High temperature deformation of Ti–(46–48)Al–2W intermetallic compounds

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Abstract

The high temperature deformation behavior of Ti–46Al–2W and Ti–48Al–2W intermetallic compounds have been investigated in isothermal compressive tests, performed at temperatures between 1000°C and 1200°C for strain rates between $10^{-3}$ and $10^{-1}$ s$^{-1}$. The stress–strain curve during high temperature deformation exhibits a peak stress which is followed by a gradual decrease into a steady state stress with increasing the strain. The flow softening behavior after the peak stress is attributed to the effects of dynamic recrystallization during deformation. The dependence of flow stress on temperature and strain rate followed a hyperbolic sine relationship using the Zener-Hollomon parameter. The activation energies, $Q$, were measured as 449 kJ mol$^{-1}$ and 394 kJ mol$^{-1}$, and the stress exponents were measured as 3.6 and 3.7 for Ti–46Al–2W and Ti–48Al–2W, respectively. The activation energy increased with decreasing Al content in TiAl-base intermetallic compounds. The coefficient between peak stress and Zener-Hollomon parameter, $A$, was not a constant, but was dependent on the activation energy. The peak stresses can be predicted well by using a normalized Zener-Hollomon parameter. The dynamic recrystallization rate and recrystallized grain size increased with increasing the temperature and with decreasing the strain rate. © 1998 Elsevier Science S.A. All rights reserved.

Keywords: High temperature deformation; TiAl-base intermetallic compounds; Isothermal compressive tests; Dynamic recrystallization

1. Introduction

TiAl-base intermetallic compounds have been investigated for aerospace engine components due to their attractive properties such as low density, good elevated temperature strength, high resistance to oxidation and excellent creep properties [1–3]. However, the poor ductility and low fracture toughness of TiAl-base intermetallic compounds at ambient temperature are two of the major limitations to practical application. It has been reported that the ductility and toughness at ambient temperature are very sensitive to the microstructure [1–10]. This has led to significant research efforts designed to improve the ductility and fracture toughness by microstructure control and alloy design [1–10].

Various thermomechanical treatments have been used for the homogenization, grain refinement and microstructure control. Duplex and lamellar microstructures have also been developed in effort to optimize the required mechanical properties [4–10]. Thermomechanical treatment can be generally divided into two steps of hot working process and subsequent heat treatment process. The final microstructure is determined by the microstructure evolution during the thermomechanical process. There have been extensive studies on the microstructure evolution during hot working of TiAl-base intermetallic compounds as a function of temperature and strain rate [11–22].

It is reported that the addition of W in TiAl-base intermetallic compound improves the high temperature strength [23], creep resistance [24] and oxidation resistance [25]. As the W decreases the stacking fault energy (SFE), the addition of W reduces the climb rate [24]. At the same time, the low diffusivity of W solute atoms reduces the kinetic of diffusion controlled deformation process [26]. Fuchs [27] reported that the strength and creep resistance increased by the addition of W in powder metallurgy (PM) processed Ti–48Al–2Cr–2Nb, however decreased in ingot metallurgy (IM) pro-
cessed Ti–48Al–2Cr–2Nb. It is explained that the inferior high temperature mechanical properties of IM processed Ti–48Al–2Cr–2Nb was attributed to the microstructural inhomogeneity caused by W segregation to dendrite cores. Martin et al. [28] also reported that the segregation of W in dendrite cores in cast ingots. The segregation was retained after homogenization heat treatment and did not show a significant improvement after forging.

The thermomechanical process is very important to obtain the homogeneous microstructure for improved mechanical properties. There have been some studies [19] on the high temperature deformation behavior and microstructure evolution of TiAl–W intermetallic compounds produced by powder metallurgy, however there have been no systematic study on high temperature deformation behavior and microstructure evolution during hot working process of TiAl–W produced by ingot metallurgy.

