Development of High Modulus Ti–Fe–Cu Alloys for Biomedical Applications

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Ti–Fe–Cu alloys with higher Young’s modulus, hardness and compressive mechanical properties than those of the existing Ti alloys were developed using a d-electrons alloy design method in order to improve Young’s modulus, hardness and compressive properties of Ti and existing Ti alloys for use as metallic stents. Their microstructures, Young’s modulus, hardness and compressive mechanical properties were investigated both before (as-cast) and after heat-treatments performed under a high-purity argon atmosphere at 1173 K for 21.6 and 86.4 ks.

The studied Ti–Fe–Cu alloys consist of the β-Fe phase and dendritic TiFe intermetallic phase. Moreover, the area fraction of the TiFe intermetallic phase increases with increasing atom ratio (Fe + Cu)/Ti of the alloys and with the heat-treatment time.

The Young’s modulus of the studied Ti–Fe–Cu alloys increases from 110 GPa (Ti₆₇Fe₂₇Cu₆ alloy) to 145 GPa (Ti₆₀Fe₂₀Cu₁₀ alloy) with increasing atom ratio (Fe + Cu)/Ti of the alloys and the area fraction of the TiFe intermetallic phase. However, the Young’s modulus is saturated or slightly decreased when the area fraction of the TiFe intermetallic phase is more than 34%. The Vickers hardness of the as-cast alloys increases from 490 HV (Ti₇₈Fe₁₈Cu₄ alloy) to 550 HV (Ti₆₃.₄Fe₃₀Cu₆.₆ alloy) with increasing atom ratio (Fe + Cu)/Ti of the alloys and area fraction of the TiFe intermetallic phase. On the other hand, the Vickers hardness of the heat-treated alloys is lower than that of the as-cast alloys, despite the increase in the area fraction of the TiFe intermetallic phase after the heat-treatment. The heat-treated alloys have better compressive properties than those of the as-cast alloys and the reported Ti–Fe–Cu alloys. The compressive strength and strain of the heat-treated Ti₆₇Fe₂₇Cu₆ alloys reach to 2131 MPa and 24.5%, respectively. [doi:10.2320/matertrans.M2012361]

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1. Introduction

Titanium (Ti) and its alloys such as commercially pure Ti (CP-Ti) and (α + β)-type Ti–6Al–4V (mass%), Ti64 ELI alloy have been widely used for orthopedic surgical implant devices including artificial hip joints¹³ and spinal fixation devices²³ as well as for dental implant devices.³ Moreover, β-type Ti alloys including Ti–29Nb–13Ta–4Zr alloy (mass%), TMTZ⁴ and Ti–24Nb–4Zr–7.9Sn alloy (mass%)⁵ have been developed as low Young’s modulus Ti alloys in order to overcome the significant problem of the bone absorption induced by stress shielding due to mismatch between Young’s modulus of the bone and implant devices.⁶ Young’s modulus of TNTZ is approximately 60 GPa,⁷ which is much closer to that of the bone (10–30 GPa)⁸ than those of existing Ti alloys.

On the other hand, a high Young’s modulus, high hardness, high compressive strength and high ductility are required for the materials used for metallic stents.⁹ This is because metallic stents must expand in the vessel and support the vessel wall. Stainless steel 316L (316L SS) and Co–Cr-based alloys have higher Young’s modulus (210 GPa for 316L SS and 200–240 GPa for Co–Cr-based alloys)¹⁰,¹¹ as compared to those of Ti and its alloys (60–120 GPa).¹² Therefore, 316L SS and Co–Cr-based alloys are mainly used for metallic stents. However, Ti and its alloys have high potential as materials for metallic stents because of their excellent biocompatibility and corrosion resistance. Moreover, metallic stents made of Ti and its alloys are expected to reduce magnetic resonance imaging (MRI) artifacts, which are caused by metallic stents made of 316L SS and Co–Cr-based alloys. This is because Ti has a lower susceptibility to MRI artifacts of (180 × 10⁻⁶)¹³ than that of 316L SS (3380 × 10⁻⁶)¹³ or Co–Cr-based alloys (1370 × 10⁻⁶ for Co–Cr–Mo alloy).¹⁴ Therefore, new Ti alloys with higher Young’s modulus, higher hardness, higher compressive strength and higher ductility as compared to those of existing Ti alloys should be developed for metallic stents.

