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Effect of hot rolling and annealing temperatures on the microstructure and mechanical properties of SP-700 alloy

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Abstract: The effect of rolling temperature on both two- and single-phase regions and annealing in a temperature range of 700–950°C on the microstructure and mechanical properties of Ti–5Al–4V–2Fe–1Mo alloy was investigated. The results indicated that the best balance of strength and ductility is obtained by rolling in the two-phase region due to the globularization of the alpha phase and increase in its volume fraction. After rolling in the two-phase region, the ductility of the specimens annealed at 700 to 800°C increased because of the finer size and globularized alpha phase, while the reduction in strength was attributed to a decrease in the alpha phase volume fraction. However, at 950°C, the strength increased and ductility dropped by the formation of acicular alpha phase due to an increase in the phase boundary area. Annealing and aging after rolling in the beta-phase region increased the strength and decreased the ductility, which is attributed to the formation of a secondary alpha phase. A combination of favorable yield strength (1113 MPa) and elongation (13.3%) was obtained through rolling at 850°C followed by annealing at 750°C and aging at 570°C.

Keywords: beta-rich two phase Ti alloy; thermomechanical processing; tensile properties; microstructure

1. Introduction

The SP-700 alloy with the Ti–4.5Al–3V–2Mo–2Fe chemical composition (in wt%) is a beta-rich two-phase titanium alloy and compared with the Ti–6Al–4V alloy, it shows higher toughness, better heat treatability, higher strength, and a wider superplasticity behavior at lower temperatures. By adding more beta-stabilizing elements, the superplastic forming temperature of the SP-700 alloy decreases to 775°C, which is approximately 100°C lower than that of the Ti–6Al–4V alloy. These advantages have made the SP-700 alloy an interesting candidate for use in the aerospace industry, power plant, transportation, and sport fields [1–3].

The thermomechanical treatment of titanium alloys has been carried out in the two-phase and single beta-phase region for improvement in mechanical properties by controlling microstructure. In a previous investigation [4] on the hot deformation of Ti–10V–2Fe–3Al alloy, stress-induced martensite transformation was found to occur after hot deformation and the maximum volume fraction of martensite was achieved at a deformation temperature of 925°C and strain rate of 1 s^{-1}. Wang *et al.* [5] reported that the dominant deformation mechanisms of Ti–35Nb–2Ta–3Zr alloy by friction stir processing are slip, deformation-induced martensite phase transformation, and deformation twining. Solution treatment and aging after rolling can be carried out to achieve favorable mechanical properties in beta titanium alloys. The strength and ductility of titanium alloys are strongly dependent on the microstructure [6]. Gunawarman *et al.* [7] studied the mechanical properties of SP-700 alloy and showed that single annealing in the α + β phase region results in higher fatigue ratio, higher specific strength, and a good combination of these properties. The tensile strength of beta titanium alloys is improved by the precipitation of the secondary alpha phase. The fine alpha precipitates can act as dislocation barriers and increase the yield strength through precipitation hardening. It has been shown that secondary alpha phase precipitate with a needle-like morphology in the beta phase during the aging of TC21 (Ti–6Al–2Sn–2Zr–3Mo–1Cr–2Nb–Si) [8], Ti–3.5Al–5Mo–6V–3Cr2Sn–0.5Fe [9],

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and Ti–6Al–4V [10–11] alloys causes an increase in the strength. The beta grain size and the primary alpha morphology determine the ductility of beta titanium alloys [12]. The studies by Terlinde *et al.* [13] indicate that the Ti–10V–2Fe–3Al alloy containing elongated alpha phase exhibits lower ductility compared to the alloy containing globular alpha phase at the same yield strength level. Besides, in another research work [14], it has been shown that the formation of grain boundary alpha leads to early crack nucleation and results in a lower ductility in Ti–5.4Al–3Mo–1V alloys. It has been reported [6,9] that the formation of fine alpha phase with globular morphology and small beta grain size increases ductility by up to 29% by reducing the effective slip length of Ti–3.5Al–5Mo–6V–3Cr–2Sn–0.5Fe alloy by rolling and annealing in the $\alpha + \beta$ region.

