Effect of $\beta$ Phase Stability at Room Temperature on Mechanical Properties in $\beta$-Rich $\alpha+\beta$ Type Ti–4.5Al–3V–2Mo–2Fe Alloy

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The stability of the $\beta$ phase at room temperature in various microstructures of a $\beta$-rich $\alpha+\beta$ type Ti–4.5Al–3V–2Mo–2Fe alloy and its relationship with the fracture toughness, hardness and tensile properties were investigated. A variety of microstructures were established by varying solution treatment temperatures in $\alpha+\beta$ field, cooling rate after solution treatment and the condition of subsequent second-step annealing treatment after air-cooling treatment. These microstructures have $\beta$ phase with lattice parameters of $\beta$ phase ranging between 0.3244 nm and 0.3221 nm. The stability of $\beta$ phase, which is indicated by decreasing lattice parameter of $\beta$ phase, is increased by either lowering cooling rate or formation of diffusional transformation products (secondary phases) in the $\beta$ phase. The $\beta$ phase with lattice parameter of $\beta$ phase around 0.3242 nm is the minimal instability of unstable $\beta$ phase at room temperature for attaining deformation-induced martensite in tensile specimens. There exists a proper degree of $\beta$ phase stability for increasing the fracture toughness, $J_{IC}$. The relatively higher fracture toughness is obtained at low or high stability of $\beta$ phase. The high fracture toughness at low stability of $\beta$ phase (unstable $\beta$) is mainly due to the deformation-induced martensite. While, the high fracture toughness at high stability of $\beta$ phase (stable $\beta$) is mainly due to the secondary phase in the $\beta$ phase that produces a prominent crack deflection toughening mechanism. However, the relatively lower fracture toughness is obtained at high stability of $\beta$ phase when the $\beta$ phase contains small amount or no secondary phase. This leads to conclude that, if only the $\beta$ phase stability is taken into account for explaining fracture mechanism, the fracture toughness would decrease monotonously with increasing stability of $\beta$ phase. The Vickers hardness is nearly independent of stability of $\beta$ phase.

KEY WORDS: mechanical properties; fracture toughness; hardness; tensile properties; stability of $\beta$ phase; microstructure; $\beta$-rich $\alpha+\beta$ titanium alloy; Ti–4.5Al–3V–2Mo–2Fe.

1. Introduction

It is well known that an unstable phase such as retained austenite in ferrous alloys, 1–4) and retained $\beta$ in titanium alloys 5–8) can transform into martensite during deformation. This transformation process, which is commonly called as transformation induced-plasticity, TRIP, or deformation-induced transformation, results in the well-known tensile properties, i.e. low yield stress, excellent ductility and high work hardening rate. 5–7) This phenomenon attracts a special interest due to certain reasons; such as the ability of the TRIP to control the behavior of shape-memory alloys. 8) It is also expected to suppress initiation and propagation of cracks by relaxation of stress concentration due to the phase transformation. 7)

For titanium alloys, the occurrence of deformation-induced martensite, DIM, strongly depends strongly on the $\beta$ phase stability of the alloys. The DIM can be observed in many less stable $\beta$ alloys such as $\alpha+\beta$ type Ti–6Al–2Sn–4Zr–6Mo (Ti-6246) 9) and metastable $\beta$ type Ti–5Al–2Sn–4Zr–4Mo–2Cr–1Fe (Ti-5246) 9) alloys. The main factor for the occurrence of DIM in the alloys is obviously the low stability of $\beta$ phase. The unstable $\beta$ phase resulting from a rapid cooling from the $\beta$ field transforms easily to hexagonal $\alpha'$ or orthorhombic $\alpha'$ during loading (straining). While, the unstable $\beta$ phase resulting from a rapid cooling from the $\alpha+\beta$ field is difficult to produce DIM because of the $\beta$ phase is stabilized by the presence of primary $\alpha$ phase and the decrease of prior $\beta$ grain size as pointed out recently by Grossidier et al. 5, 8) They have explained that the presence of primary $\alpha$ phase tends to hinder the formation of martensite by increasing the stability of $\beta$ phase and thus its martensite start temperature ($M_s$) is decreased. While, the decrease of prior $\beta$ grain size also tends to stabilize $\beta$ phase by making it more difficult to accommodate the shape change associated with the transformation. This indicates that the stability of $\beta$ phase depends on the microstructural parameters of both $\alpha$ and $\beta$ phases.
and thus critical stability of $\beta$ phase that helps to attain DIM is exist. Since the stability of $\beta$ phase corresponds linearly to the content of $\beta$ stabilizer (Mo equivalent) and the lattice parameter of $\beta$ phase is inversely related to the content of Mo equivalent, the change in the stability of $\beta$ phase can be, therefore, determined by examining the change in the lattice parameter of $\beta$ phase. The stability of the $\beta$ phase is increased with decreasing the lattice parameter of $\beta$ phase.

