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Effect of β Volume Fraction on the Dynamic Grain Growth during Superplastic Deformation of Ti_3Al-based Alloys

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The superplastic deformation behavior of Ti_3Al based (α_2+β) alloy was studied with respect to the volume fraction of α_2/β. Three alloys containing 21, 50 and 77% in volume fractions of β exhibited large tensile elongations of over 500% at 970°C with a strain rate of 2.5×10^{-4} sec^{-1}. The largest elongation was observed in the alloy with 21% of β. As the volume fraction of β phase increased, the flow stress and correspondingly, the strain-rate sensitivity values decreased. Due to the higher diffusivity of Ti in β phase than in α_2 phase, the increase in β volume fraction from 21% to 77% accelerated the dynamic grain growth, and degraded the superplasticity of the Ti_3Al-based alloys. The strain-based grain growth behavior was quantitatively analyzed and incorporated into a constitutive equation. The calculated flow curves are in agreement with the experimental ones in the stable deformation region.

Key words : dynamic grain growth, superplasticity, Ti_3Al, β volume fraction

1. INTRODUCTION

Titanium aluminides based on Ti_3Al have received a great deal of attention in recent years due to their low density and the fact that they possess a good combination of mechanical properties and oxidation resistance at high-temperatures. However, low ductility and poor formability at room temperature are the major obstacles for practical applications. These problems have been partly solved by the addition of β-stabilizing elements such as Nb, V, and Mo. The fabrication of complex components from these difficult-to-fabricate materials can also be achieved by the superplastic forming method, one of the novel processing techniques. The superplastic behavior of the Ti_3Al alloys have been recently reported; however, the microstructural effects associated with the volume fraction of β phase on the superplastic deformation have not yet been thoroughly understood [1-3].

The study was conducted to examine the superplastic behavior of Ti_3Al based (α_2+β) alloys with respect to the volume fraction of β. Two aspects are of major concern: finding the optimum volume fraction of β, temperature and strain-rate for the best superplasticity of Ti_3Al alloys, and how microstructural evolution (i.e., dynamic and static grain growth and flow hardening etc.) will occur with different volume fractions of β during superplastic deformation.

2. EXPERIMENTAL PROCEDURE

Three different alloy compositions based on Ti-13.5Al-21Nb (wt.%) alloy were produced by a vacuum arc-melting method using non-consumable electrodes. The chemical compositions of the alloys are shown in Table 1. At least two heats were made for each composition and the dimension of a bar was approximately 80 mm × 40 mm × 22 mm. To vary the volume fraction of β phase, the content of Nb was increased from 21 wt.% to 31 wt.%, and that of Al was decreased from 13.5 wt.% to 10.5 wt.%. After being melted, the pieces were homogenized at 1200°C for 12 hours, cross-rolled from 22 mm to 3 mm at 1020°C and recrystallized at the same temperature.

For the step strain rate tests, specimens of 15 mm gauge length (5 mm width and 3 mm thickness) were machined, and step strain rate tests were carried out at 950, 970, and 990°C and various strain rates from 2.5×10^{-5}/sec to 10^{-2}/sec in Ar atmosphere. For the superplastic tests, specimens with gauge dimensions of

Table 1. Chemical composition (wt.%) of the alloys produced

	Ti	Al	Nb	O_2	N_2	C
Alloy 1 (Ti-13.5Al-21Nb)*	bal.	13.54	21.37	0.0023	0.00037	0.0082
Alloy 2 (Ti-12.0Al-26Nb)	bal.	11.97	26.22	0.0018	0.00017	0.0043
Alloy 3 (Ti-10.5Al-30Nb)	bal.	10.51	30.75	0.0025	0.00020	0.0047

*Composition of alloy 1 is same as that of regular α_2 alloy.

5 mm length, 5 mm width and 3 mm thickness were made. The tests were conducted at 970°C with a constant strain rate of 2.5×10^{-4}/sec. The loading axis was perpendicular to the final rolling direction. To analyze the microstructures before and after the superplastic tests, Kroll's reagent (10% HF, 5% HNO_3 and 85% H_2O) was used.

