Texture Evolution and Mechanical Anisotropy in Dual-Phase Ti₃Al-Based Alloy Loaded at 700 °C to 1000 °C

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The super α_2 Ti₃Al-based alloy with a fine grain size of $\sim 2.2 \mu\text{m}$ exhibits superplastic elongations over 1000 pct at 920 °C to 1000 °C, 600 pct at 900 °C, 330 pct at 850 °C, and 140 pct at 750 °C. Mechanical anisotropy is observed in this alloy, and relatively lower flow stresses and higher tensile elongations are obtained in the 45 deg specimen loaded at 25 °C to 960 °C. The texture characteristics appear to impose significant influence on the mechanical anisotropy at temperatures below 900 °C (under the dislocation creep condition), and the $\{111\}\langle 2\bar{1}1 \rangle$ and $\{0001\}$ basal textures evolve in the β and α_2 phases after tensile straining. At loading temperatures higher than 900 °C (under the superplastic flow condition), the anisotropy effect is less pronounced and the grain orientation distribution becomes basically random in nature. Rationalizations for the mechanical anisotropy in terms of the Schmid factor calculations for the major and minor texture components in the β and α_2 phases provide consistent explanations for the deformation behavior at lower temperatures as well as the initial straining stage at higher temperatures.

I. INTRODUCTION

Ti₃Al-BASED intermetallic alloys have the advantages of superior high-temperature properties and low density, which are highly attractive for aerospace and aircraft applications.^[1] However, Ti₃Al-based alloys are known to be brittle at low temperatures because of the DO₁₉ ordered hexagonal crystal structure. By adding the β -stabilizing elements such as Nb, V, and Mo, or receiving thermomechanical treatments to reduce the grain sizes, the Ti₃Al-based alloys with a duplex α_2 and β phase at temperatures below the β -transus temperature would exhibit acceptable formability.

Figures 1(a) and (b) show the Ti-Al binary and Ti₃Al-Nb phase diagrams, respectively.^[1,2] The structure of pure Ti₃Al is DO₁₉, $P63/mmc$, an ordered hexagonal phase, and termed as the α_2 phase at temperatures below 1100 °C. Niobium will occupy the same sites as the Ti atoms.^[3,4,5] The structure of the β phase may be a disordered bcc structure or an ordered bcc (called the B2) structure. For the Ti₃Al alloy containing Nb, the disordered β phase present above the order-disorder transition temperature, T_{O-D} , could be transformed into the ordered β phase region during cooling. In regular α_2 (Ti-24 at. pct Al-11 pct Nb) and super α_2 alloys (Ti-25 at. pct Al-10 pct Nb-3 pct V-1 pct Mo), the T_{O-D} is in the temperature range of 1100 °C to 1200 °C.^[6-9] Since the volume fractions of the α_2 and β phases play an important role in superplasticity, the measured relative volume

fractions in the current super α_2 alloy in consistence with the equilibrium phase diagram are shown in Figure 1(c).

The main slip systems for the α_2 phase are the $\langle a \rangle$ type dislocations in the prism, pyramidal, and basal planes, including and $\{10\bar{1}0\}\langle 11\bar{2}0 \rangle$, $\{10\bar{1}\bar{1}\}\langle 11\bar{2}0 \rangle$, and $\{0001\}\langle 11\bar{2}0 \rangle$, plus the $\langle c + a \rangle$ type dislocations in the pyramidal plane such as $\{10\bar{1}\bar{1}\}\langle 11\bar{2}3 \rangle$ and $\{11\bar{2}2\}\langle 11\bar{2}3 \rangle$. With the c/a ratio less than the ideal value, the $\langle a \rangle$ type dislocations in the prism plane (rather than the basal plane) are most important. Meanwhile, the $\langle c + a \rangle$ type dislocations could improve the ductility at temperatures above 800 °C.^[10] As for the β phase, dislocations on the $\{110\}\langle 111 \rangle$ systems are active, making β a more deformable phase, which has been regarded as the strain accommodation carrier.^[5]

To date, there have been a number of reports on the superplastic behavior of Ti₃Al-based alloys,^[10,11-19] confirming that the Ti₃Al-based alloys exhibited admirable high-temperature superplasticity (HTSP) at higher temperatures between 920 °C and 1000 °C ($\sim 0.64 T_m$) and at lower strain rates between 10^{-4} and 10^{-5} s^{-1} to an elongation greater than 1200 pct.^[12,16,17,19] Lowering the temperature range to $< 0.6 T_m$ would be attractive. There have been very limited studies reporting the low-temperature superplasticity (LTSP) in Ti₃Al alloys.^[20,21] Table I summarizes the related LTSP and high-strain-rate superplasticity (HSRSP) reports.^[20-24]

The textured Ti₃Al-Nb alloys have been reported to exhibit anisotropic superplasticity at 925 °C to 1000 °C.^[15,18,25] The superplastic deformation became less anisotropy with increasing superplastic strain due to the progressive breakdown of aligned microstructure by grain boundary sliding (GBS). For conventional superplastic materials, the grain structures are equiaxed. It is expected that the deformation is more or less homogeneous, with no apparent anisotropy behavior. The limited reports on the anisotropic superplasticity might be either related to the grain or microstructure inhomogeneity, but might also result from the textures present in the processed materials.

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(a)

(b)

(c)

Fig. 1—(a) The Ti-Al binary equilibrium phase diagram,^[1] (b) the Ti₃Al-Nb phase diagram,^[2] and (c) the measured relative volume fractions of the α₂ and β phases in the current super α₂ alloy, consistent with the Ti₃Al-Nb phase diagram in (b).

After thermomechanical processes over 900 °C to 1100 °C, the Ti₃Al alloys usually exhibit a crystallographic texture,^[17,26–32] such as {0001}^[26] or {1120} texture in the α₂ phase and the {200} cubic texture in the β phase.^[27] Multiple texture components have also been reported, i.e., {001}<110>, {112}<110>, {112}<111>, and {013}<331> in β^[28,29,30] and (1210)[6060] or (1213)[1010] in α₂.^[29] Fu *et al.*^[17] found that the texture in the Ti₃Al-Nb alloy became much weaker with increasing superplastic strain. It was concluded that GBS and grain rotation proceeded during superplastic loading. Thus, texture might influence the tensile properties of Ti₃Al alloys as a result of changes in grain boundary structure.