Das et al. have reported that Ti₅₁₃Fe₄₈₅ alloy (at%),¹⁵ which has a B2-structured TiFe intermetallic phase, has a higher Young’s modulus (222 GPa) as compared to those of existing Ti alloys. Moreover, Louguine et al. have reported that Ti₆₃Fe₃₅ alloy (at%).¹⁶ Ti–Fe–Co alloys¹⁷,¹⁸ and Ti–Fe–Cu alloys¹⁹ have higher compressive strengths as compared to those of Ti and existing Ti alloys as a result of formation of the TiFe intermetallic phase in β-Ti phase. Table 1 shows the compressive strength and compressive strain of as-cast Ti₅₅Fe₃₅ alloy, as-cast Ti–Fe–Co alloys and as-cast Ti–Fe–Cu alloys reported by Louguine et al.¹⁶–¹⁹ As shown in Table 1, Ti₅₅Fe₃₅ alloy, Ti₅₃Fe₁₅Co₁₅ alloy, Ti₅₃Fe₁₇Co₁₃ alloy, Ti₅₃Fe₁₉Co₁₅ alloy, Ti₅₃Fe₁₇Co₁₇ alloy and Ti₅₃Fe₂₀Co₂₀ alloy (at%) all have compressive strengths over 2000 MPa. In addition, these Ti–Fe–Co alloys also exhibit compressive strains of over 15% along with their high compressive strength. This is because the addition of Co decreases bond order of the TiFe intermetallic phase. Although Young’s modulus of these Ti–Fe–Co alloys were not measured, these good compressive properties indicate that Ti–Fe–Co alloys can be expected to be suitable materials for metallic stents. However, there is significant concern that the susceptibility to MRI artifacts of the Ti alloys will be increased by the addition of Co, because Co is a ferromagnetic material. On the other hand, addition of Cu is even more effective in decreasing the bond order of the TiFe intermetallic phase than is Co. Moreover, the addition of Cu could also reduce the susceptibility to MRI artifacts of the Ti–Fe alloys, because Cu is a diamagnetic material.

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Therefore, Louguine et al.\textsuperscript{19} have investigated the compressive strength and strain of Ti$_{68}$Fe$_{22}$Cu$_{14}$ alloy and Ti$_{72}$Fe$_{14}$Cu$_{14}$ alloy (at%), which have Cu contents close to the Co contents of the reported Ti–Fe–Co alloys. However, the compressive strains of the Ti$_{68}$Fe$_{22}$Cu$_{14}$ alloy and Ti$_{72}$Fe$_{14}$Cu$_{14}$ alloy are less than 6%, as shown in Table 1. These results suggest that it is necessary to select the Cu contents more carefully in order to improve the ductility of Ti–Fe–Cu alloys.

The purpose of this study is to develop new Ti–Fe–Cu alloys with high ductility along with higher Young’s modulus, higher hardness and higher compressive strength as compared to those of existing Ti alloys. These mechanical properties of alloys are determined by microstructure, which is controlled by both alloy compositions and thermal or mechanical processing. In this study, the alloy compositions of the studied Ti–Fe–Cu alloys were designed using a \textit{d}-electrons alloy design method with the bond order–\textit{d}-orbital level (Bo–Md) diagram,\textsuperscript{20} as shown in Table 1 and Fig. 1. The Ti–Fe–Co alloys that have comparatively high compressive properties (over 15% compressive strain with a maximum compressive strength near or over 2000 MPa) are grouped in a certain area of Fig. 1 (2.230 < Bo < 2.256 and 0.4 < Md < 1.0). Therefore, the alloy compositions of the studied Ti–Fe–Cu alloys were selected to be inside and around this area.

