

Sintered Ti–Fe alloys with in situ synthesized TiC dispersoids

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ABSTRACT

Titanium and its alloys can be strengthened by adding carbon because of the solid solution strengthening and dispersion strengthening effects. The level of strengthening depends on the source and the amount of carbon chosen. This study compares the microstructures and mechanical properties of Ti–7Fe–xC, with the carbon provided by graphite powders and two carbonyl iron powders. All sintered compacts reached relative densities of about 97%. The compacts produced using a mixture of Ti powders and OS carbonyl iron powders, which contain SiO₂, had the highest tensile strength of 1007 MPa, a hardness of 34HRC, and an elongation of 10%. The microstructure shows that fine TiC and SiO₂ dispersoids were uniformly distributed in the Ti–7Fe matrix, contributing to strengthening. The evolution mechanism of TiC is proposed based on the microstructure observations and Thermo-Calc calculations.

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1. Introduction

Titanium and its alloys are excellent materials for advanced applications owing to their high specific strength and corrosion resistance. However, practical applications have been limited to a few advanced fields, such as the aerospace, biomedical, and chemical industries, because of the high cost in materials and manufacturing, in particular the machining. Therefore, economic production processes, such as powder metallurgy and metal injection molding, which are near-net-shape processes, have become an attractive topic in recent research on titanium alloys [1,2].

Among the titanium alloys, Ti–6Al–4V has excellent mechanical properties, with a complex microstructure of α (hcp) + β (bcc) phase after solutioning and aging [3]. However, since vanadium is costly and is not biomedically friendly, the search for other titanium alloys with α + β microstructure has drawn interest from researchers. Among the β -stabilizing elements, other than aluminum and vanadium, iron has a low cost and a high diffusion rate into titanium, and thus can be homogenized quickly during sintering and provides good mechanical properties, as compared to other elements [4]. Several previous reports have indicated that the powder metal Ti–Fe system can be sintered to high densities and that its mechanical properties can be optimized through composition adjustment [5–9], fine-tuning of sintering schedules [10], and heat treatment [11,12].

To make the Ti–Fe system even more compatible to Ti–6Al–4V in mechanical properties, the Ti–Fe matrix can be further

strengthened by TiC dispersoids, which are stable and have good bonding strength at the particle–matrix interface [13]. However, different levels of success have been reported, depending on the characteristics of the TiC. When formed in situ between titanium and carbon during sintering, the size of TiC particles is finer and the distribution is more uniform than those of using TiC powders. Various carbon sources have been applied to in situ synthesis of the TiC dispersoids during sintering of titanium and titanium alloys, including graphite powder [13–15], carbon black [16], carbon nanotubes (CNTs) [17,18], and methane atmosphere [19]. In general, the yield strength, tensile strength, and hardness can be increased due to the presence of in situ formed TiC dispersoids, and an elongation of over 20% can still be attained [16,18].

Another method of producing in situ synthesized TiC particles is to add non-TiC carbides. Cr₃C₂ was used as an additive to Ti–1.5Fe–2.25Mo–1.2Nd–0.3Al to develop TiC/titanium matrix composites [20]. The strength was improved due to the formation of TiC particles as a result of reaction between Ti and Cr₃C₂, and the solution strengthening effect of Cr. The B₄C had also been demonstrated in castings and laser claddings of titanium alloys as a good source of carbon because both TiC and TiB particles form during processing and improve the mechanical properties [21].

These previous studies have indicated that Ti–Fe is a good titanium alloy with the potential to replace Ti–6Al–4V in some applications. Moreover, carbonyl iron powders have been used successfully in several studies [6,8,12,20]. However, it was not clear whether the carbonyl iron powders used previously contained carbon or not, which depends on whether a decarburization process is employed after decomposing iron carbonyls, Fe(CO)₅, into carbonyl iron powders. Thus, it is of our interest to examine the effect of intrinsic carbon contained in the carbonyl iron powder on the

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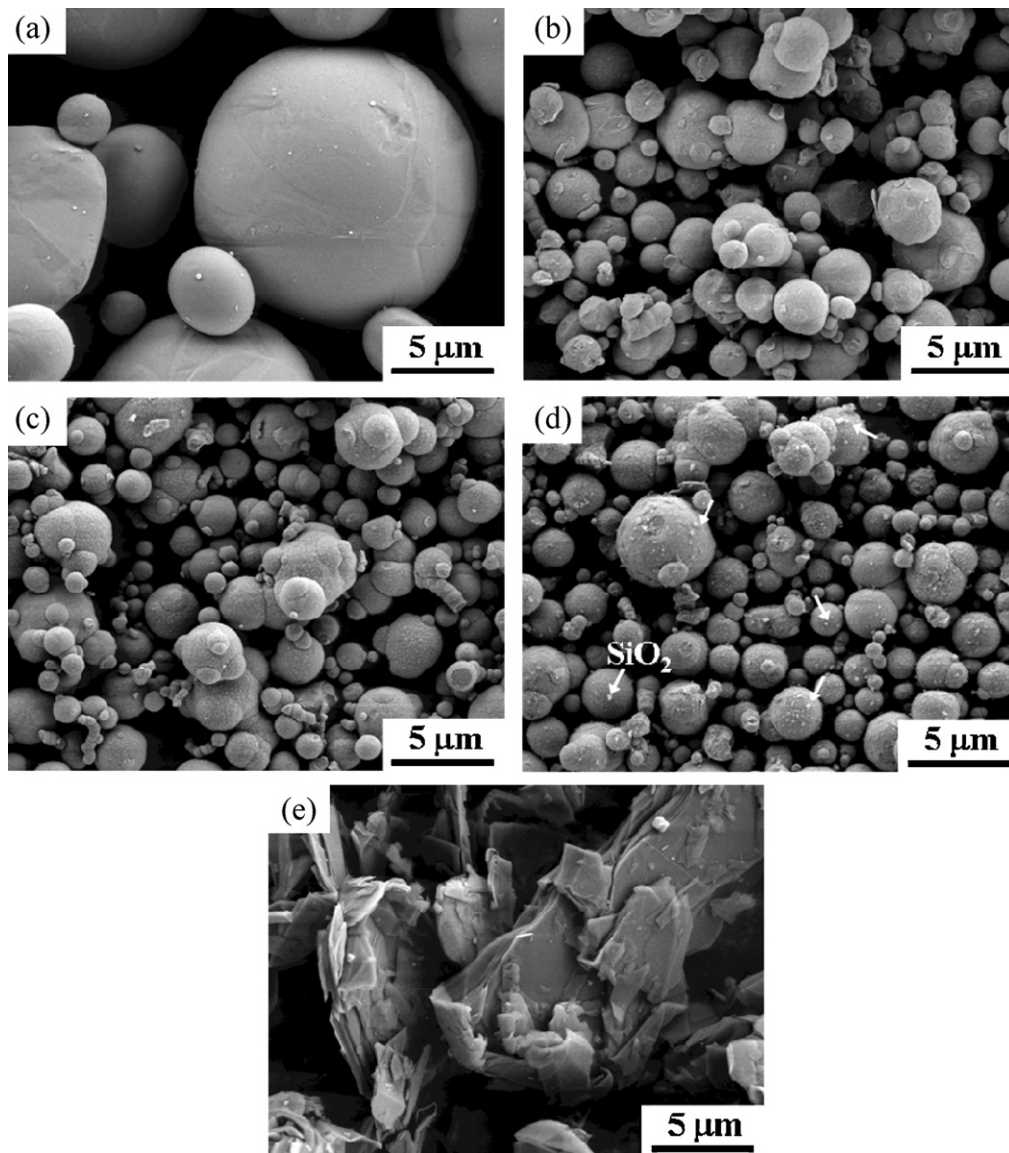


Fig. 1. The morphology of as-received powders used in this study: (a) Ti, (b) CM, (c) OM, (d) OS, and (e) graphite.

formation of TiC and the resulting microstructure and mechanical properties of Ti–Fe alloys. In this study, three different carbonyl iron powders were used. The results were compared to those of using graphite powders.