Although LTSP in TiAl- and Ti₃Al-based alloys with submicrocrystalline structure could occur at temperatures below 900 °C,^[20,21] their anisotropy behavior and texture evolution are still inadequately understood. In this article, the latter effect at higher (950 °C ± 50 °C) and lower (800 °C ± 100 °C, ~0.55 *T_m*) temperatures is explored.

II. EXPERIMENTAL

The super α₂ material was supplied by Rockwell International (Thousand Oaks, CA), originally fabricated by Timet, with a nominal composition of Ti-25Al-10Nb-2V-1Mo (at. pct). The material has been checked by inductively coupled plasma-optical emission spectrometry, which gave the composition of Ti-22.88 pct Al-10.48 pct Nb-3.07 pct V-1.01 pct Mo. The as-received Ti₃Al thin sheet specimen was 2 mm in thickness. The thermomechanical treatment for the current super α₂ alloy was conducted mostly at ~1050 °C, and then annealed at 900 °C to 1100 °C to fully recrystallize and to achieve a fine-grained two-phase (α₂ + β) microstructure.

In order to study superplasticity, the minitensile specimens with the tensile axis parallel to the rolling direction were machined from the as-received Ti₃Al thin sheet by line cutting with a gage length of 5.5 mm and gage width of 3 mm for testing at elevated temperatures. Larger specimens were machined with a gage length of 10 mm and a gage width of 3 mm for applying an extensometer during loading at room temperature. Tensile tests were conducted in accordance with the constant-crosshead-speed method. The testing temperature was controlled to be within ±3 °C. The system was performed with purging argon at a flow of 3 L/min all over the heating and superplastic loading time. Furthermore, specimens were prior heated to 100 °C and then coated with special glass on the surface with sequential air cooling before tensile tests to suppress oxidation. Argon was a protective atmosphere rather than a back pressure that was usually used for suppressing cavitations during superplastic tensile loading or superplastic forming practices. Systematic LTSP tensile tests were performed under the constant-crosshead-speed tests at 700 °C to 850 °C with initial strain rates of 8×10^{-5} , 2×10^{-4} , and 5×10^{-4} s⁻¹, respectively. For studying the anisotropic superplasticity, the tensile specimens were machined with the loading axis aligned along a direction parallel (0 deg), perpendicular (90 deg), or 45 deg to the rolling directions, respectively. Anisotropic tensile tests were performed at 25 °C, 750 °C, and 920 °C. To ensure reproducibility, at least three tests were conducted for each case, and the average data are

Table I. Summary of LTSP or HSRSP in Various Ti-Based Alloys (Compositions Given in Atomic Percent)

Materials	Process	d (μm)	Temp. ($^{\circ}\text{C}$)	Strain Rate (s^{-1})	Elong. (Pct)	Author	Year	Ref.
Ti-10Co-4Al	hot rolling	—	1000	5×10^{-2}	650 to 750	Hofmann <i>et al.</i>	1995	22
Ti-47Al-2Cr-2Nb	extrusion + annealing	4 to 5	800	2×10^{-5}	310	Nieh <i>et al.</i>	1997	23
Ti-50Al	isothermal forging + annealing	0.4	800	8×10^{-4}	225	Imayev <i>et al.</i>	1999 2000	20 21
Ti-46Al	isothermal forging + annealing	0.2	800	6.4×10^{-4}	415	Imayev <i>et al.</i>	1999 2000	20 21
Ti-46Al-2Cr-2Nb-1Ta	isothermal forging + annealing	0.3	750 800	6.7×10^{-4}	310 520	Imayev <i>et al.</i>	1999 2000	20 21
Ti-48Al-2Cr-2Nb	isothermal forging + annealing	0.3	750 800 850	8.3×10^{-4}	195 355 270	Imayev <i>et al.</i>	1999 2000	20 21
Ti-25Al	multi-directional isothermal forging + annealing	0.1	500 600	6.4×10^{-4} 6.4×10^{-4}	120 680	Imayev <i>et al.</i>	1999 2000	20 21
		0.3	600 700	6.4×10^{-4} 6.4×10^{-4}	300 780			
		0.8	600 800 800	6.4×10^{-4} 1.3×10^{-4} 6.4×10^{-4}	170 640 580			
Ti-6Al-4V	hot rolling + high-pressure torsion	0.1 to 0.2	650 725	1×10^{-3} 1×10^{-1}	575 215	Sergueeva <i>et al.</i>	2000	24

presented. To retain the microstructure of superplastically loaded specimens, the specimens should be rapidly cooled. However, to take account of safety and to prevent oxidation in air, a high-temperature furnace could not be opened at temperatures above 500 $^{\circ}\text{C}$ for cooling by liquid nitrogen. To achieve rapid cooling, it is necessary to use a thin steel pipe to insert into the furnace so that liquid nitrogen can be purged *via* the pipe at a pressure of 2 kg/cm².

The as-received and superplastically loaded specimens were milled using emery papers with water, and then polished using 0.1- to 0.3- μm Al₂O₃ suspensions. After that, the specimens were etched with the Kroll's reagent (10 pct HF + 5 pct HNO₃ + 85 pct H₂O) to attain the microstructural information by optical microscopy (OM) or scanning electron microscopy (SEM). Detailed microstructural characterizations have been previously presented.^[32]

Microtextures of both the as-received and post-SP specimens were determined by EBSD. Pole figures, inverse pole figures, orientation distribution functions (ODF), and grain misorientation angle distributions are used to investigate the texture change during superplastic deformation. The EBSD system applied is the Oxford Instrument LINKOPAL* interfaced

*LINKOPAL is a trademark of Oxford Instruments, Buckinghamshire, UK.

to a field emission gun JSM-6330TF scanning electron microscope (JEOL, Tokyo). The lattice constants for the β and α_2 phases have been determined previously.^[17]