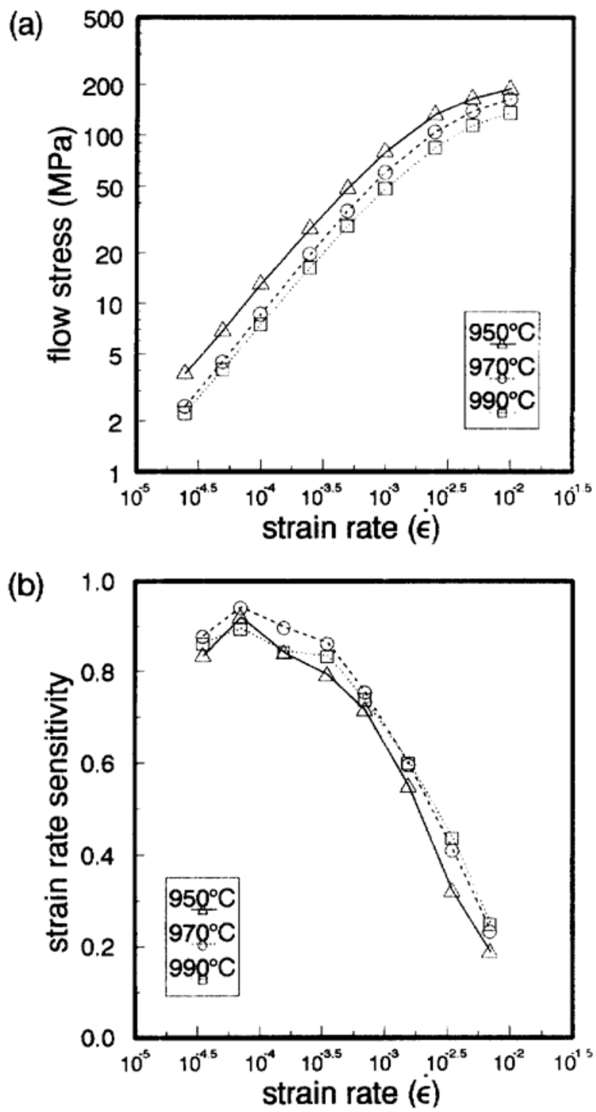

Fig. 1. Effects of temperature on the flow curve and strain rate sensitivity in Alloy 1. a) flow stress-strain curve and b) strain rate sensitivity

3. RESULTS AND DISCUSSION

3.1. Effect of β volume fraction on the flow curve

Fig. 1 shows the variation of flow stress (true stress, σ) and strain-rate sensitivity (m) with the strain rate at various temperatures for Alloy 1 (Ti-13.64Al-21.15Nb, a regular (2). It can be seen that for a strain rate of around 10^{-4}/sec, the flow stresses at 950°C, 970°C and 990°C are well below 20 MPa, and the m-values are in the range of 0.85-0.95. This fact implies that large elongations can be obtained at a strain rate of 10^{-4}/sec. The m-value is largest at 970°C and decreases slightly at 950°C and 990°C, therefore, the testing conditions for optimum superplasticity was considered to be 970°C and 10^{-4}/sec. To investigate the effect volume fraction of β on the superplasticity at 970°C, the recrystallized microstructures were annealed at 970°C for 30 minutes and then quenched in water. Fig. 2 shows the microstructures of the three alloys representing very fine equiaxed grains below 4 μm in diameter. Microstructural parameters for the three alloys annealed at 970°C were also measured and summarized in Table 2. The volume fractions of β phase varied from 21% to 77% with an increase in Nb content, while mean grain sizes of all three alloys were considered to be constant (i.e., $\simeq$ 3.5 μm).

