the contact region, mica sheets with a diameter of 20 mm and thickness of approximately 0.2 mm were used as heat insulators closely contacting with the anvil surfaces. They were placed close to the anvil surface. In addition, carbon paper was used as an aid to reduce friction owing to its high heat resistance and high lubricating effect. Microstructure observation was carried out in electron backscatter diffraction (EBSD) and transmission electron microscope (TEM).

Power dissipation coefficient index

$$\eta = \frac{2m}{m+1}$$

and

$$m = \left.\frac{\partial \ln \sigma}{\partial \ln \dot{\varepsilon}}\right|_{T,\varepsilon} \tag{1}$$

where η denotes the power dissipation coefficient, m is the strain rate sensitivity, and $\dot{\varepsilon}$ is the strain rate. In general, microstructural behaviors during hot deformation are DRX, dynamic recovery (DRV), superplasticity, void formation, precipitation from metastable phase and so on. These mean that the microstructures of workpiece change remarkably, the dissipation efficiencies have high values when the microstructural changes are beneficial to hot workability. In contrast, materials with localized deformation are generally characterized by low value of η [23].

3. Results and discussion

Fig. 1 show the true stress-true strain (σ-ε) curves obtained at T = 1000, and 1100 °C and strain rate of $10\ s^{-1}$. The σ-ε curve exhibits typical DRXed characteristics: the flow stresses reached to peaks followed by work softening and the steady-state flow is gradually attained at higher strain levels. Furthermore, all the σ-ε curves have single peak; neither the stress oscillations with multiple stress peaks nor dynamic recovery without work softening were observed. The results of η at two temperatures are plotted in Fig. 1(b) as functions of strain. In both cases η is a strong function of strain, and peak values of η in true strain range of approximately 0.4–0.8 are observed at both temperatures, implying the occurrence of drastic microstructure evolutions through DRX in these strain ranges [23]. Higher η is observed at higher temperature, indicating stronger microstructure evolution during compression.

The typical microstructures at strains of 0.2 (a point in curves of Fig. 1), 0.4 (b point in curves of Fig. 1), 1.0 (c point in curves of Fig. 1) and T = 1000 °C are shown in Fig. 2(a-b), (c), and (d), respectively. These strain levels are also marked in the corresponding curves in Fig. 1 by a, b, and c, respectively. Prior to the stress peak in σ-ε curve (point a), due to the extremely low SFE even at high temperature, as expected, deformation of parent grains is dominated by formation of planar bands from their relatively high values of kernel average misorientation (KAM) as shown in Fig. 2(a and b), which evolve into high angle boundaries (HAB) including mechanical twin boundaries (Fig. 2(a)-(b)). From the plane traces identified by EBSD, these bands are observed to mainly align in {111} planes (white straights), leading to high local orientation gradients across the defect/matrix interfaces. The planar bands are observed to intersect with each other, resulting in that the parent grains were partitioned into small parts surrounded by HAB. The alloy matrix is also characterized by low angle boundaries (LAB) as marked by the white lines, which can be confirmed parallel to the {111} planes (Fig. 2(a)). The formation of these LAB leads to significant orientation gradient as marked by the black arrow in Fig. 2(a). According to the result in Fig. 2(a-b), rotations in orientation of new grains surrounded by the bands take place during subsequent deformation (see the area pointed by arrows Fig. 2(b)), leaving behind the randomly-orientated grains with low KAM value compared to that in matrix (see the area pointed by arrows in Fig. 2(c-d)). In the refined grains, LABs (white lines), although much fewer than matrix, are still visible as shown in Fig. 2(c), suggesting that these grains are formed through not the static recrystallization (SRX) during the subsequent cooling but DRX. Most of the refined grains showed to be polygonal geometry, and their boundaries were confirmed to align in {111} planes, implying a possibly relation between the planar band and the new grain boundaries. At strain of about 0.4, over half of the microstructure is replaced by the refined grains (Fig. 2(c)). Compared to low strain level, decreasing in number of {111} mechanical twin boundaries in the matrix suggests that the twin boundaries gradually lose their coherency with further compression. At strain level of 1.0, the microstructure was almost fully replaced by the homogenously-distributed fine grains with relatively low KAM value (Fig. 2(d-e)).

In order to obtain more information about the DRX process in the current study, the boundaries in different areas of microstructure (ε = 0.4, Fig. 2(b)) were distinguished in EBSD. The first group of

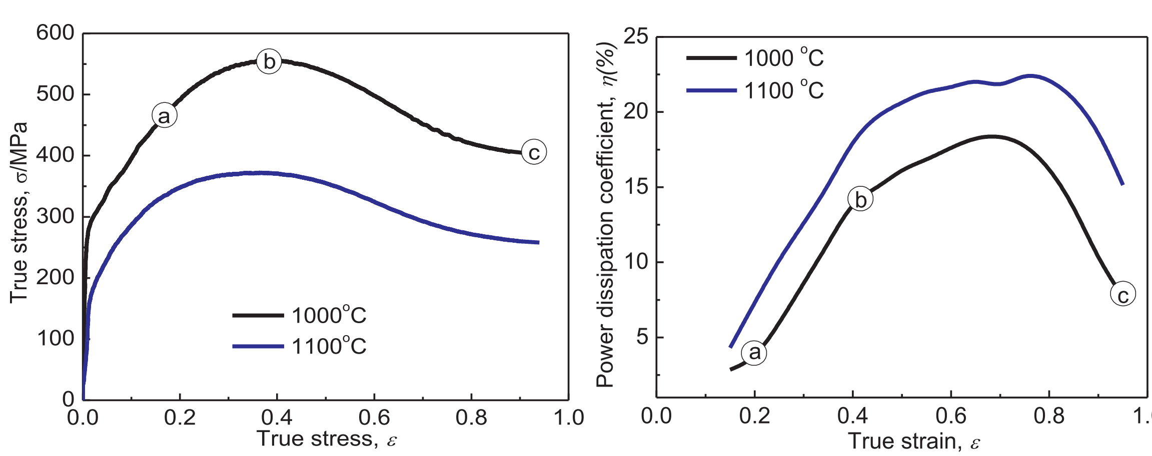

Fig. 1. (a) True stress-true strain curves at various temperatures and strain rate of $10\ s^{-1}$, and (b) Power dissipation coefficient of the corresponding curves in (a).