The as-received and post-SP specimens were milled using emery papers with water to remove about 25 pct of the specimen thickness. Then, the surface was further refined using a colloidal silica polishing suspension for more than 1 hour. If the specimen surface retains surface damage on a crystallo-

graphic scale, or has any surface contaminant, oxide, or reaction product layers left, a clear EBSD pattern formation may be suppressed. To enhance the pattern quality, the specimens were further electropolished with an electrolyte, which is the same as the twin-jet electrochemical polishing, prior to the establishment of Kikuchi patterns by EBSD. The pixel size was typically around 0.2 μm , and around 20,000 pixel points were measured for each mapping and the ODF, corresponding to an area around 25 $\mu\text{m} \times 35 \mu\text{m}$. On average, around 150 grains were included for each mapping and at least three regions were examined to ensure the reproducibility. Including more grains in the EBSD mapping would certainly improve the accuracy, but this method is costly and time-consuming. The accumulation time for one pole figure (on one region) would be roughly 4 and 0.5 hours for the regular and field emission gun SEM, respectively. The texture determination was also done with the help of CaRIne crystallography computer software.^[33]

III. RESULTS

A. Microstructure Characterization of the As-Received Material

The three-dimensional SEM microstructure of the as-received material is shown in Figure 2. The alloy consists of a mixture of α_2 (the darker phase) and β (the lighter phase) phases. The fine spherical α_2 particles (measuring $\sim 2 \mu\text{m}$) and a few coarse α_2 plates (measuring $\sim 10 \times 4 \times 2 \mu\text{m}^3$) are nearly random dispersed within the β matrix. Assuming that the α_2 grains are spherical particles, the average effective grain size of the α_2 grains is taken as 2.2 μm . Judging from

Fig. 2—SEM micrographs showing the three-dimensional microstructure of the as-received material (the darker phase is the α_2 phase and the lighter one is the β phase).

Table II. Anisotropy Tensile Test Results Performed at Room Temperature and a Strain Rate of $3 \times 10^{-3} \text{ s}^{-1}$

Specimen Orientation	YS (MPa)	UTS (MPa)	Elongation (Pct)
0 deg	950 ± 30	1100 ± 40	5 ± 1
45 deg	900 ± 40	1020 ± 50	6 ± 1
90 deg	1150 ± 30	1220 ± 50	2 ± 1

the grain size, the LTSP performance may not be as satisfactory as those possessing submicrometer grains around 0.1 to 0.3 μm .^[20,21]

B. Mechanical Tests

The engineering yield strength (YS) and the ultimate tensile strength (UTS) for the 0 deg specimen at room temperature were extracted to be $950 \pm 30 \text{ MPa}$ and $1100 \pm 40 \text{ MPa}$, respectively. The data lie between the higher values for the 90 deg specimen and the lower values for the 45 deg one, as compiled in Table II. The 45 deg specimens always show lower tensile strengths and higher elongations.

The superplastic tensile tests were performed at temperatures over 700 °C to 850 °C with initial strain rates from 8×10^{-5} to $1 \times 10^{-2} \text{ s}^{-1}$. Table III summarizes the results of tensile mechanical properties at elevated temperatures together with some of the previous results for HTSP.^[17] The dependence of superplastic elongations as a function of test temperature is shown in Figure 3. The highest LTSP elongation of 333 pct was obtained at 850 °C with an initial strain rate of $5 \times 10^{-4} \text{ s}^{-1}$. However, the elongation decreases abruptly upon lowering the loading temperatures below 750 °C.

The anisotropy tensile results conducted at 750 °C to 960 °C are listed in Table IV. Higher elongations for the 45 deg specimens are obtained at all temperatures. The anisotropy behavior in terms of tensile elongation shows strong temperature dependence, being most pronounced at 750 °C, and gradually diminishes with increasing loading temperature. Figure 4 represents the untested and tested specimens loaded at 920 °C. Note that the 1910 pct that occurred in the 45 deg specimen is even higher than the previously

Table III. Summary of the Tensile Properties of the 0 Deg Specimens Loaded at Different Elevated Temperatures and Strain Rates

Temperature (°C)	Strain Rate (s^{-1})	UTS (MPa)	Elongation (Pct)
700	8×10^{-5}	526 ± 25	28 ± 2
	2×10^{-4}	594 ± 23	20 ± 3
	5×10^{-4}	855 ± 34	26 ± 1
	1×10^{-2}	917 ± 30	16 ± 2
750	8×10^{-5}	346 ± 15	141 ± 6
	2×10^{-4}	379 ± 15	131 ± 5
	5×10^{-4}	480 ± 18	39 ± 2
	1×10^{-2}	737 ± 20	28 ± 2
800	8×10^{-5}	214 ± 8	80 ± 4
	2×10^{-4}	269 ± 8	130 ± 6
	5×10^{-4}	305 ± 10	123 ± 5
	1×10^{-2}	553 ± 15	84 ± 5
850	8×10^{-5}	114 ± 5	186 ± 10
	2×10^{-4}	183 ± 6	200 ± 8
	5×10^{-4}	178 ± 6	333 ± 16
	1×10^{-2}	395 ± 12	145 ± 7
900	8×10^{-5}	39 ± 3	643 ± 32
	2×10^{-4}	64 ± 2	595 ± 20
	5×10^{-4}	105 ± 4	451 ± 35
	1×10^{-2}	80 ± 3	1680 ± 54
920	8×10^{-5}	39 ± 2	1400 ± 75
	2×10^{-4}	60 ± 2	1350 ± 35
	5×10^{-4}	80 ± 3	1680 ± 54
	1×10^{-2}	35 ± 2	980 ± 32
960	8×10^{-5}	52 ± 3	1500 ± 55
	2×10^{-4}	63 ± 3	1200 ± 43
	5×10^{-4}	63 ± 3	1200 ± 43
	1×10^{-2}	63 ± 3	1200 ± 43

Fig. 3—Dependence of superplastic elongations as a function of loading temperature at four different initial strain rates.

Table IV. Anisotropy Tensile Elongation (in Percent) Results Obtained at Elevated Temperatures and a Strain Rate of $5 \times 10^{-4} \text{ s}^{-1}$

Specimen Orientation	750 °C	850 °C	920 °C	960 °C
0 deg	39 ± 2	333 ± 16	1680 ± 54	1200 ± 43
45 deg	155 ± 5	380 ± 15	1910 ± 25	1290 ± 50
90 deg	38 ± 3	310 ± 12	1880 ± 39	1130 ± 34

reported highest superplastic elongation of 1500 pct for the Ti_3Al -based alloy.^[17] The current temperature and strain rate of 920 °C and $5 \times 10^{-4} \text{ s}^{-1}$, respectively, are also lower and faster than the previous values of 960 °C and $2 \times 10^{-4} \text{ s}^{-1}$.