Fig. 3 represents the flow stress and strain rate sensitivity (m) values as a function of strain rate at 970°C for the three Ti_3Al alloys. It should be noted that an increase in β phase causes an increase in flow stress and a decrease in the m-value at strain rates below 5×10^{-4} sec^{-1}. As for the role of β phase in conventional $(\alpha+\beta)$ Ti alloys, it is generally believed that an increase in β volume fraction of up to 50~60% decreases the flow stress and increases the m-value [4-7]. However, in the present study of $(\alpha_2+\beta)$ Ti_3Al alloys, the reverse trend was observed. This reverse effect of β phase in the Ti_3Al alloys is mainly attributed to the four orders of higher diffusivity of Ti in β than in α. This is because it is generally accepted that the steady-state flow stress is inversely proportional to diffusivity at high temperatures and low strain rates. Therefore, the grain growth

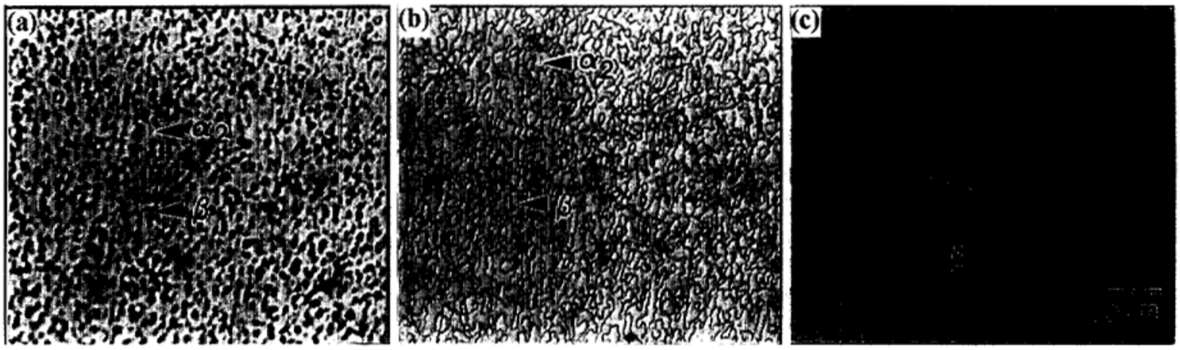

Fig. 2. Optical micrographs for three alloys after annealing at 970°C. a) Alloy 1, b) Alloy 2, and c) Alloy 3

Table 2. Microstructural parameters of the recrystallized specimens after annealing at 970°C

	Alloy 1	Alloy 2	Alloy 3
Mean grain size (μm)	3.41	3.52	3.56
Volume fraction of β (%)	21	50	77
Phase size of β (μm)	1.60	3.52	3.92
Phase size of α_2 (μm)	3.89	3.52	1.75

behavior will be more active in the specimen with the higher β volume fraction, leading to the increase in flow stress. Another possible reason is the difference in boundary sliding characteristics between the conventional α/β two phase Ti alloys and the Ti_3Al alloys containing intermetallic α_2 phase. In conventional Ti alloys, a large superplasticity when the volume fraction of β phase is close to 50% is frequently observed. This implies that the sliding resistance of α/β is much lower compared to those of α/α or β/β, which have been recently analyzed quantitatively [8]. However, in these Ti_3Al alloys that contain intermetallic α_2 phase, the relative boundary sliding resistances are thought to be different from those of conventional Ti alloys, due to the difference in the crystal structure of α_2 (DO_{19}) and α (HCP) phase. A recent investigation on dislocation structures along various boundaries [9] has also shown that sliding along α_2/α_2 boundaries as well as α_2/β boundaries plays a major role in the superplastic deformation of this alloy. Therefore, at α_2/α_2 boundaries serve as an important site for boundary sliding, inducing a large superplasticity in the specimen with high β volume fraction. Furmore quantitative analysis on the sliding resistance of each boundary is in progress. To verify the results of the step strain rate test, superplastic deformation tests were conducted at 970°C with a constant strain rate of 2.5×10^{-4}/sec, and the results are shown in