<table>
<thead>
<tr>
<th>Alloy (at%)</th>
<th>$\sigma$ (MPa)</th>
<th>$\epsilon$ (%)</th>
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</thead>
<tbody>
<tr>
<td>Ti$<em>{68}$Fe$</em>{22}$Cu$_{14}$</td>
<td>2220</td>
<td>6.7</td>
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<tr>
<td>Ti$<em>{72}$Fe$</em>{14}$Cu$_{14}$</td>
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<tr>
<td>Ti$<em>{68}$Fe$</em>{22}$Co$_{10}$</td>
<td>1890</td>
<td>8.1</td>
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<tr>
<td>Ti$<em>{72}$Fe$</em>{14}$Co$_{12}$</td>
<td>1690</td>
<td>2.3</td>
</tr>
<tr>
<td>Ti$<em>{78}$Fe$</em>{18}$Cu$_{4}$</td>
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</tr>
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<td>2160</td>
<td>15.3</td>
</tr>
<tr>
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<td>1935</td>
<td>16</td>
</tr>
<tr>
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<td>15</td>
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<td>Ti$<em>{68}$Fe$</em>{22}$Cu$_{32}$</td>
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<td>2110</td>
<td>5.5</td>
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<td>Ti$<em>{68}$Fe$</em>{22}$Cu$_{14}$</td>
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</table>

$\sigma_{\text{max}}$: Compressive strength  
$\epsilon$: Compressive strain

2. Materials and Experimental Procedures

2.1 Material design and preparation

The alloy compositions of the studied Ti–Fe–Cu alloys were selected using a \textit{d}-electrons alloy design method. Figure 1 shows the Bo–Md diagram of the reported Ti–Fe–Co alloys and the studied Ti–Fe–Cu alloys, which was calculated using a DV-Xa cluster method.\textsuperscript{21,22} Moreover, the compressive strain of reported as-cast Ti–Fe–Co alloys is shown in Fig. 1 as a basic compressive strain of the Ti–Fe–Co alloys. As shown in Table 1 and Fig. 1, the Ti–Fe–Co alloys that have comparatively high compressive properties (over 15% compressive strain with a maximum compressive strength of 2000 MPa) are grouped in a certain area of Fig. 1 (2.230 < Bo < 2.256 and 0.4 < Md < 1.0). Therefore, the alloy compositions of the studied Ti–Fe–Cu alloys were selected to be inside and around this area.

Ti$_{54}$Fe$_{23}$Co$_{23}$ alloy, Ti$_{66}$Fe$_{30}$Cu$_{12}$ alloy, Ti$_{68}$Fe$_{30}$Cu$_{4}$ alloy and Ti$_{64}$Fe$_{30}$Cu$_{6.6}$ alloy (at%) ingots were prepared from sponge Ti (99.7%), wire Fe (99%) and high-purity Cu (99.99%) using an arc-melting method under a high-purity argon atmosphere. The alloy composition and atom ratio (Fe + Cu)/Ti of each alloy are listed in Table 2. All alloys were subjected to the heat-treatment under a high-purity argon atmosphere at 1173 K for 21.6 ks followed by air cooling. Furthermore, Ti$_{68}$Fe$_{23}$Co$_{12}$ alloy, Ti$_{67}$Fe$_{27}$Cu$_{6}$ alloy, Ti$_{66}$Fe$_{30}$Cu$_{4}$ alloy, and Ti$_{63.4}$Fe$_{30}$Cu$_{6.6}$ alloy were subjected to the heat-treatment under a high-purity argon atmosphere at 1173 K for 86.4 ks followed by air cooling.