C. Texture Analyses

1. The as-received materials

The macrotextures revealed by X-ray pole figures of the as-received material are shown in Figure 5. There is a strong $\{100\}$ texture with a maximum intensity contour of 40 existing in the β phase, while there appears to be a weaker $\{11\bar{2}0\}$ texture with a maximum intensity contour of 10 existing in the α_2 phase. It could be concluded that the preferred orientation in the β matrix was a result of the previous thermo-mechanical treatments, while the α_2 grains were relatively more randomly dispersed in the continuous β matrix. The orientation distribution on the X-ray pole figures displays an incomplete examining range from 0 to ~ 70 deg due to the limitation of the X-ray diffraction instrument. Hence, it is difficult to determine the orientation direction $\langle \text{RD} \rangle$ from the X-ray pole figures.

The microtexture of the as-received material was also examined by EBSD for comparisons. Pole figures, inverse pole

figures, and ODFs are helpful to establish the texture information. In order to ensure the homogeneous sampling, five regions were randomly selected. Although slight deviations are seen occasionally, representative textures can be easily revealed. Figure 6 shows the typical EBSD pole figure results of the β and α_2 phases in the as-received specimen, respectively. Based on the pole figures and ODFs, multiple textures exist in both phases, *i.e.*, strong $\{100\}\langle 011 \rangle$ and weak $\{111\}\langle 2\bar{1}\bar{1} \rangle$ (plus minor $\{0\bar{1}1\}\langle 001 \rangle$) textures in the β phase, as well as strong $\{11\bar{2}0\}\langle 0001 \rangle$ plus weak $\{0001\}\langle \bar{1}100 \rangle$ and $\{0001\}\langle 11\bar{2}0 \rangle$ textures in the α_2 phase. The intensities are classified automatically by the computer into eight levels, and the black color implies that there is no orientation present. The results are consistent with the X-ray pole figures.

The orientation data could be displayed in the form of a pole figure, an inverse pole figure, or a Euler angle. It could be further displayed in an orientation image map (OIM) with different colors by various orientations, providing the information related to the microstructure. Figure 7 displays the OIMs in the normal, transverse, and longitudinal directions in the β and α_2 phases. By using color keys in Figures 7(d) and (h) for the cubic β and hexagonal α_2 phases, respectively, the orientation distribution could be roughly determined; for instance, the $\{100\}\langle 011 \rangle$ orientation in ND- β is displayed in red and that in RD- β in green. A strong preferred orientation is present in the β phase with most grains displayed by a single color, while multiple preferred orientations are present in the α_2 phase with plentiful colors.

a. Texture evolution during anisotropic tensile straining at 750 °C and 920 °C

For the 0 deg specimen loaded at 750 °C to a total failure strain of 0.33 (or 39 pct elongation), the EBSD pole figures for the β and α_2 phases at the fracture tip of the post-SP specimen are shown in Figures 8(a) and (b). The original $\{111\}\langle 2\bar{1}\bar{1} \rangle$ texture is still present significantly, while the $\{100\}\langle 011 \rangle$ rotated cube becomes much weaker in the β phase. The weakening of the $\{100\}\langle 011 \rangle$ texture should

Fig. 4—Representative anisotropic tensile specimens loaded at 920 °C and $5 \times 10^{-4} \text{ s}^{-1}$.

Fig. 5—X-ray pole figures of the as-received material: (a) the β phase (200) pole figure with the maximum intensity contour level of 40 against the background, and (b) the α_2 phase (1120) pole figure with the maximum contour of 10.

Fig. 6—Pole figures of the (a) β phase and (b) α_2 phase in the as-received specimen.

Fig. 7—The OIMs in the as-received specimen: (a) through (c) for the β phase and (e) through (g) for the α_2 phase. By using the orientation color key (d) and (h), orientations in the β and α_2 phases can be visualized by their colors, respectively. Note that (d) and (h) are not referred to the stereographic triangle, but only the index for color in the mapping.

be due to the enhancement of $\{111\}\langle 112 \rangle$. From Figure 8, the strongest $\{1120\}\langle 0001 \rangle$ texture present in the as-received state became weak. Instead, the strongest texture

became nearly the $\{0001\}\langle \bar{1}100 \rangle$ orientation, or more precisely, the $\{0001\}\langle \bar{4}310 \rangle$ orientation, with the overall weak $\{0001\}$ plane texture in the α_2 phase. That is, the

Fig. 8—Pole figures of the β and α_2 phases in the post-SP 0 deg specimens (a) and (b) loaded at 750 °C and (c) and (d) loaded at 920 °C. Note that the tensile loading direction is all horizontal. The RD and TD in the pole figures, preset by the SEM-EBSD software, are not referred to the actual specimen direction for the 45 and 90 deg specimens. They should be rotated by 45 and 90 deg, respectively.

α_2 grains have undergone rotation and sliding and exhibit a tendency to align the $\{0001\}$ planes parallel to the rolling plane during straining, whereas the β grains concentrate into the $\{111\} \langle 2\bar{1}1 \rangle$ orientation.

The EBSD results of the 45 deg specimen loaded at 750 °C to a total failure strain of 0.94 (or 155 pct elongation) are shown in Figures 8(a) and (b). Due to the much higher tensile elongation experienced in this 45 deg specimen, more diverse texture

components are observed. There are $\{111\}\langle\bar{3}41\rangle$ (a 45 deg rotation against $\{111\}\langle 2\bar{1}\bar{1}\rangle$), $\{001\}\langle 100\rangle$, and other multiple components in the β phase. And the $\{0001\}$ plane texture, $\{0001\}\langle\bar{4}310\rangle$ and $\{0001\}\langle 11\bar{2}0\rangle$ (a 45 deg rotation against $\{0001\}\langle\bar{4}310\rangle$) are seen in the α_2 phase. For the 90 deg specimen to a total failure strain of 0.32 (or 38 pct elongation) at 750 °C, a strong $\{111\}\langle 0\bar{1}1\rangle$ texture (a 90 deg rotation against $\{111\}\langle 2\bar{1}\bar{1}\rangle$) is present in the β phase. And strong $\{0001\}\langle 11\bar{2}0\rangle$ (a 90 deg rotation against $\{0001\}\langle\bar{1}100\rangle$), $\{11\bar{2}0\}\langle\bar{1}100\rangle$, $\{0001\}$ plane texture, and $\{11\bar{2}0\}\langle 0001\rangle$ are present in the α_2 phase.