The microstructure and mechanical properties such as the fracture toughness of the solution-treated and aged SP-700 alloy for different cooling rates have been well investigated [15–16]; however, the microstructural evolution and optimization of tensile properties of this alloy during thermomechanical processing have less been studied. In the present study, a new cycle of thermomechanical process including rolling in both the two-phase and the single beta-phase region, annealing, and subsequent aging was applied. The aim of the design of the new cycle is to compare the optimum mechanical properties cycle in the two-phase and the single beta-phase region. Thereafter, the relationship between the rolling and heat-treatment temperatures, the microstructure evolution, and the tensile properties was investigated, and a suitable thermomechanical cycle was introduced for the Ti–5Al–4V–2Fe–1Mo alloy.

2. Experimental

A Ti–5Al–4V–2Fe–1Mo ingot was manufactured through double melting in a vacuum arc melting furnace. To remove the casting structure and reduce the segregations, the ingot was homogenized for 5 h at 1150°C and then primary rolled to a reduction of 60%. The chemical composition of the alloy is presented in Table 1.

Table 1. Chemical composition of the investigated alloy

wt%

Ti	Fe	Mo	V	Al
Balance	1.98	0.78	4.24	4.96

The beta transus temperature of the alloy was estimated by metallographic method to be around (950 ± 10)°C. The schematic of the thermomechanical processing of the Ti–5Al–4V–2Fe–1Mo alloy is shown in Fig. 1. The thermomechanical processing applied is as follows: (I) homogenization at 1150°C for 5 h and subsequently primary hot rolling to a total reduction of 60% at the mentioned temperature; (II) reheating at 1000°C and 850°C for 1.5 h, followed by secondary hot rolling to a total reduction of 55% at both temperatures; (III) solution annealing for 1 h at 700, 750, 800, 850, and 950°C; (IV) aging for 3 h at 570°C. The samples were water-quenched (WQ) after first stage, but air-cooled (AC) after the subsequent stages. To prevent the formation of thermal martensite, the samples were not quenched in water. Investigation of the hot deformation behavior of SP-700 alloy [17] showed that the suitable range of hot deformation is 850–1000°C due to the occurrence of dynamic recrystallization. Previous studies [7,18] have shown that the annealing temperature of SP-700 alloy is in the range of 830–900°C. It has been reported in the AMS4899 standard [19] that SP-700 alloy should be annealed in the temperature range of 682–760°C. Therefore, the selection of annealing temperatures was adjusted according to the mentioned references. For precise investigation of the alloy, the step of annealing tem-

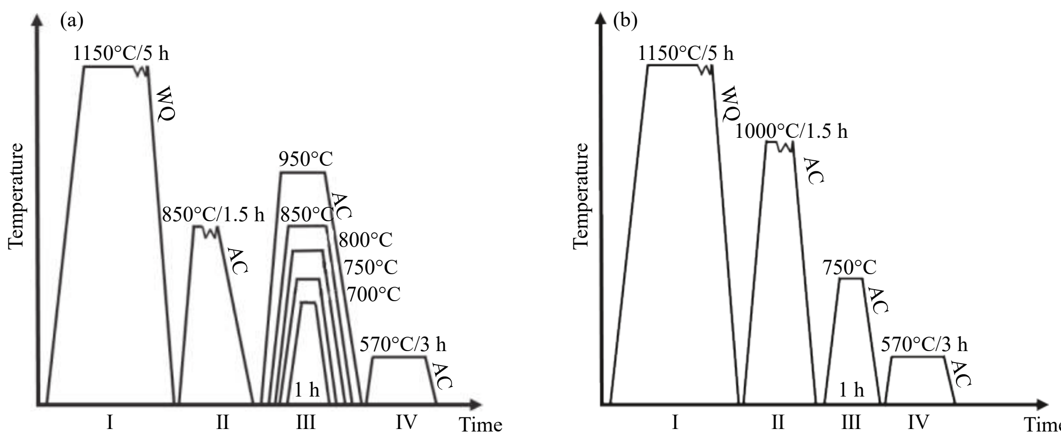

Fig. 1. Schematic of thermomechanical processing: (a) secondary rolling in alpha/beta region; (b) secondary rolling in beta-phase region.

peratures was also reduced.