2.2 Material characterization

The phase constitutions of the studied Ti–Fe–Cu alloys were investigated using an X-ray diffraction (XRD) analysis. The XRD analysis was carried out using a Cu target with an accelerating voltage of 40 kV and a tube current of 40 mA. The specimens for XRD analysis were wet-polished using...
waterproof emery papers of up to #4000 grit and a colloidal SiO₂ suspension.

The microstructures of the studied Ti–Fe–Cu alloys were observed using a scanning-electron microscopy (SEM) and an optical microscopy. The alloying element distribution was investigated using an energy dispersive X-ray spectroscopy (EDX). For microstructural observations using an optical microscopy, an SEM and an EDX, the specimens etched using a 5% HF etching solution for 10 s after wet-polished using waterproof emery papers of up to #4000 grit and buff-polished using a colloidal SiO₂ suspension were used. Furthermore, SEM and EDX were operated at an accelerating voltage of 40 kV.

2.3 Mechanical tests

The Young’s modulus measurement was carried out at room temperature in air using a free resonance method. For Young’s modulus measurements, plate specimens with a width of 10 mm, a length of 40 mm and a thickness of 1.5 mm wet-polished using waterproof emery papers of up to #1500 grit were used.

The hardness measurement was carried out using a micro-Vickers hardness tester with a load of 9.807 N for a dwell time of 15 s. The diagonal length of indentation was approximately 60 µm. For hardness measurements, square specimens with a width of 5 mm × 5 mm and a thickness of 1.5 mm wet-polished using waterproof emery papers of up to #4000 grit and buff-polished using a colloidal SiO₂ suspension were used.

The compressive properties of the studied Ti–Fe–Cu alloys were evaluated using an Instron-type testing machine with a cross-head speed of 6.0 × 10⁻⁶ m·s⁻¹ at room temperature. For mechanical compressive tests, bar specimens with a length of 7.5 mm and a diameter of 3 mm wet-polished using waterproof emery papers of up to #1500 grit were used.

3. Results and Discussion

3.1 Material characterization

3.1.1 Phase constitution

Figure 2 shows XRD profiles of the studied Ti–Fe–Cu alloys before heat-treatments (as-cast). As shown in Fig. 2, the diffraction peaks of β-Ti and, the (110), (200) and (211) diffraction peaks (hereafter peak) of the TiFe intermetallic are detected from the studied Ti–Fe–Cu alloys. This result suggests that the studied Ti–Fe–Cu alloys consist of the β-Ti phase and TiFe intermetallic phase.

Figure 3 shows the XRD profiles of the Ti₆₇Fe₂₇Cu₆ alloy, chosen as a typical Ti–Fe–Cu alloy, before (as-cast) and after heat treatments at 1173 K for 21.6 and 86.4 ks. The phase constitution of the Ti₆₇Fe₂₇Cu₆ alloy after the heat-treatment is as same as that of the as-cast alloy. However, the (110) peaks of the β-Ti of the heat-treated alloys are shifted to a lower angle than that of the as-cast alloy. This result suggests that lattice parameter of the β-Ti phase becomes larger than that of the as-cast alloy.

The compressive properties of the studied Ti–Fe–Cu alloys were evaluated using an Instron-type testing machine with a cross-head speed of 6.0 × 10⁻⁶ m·s⁻¹ at room temperature. For mechanical compressive tests, bar specimens with a length of 7.5 mm and a diameter of 3 mm wet-polished using waterproof emery papers of up to #1500 grit were used.

