Journal of Alloys and Compounds 640 (2015) 298–304

Contents lists available at ScienceDirect

Journal of Alloys and Compounds

journal homepage: www.elsevier.com/locate/jalcom

Design of low-cost titanium aluminide intermetallics

Bin Liu, Yong Liu *, Congzhang Qiu, Canxu Zhou, Jianbo Li, Huizhong Li, Yuehui He

State Key Lab of Powder Metallurgy, Central South University, Changsha 410083, PR China

ARTICLE INFO

Article history:
Received 1 August 2014
Received in revised form 5 March 2015
Accepted 30 March 2015
Available online 9 April 2015

Keywords:
TiAl alloy
Low cost
β phase
Deformation
Creep

ABSTRACT

For titanium aluminide (TiAl) alloys, the high cost mainly comes from the forming process since the workability of TiAl alloys is quite low. To optimize the workability, introduction of β phase is an effective way. In the present study, a novel TiAl material alloyed with cheap β-stabilizing elements, Mo and Fe, was developed. The deformation behavior of the new alloy was characterized, and further alloying with interstitial elements to improve the high temperature properties was explored. Results show that the designed Ti–45Al–3Fe–2Mo at.% alloy has a fine grain β-stabilized microstructure. It shows a wide forging window, providing the foundation for an industrial forging process. The good deformability is derived from dynamic recrystallization of β phase and deformation-induced $\beta \rightarrow \alpha_2 + \gamma$ transformation. With the addition of 0.5% C, the creep properties can be significantly improved, which is mainly caused by the decrease in the amount of β phase and the precipitation of P-type carbides.

1. Introduction

Up to now, TiAl alloys have practical applications in turbocharger turbine wheels for passenger vehicles [1–3] and low-pressure turbine blades for jet engines [4–6]. In order to take further advantage of the superior properties of TiAl alloys, i.e., low weight and excellent high-temperature strength, which contributes to fuel savings and the reduction of CO_2 emissions, it is necessary to apply to larger-sized and complex-shaped components. This requires a forging process similar to ordinary metallic materials. However, the workability of conventional TiAl alloys is quite low, and since only special processes such as isothermal forging have been applicable, it has been difficult to produce larger-sized parts in an industrially cost feasible manner. Therefore, the design of TiAl alloys with improved hot deformability and low processing cost is essentially necessary.

One way to improve the hot-workability of TiAl alloys is to add alloying elements to form β phase at elevated temperatures [7]. The disordered β phase with bcc lattice provides a sufficient number of independent slip systems, and is softer than the α_2 and γ phases. Thus, it may improve the deformability of TiAl alloys at elevated temperature, so that TiAl alloys could be forged under near conventional condition. A number of studies have demonstrated that β phase can be stabilized through alloying with either β-isomorphous elements [8–11], such as Nb, V, and Ta, or β-eutectoid elements, such as Cr, Fe, Mn, Ni and W. Among those elements, Mo is well known to be a strong β stabilizing element, and a modest amount addition of Mo can promote the formation of β containing microstructure in TiAl alloys. The addition of Fe is less popular, however, a small amount of Fe can also have a significant effect on extending the $\alpha_2 + \gamma$ phase field and stabilizing the β phase. Most importantly, Fe and Mo are relatively cheap, and have been widely used for reducing the costs of the titanium alloys [12]. Therefore, Fe and Mo are promising alloying elements to fabricate low-cost TiAl alloys with good workability. In our previous work [13,14], we studied the effects of Fe and Mo on microstructures and properties of TiAl alloys. The results show that, with the addition of Mo and Fe, the amount of β phase increases apparently. For example, with the increase of Mo from 1% to 4%, the volume fraction of β phase increases from 1% to 20%, and with the increase of Fe from 1% to 4%, the volume fraction of β phase also increases from 0.5% to 4%. Furthermore, mixed addition of Fe and Mo can significantly refine the microstructure other than stabilize the β phase. The grain size of the TiAl alloys can be refined to about 12 μm when the Fe and Mo contents are 3% and 2%, respectively.