In this study, the high temperature deformation behavior of Ti–46Al–2W and Ti–48Al–2W, fabricated by ingot metallurgy process, was investigated. The effects of deformation temperature, strain rate and flow stress on microstructure has been analyzed. The flow curves obtained for ingot metallurgy Ti–48Al–2W are compared with those of a powder metallurgy alloy of same composition.

2. Experimental procedure

The Ti–46Al–2W and Ti–48Al–2W intermetallic compounds were prepared by plasma-arc-melting in a cold copper hearth under static argon atmosphere. Cylindrical ingots with 15 mm in diameter and 50 mm in length were melted in a plasma arc melting furnace under a condition of 20 V/250 A. The melted ingots were sealed in quartz tube filled with argon, and then homogenized at 1250°C for 24 h. Cylindrical compressive specimens, with a diameter of 8 mm and a height of 12 mm, were machined from the homogenized ingots by electro-discharge machine. The high temperature compression tests were conducted in vacuum of $10^{-1}$ torr at temperatures between 1000°C and 1200°C with constant strain rates between $10^{-3}$ s$^{-1}$ and $10^{-1}$ s$^{-1}$. Specimens were heated by induction coils with heating rate of 5°C min$^{-1}$ and soaked for 300 s at test temperatures before performing the compression tests. The true stress–true strain curves were obtained from the load-displacement data. In order to investigate the microstructural evolution during the deformation, the specimens were quenched from the test temperatures by flowing liquid nitrogen immediately after deforming to various true strains up to 1.2. The microstructures of the deformed specimens were observed. Optical micrographs were obtained from the cross-sectional surface of the deformed specimens, cut parallel to the compression axis. The cut surfaces were ground by emery paper and polished by diamond paste, and then etched with Kroll’s reagent (1 ml HF + 3 ml HNO$_3$ + 16 ml H$_2$O). The dynamically recrystallized grain sizes were measured at various test temperatures and strain rates from the optical micrographs.

3. Results and discussion

The chemical composition of the ingots are shown in Table 1. Table 1 show that the actual composition is well consistent with the nominal composition. The microstructures of Ti–46Al–2W and Ti–48Al–2W homogenized at 1250°C are shown in Fig. 1. The optical micrograph of Ti–46Al–2W shows the near lamellar structure. The lamellar structure consisted of three phases. These were identified as γ phase, α phase and W-rich β phase from the X-ray diffraction (XRD) analysis. Semi-quantitative estimates of the compositions of each phase were obtained via energy dispersion spectroscope (EDS). The results are shown in Table 2. The amount of W in β phase was 7–8 times higher compared to that in γ phase. W was slightly enriched in α phase compare with γ phase. These results indicate that the segregation of W stabilizes the β phase. This result is consistent with the previous results that Cr, Mo, W stabilized the β phase in TiAl alloys by lowering α/β transus temperature [29]. An optical micrograph of Ti–48Al–2W shows the near γ structure with an average grain size of about 75 μm. The back scattered electron micrograph shows that the W-rich β phase was segregated to the dendrite region. The W-rich β phase was formed by the segregation of W into dendrite cores during non-equilibrium peritectic solidification [28].

The stress–strain curves obtained from the high temperature compressive tests of homogenized Ti–46Al–2W and Ti–48Al–2W are shown in Fig. 2. The flow

| Table 1 | The actual compositions of Ti–46Al–2W and Ti–48Al–2W ingot |
|-----------------|---|---|---|---|---|---|---|
| Alloy composition (at%) | Ti | Al | W | C | O | H | N |
| Ti–46Al–2W | 52.0 | 45.8 | 1.9 | 0.01 | 0.22 | 0.04 | 0.02 |
| Ti–48Al–2W | 49.9 | 47.9 | 1.9 | 0.01 | 0.22 | 0.04 | 0.02 |
stress exhibited a peak stress, then the flow stress decreased gradually to a steady state stress with increasing the strain. It has been reported that the flow softening is caused by dynamic recrystallization during high temperature deformation in TiAl-base intermetallic compounds [11–22]. The degree of flow softening generally increases with decreasing temperature and increasing strain rate. The stress–strain curves were similar for Ti–46Al–2W and Ti–48Al–2W, as shown in Fig. 2. However, the flow stress of Ti–48Al–2W alloy was slightly higher than that of Ti–46Al–2W.