As for the fracture toughness, Niinomi et al. have reported that the unstable $\beta$ phase is favorable for the increase in both static and dynamic fracture toughness in commercial titanium alloys due to either stress-induced transformation or mechanical twinning. On the contrary, Akmoulin et al. have reported that the presence of highly unstable retained $\beta$ phase leads to a decrease in the dynamic fracture toughness of Ti–6Al–4V due to a combination of a large stress–strain gradient that leads to high localized deformation of the soft unstable $\beta$ phase and a high stress triaxiality existing ahead of the crack tip. The contrary result was also found in austenitic stainless steels. Reed has reported that the fracture toughness of stainless steels decreases with increasing stability of austenite, but Tobler et al. have reported an opposite trend in Mn-containing stainless steels. This implies that the proper degree of stability may exist to improve fracture toughness of the alloys.

For $\beta$-rich $\alpha+\beta$ type Ti–4.5Al–3V–2Mo–2Fe alloy, it was found that the fracture toughness, either $K_{IC}$, or $J_{IC}$, has relatively low value when the alloy is air-cooled from solution treatment temperature of 1123 K. From the viewpoint of the fracture mechanisms, it has been explained that the reasons for the low fracture toughness are decreasing the effects of crack tip blunting and crack deflection due to the presence of $\alpha$ and local acicular $\alpha$ phase in matrix $\beta$ phase. Ishikawa et al. have reported that the relatively higher amount of retained $\beta$ phase is obtained when the alloy is water-cooled or air-cooled from the solution treatment temperature at 1123 K. The high content of the "soft" retained $\beta$ phase may cause the relatively higher damping capacity of the alloy water-quenched from 1123 K as reported by Guan et al. This factor may also contribute to the low fracture toughness of the alloy air-cooled from 1123 K. Since retained $\beta$ phase is well known to have low stability and has a potential to transform to DIM, it is necessary, therefore, to investigate the stability of $\beta$ phase and its relationship with mechanical properties of the alloy. Particularly, it is interesting to know the minimal instability of $\beta$ phase for attaining DIM and the proper degree of the stability of $\beta$ phase for increasing the fracture toughness of the alloy. Although the alloy has been originally designed to improve superplastic properties and to lower manufacturing cost of the most popular $\alpha+\beta$ type titanium alloy, Ti–6Al–4V, it has also shown excellent mechanical properties as well as good hardenability. Due to these merits, it has been applied as materials for many structural and non-structural components.

2. Experimental Procedures

2.1. Material and Heat Treatments

The material used in this study was a hot rolled plate of Ti–4.5Al–3V–2Mo–2Fe alloy of 12.5 mm in thickness. The chemical composition of the alloy is given in Table 1. The rectangular blocks with the size of 55 mm long×11 mm wide×11 mm thick were cut along the rolling direction of the plate as schematically shown in Fig. 1(a). In order to obtain various stability of $\beta$ phase, the blocks were then solution treated at temperatures between 1103 K and 1173 K in $\alpha+\beta$ field for 3.6 ks (1 h) and cooled to room temperature at various cooling rates. The cooling methods were water-quenching, WQ, air-cooling, AC, furnace-cooling, FC, slow furnace cooling, SFC, with cooling rates around 200 K s$^{-1}$, 10 K s$^{-1}$, 0.1 K s$^{-1}$ and 0.05 K s$^{-1}$, respectively. The cooling rates were calculated by dividing the half temperature of the difference between the solution treatment temperature and room temperature, and the cooling time required for reaching the half temperature. This is the relatively linear part of cooling temperature–time curve. Some air-cooled specimens were subsequently heat treated at 993 K for 3.6 ks (duplex-annaling, DA) as shown in schematic illustrations of heat treatments combined with the schematic phase diagram of titanium alloys in Fig. 1(b). Solution treatment was carried out using vacuum quartz tubes in which some blocks encapsulated.