2. Experimental procedure

Gas atomized titanium powder with low oxygen and carbon contents was used as the baseline powder in this study. To compare the effect of different carbon sources on the characteristics of in situ formed TiC, three carbonyl iron powders (CM, OM and OS grades) and graphite powders were used. The characteristics of these powders are given in Table 1 and Fig. 1. The carbon contents of the as-received OM and OS powders were 0.790 and 0.783 wt.%, respectively. The as-received OS powder also had 0.7 wt.% SiO₂ powder, as shown in Fig. 1(d), which was added in the powder-making process to alleviate the agglomeration problem and to improve the flowability of the powder. Both OM and OS powders had an onion-ring structure with interlayers of iron and carbon, as shown in Fig. 2(a) and (b), as a result of the decomposition process of

iron carbonyls. The CM powder, which had been reduced under a hydrogen-containing atmosphere, contained only 0.006 wt.% carbon, and no onion-ring structure was observed, as shown in Fig. 2(c).

To produce Ti–7Fe specimens, four different mixed powders were prepared. For carbon-free specimens, titanium powders were mixed with 7 wt.% CM carbonyl iron powder. To examine the effect of using graphite powder as the source of carbon on forming TiC, the Ti–7CM powder was also mixed with 0.05 wt.% graphite powder, and the mixed powder was designated as Ti–7CM–0.05G. To provide carbon from carbonyl iron powders, OM and OS iron powders were used to prepare the specimens, which were designated as Ti–7OM and Ti–7OS, respectively. All mixing was carried out using a turbula mixer (T2C, Willy A. Bachofen AG Maschinenfabrik, Basel, Switzerland) for 1 h. The admixed powder was compacted into 80% dense compacts, 40 mm × 15 mm × 2.8 mm, under a pressure of about 350 MPa. No lubricants were used. The green compacts were placed on a zirconia plate, covered with a zirconia cup, and surrounded by Ti powders, which acted as a getter for the residual carbon, oxygen, and nitrogen in the vacuum

Table 1
Characteristics of the titanium, iron, and graphite powders used in this study.

Designation	Ti	Fe			Graphite
	TILOP-45	CM	OM	OS	1651
Particle size distribution, μm (laser scattering)	$D_{90} = 49.0$ $D_{50} = 25.2$ $D_{10} = 9.8$	$D_{90} = 24.9$ $D_{50} = 8.6$ $D_{10} = 3.5$	$D_{90} = 7.5$ $D_{50} = 4.1$ $D_{10} = 2.1$	$D_{90} = 7.6$ $D_{50} = 3.9$ $D_{10} = 1.9$	$D_{90} = 21.6$ $D_{50} = 9.2$ $D_{10} = 3.5$
Shape	Spherical	Spherical	Spherical	Spherical	Flaky
True density, g/cm^3 (Pycnometer)	4.50	7.58	7.63	7.61	2.45
Carbon content, wt.%	0.006	0.006	0.790	0.783	–
Oxygen content, wt.%	0.134	0.255	0.230	0.662	–
SiO_2 , wt.%	–	–	–	0.7	–
Supplier	OSAKA Titanium Technologies	BASF	BASF	BASF	Asbury Carbons

furnace during sintering. The specimens were heated at a heating rate of $5^\circ\text{C}/\text{min}$ to 1150°C , followed by sintering for 2 h under a vacuum of 6.67×10^{-3} Pa. The sintered compacts were furnace-cooled by turning off the power between 1150°C and 900°C and then fan-cooled between 900°C and 30°C . Additional post-sintering heat treatment was applied for Ti–70M and Ti–70S compacts; at the end of the isothermal sintering period, these specimens were furnace-cooled from 1150°C to 740°C , held for 24 h, and then fan-cooled to room temperature [5]. To understand the evolution of the in situ synthesized TiC particles during the sintering cycle, green compacts were sealed in quartz tubes without a getter under a vacuum of 5×10^{-3} Pa and were heated to six different stages of the sintering cycle and then water quenched. The first, second, and third specimens were heated to 700°C , 900°C , and 1150°C , respectively. The fourth specimen was heated to 1150°C and held at that temperature for 2 h prior to water quenching. The fifth and sixth specimens were sintered at 1150°C for 2 h and then furnace cooled to 900°C and 700°C , respectively, and then quenched.

The sintered density was measured using the Archimedes' method, as per Metal Powder Industries Federation (MPIF) standard 42. The total amount of carbon in the as-received powders and sintered compacts was measured using a carbon/sulfur analyzer (EMIA-220V, HORIBA, Kyoto, Japan), and for oxygen, using an oxygen/nitrogen analyzer (TC-136, LECO Co., St. Joseph, Michigan, USA). The hardness was measured with a Rockwell hardness tester (ARK-600, Mitutoyo Co., Tokyo, Japan). To examine the tensile properties, sintered plates were machined into tensile test specimens using a wire-electrical-discharge machine. The tensile strength and elongation were measured with a universal tensile tester (AG-10TE, Shimadzu Co., Kyoto, Japan).

For microstructure analysis using a scanning electron microscope (FE-SEM, LEO-1530, LEO Electron Microscopy Co., Cambridge, UK), specimens were prepared following the standard grinding and polishing procedure and then etched using a mixed solution of 6% H_2O_2 , 2% HF, and 92% H_2O . The grain size was measured using the line intercept method. The data reported are averages of about 200 intercepts measured from twenty lines each in three different metallographs with a magnification of $100\times$. The volume fraction

of in situ formed TiC was averaged from three metallographs at a magnification of $500\times$ using the point-counting method. For transmission electron microscopy examinations (TEM, Tecnai G² LaB₆, 200 kV, FEI Company, Hillsboro, OR, USA), thin foil specimens were prepared by mechanical grinding and polishing, followed by argon ion milling (PIPS 691, Gatan Inc., Pleasanton, CA, USA). The chemical composition of the phases present and the distribution of Ti, Fe, and C in sintered compacts were examined using an electron probe microanalyzer (EPMA, JXA-8200SX, JEOL Co., Tokyo, Japan).

In order to understand the evolution of the TiC compound during sintering and to explain the differences in the microstructure of the four specimens, the solubility of carbon in Ti–7Fe was calculated using thermodynamics software and a SSOL4 database (Thermo-Calc Software, Stockholm, Sweden).