The variation of peak flow stress with varying temperature and strain rate of Ti–46Al–2W and Ti–48Al–2W are shown in Fig. 3. The dependence of peak flow stress on strain rate at a fixed temperature is expressed in Eq. (1).

$$\sigma_p = K\dot{\varepsilon}^m$$  \hspace{1cm} (1)

where $\sigma_p$ is peak flow stress, $\dot{\varepsilon}$ is strain rate, $K$ is constant and $m$ is strain rate sensitivity. The strain rate sensitivities, $m$, which is known as $d \log \sigma_p/d \log \dot{\varepsilon}$, were obtained from the slope of curves in Fig. 3. The strain rate sensitivities were measured as 0.13, 0.18 and 0.21 in Ti–46Al–2W and 0.15, 0.18 and 0.23 in Ti–48Al–2W at 1000, 1100 and 1200°C, respectively. The dependence of strain rate and temperature on the flow stress during high temperature deformation can be described by the power-law relationship at low stress regime and exponential relationship at high stresses regime as following,

$$Z = \dot{\varepsilon} \exp(Q/RT) = A'\sigma^n$$ : low $\sigma$ regime  \hspace{1cm} (2)

$$Z = \dot{\varepsilon} \exp(Q/RT) = A'' \exp(\beta\sigma)$$ : high $\sigma$ regime  \hspace{1cm} (3)

where $Z$ is Zener-Hollomon parameter, $Q$ is activation energy, $n$ is stress exponent, and $A'$, $A''$, $\beta$ are constants. These equations could be combined into a hyperbolic sine relationship as following,

$$Z = \dot{\varepsilon} \exp(Q/RT) = A \{\sinh(x\sigma)\}^n$$  \hspace{1cm} (4)

which reduces to Eq. (2) in the low stress regime when $x\sigma < 0.8$, and reduces to Eq. (3) in the high stress regime.
Fig. 2. True stress–true strain curves of Ti–46Al–2W and Ti–48Al–2W obtained during the compression tests at 1000°C (a) and 10^{-3} s\(^{-1}\) (b).

regime when z\sigma > 1.2. The apparent activation energies were measured as 449 kJ mol\(^{-1}\) and 394 kJ mol\(^{-1}\) for Ti–46Al–2W and Ti–48Al–2W, respectively. The measured activation energies are comparable to the reported values for high temperature deformation of TiAl-base alloys alloy showing flow softening behavior such as 410 kJ mol\(^{-1}\) in Ti–47Al–1V [13], 465 kJ mol\(^{-1}\) in Ti–47Al–2V [11], 343 kJ mol\(^{-1}\) in Ti–48Al [16], 355 kJ mol\(^{-1}\) in Ti–48Al–2Cr–2Nb [15], 327 kJ mol\(^{-1}\) in Ti–49.5Al–2.5Nb–1.1Mn [14] and 417 kJ mol\(^{-1}\) in Ti–45.5Al–2Nb–2Cr [21]. Seetharaman and Lombard [14] reported that the significant amount of flow softening is observed in the temperature range 1000–1250°C due to the occurrence of dynamic recrystallization in Ti–49.5Al–2.5Nb–1.1Mn. Nobuki et al. [11,16] reported that the dynamic recrystallization occurred during the compressive deformation of Ti–(43–52)Al, although the recrystallization was incomplete for fully lamellar structure. Also, Fujitsuna et al. [13] and Shih and Scarr [15] reported that the dynamically recrystallized structure was observed in Ti–47Al–1V and Ti–48Al–2Cr–2Nb after deformation in the temperature range 1000–1200°C. The measured constants and activation energies are listed in Table 3. The calculated values of z\sigma are in the range between 0.5–2.9 for Ti–46Al–2W, and 0.4–2.5 for Ti–48Al–2W. The transition stress between the power-law relationship and the exponential relationship was calculated as 229 MPa and 278 MPa in Ti–46Al–2W and Ti–48Al–2W, respectively.

