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Effect of Fe addition on microstructure and mechanical properties of Ti-25V-15Cr-2Al-0.2C alloy①

LEI Li-ming(雷力明), HUANG Xu(黄 旭), SUN Fu-sheng(孙福生), WU Xue-ren(吴学仁), CAO Chun-xiao(曹春晓)
(Beijing Institute of Aeronautical Materials, Beijing 100095, China)

Abstract: The effect of 2% Fe addition on the microstructure and mechanical properties of Ti-25V-15Cr-2Al-0.2C alloy (mass fraction) was studied. It is found that the addition of 2% Fe seems to have no obvious effect on the microstructure of the alloy, but results in a significant change in mechanical properties. Compared with the alloy without Fe addition, the alloy with 2% Fe addition exhibits remarkable higher tensile strength and creep resistance, whereas the ductility is relatively lower at room temperature. The significant changes in mechanical properties can be rationalized by the decrease of stacking fault energy caused by the addition of 2% Fe.

Key words: non-burning titanium alloy; iron; microstructure; mechanical property; deformation structure
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1 INTRODUCTION

Titanium alloys have been extensively applied to aerospace industries due to their excellent specific strength, good corrosion resistance and mechanical properties. However, conventional titanium alloys are highly reactive and can undergo sustained combustion under conditions encountered in gas turbine engines compressors. In the early 1990s, Pratt & Whitney[1-3] in the USA developed a new β titanium alloy with good burn resistance, designated Alloy C (Ti-35V-15Cr). The production cost of this alloy is very high because of a large amount of expensive vanadium addition. In recent work undertaken jointly by the IRC and Rolls-Royce in UK[4-6], Al has been introduced into Ti-V-Cr system in order to improve the cost-competitiveness of the alloy. The Al content should be kept lower than 3% to avoid the formation of ordered α_2 phase detrimental to the ductility. Li et al[7] found that carbon addition significantly improved the ductility of the alloys containing Al. α precipitation can be reduced through gathering of oxygen by titanium carbide, thus the β matrix becomes ductile. The alloy with nominal composition of Ti-25V-15Cr-2Al-0.2C is a non-burning titanium alloy with an optimized carbon addition[8-12]. Previous work[12] showed that under the expected application conditions the creep resistance of the alloy is relatively low. Numerous approaches have been carried out to improve the mechanical properties of the alloy by variation of heat treatment and composition. The purpose of the present work is to study the effect of 2% Fe addition on the microstructure and mechanical properties of Ti-25V-15Cr-2Al-0.2C alloy.

2 EXPERIMENTAL

Two alloys used for this study have nominal compositions of Ti-25V-15Cr-2Al-0.2C and of Ti-25V-15Cr-2Al-2Fe-0.2C (named TF3a and TF3b in this paper, respectively). After mixing and compacting raw materials of the two alloys, 15 kg ingots were produced by vacuum arc remelting (VAR). The ingots were melted for 2 - 3 times to ensure chemical homogeneity. The chemical compositions of the ingots are shown in Table 1.

Table 1 Chemical compositions of experimental alloys (mass fraction, %)

Alloy	Ti	V	Cr	Al	Fe	C	O	N
TF3a	Bal.	25.6	14.7	2.03	–	0.2	0.088	0.008
TF3b	Bal.	25.5	14.7	1.92	1.88	0.2	0.100	0.012

The ingots were forged and finally hot-rolled into bars with 18.5 mm in diameter. As-rolled samples were solution-treated for 30 min at 1 050 ℃ followed by air cooling and then aged for 4 h at 700 ℃. In order to investigate the thermal stability of the alloys, the heat-treated samples were exposed for 100 h at 540 ℃.

All samples were cut, polished and etched for microstructural examination. The etchant consisted of 5% HF, 10% HNO_3 and 85% H_2O (volume frac-

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Correspondence: LEI Li-ming; Tel: +86-10-62458116; E-mail: liming.lei@biam.ac.cn

tion). Thin foils for transmission electron microscopy (TEM) observation were prepared by a twin-jet electropolishing technique, using an electrolyte composed of 6% perchloric acid, 34% *n*-butanol and 60% methanol at about 40 V and − 30 ℃. TEM observations were carried out on JEOL 200CX and HITACHI H-800 transmission electron microscope operated at 200 kV.

Tensile tests were performed on smooth cylindrical specimens with 5 mm in diameter and 25 mm in gauge length using an Instron hydraulic-driven testing machine with a strain rate of $10^{-3} s^{-1}$. Creep tests were conducted at 540 ℃ in constant load machines at a stress of 250 MPa. Creep specimens of 10 mm in diameter and 50 mm in gauge length were machined. Fracture surfaces were observed using a JSM-5600LV scanning electron microscope equipped with a LINK ISIS300 EDX analytical system operated at 20 kV.