The corrosion resistance of sintered Ti and Ti–70S compacts in this study was compared using the polarization curve, along with the Grade 2 CP Ti (ASTM B348 Gr. 2) and Grade 5 Ti–6Al–4V (ASTM B348 Gr. 5), both supplied by S-Tech Corp., Tainan, Taiwan. The electrochemical polarization test was conducted in 1 N HCl solution at 60°C . The conventional three-electrode-cell was used with a saturated calomel electrode (SCE) as the reference electrode and a platinum plate as the counter electrode. The potential was increased continuously from -750 mV to 4000 mV at a scanning rate of 10 mV/s using an EG&G 263A Galvanostat/Potentiostat.

3. Results

3.1. Sintered density

Fig. 3 shows that, after being sintered at 1150°C for 2 h, relative densities of all compacts were between 97% and 98%, regardless of the powder used. Although it was expected that the fine in situ synthesized TiC could retard the grain growth and thus enhance densification, the Ti–7CM, which is free of TiC, still reached a density of 97%. This result suggests that the fine titanium and iron powders and the high interdiffusion rate are the main driving forces for densification, and the grain-boundary-pinning effect by

Table 2
Contents of carbon and oxygen in admixed powders and sintered compacts.

	Carbon content, wt.%		Oxygen content, wt.%	
	Admixed powder	Sintered compact	Admixed powder	Sintered compact
Ti–7CM	0.006	0.004	0.142	0.155
Ti–7CM–0.05G	0.056	0.055	0.142	0.130
Ti–70M	0.061	0.064	0.141	0.121
Ti–70S	0.060	0.058	0.171	0.187

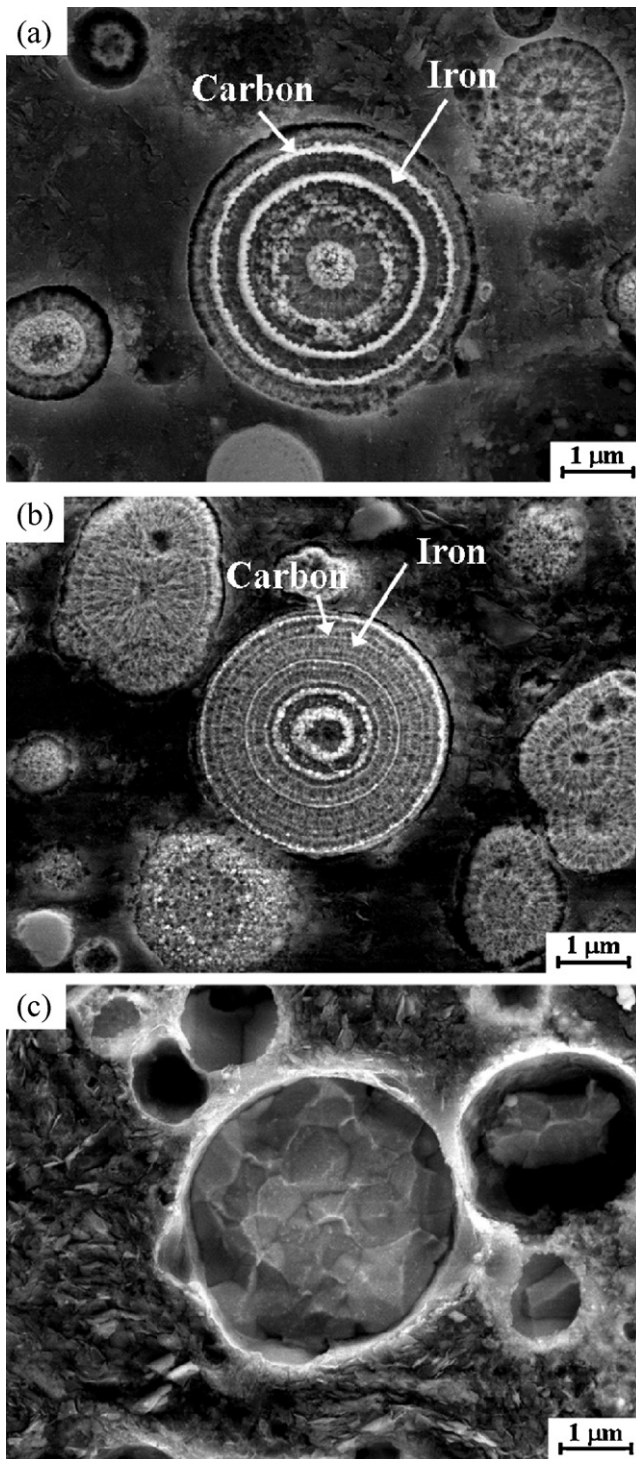


Fig. 2. The cross-section of carbonyl iron powders used in this study: (a) OM, (b) OS, and (c) CM.

in situ synthesized TiC was limited in the Ti-7CM-0.05G, Ti-7OM, and Ti-7OS compacts.

3.2. Carbon and oxygen contents

Table 2 shows the contents of carbon and oxygen in admixed powders and sintered compacts of Ti-7CM, Ti-7CM-0.05G, Ti-7OM, and Ti-7OS. The carbon contents of sintered parts were close to those of admixed powders, suggesting that the

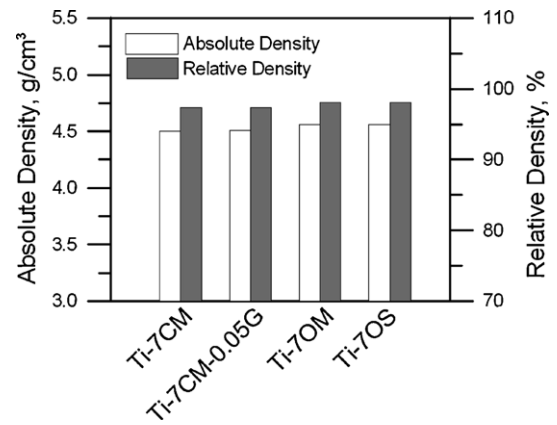


Fig. 3. The absolute and relative densities of sintered TiC/Ti-7Fe compacts, showing that all sintered densities are between 97% and 98%.

contamination of carbon from the sintering atmosphere was minimal. The oxygen contents of sintered compact were also close to those of admixed powders.

3.3. Microstructure

Fig. 4 shows the microstructure of all sintered compacts. The grain sizes of the Ti-7CM, Ti-7CM-0.05G, and Ti-7OM compacts were similar, about 147–156 μm , as shown in Table 3. However, when OS powder was used, the grain size decreased to 128 μm . These compacts were also examined using SEM at high magnifications. Fig. 5(a) shows $\alpha + \beta$ lamellar structure in sintered Ti-7CM compacts. No TiC particulates were observed due to its low carbon content of 0.004 wt.%. When graphite powder or carbonyl iron powder was employed, in situ synthesized TiC particles were observed, as shown in Fig. 5(b)–(d) for Ti-7CM-0.05G, Ti-7OM, and Ti-7OS compact, respectively. Table 3 shows that the mean particle sizes of the TiC in Ti-7CM-0.05G, Ti-7OM, and Ti-7OS compacts were close, between 3.3 and 3.6 μm . The volume fractions of TiC in these three compacts were also similar. Thus, the smaller grain size in the Ti-7OS compact must be caused by the fine SiO_2 particles in the OS carbonyl iron powder.