As for the 0 deg specimen loaded at 920 °C to a much higher true strain of 2.9 (or 1680 pct elongation), multiple orientations are present in the β and α_2 phases and the texture intensities are greatly decreased (Figures 8(c) and (d)). Poles are distributed uniformly over the projections for the β phase, and there is almost no apparent preferred orientation. In comparison, weak and multiple orientations are present in the α_2 phase and even the poles are nearly randomly distributed over the $(11\bar{2}0)$ and $(10\bar{1}0)$ projections. Weak $\{0001\}\langle\bar{4}310\rangle$ and $\{0001\}$ plane textures can still be traced. For the 45 deg specimen loaded at 920 °C to a true strain of 3.0 (or 1910 pct elongation), the β grains become almost completely random oriented and the textures are difficult to define (Figure 8(c)), while the α_2 grains rotate and slide to result in the weak $\{0001\}$ plane and other extremely weak textures (Figure 8(d)). Similar results are obtained in the 90 deg specimen (true strain of 3.0 or 1880 pct elongation) at 920 °C. The orientation distribution becomes almost random in the β and α_2 phases,

with very weak $\{111\}\langle 0\bar{1}1\rangle$ in β and very weak $\{0001\}$ plane or $\{0001\}\langle 10\bar{1}0\rangle$ occasionally observed in α_2 .

b. *Texture evolution at 850 °C to $\varepsilon \sim 0.3, 1.0, 1.2$, and 1.5*

The gradual orientation evolution during superplastic deformation under the optimum LTSP condition at 850 °C and $5 \times 10^{-4} \text{ s}^{-1}$ is investigated. The tensile tests were stopped at different strain levels and the post-SP 0 deg specimens were examined. Figure 9 shows the texture evolution. In the initial straining stage to $\varepsilon = 0.3$, the $\{111\}\langle 2\bar{1}\bar{1}\rangle$ and $\{100\}\langle 011\rangle$ with a slight deviation of around ± 10 deg, which remain from the as-received state, are both strongly present. At $\varepsilon = 1.0$ -1.2, more β grains become concentrated at $\{111\}\langle 2\bar{1}\bar{1}\rangle$, but the $\{100\}\langle 011\rangle$ rotated cube texture still remains, along with some other minor components. At $\varepsilon = 1.5$, some other minor components appear, but weak $\{100\}\langle 011\rangle$, $\{111\}\langle 2\bar{1}\bar{1}\rangle$, and $\{110\}\langle\bar{1}10\rangle$ are still seen, and the β orientation becomes much more random. Upon increasing the strain, the β orientation continuously becomes more random due to gradual operation of GBS at later stages.

In the initial stage at $\varepsilon = 0.3$, the α_2 orientation still maintains the original $\{11\bar{2}0\}\langle\bar{1}100\rangle$ and becomes even stronger, accompanied by the weak $\{0001\}\langle\bar{1}100\rangle$ and $\{0001\}\langle 11\bar{2}0\rangle$ textures, not much different from the as-received specimen. At $\varepsilon = 1.0$ to 1.2, $\{11\bar{2}0\}\langle\bar{1}100\rangle$ becomes weakened and some of the $\{0001\}\langle\bar{1}100\rangle$ and $\{0001\}\langle 11\bar{2}0\rangle$ textures form the $\{0001\}$ plane texture. In addition, with the appearance of other weak texture components, the overall tex-

Fig. 9—Texture evolution in (a) the β and (b) the α_2 phases as a function of strain in the 0 deg specimens loaded at 850 °C and $5 \times 10^{-4} \text{ s}^{-1}$ to a true strain of $\varepsilon = 0.3$, $\varepsilon = 1.0$, $\varepsilon = 1.2$, and $\varepsilon = 1.5$ (failure). The tensile loading axis is along the horizontal direction.

ture intensities are weakened. At $\epsilon = 1.5$, multiple weak texture components are simultaneously present, and the $\{0001\}$ fiber orientation is no longer distinctly dominant. Orientation distributions become much more random (Figure 9(b)).

2. Grain misorientation distributions under various conditions

The grain orientation distribution could be further clarified from the misorientation angle distribution, which provides the overall angle distribution information. In general, the ranges of low-angle boundary (LAB), medium-angle boundary (MAB), and high-angle boundary (HAB) in the cubic system are often defined to be 0 to 10 deg, 10 to 30 deg, and greater than 30 deg, respectively. However, the boundary definition has not yet been well defined in the hexagonal system. The simulated random distributions in the cubic and hexagonal systems are shown in Figure 10.^[34,35] Assuming that the areas of LAB, MAB, and HAB occupied the same area proportions in the cubic and hexagonal systems, the LAB, MAB, and HAB could be roughly assigned to be 0 to 13 deg, 13 to 35 deg, and over 35 deg, respectively, in the completely random hexagonal system. The misorientation angle distributions in the α_2 and β phases in the as-received alloy, and the post-SP 0 deg specimen at the fracture tip loaded at 750 °C and 920 °C are shown in Figure 11. In the as-received alloy, LAB occupies a significant amount (69 pct) in the β phase, with a considerably low HAB fraction (13 pct), as listed in Table V. This is evident from the OIM images shown in Figure 7. For the α_2 phase, a bimodal distribution is displayed and an appreciable HAB fraction (46 pct) is observed due to the presence of multiple textures. This implies that α_2/α_2 is more available than β/β for GBS for the initial straining stage.

In the post-SP 0 deg specimen loaded at 750 °C, even though the texture evolves to result in a stronger $\{111\} \langle 2\bar{1}1 \rangle$ and a weaker $\{100\} \langle 011 \rangle$ in β , the LAB population (Figure 11) still dominates (decreases from the as-received 69 to 53 pct) yet with much higher HAB (31 pct). With the extensive dislocation motion, α_2 grains rotate to align the

$\{0001\}$ planes parallel to the rolling plane during straining, and the bimodal distribution becomes weakened in the α_2 phase. In comparison, in the post-SP 0 deg specimen loaded

Fig. 10—Random grain misorientation distributions in the cubic and hexagonal systems; redrawn from Refs. 34 and 35.