The tensile tests for rolled and heat-treated samples were carried out at room temperature by an Instron 8502 testing machine at a constant cross head speed of 2 mm/min. The flat tensile specimens were cut according to the ASTM E8M standard, with a 28 mm gage length, 6.25 mm width, and 3 mm thickness. For microstructural investigations, the samples were polished with 100–3000 SiC grinding paper and then etched using Kroll solution (4 mL HF, 2 mL HNO_3, 14 mL H_2O). The microstructural observations were carried out on an Olympus BX 51 optical microscope and a Tescan VEGA3 scanning electron microscope (SEM) in back-scattered electron mode. The SEM was operated under a voltage of 20 kV. The volume fraction and grain size of the alpha and beta phases were measured using ImageJ analysis software according to ASTM E1181 [20].

3. Results and discussion

3.1. Effect of thermomechanical treatment on the microstructure

The microstructure of the secondary hot-rolled Ti–5Al–4V–2Fe–1Mo alloy at 850 and 1000°C after air cooling is depicted in Fig. 2. As seen, the alpha plates with a 1.4 μm width were elongated in the rolling direction and some globular alpha were also formed in the beta matrix due to dynamic globularization during deformation at 850°C. The dynamic globularization of the alpha phase is attributed to the instability of the α/β and α/α phase boundaries [21–22]. Semiatin *et al.* [21] and Weiss *et al.* [22] reported that depending on the applied strain, high density of dislocations accumulate in the alpha layers during the dynamic globularization of alpha. As a result of deformation, shear bands are formed in the alpha layers, and consequently, α/α phase boundary forms in the alpha layers. However, surface tension requirements do not permit a 180° dihedral angle to form between the α/α phase boundary and the beta boundary. Therefore, a driving force is provided for the penetration of the beta phase into α/β boundaries and the simultaneous rotation of the α/β phase boundaries. Then, two recrystallized grains are separated through the penetration of the beta phase, and consequently, the globularization of the alpha phase occurs.

The microstructure of the specimen rolled at 1000°C shows coarse beta grains (264.0 μm), grain boundary alpha, and alpha layers (Fig. 2(b)). When titanium alloys are cooled at a low cooling rate from the single beta-phase region, the alpha phase first nucleates preferentially at beta grain boundaries and forms a continuous alpha layer along the beta grain boundary. As the cooling continues, the alpha plates grow either at the interface of continuous alpha layers or at the beta grain boundaries and move into the beta grains as parallel plates.

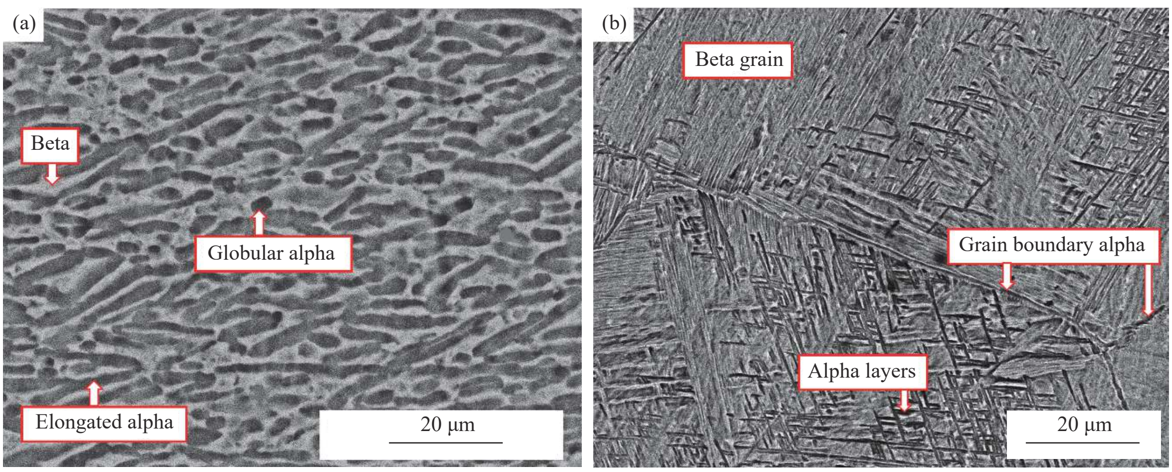

Fig. 2. SEM micrographs of the specimen after secondary hot rolling at (a) 850°C and (b) 1000°C.