Fig. 5 also shows that all three TiC-containing compacts have blocky α and β phases, with some lamellar $\alpha + \beta$ areas situated in between. The reason is that as the α phase precipitated and coarsened during cooling, the iron element in the α phase was repelled to the surrounding region. Moreover, as TiC formed, the relative content of the iron in the neighboring areas also increased, as shown by the line scan in Fig. 6 using EPMA. With a high iron content of more than 7 wt.% in the areas surrounding the α phase and TiC particles, the β phase was stabilized and $\alpha + \beta$ lamellae, finer than those in the Ti-7CM compact, were therefore observed.

The microstructures of Ti-7CM-0.05G and Ti-7OS were also examined using TEM. The high magnification images, as shown in Fig. 7, reveal that nano-scale in situ synthesized TiC particles were

Table 3

The particle size and content of in situ synthesized TiC particles in sintered TiC/Ti-Fe compacts.

	Grain size, μm	TiC particle size, μm	TiC content, vol.%
Ti-7CM	156	–	–
Ti-7CM-0.05G	156	3.6	0.33
Ti-7OM	147	3.3	0.43
Ti-7OS	128	3.4	0.36

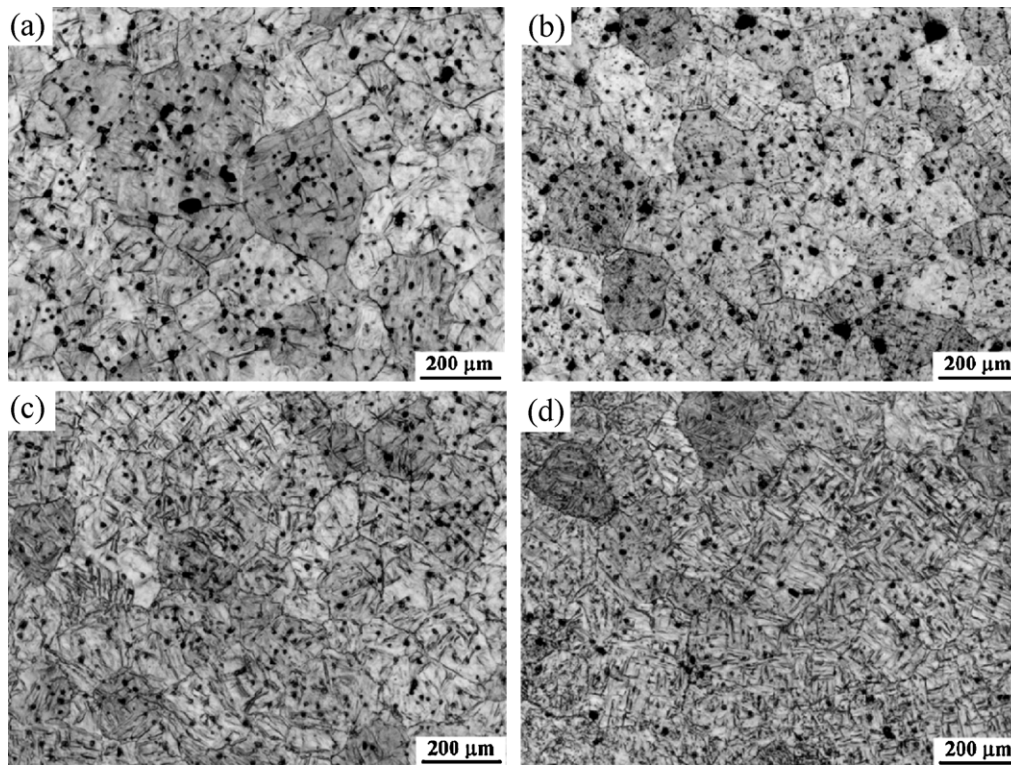


Fig. 4. The metallographs of (a) Ti-7CM, (b) Ti-7CM-0.05G, (c) Ti-70M and (d) Ti-70S, showing that the grain size of the Ti-70S was the smallest.

present within the matrix when the carbon source was carbonyl iron powders. This indicates that the carbon distribution in the specimen prepared with carbonyl iron powders was more homogeneous than occurred with the use of graphite powders, and could form nano-sized TiC particles, in addition to the micron-sized TiC, as shown in Fig. 5.

3.4. Evolution of TiC carbides

To further understand the evolution of in situ synthesized TiC particles, Ti-70S compacts were quenched at different stages of the sintering cycle. Fig. 8(a) shows that TiC particles formed after reaching 700 °C. With heating to 900 °C, the amount of in situ formed TiC

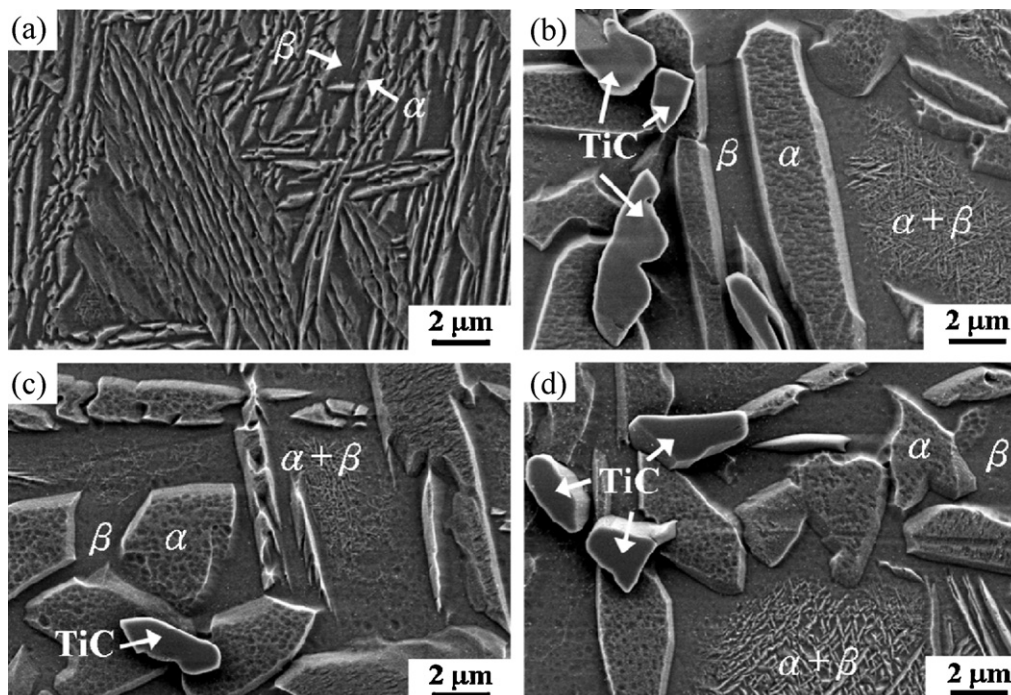


Fig. 5. Microstructures of sintered compacts: (a) Ti-7CM, (b) Ti-7CM-0.05G, (c) Ti-70M, and (d) Ti-70S.