The relationships between the measured flow stress and the calculated Zener-Hollomon parameter using

Table 3

<table>
<thead>
<tr>
<th>Alloy composition (at%)</th>
<th>A</th>
<th>x</th>
<th>n</th>
<th>Q (kJ mol(^{-1}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti–46Al–2W</td>
<td>1.31 × 10^{14}</td>
<td>4.37 × 10^{-3}</td>
<td>3.6</td>
<td>449</td>
</tr>
<tr>
<td>Ti–48Al–2W</td>
<td>2.25 × 10^{12}</td>
<td>3.60 × 10^{-3}</td>
<td>3.7</td>
<td>394</td>
</tr>
</tbody>
</table>
the obtained parameters of $A$, $x$, $n$ and $Q$ are shown in Fig. 4. Fig. 4 shows that the flow stresses of Ti–46Al–2W and Ti–48Al–2W are fitted well by the hyperbolic sine relationship. The measured parameters related to the hyperbolic sine relationship were compared to the previous results showing flow softening in Table 4. $Q$ in Table 4 is the activation energy for flow softening during high temperature deformation. Table 4 show that the stress exponent, $n$, and constant, $x$, were almost similar, however the activation energy, $Q$, and constant, $A$, were sensitively varied with varying the alloy composition and microstructure. The average stress exponent, $n$, and $x$ of TiAl-base intermetallic compounds are calculated as $3.5 \pm 0.4$ and $4.5 \pm 0.8 \times 10^{-3}$ MPa$^{-1}$. It is noted that the activation energy increased with decreasing Al content and with increasing lamellar volume fraction. When the initial microstructure is $\gamma$ or near $\gamma$, the apparent activation

### Table 4

The material constants of TiAl-base intermetallic compounds during high temperature compression deformation Q is the activation energy for flow softening.