Fig. 11—Misorientation angle distributions of the β and α_2 phases (a) and (b) in the AR alloy, (c) and (d) in the post-SP 0 deg specimens loaded at 750 °C, and (e) and (f) in the post-SP 0 deg specimens loaded at 920 °C.

Table V. The Relative Fractions of the LAB, MAB, and HAB in Grain Misorientation Angle Distribution Histograms for the Simulated Completely Random Case, As-Received Specimen, and Post-SP 0 Deg Specimens Loaded at 750 °C ($\epsilon = 0.3$), 850 °C ($\epsilon = 1.5$), and 920 °C ($\epsilon = 2.9$)

Specimen Condition	β Phase			α_2 Phase		
	LAB	MAB	HAB	LAB	MAB	HAB
Completely random	1 pct	18 pct	82 pct	1 pct	18 pct	82 pct
As-received	69 pct	18 pct	13 pct	29 pct	25 pct	46 pct
Post-SP at 750 °C	53 pct	16 pct	31 pct	29 pct	28 pct	43 pct
Post-SP at 850 °C	22 pct	18 pct	60 pct	23 pct	24 pct	53 pct
Post-SP at 920 °C	12 pct	21 pct	67 pct	19 pct	13 pct	68 pct

at 920 °C, phase and GBS have proceeded to maintain the equiaxed shape and to yield a nearly random misorientation distribution (Figure 11). The HAB fractions for the β and α_2

Fig. 12—Misorientation angle distribution of the β and α_2 phases for $\epsilon \sim 0.3, 1.0, 1.2$, and 1.5 in the post-SP 0 deg specimen loaded at 850 °C and $5 \times 10^{-4} \text{ s}^{-1}$.

Table VI. The Misorientation Angle Distribution in the Theoretical Random Case, and the Post-SP 0 Deg Specimen Performed at 850 °C and $5 \times 10^{-4} \text{ s}^{-1}$ to Different True Strains ϵ

Specimen Condition	β Phase			α_2 Phase		
	LAB	MAB	HAB	LAB	MAB	HAB
Random distribution	1 pct	18 pct	82 pct	1 pct	18 pct	82 pct
$\epsilon = 0.3$	56 pct	6 pct	38 pct	19 pct	32 pct	49 pct
$\epsilon = 1.0$	27 pct	30 pct	43 pct	20 pct	27 pct	53 pct
$\epsilon = 1.2$	22 pct	9 pct	69 pct	28 pct	18 pct	54 pct
$\epsilon = 1.5$	22 pct	18 pct	60 pct	23 pct	24 pct	53 pct

phases reach 67 and 68 pct, respectively, both approaching the ideal fraction for completely random distribution (82 pct).

The gradual evolution of misorientation distribution for the α_2 and β phases as a function of true strain from 0 to 1.5 was traced in the specimens loaded at 850 °C and $5 \times 10^{-4} \text{ s}^{-1}$. The variations of the LAB and HAB fraction are shown in Figure 12 and Table VI. An apparent decreasing trend for LAB (from 69 to 22 pct) and an increasing trend for HAB (from 13 pct 60 pct) can be seen for the β phase; whereas only moderate evolution is observed for the α_2 phase (LAB: from 29 to 23 pct; HAB: from 46 to 53 pct). This is mainly due to the fact that the initial α_2 misorientation angle distribution in the as-received specimen was already more random; the α_2 grains were more randomly dispersed in the β matrix.

IV. DISCUSSION

A. Texture Evolution during Processing and SP Loading

The rolling texture in the β phase was also investigated by Suwas and Ray.^[30] The as-cast Ti_3Al rolled at 1100 °C possessed strong multiple $\{011\} \langle 100 \rangle$, $\{011\} \langle 011 \rangle$, and $\{013\} \langle 110 \rangle$ texture components, plus other orientations such as $\{112\} \langle 111 \rangle$, $\{112\} \langle 110 \rangle$, $\{011\} \langle 110 \rangle$, and $\{001\} \langle 230 \rangle$. Knorr and Stoloff^[27] reported that the $\{200\}$ cubic texture would appear through recrystallization. In the present study, the strong $\{100\} \langle 011 \rangle$ rotated cube texture should be a result of recrystallization, while the $\{111\} \langle 2\bar{1}\bar{1} \rangle$ and $\{110\} \langle 001 \rangle$ textures should be the retained ones originated from thermomechanical treatment involving complicated direct- and cross-rolling practices.^[36] From the literature,^[26,28] it was suggested that the rolling texture in the α_2 phase is mainly the basal $\{0001\}$ fiber texture. Similarly, in our current study, the basal texture is also retained from the rolling texture and deviated the perfect $(0001) \langle uvw \rangle$ within 10 deg. The strong $\{11\bar{2}0\} \langle \bar{1}100 \rangle$, $\{0001\} \langle 11\bar{2}0 \rangle$ and $\{0001\} \langle \bar{1}100 \rangle$ basal textures, as well as $\{11\bar{2}0\} \langle 0001 \rangle$, and other multiple non-basal textures coexisting in our starting material should result from recrystallization.

After superplastic loading at 750 °C, the orientation distributions in the β phase become more random than those in the α_2 phase during LTSP. Some of the β grains concentrate rapidly into the $\{111\} \langle 2\bar{1}\bar{1} \rangle$ orientation, a characteristic deformation texture resulting from dislocation deformation in bcc structures.^[36] Since the major deformation mode is postulated to be governed by dislocation slip during the dislocation creep regime at 750 °C, several slip systems are activated simultaneously in this bcc type structure. Subsequently, the β grains concentrate into the $\{111\} \langle 2\bar{1}\bar{1} \rangle$ orientation. Upon increasing the loading temperature to 850 °C, or even 920 °C, GBS is thought to become increasingly dominant, and the $\{111\} \langle 2\bar{1}\bar{1} \rangle$ intensity is decreased. At a later straining stage at 920 °C, there is basically no apparent preferred orientation.

As for the α_2 grains, the $\{11\bar{2}0\} \langle 0001 \rangle$ recrystallization texture present in the starting material becomes weakened upon loading at 750 °C. The α_2 grains have a tendency to lead the (0001) plane parallel to the rolling plane (similar to the situation during rolling), and the near $\langle \bar{1}100 \rangle$ orientation, or more precisely the $\langle \bar{4}130 \rangle$ orientation, tends to align along the tensile axis. With increasing loading temperature to 850 °C, or even to 920 °C, $\{11\bar{2}0\} \langle 0001 \rangle$ soon becomes diminished

with minor basal type textures ($\{0001\}$ plane, $\{0001\} \langle \bar{4}310 \rangle$, $\{0001\} \langle 11\bar{2}0 \rangle$, and $\{0001\} \langle 10\bar{1}0 \rangle$). At a later straining stage at 920 °C, even the basal textures evolve to extremely weak.