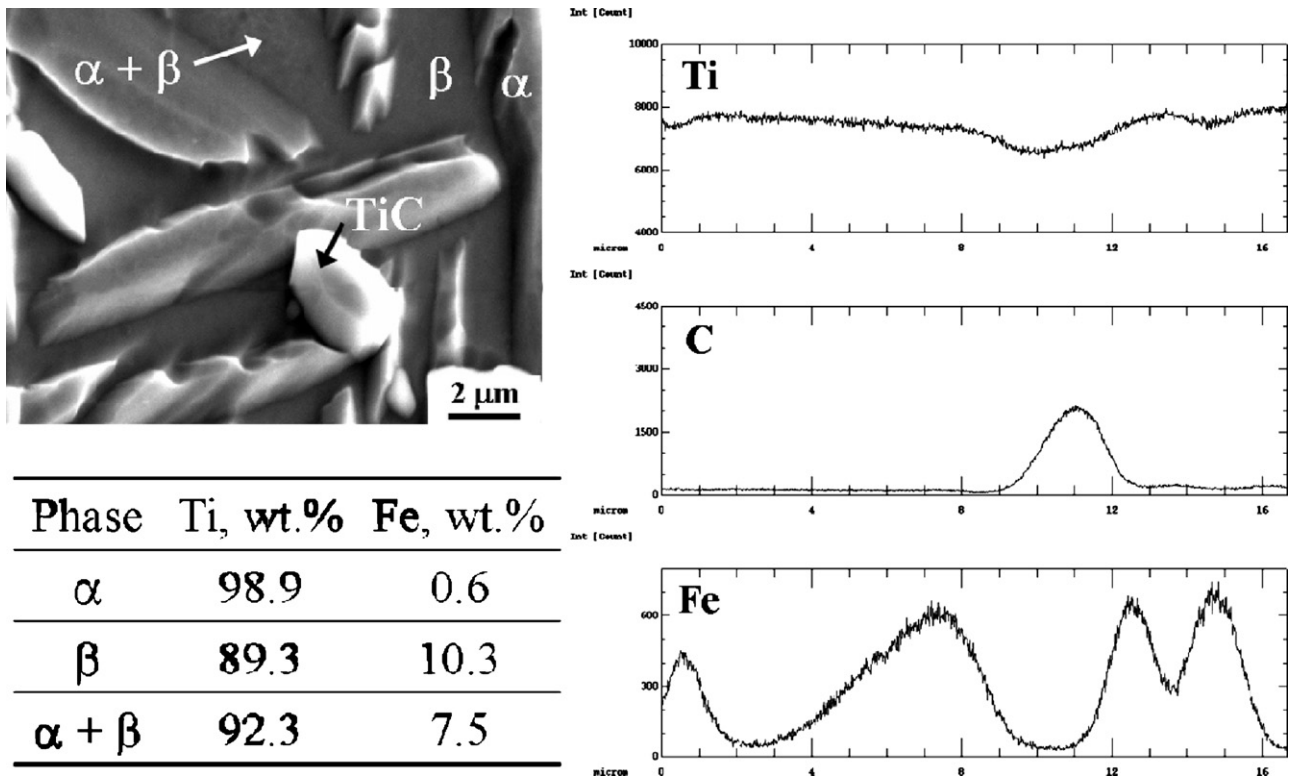


Fig. 6. EPMA quantitative and line scan analysis of sintered-and-fan-cooled Ti-7OS, showing that iron element in the α phase was repelled to the surrounding region.

particles increased, as shown in Fig. 8(b). With heating to 1150 °C, most TiC carbides were distributed at grain boundaries, as shown in Fig. 8(c). However, after sintering for 2 h at 1150 °C, most TiC particles dissolved back into the Ti-7Fe matrix, as shown in Fig. 8(d). When sintered Ti-7OS compacts were furnace cooled, the TiC particles grew in size, as shown in Fig. 8(e) and (f).

The difference in the evolution of TiC during sintering between Ti-7CM-0.05G, Ti-7OM, and Ti-7OS were also examined. The TiC in the Ti-7OM compact evolved similarly to that in the Ti-7OS, but not the Ti-7CM-0.05G. Fig. 9(a) shows that, after heating to 1150 °C, the Ti-7CM-0.05G specimen had a high carbon content in the TiC particles, and the atomic Ti/C ratio was 37.6/62.4, lower than the 62.6/37.4 in the Ti-7OS. This suggests that these particles formed at the original sites of graphite powders. When the specimens were quenched after being sintered at 1150 °C for 2 h,

the amount of in situ formed TiC particles decreased, as shown in Fig. 9(b). Moreover, the Ti/C ratio increased to 57.6/42.4, similar to the 56.4/43.6 in the Ti-7OS specimen.

3.5. Mechanical properties

Table 4 shows that the tensile strength, hardness, and elongation of the Ti-7CM compact were 883 MPa, 26HRC, and 13%, respectively. When 0.05 wt.% graphite powder was added, the tensile strength slightly increased to 915 MPa and the hardness increased to 28HRC due to the in situ formation of TiC particles. When carbon was provided by the OM carbonyl iron powder, the tensile strength and hardness increased to 962 MPa and 33HRC, respectively. The tensile strength and hardness were further improved by using OS carbonyl iron powders, reaching 1007 MPa

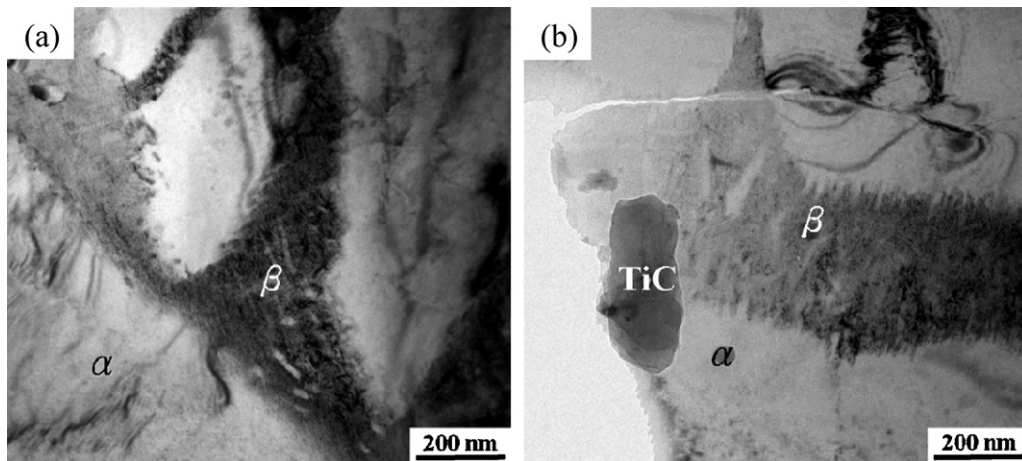


Fig. 7. TEM images of sintered TiC/Ti-7Fe using different carbon sources: (a) Ti-7CM-0.05G and (b) Ti-7OS, showing no TiC particles in the $\alpha + \beta$ region of the Ti-7CM-0.05G compact, but with small TiC particles in the Ti-7OS compact.