<table>
<thead>
<tr>
<th>Composition</th>
<th>Initial Structure</th>
<th>T (°C)</th>
<th>$\dot{\epsilon}$ (s$^{-1}$)</th>
<th>$A$ (10$^{-13}$)</th>
<th>$x$ (10$^{-1}$)</th>
<th>$n$</th>
<th>$Q$ (kJ mol$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti–43Al [11]</td>
<td>Lamellar</td>
<td>927–1203</td>
<td>7.5 × 10$^{-4}$ – 7.5 × 10$^{-1}$</td>
<td>1.67 × 10$^{15}$</td>
<td>4.95</td>
<td>2.94</td>
<td>528</td>
</tr>
<tr>
<td>Ti–47Al–2W [11]</td>
<td>Lamellar</td>
<td>927–1203</td>
<td>7.5 × 10$^{-4}$ – 7.5 × 10$^{-1}$</td>
<td>3.60 × 10$^{15}$</td>
<td>3.61</td>
<td>3.60</td>
<td>465</td>
</tr>
<tr>
<td>Ti–51Al [11]</td>
<td>$\gamma$</td>
<td>927–1203</td>
<td>7.5 × 10$^{-4}$ – 7.5 × 10$^{-1}$</td>
<td>1.53 × 10$^{15}$</td>
<td>4.90</td>
<td>3.63</td>
<td>416</td>
</tr>
<tr>
<td>Ti–52Al [11]</td>
<td>$\gamma$</td>
<td>927–1203</td>
<td>7.5 × 10$^{-4}$ – 7.5 × 10$^{-1}$</td>
<td>6.33 × 10$^{12}$</td>
<td>4.56</td>
<td>3.74</td>
<td>398</td>
</tr>
<tr>
<td>Ti–47Al–1V [13]*</td>
<td>Near lamellar</td>
<td>1000–1200</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{9}$</td>
<td>2.93 × 10$^{15}$</td>
<td>3.69</td>
<td>3.8</td>
<td>404</td>
</tr>
<tr>
<td>Ti–43.8Al [16]</td>
<td>Lamellar</td>
<td>927–1323</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>1.41 × 10$^{25}$</td>
<td>6.32</td>
<td>3.13</td>
<td>672</td>
</tr>
<tr>
<td>Ti–44.9Al [16]</td>
<td>Lamellar</td>
<td>927–1323</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>1.30 × 10$^{16}$</td>
<td>4.52</td>
<td>3.74</td>
<td>496</td>
</tr>
<tr>
<td>Ti–48.2Al [16]</td>
<td>Duplex</td>
<td>927–1323</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>5.52 × 10$^{16}$</td>
<td>4.01</td>
<td>3.70</td>
<td>343</td>
</tr>
<tr>
<td>Ti–49.5Al [16]</td>
<td>$\gamma$</td>
<td>927–1323</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>3.68 × 10$^{16}$</td>
<td>5.57</td>
<td>3.03</td>
<td>330</td>
</tr>
<tr>
<td>Ti–50.2Al [16]</td>
<td>$\gamma$</td>
<td>927–1323</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>1.09 × 10$^{16}$</td>
<td>4.79</td>
<td>3.57</td>
<td>354</td>
</tr>
<tr>
<td>Ti–49.5Al–2.5Nb–1.1Mn [14]</td>
<td>Near $\gamma$</td>
<td>1000–1250</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>8 × 10$^{10}$</td>
<td>3.2</td>
<td>3.9</td>
<td>327</td>
</tr>
<tr>
<td>Ti–48Al–2Cr–2Nb [15]*</td>
<td>Duplex</td>
<td>975–1200</td>
<td>3 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>2.7 × 10$^{9}$</td>
<td>5.38</td>
<td>2.7</td>
<td>324</td>
</tr>
<tr>
<td>Ti–47Al–2Cr–4Nb [22]</td>
<td>Near $\gamma$</td>
<td>1000–1200</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>4.26 × 10$^{9}$</td>
<td>4.24</td>
<td>3.1</td>
<td>295</td>
</tr>
<tr>
<td>Ti–46Al–2W</td>
<td>Near lamellar</td>
<td>1000–1200</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>1.31 × 10$^{10}$</td>
<td>4.37</td>
<td>3.6</td>
<td>449</td>
</tr>
<tr>
<td>Ti–48Al–2W</td>
<td>Near $\gamma$</td>
<td>1000–1200</td>
<td>1 × 10$^{-3}$ – 1 × 10$^{-1}$</td>
<td>2.25 × 10$^{12}$</td>
<td>3.60</td>
<td>3.7</td>
<td>394</td>
</tr>
</tbody>
</table>

* Recalculated by Eq. (4).
energy was measured as $350 \pm 40 \text{ kJ mol}^{-1}$ and was similar to the activation energy for creep deformation of TiAl-base intermetallic compounds [30]. However, when the initial microstructure is lamellar or near lamellar, the activation energy increased with decreasing Al content and increasing the amount of $\alpha$ phase. The general linear relationship between the activation energy, $Q$, and the ratio of Ti to Al content, Ti/Al, exists as shown in Fig. 5. The relationship in $\gamma + \alpha$ two phase region can be formulated as following equation from the correlation in Fig. 5.

$$Q (\text{kJ mol}^{-1}) = -460 + 800 (\text{Ti/Al})$$  \hspace{1cm} (5)

The activation energy for flow softening of stoichiometric composition is estimated as $340 \text{ kJ mol}^{-1}$ from Eq. (5), and is consistent well with the previous experimental results [16]. It is well established that the flow softening of $\gamma$ single phase or near $\gamma$ phase is due to dynamic recrystallization of $\gamma$ phase [11–22]. Therefore, the measured activation energy for flow softening correspond to that for initiate dynamic recrystallization. The reason for the higher activation energy in lamellar structure is still unclear, but it is suggested that the flow softening mechanism is different in lamellar structure. This means that the Zener-Hollomon parameter is a function of not only strain rate and temperature but the chemical composition and microstructure. From Eq. (3), the $\sigma_p$ is expressed as:

$$\sigma_p = \frac{1}{\alpha} \sinh^{-1} \left[ \frac{Z}{A} \right]^\frac{1}{2/n}$$  \hspace{1cm} (6)