Table 1. Chemical composition of Ti–4.5Al–3V–2Mo–2Fe alloy (mass%)

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>V</th>
<th>Mo</th>
<th>Fe</th>
<th>O</th>
<th>C</th>
<th>N</th>
<th>H</th>
</tr>
</thead>
<tbody>
<tr>
<td>content</td>
<td>4.47</td>
<td>3.00</td>
<td>1.86</td>
<td>1.96</td>
<td>0.1</td>
<td>0.01</td>
<td>0.01</td>
<td>0.0036</td>
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</table>

Fig. 1. Schematic illustrations of (a) sample blocks in the plate, (b) phase diagram and heat treatment schedules of the blocks and (c) geometry of specimens. WQ = water-quenching, AC = air-cooling, FC = furnace-cooling, SFC = slow furnace cooling, and DA = duplex-annaling.
2.2. Mechanical Tests

Three point bending fracture toughness test specimens with the size of 55 mm long×10 mm wide×10 mm thick and tensile test specimens with a gauge diameter of 4 mm and a gauge length of 20 mm as schematically shown in Fig. 1(c) were machined from the heat-treated blocks. The fatigue pre-crack was introduced on the slit tip for $a_o/W$ to be 0.53, where $a_o$ and $W$ are the initiation crack length and the specimen width, respectively. Static fracture toughness tests were carried out using an Instron-type testing machine according to ASTM E813 at room temperature in air. The crack initiation point was detected using an AC electrical potential method. Evaluation procedures to obtain the values of fracture toughness, $J_{IC}$, are given in a separate article. Tensile tests were conducted using an Instron-type testing machine at a cross speed of $8.3 \times 10^{-3}$ m/s at ambient temperature which corresponds to strain rate of $0.417 \times 10^{-3}$ s$^{-1}$. The strain was detected by both clip gauge and foil gage directly attached on the specimens. The displacement was measured directly from the tensile tested specimens. This is to avoid miscalculation in the displacement and elongation due to the slip on the specimens at the grip during testing, which causes an excessive displacement in the load–displacement curve. Vickers hardness tests were carried out on the fracture toughness tested specimens with 20 kg loads for 15 s using a hardness-testing machine.

2.3. Microstructural Observations

Using a scanning electron microscope (SEM) and an X-ray diffractometer (XRD) microstructures and phase constitutions of the specimens were characterized. Deep chemical etching in solution of 5% nitric acid, 3% hydrofluoric acid and 92% water was applied after grinding with emery paper and buff polishing to minimize the effect of deformation resulting from the grinding and polishing processes. XRD analysis was conducted on the longitudinal direction of the specimens using Cu-K$_\alpha$ line ($\lambda = 0.15406$ nm) under the conditions of voltage and current of 40 kV and 30 mA, respectively.

3. Results and Discussion

3.1. Effect of Heat Treatment Conditions on Microstructures

Typical SEM micrographs and XRD patterns of the alloy resulting from the heat treatment schedules as already shown in Fig. 1 are shown in Figs. 2 and 3, respectively. All microstructures contain hexagonal closed-packed, HCP, primary α phase.
mary α phase (the dark phase in Fig. 2) and different types of transformation products in body centered cubic, BCC, β phase (the bright phase in Fig. 2). The types of transformation products, which are also called as secondary phases hereinafter, are martensite (orthorhombic α’), acicular α (HCP) and plate-like α (HCP) phases observed in WQ, AC and DA, and FC specimens, respectively. Both the secondary phase and β phase are found within prior β grain as can be seen in Figs. 2(a)–2(c) and 2(e). While, SFC specimen has β phase that contains no secondary phase as shown in Fig. 2(d). This β phase is found in the form of intergranular structure.