3 RESULTS AND DISCUSSION

3.1 Microstructure after heat-treatment

Fig. 1(a) and (b) show the microstructures of TF3a and TF3b alloys after heat-treatment, respectively. According to our previous investigation[12], the heat-treated microstructure of the alloys consists of equiaxed β grains along with globular or short-rod carbide particles (2 − 10 μm) and a small amount of very fine α precipitation (< 1 μm). It can be seen from the micrographs that the differences in the microstructures between TF3a and TF3b alloy seem to be subtle, indicating that the addition of 2% Fe has no obvious effect on the microstructure of Ti-25V-15Cr-2Al-0.2C alloy.

3.2 Mechanical properties

The results of mechanical properties of TF3a and TF3b alloys are listed in Table 2. Compared with TF3a alloy, at room temperature the tensile strength of TF3b alloy is higher, however, the ductility is considerably lower. It is noteworthy that at 540 ℃ the ductility of TF3b alloy is as good as that of TF3a alloy, although its strength is still higher. As listed in Table 2, after creep at 540 ℃, 250 MPa for 100 h the residual strain of TF3b alloy is significantly smaller than that of TF3a alloy, suggesting that TF3b alloy owns higher creep resistance.

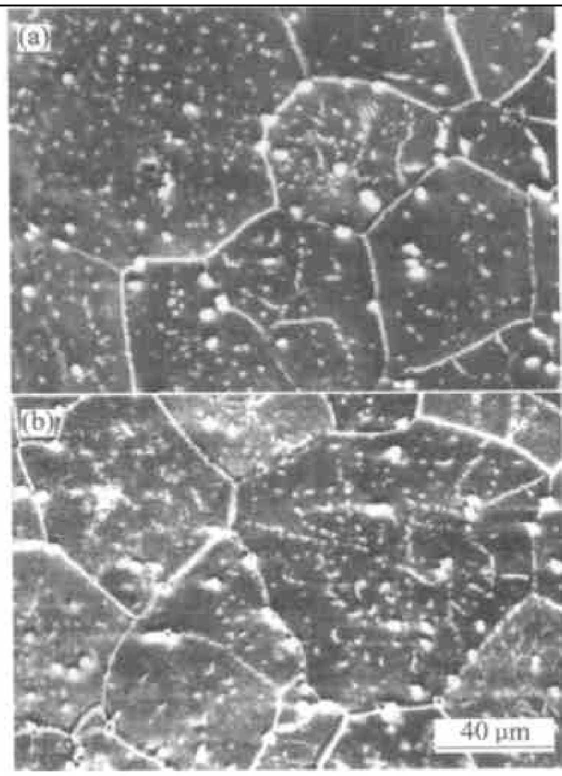

Fig. 1 SEM micrographs showing microstructures of two alloys after heat-treatment
(a) —TF3a; (b) —TF3b

Fig. 2 shows the tensile fracture surfaces of two alloys at room temperature and 540 ℃. It can be observed from Fig. 2(a) that TF3a alloy fractures by a mixture of ductile transgranular and intergranular modes at room temperature. In Fig. 2(b), the fracture surface of TF3b alloy at room temperature shows extensive cleavage facets except for a small number of dimples and intergranular cracks, indicating a brittle fracture behavior. As shown in Fig. 2(c) and (d), at 540 ℃ both TF3a and TF3b alloys fail by a ductile manner, and dimples can be clearly seen throughout the fracture surfaces.

3.3 TEM observations

Fig. 3 shows the deformation structures of TF3a and TF3b alloys after creep at 540 ℃, 250 MPa for 100 h. In terms of previous work[12], the dislocation loop group (DLG) is the typical dislocation configuration in TF3a alloy (Fig. 3(a)). However, TEM observations show that some stacking faults (SFs, Fig. 3(b)) and twins (Fig. 3(c)) occur in the creep deformation structure of TF3b

Table 2 Mechanical properties of TF3a and TF3b alloys

Alloy	σ_b/MPa		$\sigma_{0.2}$/MPa		δ_5/%		ϕ/%		Residual creep strain(540 ℃, 250 MPa, 100 h)
	RT	540 ℃	RT	540 ℃	RT	540 ℃	RT	540 ℃	
TF3a	1 025	855	998	695	17	19	36	32	4.029
TF3b	1 050	900	1 040	739	7	17	10	31	0.217