Fig. 6 shows the variation of peak stress with varying the $Z/A$ ratio of several TiAl-base intermetallic compounds. The line in Fig. 6 is plotted according to Eq. (6). The average values of the stress exponent, $n$, and $\alpha$ in Table 4 were used for Eq. (6). Fig. 6 shows that the peak stresses are fitted well by the normalized Zener-Hollomon parameter which is defined as Zener-Hollomon parameter divided by constant $A$ in TiAl-base intermetallic compounds. This means that $Z$ is not a sufficient parameter to express the dependence of peak stress on the strain rate and temperature. The good fit of peak stress in Fig. 6 indicates that the $A$ is not a constant, but is dependent on the activation energy, $Q$. The Fig. 7 indicates that a linear relationship exists between activation energy, $Q$, and $A$, and the relationship can be expressed as Eq. (7) from the linear regression.

$$A (\text{s}^{-1}) = 4.8 \times 10^{-2} \exp(0.082Q)$$  \hspace{1cm} (7)
It is also reported that the dependence of $A$ on activation energy in low and microalloyed steels [31]. If the chemical composition of TiAl is determined, then the activation energy and $A$ can be estimated by Eqs. (5) and (7). By inserting the measured values of $a$, $n$ and the expressions of $Z$ and $A$ into Eq. (6), the peak stress, $\sigma_p$, can be predicted at test temperature and strain rate in TiAl-based intermetallic compounds.

The comparison of the flow curves of Ti–48Al–2W produced by powder metallurgy (PM) process and ingot metallurgy (IM) process at 1100°C is shown in Fig. 8. The flow curves of IM Ti–48Al–2W were obtained from this study, while the flow curves of PM Ti–48Al–2W were obtained from the results reported by Beddoes et al. [19]. The flow curves of the IM Ti–48Al–2W exhibit substantially higher peak stress and greater flow softening behavior compared to PM Ti–48Al–2W. However, at larger strains, when the steady state is reached, the curves were close to each other. The difference in the peak stresses and flow softening rates may be ascribed to the differences in the initial microstructures. The IM Ti–48Al–2W alloy had an inhomogeneous dendritic microstructure (Fig. 1(b, d)). The average grain size was ~75 μm, and the W-rich $\beta$ particles were present mostly in the dendrite region. In contrast, the PM Ti–48Al–2W exhibited a finer microstructure with average grain size

Fig. 9. Microstructures of Ti–46Al–2W after compressive deformation up to a true strain of 1.2 with strain rate of $10^{-3}$ s$^{-1}$ at 1200°C (a and b), 1100°C (c and d), and 1000°C (e and f). (a), (c) and (e) are the optical micrographs and (b), (d) and (f) are the SEM micrographs in back scattered electron mode.
of 13 μm. Thus, it could be concluded that the coarse and inhomogeneous microstructure of the IM Ti–48Al–2W alloy are the main reason for higher peak stress and greater degree of flow softening in IM Ti–48Al–2W compared to PM Ti–48Al–2W. The initial grain size effect on the peak stress is consistent with previous result obtained from Cu [32]. It is reported that the larger initial grain size resulted in higher peak stress during high temperature compression deformation [32]. The nucleation initiates at pre-existing grain boundaries by local strain induced grain boundary migration [22]. The nucleation occurred at the interface between recrystallized grain and unrecrystallized grain and continues until the sites at initial grain boundaries have been exhausted. This sequence of nucleation continues until all grains have been recrystallized. It is expected that the rate of dynamic recrystallization increases with decreasing the initial grain size. As a result, it is concluded that the greater degree of flow softening in PM Ti–48Al–2W is resulted from the faster recrystallization rate due to smaller grain size.