A crucial requirement for TiAl alloys as a structural material is the high creep strength at high temperatures. The introduction of the β phase may deteriorate creep properties since the β phase is soft at elevated temperatures. One way to solve this dilemma is to eliminate the β phase by subsequent heat treatments after the hot working [15], but it is difficult to eliminate the β phase completely since it usually contents stabilizing elements. Addition of interstitial elements, such as C, is another way to increase the high temperature properties of TiAl alloys. A plenty of works have demonstrated that addition of C can improve the

* Corresponding author. Tel.: +86 731 88836939; fax: +86 731 88710855.
E-mail address: yonliu@csu.edu.cn (Y. Liu).

http://dx.doi.org/10.1016/j.jallcom.2015.03.239
0925-8388/

creep performance of γ-TiAl alloys [16–18]. However, for the β stabilized TiAl alloys, the effect of C on the creep performance is still unclear.

The present work is to report the development of a new β-stabilized TiAl alloy with additions of Mo and Fe. To further improve the high temperature performance of the new alloy, the effects of C-addition on the creep performance were also studied.

2. Experimental

Ingots with nominal compositions of Ti–45Al–3Fe–2Mo at.% and Ti–45Al–3Fe–2Mo–0.5C at.% were prepared by vacuum arc melting and suck casting. The dimension of the ingots is 58 mm in diameter and 120 mm in height, and the total amount of interstitial impurities is less than 750 ppm by mass. Cylindrical compression specimens with diameter of 8 mm and height of 10 mm were cut from the ingots. Compressive tests were carried out in the temperature range of 1000–1200 °C and the strain rate range of 0.001–10 s^{-1} on a computer-aided Gleeble 3800 equipment. After deformation, all specimens were quenched in water in order to preserve the as-deformed microstructure. Rolling experiments were conducted on a conventional rolling mill. The rolling strain rate is about 0.05 s^{-1} and the temperature is 1100 °C.

Creep tests were carried out for 300 h under constant loading of 150 MPa at 750 °C with a MTS-GWT2105 creep testing machine. The dimension of the gauge area is 30 mm in length and 6 mm in diameter. The experimentally derived creep curves were fitted with a function based on the Garofalo equation [19]. The microstructures of the samples were examined by using Quanta 600 environmental scanning electron microscope (SEM), electron back scattered diffraction (EBSD) and JEOL-2100F transmission electron microscope (TEM). The EBSD data were analyzed with the TSL OIM 5.31. The specimens were electrolytically polished by using 30 ml nitric acid and 70 ml methanol solution at 30 °C. TEM specimens were prepared by twin jet-polishing in an electrolyte solution containing 5% volume of perchloric acid, 35% volume of butanol and 60 vol.% of methanol at 30 °C and 20 V.

3. Results

3.1. Microstructure and mechanical property of Ti–45Al–3Fe–2Mo alloy

Fig. 1 represents the typical as-cast microstructure of the Ti–45Al–3Fe–2Mo alloy. The average grain size is about 12 μm. EDX analysis shows that the bright areas were enriched by Mo and Fe, suggesting that these areas are the β phase. Therefore, the microstructure consists of γ phase (dark), α_2 phase (gray) and β phase (bright). The volume fraction of the β phase in the as-cast condition is about 25.5%. The enrichment of the β phase by Mo was also observed by Morris et al. in Ti–44Al–2Mo alloy [20].