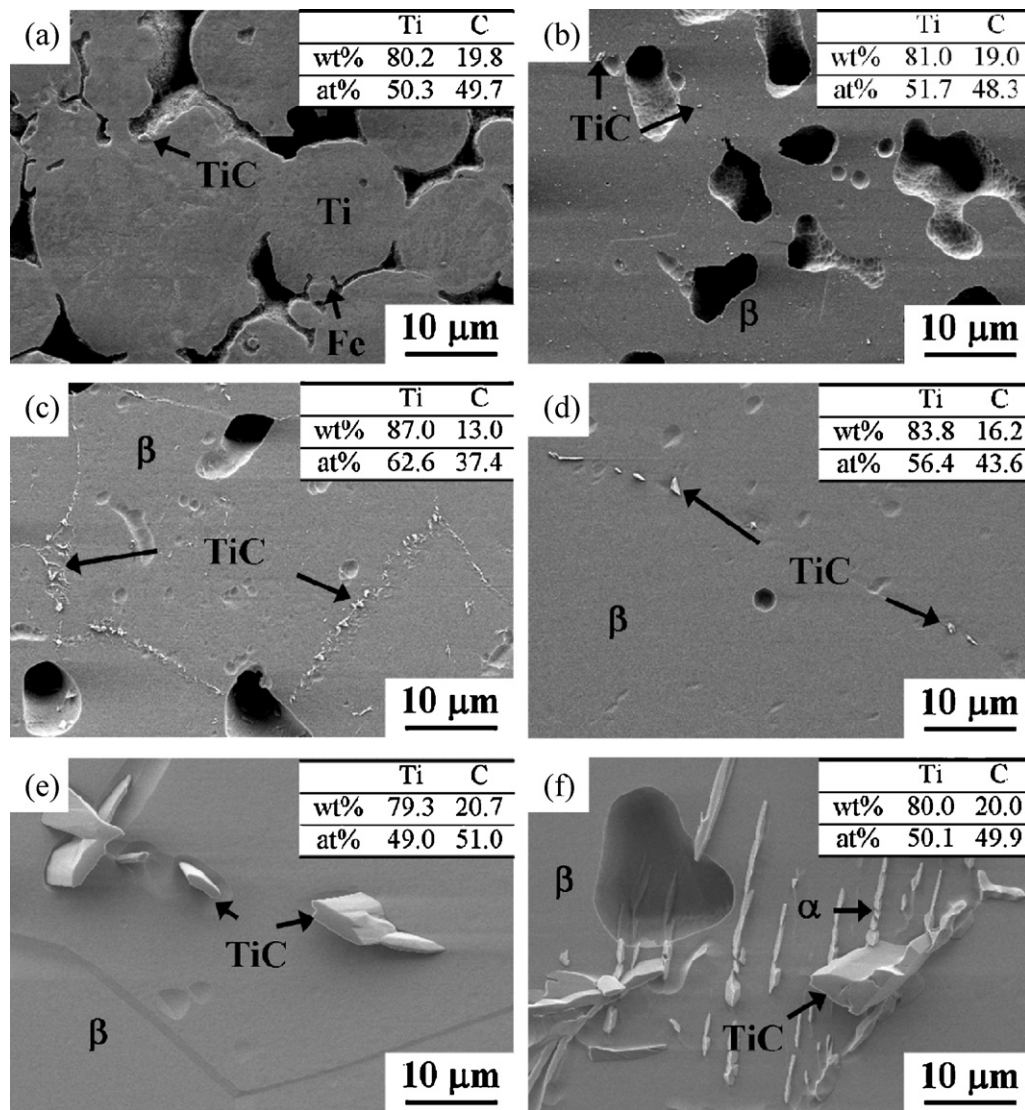


Fig. 8. SEM images of sintered Ti-7OS compacts using water quenching at different stages; (a) after reaching 700 °C, (b) after reaching 900 °C, (c) after reaching 1150 °C, (d) after sintering at 1150 °C for 2 h, (e) after cooling to 900 °C, and (f) after cooling to 700 °C.

and 34HRC, respectively. The elongation of the Ti-7CM specimen was 13%. When carbon was added to form TiC, the elongation of Ti-7CM-0.05G, Ti-7OM, and Ti-7OS specimens decreased slightly, but remained above 10%. These properties are comparable to those of the annealed wrought Ti-6Al-4V (970 MPa tensile strength and 16% elongation) [22]. The sintered Ti-7OM and Ti-7OS specimens

were also heat treated by holding at 740 °C for 24 h, followed by fan-cooling [5]. Table 5

shows that the tensile strength and hardness further increased.

The fracture surface of the Ti-7CM, Ti-7CM-0.05G, Ti-7OM, and Ti-7OS tensile specimens all had dimple structures, as shown in Fig. 10. The dimple sizes of the Ti-7OM and Ti-7OS compacts seem to be smaller than those of the Ti-7CM and Ti-7CM-0.05G

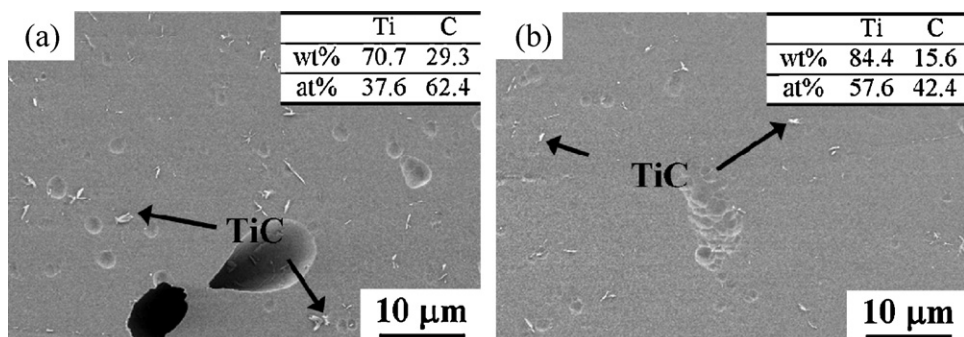


Fig. 9. SEM images of sintered Ti-7CM-0.05G water quenched (a) after reaching 1150 °C and (b) after sintering at 1150 °C for 2 h.