3.1.2 Microstructure and alloying element distribution

Figure 4(a) shows an SEM image of the Ti₆₇Fe₂₇Cu₆ alloy before the heat-treatment (as-cast). And Figs. 4(b), 4(c) and 4(d) show the corresponding EDX mapping images for Ti, Fe and Cu, respectively. As shown in Fig. 4(a), two different areas (light gray area and dark gray area) exist in the studied Ti–Fe–Cu alloys. The mapping images for Ti, Fe and Cu demonstrate that the round light gray area has a high Ti content, whereas the dark gray area has high Fe and Cu contents. These alloying element distributions and XRD results suggest that the light gray area is the β-Ti phase and the dark gray area is the TiFe intermetallic phase.
of the Ti$_{67}$Fe$_{27}$Cu$_6$ alloy measured for the $\beta$-Ti phase and TiFe intermetallic phase as a function of the heat-treatment time. The Ti and Fe contents of the $\beta$-Ti phase and TiFe intermetallic phase are almost constant even after heat-treatments at 1173 K for 21.6 and 86.4 ks. On the other hand, the Cu content decreases by approximately 10.1\% in the $\beta$-Ti phase area and slightly increases by approximately 1.3\% in the TiFe intermetallic phase area after the thermal treatment at 1173 k for 86.4 ks. These differences between the Cu contents of the as-cast alloy and heat-treated alloys affect the lattice parameter of the $\beta$-Ti phase, which was observed to change using an XRD.

Optical micrographs of the studied Ti–Fe–Cu alloys before the heat-treatment (as-cast) are shown in Fig. 6. The white areas with dendritic structures in the optical micrographs correspond to the dark gray area in the SEM image. In other words, the white areas in the optical micrographs correspond to the TiFe intermetallic phase. Figure 7 shows area fractions of the TiFe intermetallic phase for the studied Ti–Fe–Cu alloys before (as-cast) and after the heat-treatment, as determined from the optical micrographs. The area fraction of the TiFe intermetallic phase increases significantly from 1.5\% (Ti$_{78}$Fe$_{18}$Cu$_4$ alloy) to 49.2\% (Ti$_{63.4}$Fe$_{30}$Cu$_{6.6}$ alloy) with increasing atom ratio (Fe + Cu)/Ti of the alloys. This result is in good agreement with the XRD results. Moreover, area fraction of the TiFe intermetallic phase increases after the heat-treatment for all studied Ti–Fe–Cu alloys. In a typical example, the area fraction of the TiFe intermetallic phase of the Ti$_{67}$Fe$_{27}$Cu$_6$ alloy increases from 40.8 to 44\% after the heat-treatment for 86.4 ks.

### 3.2 Mechanical properties

#### 3.2.1 Young's modulus

Figure 8 shows the Young’s moduli of the studied Ti–Fe–Cu alloys (a) before (as-cast) and after the heat-treatment (b) as a function of the area fraction of the TiFe intermetallic phase. Note that the Young’s moduli of the as-cast and heat-treated Ti$_{78}$Fe$_{18}$Cu$_4$ alloy and the as-cast Ti$_{63.4}$Fe$_{30}$Cu$_{6.6}$ alloy were not investigated here. This is because these alloys are much more brittle than the other studied Ti–Fe–Cu alloys, which made it impossible to prepare specimens for the Young’s modulus measurement. D. Dyja et al. reported hardness and brittleness of alloys are increased by formation of nano-size precipitates phase.\(^{23}\) This report suggests that there is possibility of formation of nano-size precipitates, which is difficult to verify using a SEM observation.

As shown in Fig. 8(a), the measured Young’s moduli of the studied Ti–Fe–Cu alloys are from 110 to 145 GPa. The highest is found to be that of the as-cast Ti$_{68}$Fe$_{30}$Cu$_2$ alloy (145 GPa). This value is less than those of the main materials used for metallic stents, ASTM F90 (230 GPa)\(^{24}\) and 316L SS (210 GPa).\(^{10}\) However, this value is much higher than those of CP-Ti (105 GPa)\(^{29}\) and Ti–64 ELI (110 GPa).\(^{30}\) Das et al. have reported that Young’s modulus of Ti$_{51.5}$Fe$_{48.5}$ (which has a single TiFe intermetallic phase) is 222 GPa.\(^{15}\) This report supports the conclusion that the high Young’s modulus of the Ti$_{68}$Fe$_{30}$Cu$_2$ alloy is due to the formation of the TiFe intermetallic phase. However, the changes in Young’s moduli of the studied Ti–Fe–Cu alloys do not depend only on the area fraction of the TiFe intermetallic
phase. The Young’s moduli are saturated or slightly decreased when the area fraction of the TiFe intermetallic phase is more than 34\% , as shown in Fig. 8(b).
alloy are less than that of the as-cast Ti68Fe30Cu2 alloy, despite the high area fraction of the TiFe intermetallic phase. This is because the bond order of the Ti68Fe30Cu2 alloy is higher than that of the other studied Ti-Fe-Cu alloys, as shown in Fig. 1. Moreover, the decrease in the Young’s modulus after the heat-treatment is caused by a decrease in the bond order of the TiFe intermetallic phase associated with the increase in Cu content. These results indicate that the effect of the change in the bond order on Young’s modulus is more significant than the effect of the change in area fraction of the TiFe intermetallic phase.