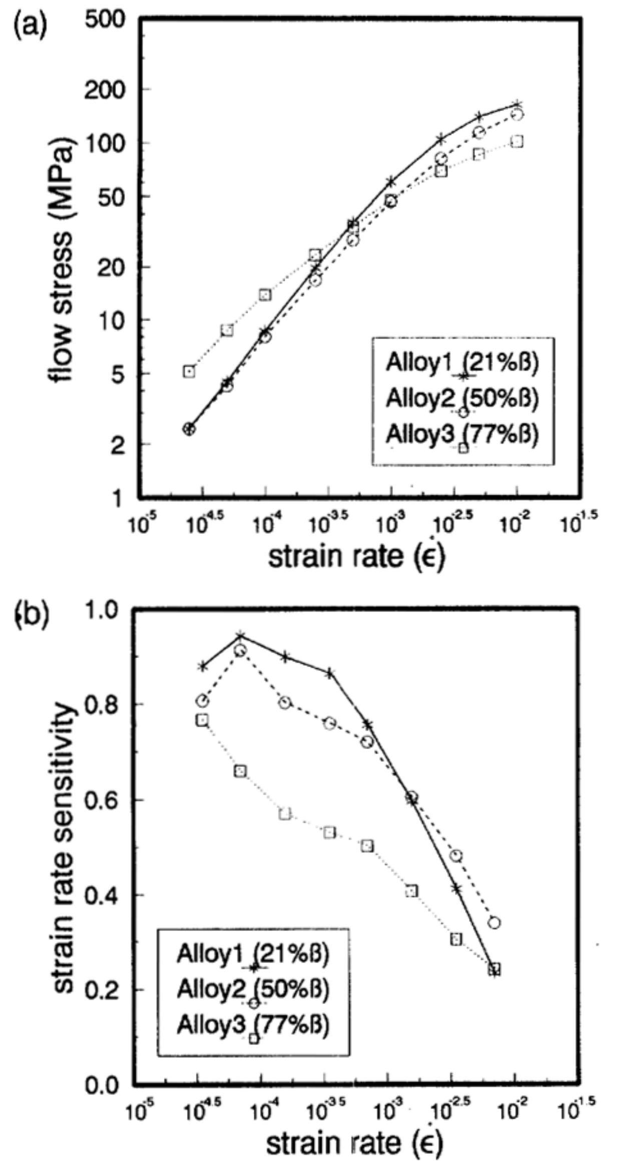

Fig. 3. Effects of β volume fraction on the flow curve and strain rate sensitivity at 970°C. a) flow stress-strain curve and b) strain rate sensitivity

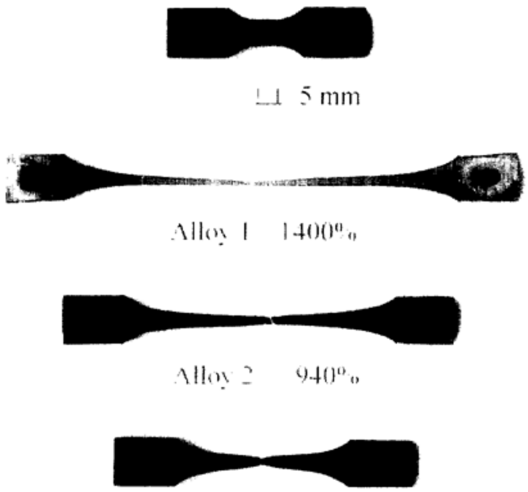

Fig. 4. Superplastically deformed specimens of Alloys 1, 2 and 3 at 970°C and a strain rate of 2.5×10^{-4}/sec.

Fig. 4. As expected, large elongations of over 500% were obtained in all three alloys. It was also noted that the largest elongation of about 1400% was obtained in Alloy 1 which contained 21% of β.