The SEM micrographs of the annealed specimen at 700, 750, 800, 850, and 950°C after rolling in two-phase region are shown in Fig. 3. The samples were annealed for 1 h and then air cooled. As shown in Fig. 3(a), the microstructure of the annealed specimen at 700 and 750°C contained elongated and globular alpha phase formed in the beta matrix. The static globularization of the alpha phase occurred in the temperature range of 750 to 850°C, and the globular alpha size increased from 2.2 to 4.5 μm with the annealing temperature increasing from 750 to 850°C. At annealing temperatures of 700 and 750°C, the size and volume fraction of the elongated alpha decreased from 6.4 μm and 10% to 4.5 μm and 5%, respectively (Fig. 4). It has been reported that the static globularization involves two stages of boundary splitting and mass transformation [23–24].

By increasing the annealing temperature up to 950°C, the primary alpha phase completely transformed to beta phase, and then, the acicular alpha phase formed in beta grains dur-

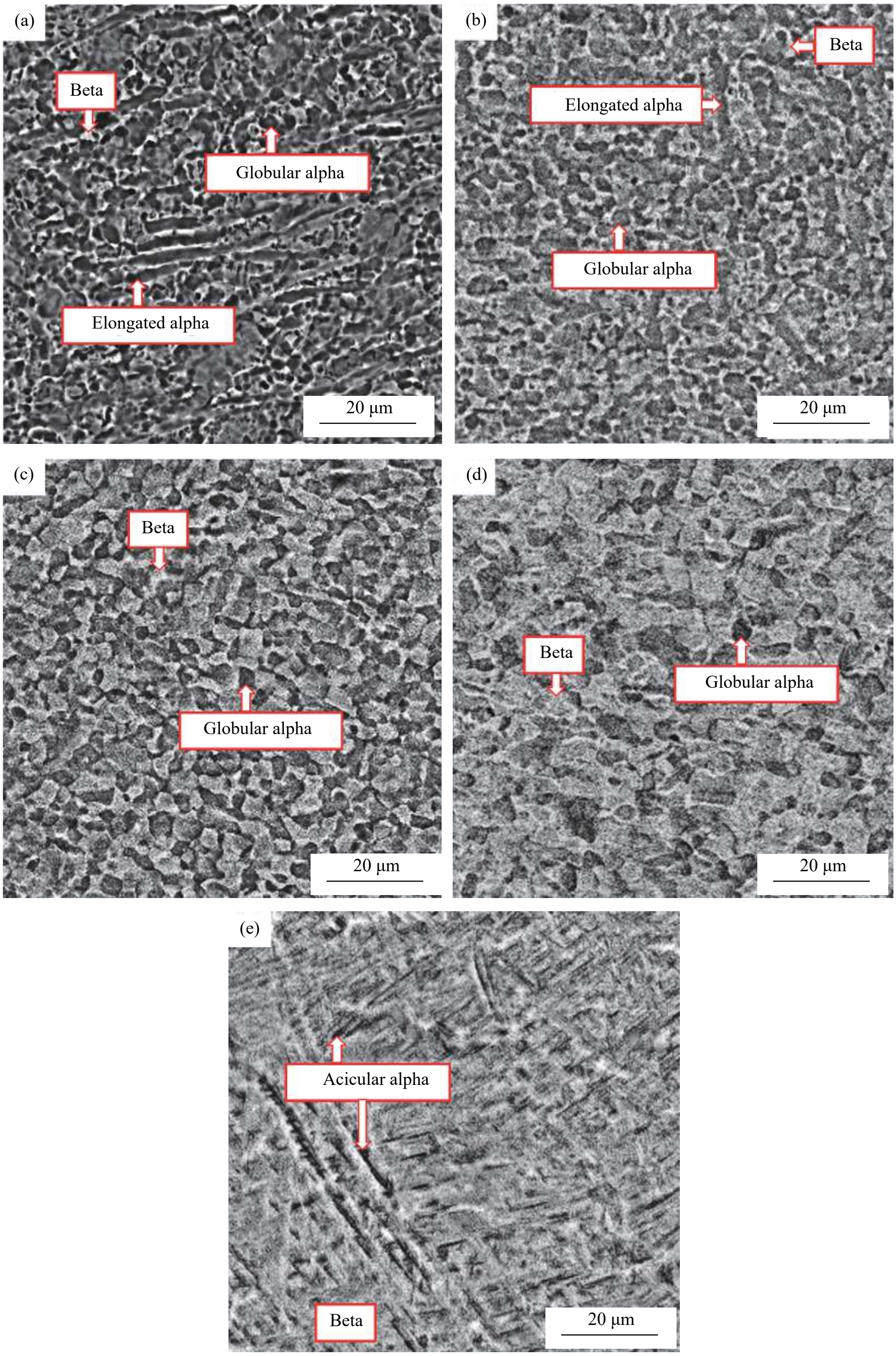

Fig. 3. SEM micrographs of the specimen after rolling at 850°C and annealing at different temperatures: (a) 700°C; (b) 750°C; (c) 800°C; (d) 850°C; (e) 950°C.