3.2.2 Hardness

Figure 9(a) shows the Vickers hardness of the studied Ti-Fe-Cu alloys before (as-cast) and after the heat-treatment. It is obvious that the Vickers hardness of the studied Ti-Fe-Cu alloys is much higher than those of CP-Ti (145 HV),25) Ti-64 ELI (320 HV),26) and the main materials used for metallic stents, Co–20Cr–15W–10Ni alloy (ASTM F90) (250 HV),24) and 316L SS (220 HV).27) The Vickers hardness of the as-cast alloys increases from 490 HV (Ti78Fe18Cu4 alloy) to 550 HV (Ti63.4Fe30Cu6.6 alloy) with increasing atom ratio (Fe + Cu)/Ti of the alloys. On the other hand, the Vickers hardness decreases by approximately 6% after the heat-treatment for all alloys. Figure 9(b) shows the Vickers hardness of the studied Ti-Fe-Cu alloys as a function of the area fraction of the TiFe intermetallic phase. As shown in Fig. 9(b), the hardness of as-cast Ti38Fe14Cu4 alloy is much higher (490 HV) than that of β-type Ti alloys such as TNTZ (270 HV) despite the low area fraction of TiFe intermetallic phase (1.5%). As abovemention, there is possibility of formation of nano-size TiFe intermetallic phase, which is difficult to verify using a SEM observation. Moreover, the increase in the Vickers hardness of the studied Ti-Fe-Cu alloys depends on the area fraction of the TiFe intermetallic phase, which was determined from the optical micrographs. These results indicate that the comparatively high hardness of the studied Ti-Fe-Cu alloys is due to the formation of the TiFe intermetallic phase. However, the Vickers hardness of the heat-treated alloys is lower than that of the as-cast alloy, despite the increase in the area fraction of the TiFe intermetallic phase after the heat-treatment. This is caused by the decrease in the bond order of the TiFe intermetallic phase associated with the increase in Cu content.
3.2.3 Compressive properties

Figure 10 shows (a) the compressive strength $\sigma_{\text{max}}$ and (b) the compressive strain $\varepsilon_c$ of the Ti$_{63}$Fe$_{30}$Cu$_6$ alloy, Ti$_{60}$Fe$_{30}$Cu$_4$ alloy, and Ti$_{63.4}$Fe$_{30}$Cu$_{6.6}$ alloy before (as-cast) and after the heat-treatment. These alloys all have comparatively high Young’s moduli. The compressive strengths of the as-cast alloys are lower than those of the Ti$_{60}$Fe$_{30}$Co$_{15}$ alloy and Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloy reported by Louzguine et al. However, the compressive strengths of all studied alloys increase significantly after the heat-treatment (86.4 ks). The highest compressive strength of all the studied Ti–Fe–Cu alloys is 2131 MPa, which is observed for the heat-treated (86.4 ks) Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloy. This compressive strength is lower than that reported for Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloy (2350 MPa), but it is higher than that reported for Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloy (2110 MPa). The compressive strengths of the studied as-cast Ti–Fe–Cu alloys are higher than those of the Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloys (5.5%). Moreover, the compressive strengths of the Ti$_{60}$Fe$_{30}$Cu$_6$ alloy and Ti$_{67}$Fe$_{27}$Cu$_6$ alloy increase significantly after the heat treatment for 86.4 ks. In particular, the compressive strain of the Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloy reaches 24.5%. This compressive strain is higher than that reported for Ti$_{60}$Fe$_{25}$Co$_{15}$ alloy (16.5%). This increase in the compressive strain of the heat-treated Ti–Fe–Cu alloys is due to the decrease in the hardness, which is caused by the decrease in the bond order of the TiFe intermetallic phase associated with the increase in Cu content.