3.2. Microstructural evolution during superplastic deformation

Fig. 5 shows the strain dependence of grain growth occurring during the superplastic deformation of Alloy 1. Static and dynamic grain growth occur significantly in both the α_2 and β phases. To analyze the effect of β volume fraction on grain growth kinetics and on superplasticity behavior, a simple strain-based grain growth model [10] was adopted. In this model, if a grain with an initial size L_0 grows to L after a strain of ε, the logarithm of grain size at any instantaneous strain (L) is linearly proportional to the dynamic growth term given by the product of the grain growth coefficient (C) and the strain (ε), i.e.,

$$\ln(L/L_s) = C\varepsilon\,, \tag{1}$$

where L_s is the grain size after static grain growth. For the precise measurement of grain size, image analyzer was used and the individual grain size of α_2 and β phase were carefully measured and incorporated to obtain the mean grain size L.

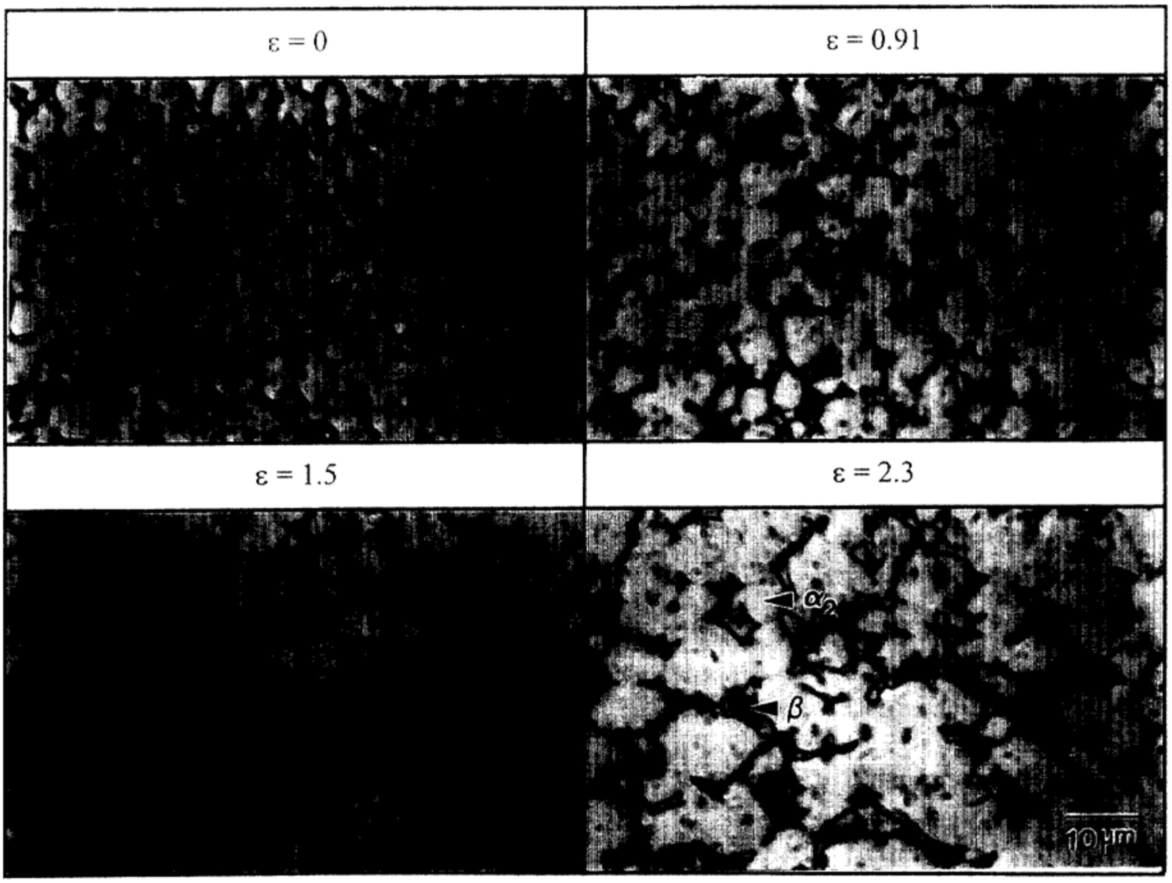

Fig. 5. Optical micrographs showing the grain growth with the increase of strain in Alloy 1.