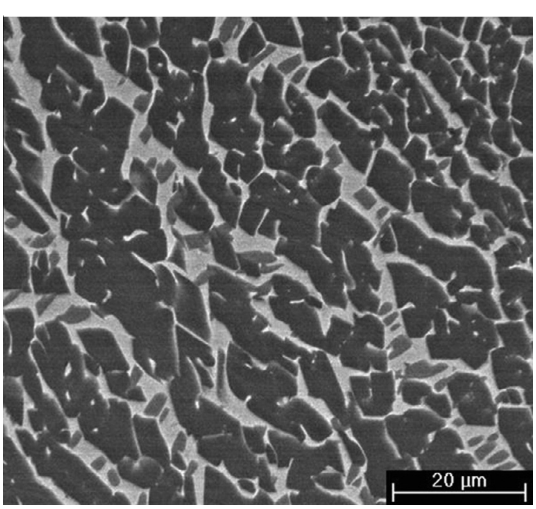

Fig. 1. SEM microstructure of the as-cast Ti–45Al–3Fe–2Mo alloy.

3.2. Hot deformation behavior of Ti–45Al–3Fe–2Mo alloy

The true strain–true stress (S–S) curves deformed from 800 °C to 1100 °C with different strain rates are shown in Fig. 2. All the curves exhibit a peak flow stress at a relatively low strain (<0.1), and followed by a flow softening to a steady-state stress regime. The flow stress increases with the increase of the deformation temperature and the decrease of the strain rate. Remarkably, almost all the deformed samples appeared to be sound and did not have any penetrating cracks, indicating that the alloy has a good deformability.

The flow stress data were analyzed using the approach of processing map to evaluate the constitutive behavior during the hot deformation. The power dissipation efficiency η in the processing map is given by the following equation [21]:

$$\eta = \frac{2m}{(m+1)} \tag{1}$$

where m is the strain rate sensitive coefficient, which represents the pattern in which the input power is dissipated by the plastic deformation through microstructural evolution rather than by heat. The three-dimensional variation in η with temperature and strain rate constitutes a power dissipation map that identifies the characteristic domains of various deformation and damage processes. Simultaneously, the microstructural instability regimes during the plastic flow are evaluated by using Zilgler's criteria [21]:

$$\xi = \frac{\partial \log[m/(m+1)]}{\partial \log \dot{\varepsilon}} + m \leqslant 0 \tag{2}$$

The processing map obtained at a strain of 0.7 is shown in Fig. 3. The processing map has two domains that exhibit peak values of the power dissipation efficiency: (1) at strain rates of 0.001–0.01 s^{-1} and temperatures of 870–920 °C with a peak of 53% at 900 °C/0.001 s^{-1} and (2) at strain rates of 0.01–0.05 s^{-1} and temperature of 1050–1100 °C with a peak of 45% at 1100 °C/0.01 s^{-1}. Fig. 4 shows the microstructures of the Ti–45Al–3Fe–2Mo alloy deformed at the two peaks, which show typical equiaxed microstructures produced by dynamic recrystallization, and the volume fractions of β phase are about 28% (Fig. 4b) and 39% (Fig. 4d), respectively. Therefore, these areas are corresponding to the optimum conditions for hot working of this material. It is worthwhile to note that the morphology of γ grains does not change obviously, which suggests that most of the deformation are occupied by the β phase. From the Eq. (2), an instability regime occurs at strain rates higher than approximately 0.05 s^{-1}. Thus, the high strain rate regime is not favorable for the hot working of this alloy.

The processing map gives an appropriate deformation parameters, which are located in the relatively low strain rate region. However, in the actual conditions of industrial production, high strain rates and low temperatures are usually preferred for the concern of reducing the costs. Therefore, taking all the various factors into account, 1100 °C/0.05 s^{-1} is chosen for the hot rolling experiments, where the rolling condition is located in the stable flow region in the processing map. Fig. 5 shows the rolled sheet with a dimension of 70 mm × 110 mm × 1 mm after a total deformation ratio of over 80%, where sound rolling can be seen without any cracks. The results validate the processing map, and indicate that the alloy has a good workability.

3.3. Improvement of creep performance through C-doping

Fig. 6 shows the SEM microstructure of the as-cast Ti–45Al–3Fe–2Mo–0.5C alloy. The microstructure consists of equiaxed γ grains, α_2/γ lamellar colonies and β phases. It is clear that,