The compressive strength of as-cast alloys is lower than that of heat-treated alloys despite the comparatively high hardness. This is because that fracture of as-cast alloys does not occur in plastic region but occurs in elastic region. After the heat-treatment, because the hardness decreases, the fracture occurs in the plastic region. Therefore, the compressive strength and compressive strain increase. This is well-known phenomenon for brittle alloys, which have intermetallic phase.

The results of this study indicate that heat-treated Ti$_{60}$Fe$_{30}$Cu$_6$ alloy has significant potential as a material for metallic stents because of its high Young’s modulus, excellent hardness, and excellent compressive strength, and compressive strain.

4. Conclusions

In order to develop Ti alloys for metallic stents, Ti$_{73}$Fe$_{22}$Cu$_5$ alloy, Ti$_{78}$Fe$_{18}$Cu$_4$ alloy, Ti$_{68}$Fe$_{30}$Cu$_2$ alloy, Ti$_{67}$Fe$_{25}$Cu$_{12}$ alloy, Ti$_{63.4}$Fe$_{30}$Cu$_{6.6}$ alloy were designed using a d-electrons alloy design method. The microstructures and mechanical properties of the as-cast and heat-treated above designed Ti–Fe–Cu alloys were investigated. The following results were obtained:

(1) The studied Ti–Fe–Cu alloys consist of the $\beta$-Ti phase and dendritic TiFe intermetallic phase. Moreover, the area fraction of the TiFe intermetallic phase increases with increasing atom ratio (Fe + Cu)/Ti of the alloys.

(2) The area fraction of the TiFe intermetallic phase and Cu content in the TiFe phase of the studied Ti–Fe–Cu alloys increase after they are subjected heat-treatment under a high-purity argon atmosphere at 1173 K for 21.6 and 86.4 ks.

(3) The Young’s moduli of the studied Ti–Fe–Cu alloys increase from 110 GPa (Ti$_{78}$Fe$_{18}$Cu$_4$ alloy) to 145 GPa (Ti$_{63.4}$Fe$_{30}$Cu$_{6.6}$ alloy) with increasing atom ratio (Fe + Cu)/Ti of the alloys and with increasing area fraction of the TiFe intermetallic phase. However, the Young’s moduli are saturated or slightly decreased when the area fraction of the TiFe intermetallic phase is more than 34%.

(4) The Vickers hardness of the as-cast studies Ti–Fe–Cu alloys increase from 110 HV (Ti$_{78}$Fe$_{18}$Cu$_4$ alloy) to 145 HV (Ti$_{63.4}$Fe$_{30}$Cu$_{6.6}$ alloy) with increasing atom ratio (Fe + Cu)/Ti of the alloys and with increasing area fraction of the TiFe intermetallic phase. On the other hand, the Vickers hardness of the heat treated studied alloys is lower than that of the as-cast alloys, despite the increase in the area fraction of the TiFe intermetallic phase after the heat-treatment.

(5) The heat-treated Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloy has better compressive properties than those of the other studied Ti–Fe–Cu alloys and reported Ti–Fe–Cu alloys. The compressive strength and compressive strain of the heat-treated Ti$_{60}$Fe$_{25}$Cu$_{12}$ alloy are 2131 MPa and 24.5%, respectively.
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