Table 3. Variation of grain growth coefficient (C) in three alloys with different β volume fractions

Alloy No.	Alloy 1	Alloy 2	Alloy 3
grain growth coefficient (C)	0.277	0.295	0.327

The values of C in the above Eq. 1 were measured for the three alloys and listed in Table 3. Since the percentages of static grain growth are less than 15%, it can be deduced that the overall grain growth is mainly dominated by the dynamic grain growth. The grain growth coefficient (C) in Table 3 increases with the increase in the volume fraction of β, which implies that the extent of grain growth during the superplastic deformation is accelerated by an increase of the β volume fraction. We should be able to observe the enhanced grain growth in the alloy with higher β volume fraction since the diffusivity of Ti in β phase is about four orders higher than that in the α_2 phase at superplastic temperatures [5, 11].

It is worthwhile to discuss the optimum β volume fraction for the best superplasticity in Ti_3Al alloys. As shown in Fig. 4, Alloy 1, containing 21% of β, showed the largest elongations of about 1400%. These volume fractions of β, which show excellent superplasticity, are considerably smaller when compared to those of Ti-6Al-4V ($\approx$40%) and Ti-10V-2Fe-3Al ($\approx$70%) alloys. Similar results were also been observed in the previous investigations [1-3]. As previously mentioned, the relatively lower optimum β volume fraction of Ti_3Al based alloys is partly related with the lower sliding resistance of the α_2/α_2 boundary. However, fast grain growth behavior in the β phase is another reason why the optimum β volume fraction for the best superplasticity of Ti_3Al (α_2+β) alloy is relatively lower than other (α+β) Ti alloys. Generally, the higher the amounts of β phase, the larger the diffusion activity of the alloy which induces an increase in the accommodation capability of phase boundary sliding and/or grain boundary sliding. In Ti_3Al alloys, however, such an increase in capability seems to be strongly counterbalanced by the dynamic grain growth, which leads to the deterioration of superplasticity with higher β volume fractions.

3.3. Flow curve at high temperatures

To describe the flow behavior at high temperature, a modified Dorn's equation [12] is frequently used as a constitutive equation,

$$\dot{\varepsilon} = AGb/kT\,(b/d)^p\,(\sigma/G)^{1/m}D\,, \tag{2}$$

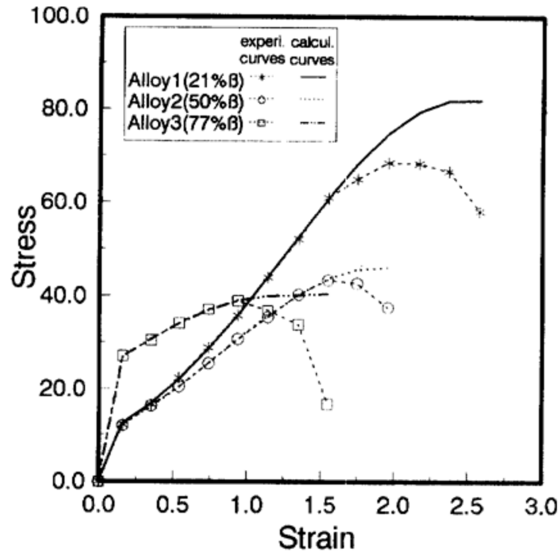

Fig. 6. Stress-strain curves for Alloys 1, 2, 3 at 970°C and a stain rate of 2.5×10^{-4}/sec.

where A is a constant, d is the grain size, G is the shear modulus, b is Burger's vector, D is the diffusivity, p is the grain size exponent, and m is the strain rate sensitivity. However, due to the considerable amounts of grain growth occurring during superplastic deformation, the grain size, d, will increase with the increase in strain. Therefore, Eq. 2 needs to be modified in such a way that L (grain size after the dynamic growth) in Eq. 1 instead of L_0 (initial grain size) should be used for the grain size, d. Rearranging the Eq. 2 leads to the Eq. 3 as follows [10].

$$\sigma = K\,(L_s/L_0)^{mp}\,\exp(\alpha mp\varepsilon)\,\dot{\varepsilon}^m\,, \tag{3}$$

where the term $(L_s/L_0)^{mp}$ term represents the static flow hardening and the term exp(αmpε) represents the dynamic one.