Microstructural parameters of primary α, prior β and secondary phases, which were reported completely in separate articles \(^1\,^1\,^2\) showed that volume fraction of primary β phase decreases and volume fraction of prior β phase increases with increasing solution treatment temperature. Prior β grain size increases and grain size of primary α phase is nearly constant with increasing solution treatment temperature. While, volume fraction and size (width) of the secondary phase increase with increasing solution treatment temperature.

### 3.2. Effect of Heat Treatment Conditions on Stability of β Phase

The lattice parameter of β phase at room temperature together with the obtained mechanical properties of the heat-treated specimens is tabulated in Table 2. The lattice parameter of β phase, \(a_β\), was calculated from the (110)β diffraction peak. In order to clearly show the effect of heat treatment condition on the stability of β phase, the data in Table 2 are plotted in the form of lattice parameter versus solution treatment temperature and cooling rate as shown in Figs. 4 and 5, respectively. Decreasing lattice parameter of β phase indicates increasing stability of β phase.

It is shown in Fig. 4 that the stability of β phase of WQ specimen is relatively low and increases slightly with decreasing solution temperature. It is well known that a rapid

<table>
<thead>
<tr>
<th>Heat treatment</th>
<th>Solution treatment temperature (K)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (% of length)</th>
<th>Hardness (HV)</th>
<th>Fracture toughness (kN/m)</th>
<th>Lattice parameter (nm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>WQ</td>
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<td>903</td>
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<tr>
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<td>938</td>
<td>17.0</td>
<td>365</td>
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<tr>
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<td>909</td>
<td>17.8</td>
<td>323</td>
<td>47</td>
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<tr>
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<td>1152</td>
<td>602</td>
<td>17.8</td>
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<tr>
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<tr>
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<td>1002</td>
<td>894</td>
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<tr>
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<td>930</td>
<td>892</td>
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<td>321</td>
<td>19</td>
</tr>
<tr>
<td>SFC</td>
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<td>1035</td>
<td>973</td>
<td>14.8</td>
<td>342</td>
<td>35</td>
</tr>
<tr>
<td>DA</td>
<td>1143</td>
<td>1089</td>
<td>993</td>
<td>12.7</td>
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<td>1008</td>
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<td>378</td>
<td>65</td>
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</tbody>
</table>

Fig. 4. Variation of lattice parameter of β phase, \(a_β\), as a function of solution treatment temperature for the indicated subsequent treatments.

Fig. 5. Variation of lattice parameter of β phase, \(a_β\), as a function of cooling rate for the indicated solution treatment temperatures.
cooling treatment retains an unstable $\beta$ phase at room temperature because this treatment does not provide enough time for diffusional transformation process. As a result, there exists element partition where the $\beta$ stabilizing elements (Mo, V and Fe) partition to the $\beta$ phase and $\alpha$ stabilizing element (Al) partitions to $\alpha$ phase. The $\beta$ stabilizer contents, in particular Mo, in the $\beta$ phase decreases with decreasing solution treatment temperature and, therefore, the $\beta$ phase in WQ specimens is stabilized with decreasing solution treatment temperature. While, the stability of $\beta$ phase of AC specimens increases significantly with increasing solution treatment temperature when the volume fraction of acicular $\alpha$ obtained in (matrix) $\beta$ phase is also increased in the same order. This indicates that the main factor that leads to stabilize $\beta$ phase is the acicular $\alpha$ phase. Unlike AC, DA and FC specimens, the stability of $\beta$ phase of SFC specimens decreases slightly with increasing solution treatment temperature when the observed $\alpha$ phase also decreases slightly with increasing solution treatment temperature. This supports above findings that the $\beta$ phase is stabilized by the $\alpha$ phase. However, it is interesting to discuss why the stability of $\beta$ phase does not increase with increasing solution temperature in SFC specimens. This may relate to the absence of either secondary phase in $\beta$ phase (see Fig. 2(c)). Unlike AC, DA and FC specimens, the stability of $\beta$ phase of SFC specimens decreases slightly with increasing solution treatment temperature when the observed $\alpha$ phase also decreases slightly with increasing solution treatment temperature.