It should be noted that, at 850 °C and $\varepsilon \sim 0.3$, the initial increment of the $\{11\bar{2}0\} \langle \bar{1}100 \rangle$ texture, rather than the $\{0001\}$ type textures, is a result of the β - α_2 transformation at this intermediate temperature (800 °C to 850 °C) originated by the driving force to approaching the equilibrium phase partition, as discussed thoroughly in our previous article.^[32] The diffusion rate is too sluggish to undergo this diffusion-controlled transformation at lower temperatures of 700 °C to 750 °C, and the driving force is too small at the higher temperatures of 900 °C to 1000 °C since the existing phases are already close to the equilibrium predictions. Previous TEM and EBSD studies^[32] have confirmed the Burgers relationship, namely, $\{011\}_\beta // \{0001\}_{\alpha_2}$ and $\langle 111 \rangle_\beta // \langle 11\bar{2}0 \rangle_{\alpha_2}$. Under this orientation relationship, the initial increment of $\{11\bar{2}0\} \langle \bar{1}100 \rangle$ in α_2 possesses the direct relationship with the $\{111\} \langle 2\bar{1}\bar{1} \rangle$ texture in β . The latter is a result of dislocation slip deformation. It appears that the newly transformed α_2 phase, in accordance with the Burgers relationship, exhibits the initially strong $\{11\bar{2}0\} \langle \bar{1}100 \rangle$ texture. Upon increasing the strain, dislocation activities and GBS in α_2 result in a weak $\{0001\}$ plane texture and other more random orientations.

In summary, grain rotations have operated at 750 °C through dislocation activities and dynamic recrystallization (with minor GBS) for all of the 0, 45, and 90 deg specimens. Due to the shorter c -axis of the DO_{19} structure, the α_2 grains exhibit a greater tendency to align $\{0001\}$ parallel to the rolling plane. Under the severe dislocation actions, some of the β grains concentrate into $\{111\} \langle 2\bar{1}\bar{1} \rangle$ orientation during loading at such lower temperatures. At 920 °C, the textures are much more weakened in both β and α_2 phases after pronounced SP deformation through the smooth operation of GBS.

The current EBSD software would automatically classify the texture intensities into eight relative contour levels, as shown in Figures 6 through 9, independent of the absolute intensity value against background. It sometimes becomes difficult to differentiate the texture intensities in various specimens. This can be overcome by judging the sharpness of the contours in color prints. Table VII presents the rating. The texture contour sharpness in the lower temperature specimens is always higher than that in the higher temperature ones.

B. Influence from Textures on the Mechanical Anisotropy

Grains with equiaxed shape are suitable for GBS and would not give rise to pronounced superplastic anisotropy. On the other hand, elongated and hard grains acting as barriers to GBS would give rise to superplastic anisotropy.^[25] In the current starting alloy, the grain structures are mostly nearly equiaxed; the microstructure grain shape effect on the mechanical anisotropy is postulated to be minor.

Dislocation mobility would significantly affect the ductility at temperatures below 800 °C (under the dislocation creep condition), and during the initial stage of the deformation at higher temperatures (under the superplastic flow condition). To consider the dislocation mobility in the β and α_2 phases, the Schmid factor is an important indicator.^[37] Based on our

Table VII. The Textures Sharpness Measured from the AR Alloy and the Post-SP Specimens

Specimen Condition		Contour Level in Average	
		α_2 Phase	β Phase
As-received alloy		sharp	extremely sharp
Post-SP			
specimens	0 deg	sharp	sharp
at 750 °C	45 deg	sharp	highly sharp
($\varepsilon = 0.3$)	90 deg	sharp	sharp
Post-SP			
specimens	0 deg	highly flat	extremely flat
at 920 °C	45 deg	highly flat	extremely flat
($\varepsilon = 2.9$)	90 deg	highly flat	extremely flat
Post-SP			
specimens	$\varepsilon = 0.3$	sharp	highly sharp
at 850 °C	$\varepsilon = 1.0$	sharp	sharp
(0 deg)	$\varepsilon = 1.2$	flat	flat
	$\varepsilon = 1.5$	flat	flat

Table VIII. Schmid Factors Calculated for the β Phase

Orientation to Rolling Direction	Schmid Factors of the β Phase		
	Major Textures		Minor Texture
	$\{100\} \langle 110 \rangle$	$\{111\} \langle 2\bar{1}\bar{1} \rangle$	$\{0\bar{1}1\} \langle 100 \rangle$
Acting on the $\{110\} \langle \bar{1}11 \rangle$ Slip System			
0 deg	0.14	0.18	0.27
45 deg	0.27	0.17	0.17
90 deg	0.14	0.14	0.14

previous TEM microstructure characterization,^[38] abundant dislocations were left within the β grains implying the dislocation activities during deformation at 700 °C to 800 °C. It is well known that dislocations would tend to slip along the system with a higher Schmid factor. The bcc β phase is more deformable than the hexagonal α_2 phase. The dislocation motion and accommodation process would be operative more readily in the β phase. In addition, the Schmid factor effect in the β phase may be more important than that in the α_2 phase due to the higher initial volume fraction (~ 60 pct) of the β phase. In this study, only the simplistic calculation is made for the applied uniaxial stress. The complicated triaxial stresses, possibly induced by the thermal expansion mismatch or the internal stress in the two-phase material, are not considered. Since the deformation occurs at elevated temperatures, such internal stresses are postulated to be minor.

For the bcc β phase, the Schmid factor analysis was only carried out on the dislocations acting on the main $\{110\} \langle \bar{1}11 \rangle$ slip system. Other slip systems, *i.e.*, $\{211\} \langle \bar{1}11 \rangle$ and $\{321\} \langle \bar{1}11 \rangle$ could also offer dislocation slip, but they are usually less favorable than the main slip system. For the texture existing in β , *i.e.*, $\{100\} \langle 011 \rangle$, $\{111\} \langle 2\bar{1}\bar{1} \rangle$, and $\{0\bar{1}1\} \langle 001 \rangle$, the overall Schmid factors are highest when the stress acts along the 45 deg direction against the RD direction (Table VIII), leading to a lower flow stress and the highest elongation. There are 12 slip systems of the type in the β phase. If one set of dislocations were blocked while loading, dislocations would easily continue to cross-slip into another set of slip systems.