Figs. 9 and 10 show the microstructure deformed up to a strain of 1.2 in Ti–46Al–2W and Ti–48Al–2W, respectively, with varying temperature at a strain rate of $10^{-3} \text{s}^{-1}$. The figures show that the γ grains were refined due to the dynamic recrystallization that occurred during the deformation. The recrystallized grain size increased
with increasing the temperature. The back-scattered electron (BSE) micrographs of deformed microstructure exhibited different morphologies with varying temperature. In Ti–46Al–2W, the lamellar phase in dendrite region was fully recrystallized, while the $\beta$ phase was spheroidized and uniformly distributed at 1200°C. The microstructures developed at 1100°C were partially recrystallized. The lamellar phase still exists in dendrite region and the $\beta$ phase is not fully spheroidized. At 1000°C, the material was mostly unrecrystallized, and the lamellar and $\beta$ phase were deformed and elongated along the direction perpendicular to compression axis. In Ti–48Al–2W, the $\gamma$ grains were fully recrystallized and the $\beta$ phase was spheroidized at 1200°C. The original dendritic regions remained after deformation at 1100 and 1000°C. At the same time, the unrecrystallized regions increased with decreasing temperature.

It is observed that the changes in microstructural evolution that occurred with increasing strain rate at a fixed temperature were similar to those that occurred with decreasing temperature at a fixed strain rate. The recrystallized grain size increased with increasing the temperature and with decreasing the strain rate, i.e. with decreasing the Zener-Hollomon parameter. The $\gamma$ grains were fully recrystallized at 1200°C with strain rate of $10^{-3}$ s$^{-1}$. The variation of recrystallized grain size with varying the Zener-Hollomon parameter is plotted in Fig. 11. The dependence of recrystallized grain size on Zener-Hollomon parameter is formulated in Eq. (8) from the linear regression in Fig. 11,

$$D_{\text{rex}} = kZ^{-0.32}$$

where $D_{\text{rex}}$ is recrystallized grain size, $Z$ is Zener-Hollomon parameter, $k$ is constant. As the increase of Zener-Hollomon parameter results in a larger driving force for recrystallization due to the higher dislocation density, the recrystallized grain size becomes finer with increasing the Zener-Hollomon parameter [33]. However, the fraction of unrecrystallized grains increased with increasing the strain rate and with decreasing the temperature. The banded structure, consisting of an inhomogeneous mixture of coarse unrecrystallized and fine recrystallized grains, increased with decreasing temperature and increasing the strain rates as shown in Fig. 10.

4. Conclusions

The high temperature deformation behavior of Ti–46Al–2W and Ti–48Al–2W intermetallic compounds have been investigated by isothermal compressive tests at temperatures between 1000 and 1200°C, and strain rates between $10^{-3}$ s$^{-1}$ and $10^{-1}$ s$^{-1}$.

(1) The flow stress exhibited a peak stress before decreasing gradually to a steady state level with increasing the strain. The observed flow softening behavior was attributed to the dynamic recrystallization during high temperature deformation.

(2) The dependence of flow stress on temperature and strain rate was fitted to a hyperbolic-sinusoidal relationship using the Zener-Hollomon parameter. The activation energies were measured as 449 and 394 kJ mol$^{-1}$, and the stress exponents were measured as 3.6 and 3.7 for Ti–46Al–2W and Ti–48Al–2W, respectively. It is suggested that the measured activation energies correspond to the activation energy for dynamic recrystallization. Using a normalized Zener-Hollomon parameter, which is defined as Zener-Hollomon parameter divided by constant $A$, it is possible to predict the peak stress for a given temperature and strain rate.

(3) The dynamically recrystallized grain size decreased with decreasing the temperature and increasing the strain rate. However, the fraction of unrecrystallized grains increased with increasing the strain rate and with decreasing the temperature.

References