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3.3. Effect of Stability of $\beta$ Phase on Mechanical Properties

In order to clearly show the effect of stability of $\beta$ phase at room temperature on the mechanical properties, the data in Table 2 are plotted in the form of mechanical properties versus lattice parameter of $\beta$ phase, irrespective of heat treatment conditions. The results are shown in Figs. 6, 9 and 10. Decreasing lattice parameter of $\beta$ phase indicates increasing stability of $\beta$ phase.

3.3.1. Effect of Stability of $\beta$ Phase on Tensile Properties

Scattered data in Figs. 6(a)-(c) indicate that the effect of stability of $\beta$ phase on tensile strength, 0.2% proof stress and elongation, respectively, forms two trends of data as indicated by the solid lines in each figure. It is seen that there exists a critical point in the relationship between the tensile properties and stability of $\beta$ phase at a lattice parameter of 0.3242 nm that is indicated by a dramatic change in the tensile properties at this point. For lattice parameter above 0.3242 nm, 0.2% proof stress decreases considerably with decreasing the $\beta$ phase stability. While, both tensile strength and elongation increase with decreasing the $\beta$ phase stability, which is a usual trend in some alloys in which stress-induced transformation occurs. As mentioned in introduction, the unstable $\beta$ can be transformed to martensite during deformation resulting in the excellent ductility in addition to the low yield stress, high work hard-
The β phase with lattice parameter above 0.3242 nm can be, therefore, called as “unstable β”. For lattice parameter below 0.3242 nm, 0.2% proof stress is relatively constant with increasing stability of β phase. While, tensile strength decreases when the elongation increases, which is a usual trend in alloys in which the DIM does not occur. The β phase in this region, therefore, can be called as “stable β”.

Figures 6(b) and 6(c) show the two characteristics the occurrence of DIM in the unstable β phase regime, that is, low yield stress and excellent ductility, respectively. Another characteristic, i.e. work hardening rate, can be calculated from stress–stress curves from a plastic strain of 2.5%, for example, to the onset of necking. In fact, it is difficult to avoid slip on the specimen at the grip during testing. As mentioned in experimental procedure, the displacement was measured directly from the fractured specimens. Since the high tensile strength is an additional characteristic observed in the present alloy (Fig. 6(a)), the third characteristic of DIM can be seen easily by using work hardening stress, WHS, i.e. the difference between tensile strength and 0.2% proof stress, instead of the work hardening rate.

The WHS as a function of stability of β phase is shown in Fig. 7. It is seen that, with increasing stability of β phase, the WHS decreases sharply for lattice parameter above 0.3242 nm and is relatively constant for lattice parameter below 0.3242 nm. This clearly indicates that the DIM occurs for lattice parameter of β phase above 0.3242 nm.

In order to directly see the DIM in microstructure, the deformed area of tensile tested specimens that have lattice parameters of β phase greater than 0.3242 nm (WQ specimens, see Table 2) was observed using an SEM and the result is shown in Fig. 8. It can be seen in this figure that the DIM, which is confirmed as orthorhombic α" using XRD analysis, observed in all microstructures. The amount of DIM (α") qualitatively decreases with increasing stability of β phase (as indicated by decreasing lattice parameter of β phase, a_b) when the observed content of primary α phase increases and prior β grain size in average diameter decreases (Table 3). This completely supported the results of Grossdidier et al.6,8 that, as mentioned in introduction, the increase of primary β phase content and the decrease of prior β grain size tend to stabilize β phase. For specimens that have lattice parameter of β phase below 0.3242 nm, it was found that no DIM is observed in the deformed area of their tensile tested specimens. This means that the β phase with lattice parameter lower than 0.3242 nm is not transformed during testing.

From above results and discussion, it can be said that the β phase with the lattice parameter of 0.3242 nm is the minimal instability of unstable β phase at room temperature for obtaining DIM for the tensile test specimens of the present alloy.

### 3.3.2. Effect of Stability of β Phase on Vickers Hardness

Figure 9 shows that the profile of scattered data of Vickers hardness looks almost similar to that of tensile strength, indicating a close relationship between them as commonly observed in an alloy. By employing the critical stability described above in Fig. 9, it can be seen that, the hardness tends to decrease with increasing stability of β phase.