Fig. 6 represents the measured and calculated (by Eq. 3) stress (σ)-strain (ε) curves for the three alloys at a constant strain rate ($\dot{\varepsilon}$=2.5×10^{-4}/sec) at 970°C. As expected from Fig. 3 and Table 3, three alloys containing 21, 50, and 77% β volume fractions represented very large tensile elongations over 500% ($\varepsilon\approx1.5$) at 970°C with a strain rate of 2.5×10^{-4}/sec. The largest elongation of 1400% ($\varepsilon\approx2.5$) was obtained in the alloy with 21% β volume fraction, due to its high m-value, low flow stress, and low grain growth rate occurring during the deformation. It is also shown in Fig. 6 that the experimental and predicted flow curves agreed well with each other until the strain reached to 1.0~1.5, then they deviated considerably due to the necking. For the simplicity of the curve fitting procedure, the grain size exponent (p) was assumed to be 3 and the strain rate sensitivity, m, was determined from the flow stress at each strain. However, the values of the microstructural term, A, for the three different volume fractions of β were calculated from the Eq. 3 using the flow stress and strain rate in the region of stable deformation ($\varepsilon \leq 0.5$).

4. CONCLUSIONS

1. The Ti_3Al-based alloys containing 21, 50, and 77% β volume fractions represented very large tensile elongations of over 500% at 970°C with a strain rate of 2.5×10^{-4}/sec. The largest elongation of 1400% was obtained in the alloy with 21% β volume fraction.

2. Due to the higher diffusivity of Ti in the β phase, the increase in β volume fraction from 21% to 77% accelerated the dynamic grain growth, resulting in the degradation of superplasticity in Ti_3Al-based alloys.

3. The strain-based grain growth behavior was quantitatively analyzed for the alloys with different β volume fractions and was incorporated into the constitutive equation. The calculated flow curves were in good agreement with the experimental ones in the stable deformation region.

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REFERENCES

1. A. Dutta and D. Banerjee, *Scripta metall.* **24**, 1319 (1990).
2. H. S. Yang, P. Jin and A. K. Mukherjee, *Mater. Sci. Eng.* **A153**, 457 (1992).
3. D. Jobart and J. J. Blandin, *J. Mater. Sci.* **31**, 881 (1996).
4. N. E. Paton and C. H. Hamilton, in *Titanium Science and Technology* (eds., G. Lütjering V. Zwicker, and W. Bunk), p. 649, Proc. 5th Int. Conf. on Ti, Munich (1994).
5. J. R. Leader, D. E. Neal and C. Hammond, *Metall. Trans.* **17A**, 93 (1986).
6. M. L. Meier, D. R. Lesuer and A. K. Mukherjee, *Mater. Sci. Eng.* **A154**, 165 (1992).
7. P. G. Partridge, D. S. McDarmaid and A. W. Bowen, *Acta metall.* **33**, 571 (1985).
8. J. S. Kim, Y. W. Chang and C. S. Lee, *Metall. Trans.* **29A**, 217 (1998).
9. J. H. Kim, C. K. Park, T. K. Ha and Y. W. Chang, *J. Kor. Inst. Met. & Mater.* **35**, 1130 (1997).
10. E. Sato and K. Kuribayashi, *ISIJ Int.* **33**, 825 (1993).
11. J. Rusing and C. Herzig, *Scripta metall.* **33**, 561 (1995).
12. O. D. Sherby and J. Wadsworth, *Progress in Mater. Sci.* **33**, 169 (1989).