RT —Room temperature

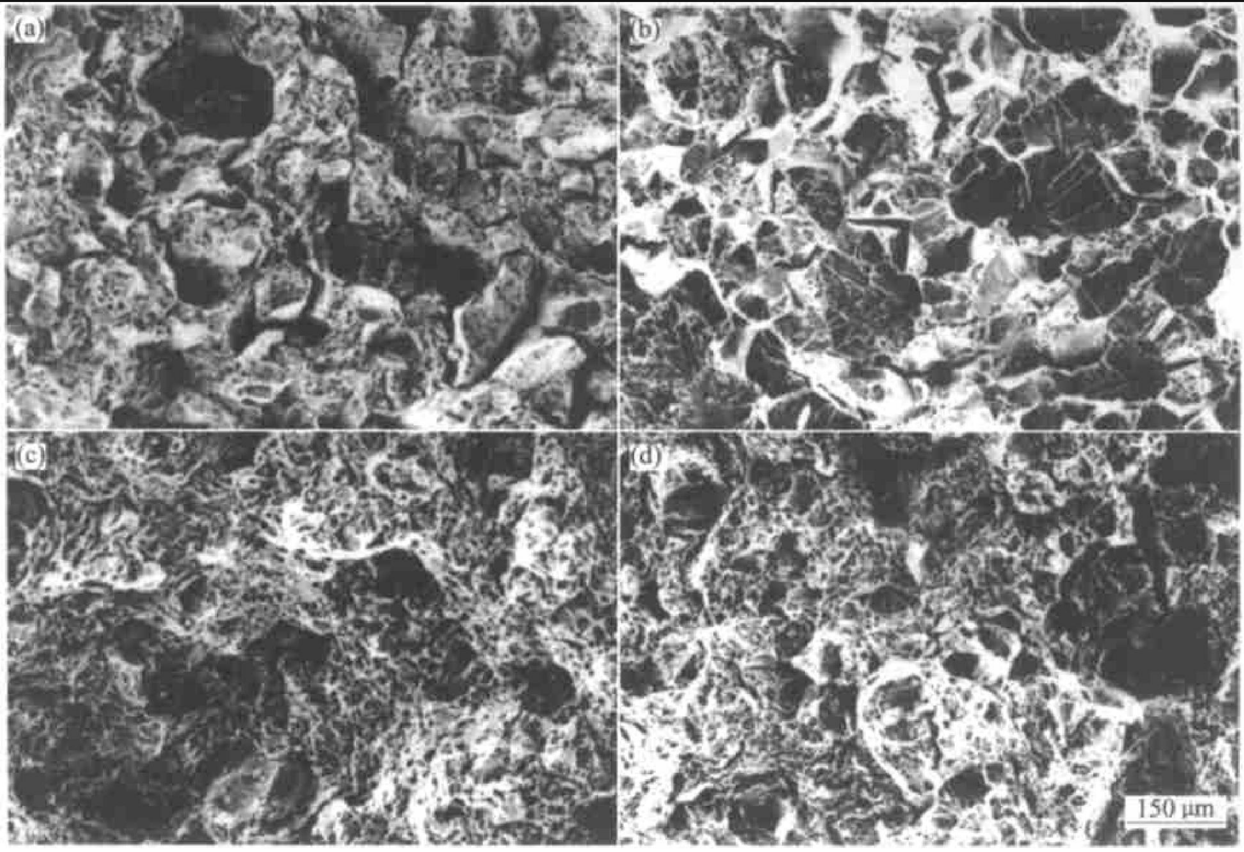

Fig. 2 SEM fractographs showing tensile surfaces of two alloys
(a) —TF3a, room temperature; (b) —TF3b, room temperature; (c) —TF3a, 540 ℃; (d) —TF3b, 540 ℃