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Lattice parameter of the β phase (nm)</th>
<th>Volume fraction of primary α phase (%)</th>
<th>Prior β grain size (μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1173WQ</td>
<td>0.3249</td>
<td>7</td>
<td>11.2</td>
</tr>
<tr>
<td>1133WQ</td>
<td>0.3246</td>
<td>17</td>
<td>7.2</td>
</tr>
<tr>
<td>1103WQ</td>
<td>0.3244</td>
<td>21</td>
<td>5.7</td>
</tr>
</tbody>
</table>

# Solution treated at temperature of 1173 K and followed by water-quenching (WQ)
phase for unstable $\beta$ phase, and to be independent of the stability of $\beta$ phase with further increasing stability of $\beta$ phase for stable $\beta$ phase. However, by considering that the scattered data of hardness is relatively narrow, that is around 65 Hv (314 Hv min and 379 Hv max) as can be seen in Table 2 or Fig. 9, and data error is found in the range of 5–10 Hv, it is more exact to say that the hardness is relatively constant against stability of $\beta$ phase as indicated by the data band in Fig. 9.

3.3.3. Effect of Stability of $\beta$ Phase on Fracture Toughness

The fracture toughness, $J_{IC}$, as a function of stability of $\beta$ phase is shown in Fig. 10. The fracture toughness appears to decrease with increasing stability of $\beta$ phase up to lattice parameter of $\beta$ phase around 0.323 nm and then to be independent of the stability of $\beta$ phase with further increasing stability of $\beta$ phase. The $\beta$ phase with lattice parameters below 0.323 nm have relatively low fracture toughness as indicated by the data with a band. However, the upper group of data indicates the $\beta$ phase with the lattice parameter below 0.323 nm can also have relatively high fracture toughness. Corresponding to each datum (see Table 2) for the lattice parameter of $\beta$ phase below 0.323 nm with related microstructure, it is found that the low fracture toughness values are belonged to the specimens with microstructures containing less and no secondary phases in the $\beta$ phase. While, the high fracture toughness values are belonged to the specimens with microstructures containing high amount and coarse secondary phase in the $\beta$ phase.

By employing the terms of unstable $\beta$ and stable $\beta$ for lattice parameter of $\beta$ phase greater and lower than 0.3242 nm, respectively, observed in tensile specimen in Fig. 10, it can be seen that the relatively higher fracture toughness can be obtained at low stability of $\beta$ phase (unstable $\beta$ phase) observed in water-quenched specimens and at high stability of $\beta$ phase (stable $\beta$ phase) observed in furnace-cooling and duplex-annealing specimens.

The relatively higher fracture toughness at low stability of $\beta$ phase must be related to the DIM. XRD analysis shows that the retained $\beta$ almost transforms completely to $\alpha$ during testing for the specimen having the highest instability of $\beta$ phase ($a_\beta=0.3249$ nm, 1173WQ specimen) as indicated by the high reduction of intensity of (110)$\beta$ diffraction peak of the fracture surface of the specimen as compared with that of as heat-treated specimen as shown in Fig. 11(a). The unstable $\beta$ phase transforms partly to DIM ($\alpha'$) for the specimen with intermediate instability of $\beta$ phase ($a_\beta=0.3246$ nm, 1133WQ specimen) as can be seen in Fig. 11(b). However, no DIM is observed for the specimen having $a_\beta=0.3244$ nm (1103WQ specimen) as can be seen in Fig. 11(c). The increase of fracture toughness with increasing instability of $\beta$ phase (see unstable $\beta$ regime in Fig. 10) when the DIM is also increased in the same order indicates that the increase of fracture toughness is mainly due to the DIM as has also been observed by Niinomi et al.
The relatively higher fracture toughness of the specimens having lattice parameter of $\beta$ phase larger than 0.323 nm (stable $\beta$ phase) when the observed $\beta$ phase contains coarse secondary phase indicates that the high fracture toughness in this case is mainly due to the secondary phase. It has been explained that the coarse secondary phase increases the effect of extrinsic toughening mechanism such as crack deflection effect, crack branching and micro-cracking.