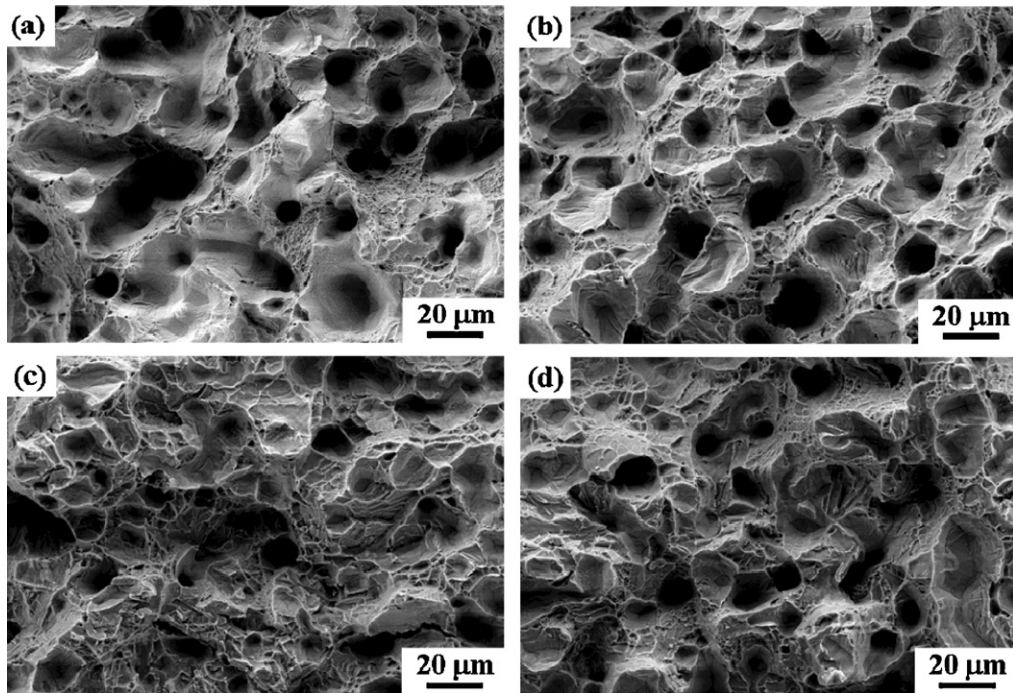


Fig. 10. The fracture surface of (a) Ti-7CM, (b) Ti-7CM-0.05G, (c) Ti-7OM and (d) Ti-7OS, showing smaller dimple size in the Ti-7OM and Ti-7OS compacts.

Table 4

Mechanical properties of sintered TiC/Ti-Fe TMCs.

	Tensile strength, MPa	Hardness, HRC	Elongation, %
Ti-7CM	883	26	13
Ti-7CM-0.05G	915	28	12
Ti-7OM	962	33	11
Ti-7OS	1007	34	10

compacts, probably due to the slightly smaller grain size. A higher magnification was also taken for Ti-7OS, and fine SiO_2 particles were observed in the dimples, as shown in Fig. 11.

3.6. Polarization curve

Although titanium and its alloys are known to have excellent corrosion resistance, it is possible that the addition of iron to the matrix may have a deleterious effect. Majima has demonstrated that the corrosion resistance of Ti-4Fe is similar to that of Ti-6Al-4V and Ti-5Al-2.5Fe [11]. With a high 7 wt.% Fe used in this study, it is of interest to check if the extra iron will impair the corrosion resistance. The polarization curves, as shown in Fig. 12, indicate that pure Ti specimens (CP Ti Grade 2 and sintered Ti) had a slightly better corrosion resistance than those of Ti alloys (Ti-6Al-4V Grade 5 and sintered Ti-7OS). The corrosion potential, breakdown potential, and passivation range of sintered Ti-7OS, as shown in Table 6, were close to those of the Ti-6Al-4V Grade 5 specimen.

4. Discussion

The results above show the changes in the amount and size of the TiC during the sintering cycle. To better understand the

Table 5

Mechanical properties of heat treated Ti-7OM and Ti-7OS.

	Yield strength, MPa	Tensile strength, MPa	Hardness, HRC	Elongation, %
Ti-7OM	880	998	36	10
Ti-7OS	913	1030	36	9

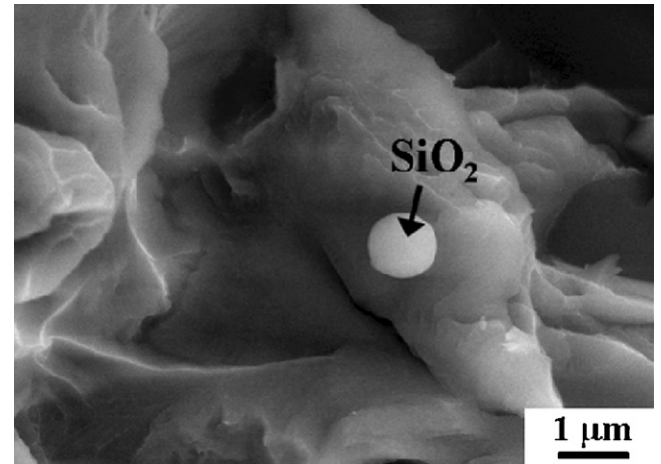


Fig. 11. High magnification of the fracture surface of Ti-7OS tensile specimen, showing SiO_2 particle in the dimple.

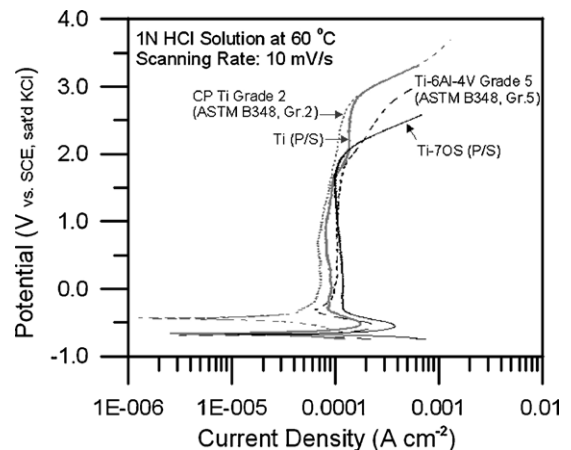
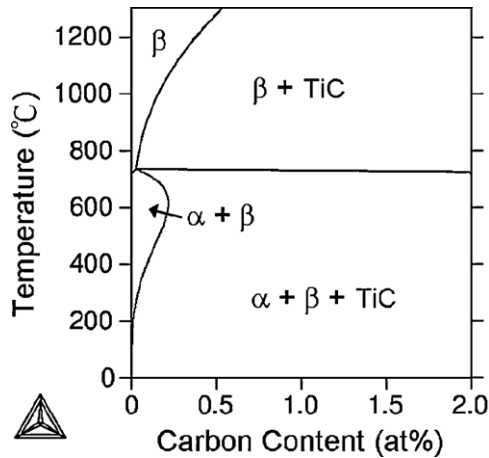


Fig. 12. Polarization curves of Ti and Ti alloys showing that the corrosion behavior of sintered Ti-7OS was close to that of Ti-6Al-4V.

Table 6

The corrosion potential and corrosion current density of the sintered Ti and Ti alloys and their wrought counterparts.