Table IX. Schmid Factors Calculated for the α_2 Phase

Orientation to Rolling Direction	Schmid Factors of the α_2 Phase											
	Major Texture			Minor Textures								
	{11 $\bar{2}$ 0}<0001>			{0001}< $\bar{1}$ 100>			{0001}<11 $\bar{2}$ 0>			{11 $\bar{2}$ 0}< $\bar{1}$ 100>		
	Acting on the Main Slip Systems											
	{10 $\bar{1}$ 0} <11 $\bar{2}$ 0>	{0001} <11 $\bar{2}$ 0>	{10 $\bar{1}\bar{1}$ } <11 $\bar{2}$ 0>	{10 $\bar{1}$ 0} <11 $\bar{2}$ 0>	{0001} <11 $\bar{2}$ 0>	{10 $\bar{1}\bar{1}$ } <11 $\bar{2}$ 0>	{10 $\bar{1}$ 0} <11 $\bar{2}$ 0>	{0001} <11 $\bar{2}$ 0>	{10 $\bar{1}\bar{1}$ } <11 $\bar{2}$ 0>	{10 $\bar{1}$ 0} <11 $\bar{2}$ 0>	{0001} <11 $\bar{2}$ 0>	{10 $\bar{1}\bar{1}$ } <11 $\bar{2}$ 0>
0 deg	0	0	0	0.29	0	0.20	0.29	0	0.20	0.29	0	0.20
45 deg	0.15	0.29	0.21	0.33	0	0.23	0.33	0	0.23	0.15	0.29	0.21
90 deg	0.29	0	0.20	0.29	0	0.20	0.29	0	0.20	0	0	0

The main slip systems in the DO_{19} α_2 phase include mainly (1) prismatic {10 $\bar{1}$ 0}<11 $\bar{2}$ 0>, (2) basal {10 $\bar{1}$ 0}<11 $\bar{2}$ 0>, and (3) pyramidal {10 $\bar{1}$ $\bar{1}$ ><11 $\bar{2}$ 0>. In the Ti₃Al alloy, the primary slip system is the <a> type dislocation acting on the prism plane, i.e., {10 $\bar{1}$ 0}<11 $\bar{2}$ 0> since the *c/a* ratio is less than the ideal value.^[10,39] From the calculated Schmid factor values listed in Table IX, the Schmid factors for the α_2 phase in average are also highest in the 45 deg specimens, also favoring a lower flow stress and the highest elongation. It could be seen that the Schmid factors in the major textures in the 0 deg specimen are zero. As a result, dislocation motions could only occur in the grains of minor textures or after the α_2 grains rotate into a position with the (0001) plane parallel to the rolling plane.

The {0001} basal texture formation during tensile loading might be similar to that in the basal rolling texture. The <11 $\bar{2}$ 0> <a> type dislocations can slip on the {0001}, {10 $\bar{1}$ 0}, and {10 $\bar{1}$ $\bar{1}$ } planes, leading to the formation of the basal texture.^[40] After the α_2 grains rotate into the {0001} <uvw> orientation, the Schmid factors increase. As tabulated in Table IX, the Schmid factors of <a> type dislocations acting on the three main slip systems in the basal {0001} < $\bar{1}$ 100> and {0001}<11 $\bar{2}$ 0> textures are greater than those in the {11 $\bar{2}$ 0}<0001> and {11 $\bar{2}$ 0}< $\bar{1}$ 100> textures. Consequently, once the basal type textures are formed in the α_2 grains during tensile loading, dislocation glide and climb can proceed more thoroughly until fracture over the dislocation creep dominating stage at 700 °C to 850 °C.

At 900 °C to 1000 °C under the smooth GBS stage, the rotation of α_2 grains is continuously operated and the orientations are distributed more and more randomly. However, the texture effects should still be considered even in the superplastic flow condition.^[37] The textured Ti alloys have been shown to exhibit slightly greater ductility as compared with the textureless alloy under the GBS-controlling regime. Nevertheless, the existing texture can only impose influence during the initial stage; the extensive GBS would eventually lead to random grain orientation at a later stage. The mechanical anisotropy can at most account for ± 10 pct in flow stress or SP elongation under the optimum superplastic region.

V. CONCLUSIONS

1. The super α_2 alloy exhibited tensile elongations of 1500 pct at 960 °C, 600 pct at 900 °C, 330 pct at 850 °C, and 140 pct at 750 °C. Higher elongations are obtained in the 45 deg specimens as they are tensile loaded at 25 °C to 960 °C.

2. In the as-received alloy, strong {100}<011> rotated cube, strong {111}<2 $\bar{1}$ $\bar{1}$ >, and weaker {110}<001> Goss textures are present in the β phase; strong {11 $\bar{2}$ 0}<0001> and the minor {0001} fiber, {11 $\bar{2}$ 0}< $\bar{1}$ 100>, {0001}< $\bar{1}$ 100>, and {0001}<11 $\bar{2}$ 0> are present in the α_2 phase.
3. After superplastic loading at 750 °C to 920 °C, the {11 $\bar{2}$ 0}<0001> texture present in the α_2 phase of the starting material becomes extremely weaker, and the (0001) plane tends to lie parallel to the rolling plane and < $\bar{2}$ 130>, < $\bar{4}$ 130>, or < $\bar{1}$ 100> directions tend to align along the tensile direction.
4. After LTSP straining, the orientations in the β phase are distributed more randomly than the α_2 phase and some β grains tend to concentrate into the {111}<2 $\bar{1}$ $\bar{1}$ > orientation.
5. The microstructure effects to the mechanical anisotropy appear to be minor due to the equiaxed and some randomly dispersed elongated α_2 grains in the as-received specimen.
6. The Schmid factors for the 45 deg specimens are the highest in the α_2 and β phases. This provides the reasonable origin for the mechanical anisotropy. The anisotropy behavior in terms of tensile elongation shows strong temperature dependence, being most pronounced at 750 °C and gradually diminishing with increasing loading temperature.

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