Figure 10 shows that the fracture toughness is kept to decrease with further increasing stability of $\beta$ phase in the range of lattice parameter of $\beta$ phase between 0.324 nm and 0.323 nm. The $\beta$ phase with lattice parameter in this range is called as intermediate stable $\beta$ phase hereinafter. XRD analysis shows that the specimens with intermediate stable $\beta$ phase, i.e. AC specimens with solution treatment temperature in the range between 1103 K and 1143 K (Table 2), has no DIM. This is obviously because the $\beta$ phase instability of the specimens are below the minimal instability for obtaining DIM ($\alpha_f=0.3242$ nm). Furthermore, the specimen having $\beta$ phase with intermediate stability contains fine and local acicular $\alpha$ phase in the $\beta$ phase. It has been reported that the fine and local acicular $\alpha$ phase cannot deflect a crack and thus cannot increase the fracture toughness.\(^7\)\(^1\)\(^2\) In the case of Ti-6Al-4V, it was reported that the low dynamic fracture toughness of this alloy is obtained when the microstructure having $\beta$ phase with lattice parameter of 0.3239 nm.\(^7\)\(^3\) This indicates that the low dynamic fracture observed in Ti-6Al-4V may also related to the intermediate stability of the $\beta$ phase.

At lattice parameter of $\beta$ phase larger than 0.323 nm, the specimens are kept to have low fracture toughness when the microstructure contains only small amount of transformation products (acicular $\alpha$ or coarse plate-like $\alpha$ phase) in the $\beta$ phase. The lowest fracture toughness is obtained when the stable $\beta$ phase containing no secondary phase as observed in slow furnace-cooled, SFC, specimen (see Table 2 or Fig. 4). These indicate that the high stable $\beta$ phase solely cannot increase fracture toughness. Therefore, if only $\beta$ phase stability is taken into account for explaining fracture mechanism, it is considered that the fracture toughness decreases monotonously with increasing stability of $\beta$ phase as indicated by the dotted line in Fig. 10.

4. Conclusions

Fracture toughness, tensile properties and Vickers hardness of various microstructures resulting from various heat treatment conditions in $\beta$-rich $\alpha'$+$\beta$ type Ti-4.5Al-3V–2Mo–2Fe alloy were investigated in terms of the stability of $\beta$ phase at room temperature. The following results were obtained.

1) The $\beta$ phase is stabilized significantly by either lowering cooling rate or formation of diffusional transformation products, i.e. acicular $\alpha$ and plate-like $\alpha$ phases, in the $\beta$ phase. While, the formation of diffusionless transformation product, i.e. martensite, in the $\beta$ phase seems to have no effect on the stability of $\beta$ phase.

2) There exists critical stability at lattice parameter of 0.3242 nm in the relationship between tensile properties and stability of $\beta$ phase, which is indicated by a dramatic change in the tensile properties versus the stability of $\beta$ phase at this stability. The high work hardening stress, i.e. the difference between tensile strength and yield stress, and the observed martensite in the microstructure after testing lead to conclude that the $\beta$ phase is unstable at lattice parameter above 0.3242 nm. While, the low work hardening stress and the observed no martensite in the microstructure after testing lead to conclude that the $\beta$ phase is stable at lattice parameter below 0.3242 nm.

3) There exists a proper degree of $\beta$ phase stability for increasing the fracture toughness, $J_{IC}$. The fracture toughness increases with increasing instability of unstable $\beta$ phase mainly due to the deformation-induced martensite. The fracture toughness also increases with increasing stability of stable $\beta$ phase when the $\beta$ phase contains coarse secondary phase, mainly due to increasing extrinsic toughening mechanism such as crack deflection effect as a result of the presence of coarse secondary phase in the $\beta$ phase.

4) The relatively lower fracture toughness is also obtained for stable $\beta$ when the $\beta$ phase contains small amount or no secondary phase in the $\beta$ phase. This leads to conclude that, if only the $\beta$ phase stability is taken into account for explaining fracture mechanism, the fracture toughness is considered to decrease monotonously with increasing stability of $\beta$ phase.

5) The hardness appears to be nearly independent of stability of $\beta$ phase.

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