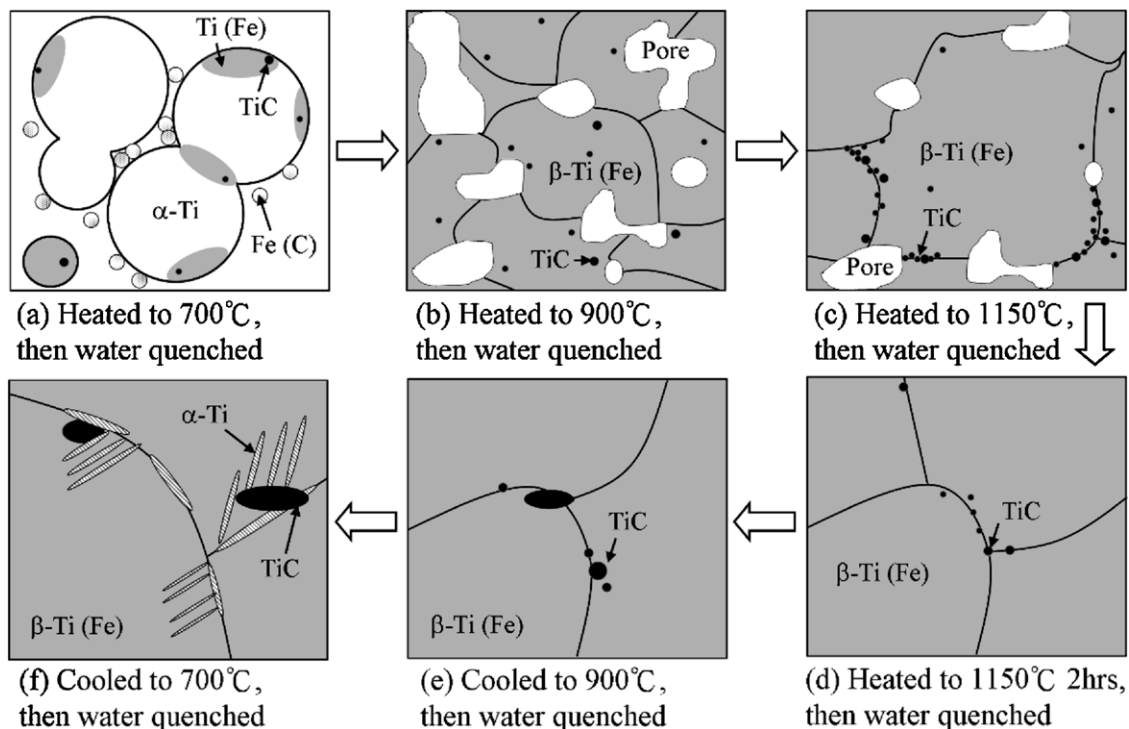
	E_{corr} , mV vs. SCE	I_{corr} , A cm ⁻²	I_{pass} , A cm ⁻²	E_{br} , mV vs. SCE
CP Ti Grade 2	-439	1.23×10^{-5}	6.53×10^{-5}	2724
Sintered Ti (P/S)	-670	3.33×10^{-5}	8.51×10^{-5}	2709
Ti-6Al-4V Grade 5	-687	9.40×10^{-5}	9.21×10^{-5}	1869
Sintered Ti-7OS	-671	1.56×10^{-4}	1.17×10^{-4}	1908

 E_{corr} : corrosion potential, E_{br} : breakdown potential, I_{corr} : corrosion current density, I_{pass} : passivation current density.**Fig. 13.** The pseudo-binary phase diagram of Ti-7Fe-xC.

mechanisms behind these changes, the pseudo-binary phase diagram of Ti-7Fe-xC was calculated using the Thermo-Calc software. Fig. 13 indicates that as the temperature increases, the solubility of carbon in the Ti-7Fe increases, reaching a maximum of 0.053 wt.% (0.214 at.%) at about 610 °C, and then decreases. As the compact is further heated, the solubility of carbon in the β phase increases with increasing temperature, reaching 0.075 wt.% (0.300 at.%) at the sintering temperature of 1150 °C. Since the total carbon content in

sintered Ti-7CM-0.05G, Ti-7OM, and Ti-7OS compacts was about 0.060 wt.% (0.240 at.%), the final microstructure in equilibrium contains α , β , and TiC phases.

From the microstructure observations and the calculated phase diagram, the evolution of the TiC dispersoids during the sintering cycle can be established. As depicted in Fig. 14, when compacts were heated to 700 °C, the homogenization of Fe and C was still incomplete, as was confirmed by the X-ray mapping using EPMA. The fast-diffusing carbon reacts with titanium powder and forms TiC, as shown in Figs. 8(a) and 14(a). These TiC particles are quite stable, since the calculated solubility of carbon in Ti-7Fe at 700 °C is 0.035 wt.% (0.140 at.%), lower than the carbon content in sintered Ti-7OM/Ti-7OS, 0.064 wt.% and 0.058 wt.%, respectively, as shown in Table 2. Thus, TiC particles formed at that temperature, as observed in Fig. 8(a). As the temperature increased to 900 °C, the calculated solubility of carbon in Ti-7Fe slightly decreased to 0.020 wt.% (0.080 at.%), which caused more TiC particles to form, as shown in Figs. 8(b) and 14(b). As sintering temperature continued to increase, the solubility of carbon in Ti-7Fe increased and reached 0.075 wt.% (0.300 at.%) at 1150 °C, as shown in Fig. 13. This caused some TiC particles to dissolve back into the Ti-7Fe matrix, and only some TiC particles were left at grain boundaries, as shown in Figs. 8(c),(d) and 14(c),(d). As a result, there was little grain growth inhibition effect by TiC in Ti-7CM-0.05G, Ti-7OM, and Ti-7OS compacts, and their grain sizes were all larger than 100 μm . When sintered Ti-7OS compacts were furnace cooled to 900 °C and 700 °C, the solubility of carbon in Ti-7Fe decreased, and thus the saturated

**Fig. 14.** The schematic of the TiC evolution mechanism in sintered Ti-7OS/Ti-7OM compact.

carbon formed TiC particles again and grew in size, as shown in Figs. 8(e),(f) and 14(e),(f).

The preceding results show that carbon addition increases the tensile strength and hardness of the Ti–7Fe alloy. The mechanism for the improvement could be solid solutioning and/or dispersoid strengthening. It has been shown that interstitial carbon can increase the strength of titanium and its alloys through solid solution strengthening effect [23,24]. Ouchi et al. demonstrated that each 1 wt.% dissolved in pure titanium and SP-700 alloy (Ti–4.5Al–3V–2Fe–2Mo) can cause 590 and 1020 MPa increments for tensile strength, respectively. To determine whether the increase in tensile strength and hardness of the TiC/Ti–7Fe obtained in this study was attributed to solid solutioning or dispersoid strengthening, the data in Tables 2 and 3 were analyzed. The calculations from the TiC contents show that almost all carbon formed TiC after sintering and cooling. This suggests that the improved tensile strength and hardness of the Ti–7CM–0.05G and Ti–7OM compacts are mainly attributed to dispersoid strengthening of TiC. It was observed that the Ti–7OS has a higher strength and hardness than those of the Ti–7OM. Since the characteristics of the TiC particles and the total contents of carbon in these two compacts were similar, as shown in Tables 2 and 3, the plausible explanation for the difference in mechanical properties could be the presence of silica particles in the Ti–7OS compact. These silica particles were quite fine, smaller than 1 μm . They remained well distributed in the matrix after sintering and provided the beneficial strengthening effect.

5. Conclusions

This study focused on the effects of different carbon sources, graphite powder and carbonyl iron powder, on the microstructure and mechanical properties of TiC/Ti–7Fe alloys. The sintered densities of Ti–7CM, Ti–7CM–0.05G, Ti–7OM, and Ti–7OS compacts were all above 97%. The tensile strength and hardness of Ti–7OS compacts increased to 1007 MPa and 34HRC, respectively, and the elongation was 10%, which are comparable to those of the annealed wrought Ti–6Al–4V (970 MPa and 16% elongation). The improved mechanical properties of the TiC/Ti–7Fe compact prepared from pure titanium and OS carbonyl iron powders is attributed to the presence of SiO_2 and in situ formed TiC dispersoids.

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