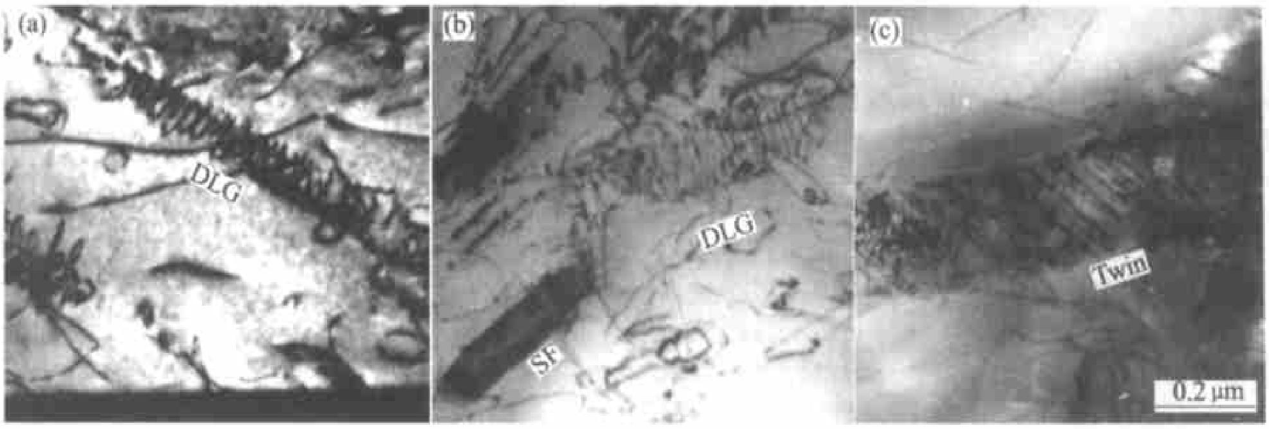

Fig. 3 TEM micrographs showing deformation structures in two alloys after creep at 540 ℃, 250 MPa for 100 h
(a) —TF3a; (b), (c) —TF3b
(DLG —Dislocation loop group; SF —Stacking faults)

alloy besides the DLG. It is well known that SFs and twins are closely related to stacking fault energy (SFE), i.e. the lower the SFE, the higher the level of SFs and twins. Therefore, the occurrence of both features in TF3b alloy seems to suggest that the addition of 2% Fe results in the decrease in the SFE of the alloy. The presence of SFs and twins is favorable to the improvement in creep resistance, due to restrained cross-slip and many interfacial barriers for dislocation motion. As a result, it can be easily understood that the creep resistance of TF3b alloy is significantly higher than that of TF3a alloy.

The cleavage fracture in TF3b alloy (Fig. 2(b)) is also correlative with the lower stacking fault energy. In general, the appearance of cleavage fracture requires a mechanism for creating a sharp crack which would be unstable and propagate before blunting by plasticity. As described previously, at room temperature the activation of cross-slip is hard in TF3b alloy because of its lower SFE. This means the absence of substantial blunting of the crack tip by cross-slip. Therefore, a sharp crack occurring in TF3b alloy tends to become cleavage initiation.

Fig. 4(a) and (b) show the tensile deforma-

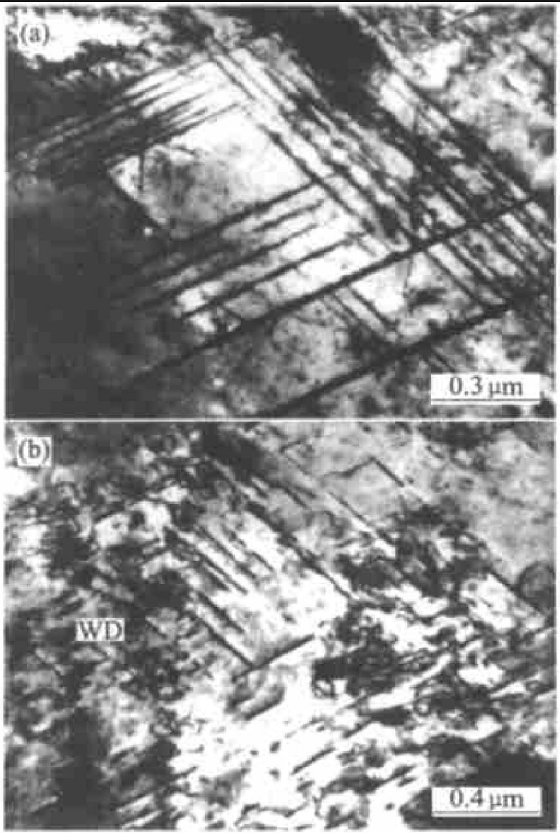

Fig. 4 TEM micrographs showing tensile deformation structures in two alloys at 540 ℃
(a) —TF3a; (b) —TF3b
(WD —Weave dislocation)

tion structures of TF3a and TF3b alloys at 540 ℃, respectively. Cross-slip bands or/and weave dislocations (WDs) can be clearly seen from the micrographs, indicating that cross-slip activity occurs during tensile deformation at 540 ℃. Obviously, the excellent ductility of both alloys at 540 ℃ results from the activation of cross-slip. The shorter cross-slip bands (Fig. 4(b)) in TF3b alloy compared with those of TF3a alloy suggest the lower cross-slip activity.

4 CONCLUSIONS

The addition of 2% Fe seems to have no obvious effect on the microstructure of Ti-25V-15Cr-2Al-0.2C alloy, but results in a significant change in mechanical properties. Compared with the alloy without Fe addition, the alloy with 2% Fe exhibits remarkable higher tensile strength and creep resistance, whereas the ductility is relatively lower at room temperature . The significant change in mechanical properties can be rationalized by the decrease of stacking fault energy caused by the addition of 2% Fe.

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