ing air cooling (Fig. 3(e)). The alpha phase is rich in alpha-stabilizing elements. This phase does not transform until it reaches the beta transformation temperature. The alpha to beta transformation occurs by increasing the volume fraction of beta phase due to migration of the α/β interface [25]. Previous studies [7,15,26] on the Ti–4.5Al–3V–2Mo–2Fe alloy have also reported the formation of acicular alpha phase in beta grains after air cooling of the solution-treated alloy.

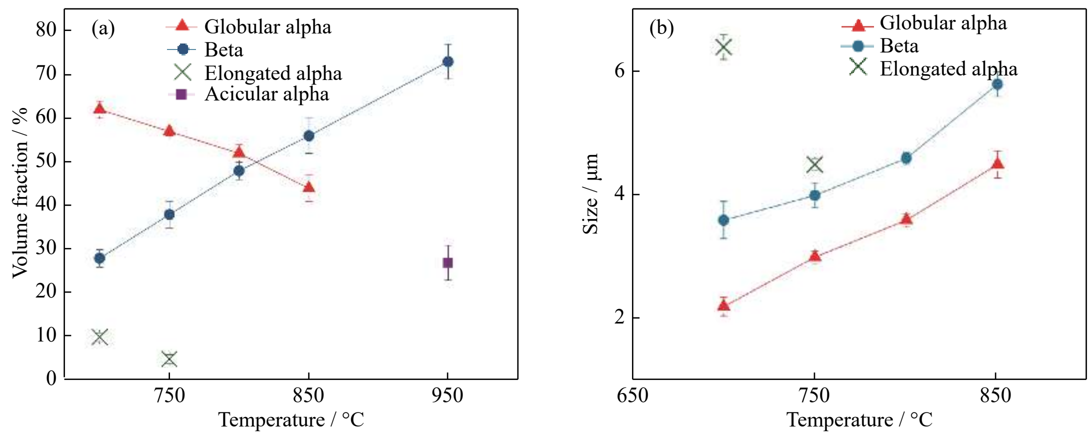

Fig. 4. Variation of (a) the volume fraction and (b) the size of globular and elongated alpha and beta phases as a function of annealing temperature for the Ti–5Al–4V–2Fe–1Mo alloy after rolling at 850°C.

The variation in the size and volume fraction of the phases at different annealing temperatures after rolling at 850°C is presented in Fig. 4. The raising of the annealing temperature led to an increase in the size of globular alpha and beta phases. The volume fraction changes in the alpha and beta phases caused the partitioning of the alloying elements during the solution treatment. Beta-stabilizing elements preferentially migrate to the beta phase and alpha-stabilizing ones to the alpha phase. Thus, the beta phase is enriched by alloying elements with increasing the annealing temperature and becomes more stable. A low alpha phase volume fraction and high self-diffusion rate of the beta phase promote the growth of beta phase by increasing the annealing temperature. The presence of alpha phase on beta grain boundaries can hinder the beta grain growth. Therefore, with a decrease in the alpha volume fraction as a result of the increase in annealing temperature from 750 to 850°C, the barriers to beta grain growth were decreased and the beta grain size grew faster.

Beta grain size increased from 5.8 to 311.0 μm by increasing the annealing temperature from 850 to 950°C. As shown in Fig. 3(d) and Fig. 3(e), this notable grain growth of the beta phase is due to the removal of the alpha phase. In a research work done on Ti–3.5Al–5Mo–6V–3Cr–2Sn–0.5Fe [9], it was shown that by rolling in the two-phase region (790°C) and solution-treating in the single beta-phase region (830°C), the beta grain size increased to about 100 μm. Due to the lower content of alloying elements such as Mo, V, and Cr in the present alloy compared to the above mentioned alloy, it is expected that the diffusion rate of alloying elements in the beta phase should be higher than that in the Ti–3.5Al–5Mo–6V–3Cr–2Sn–0.5Fe alloy, which is in agreement with the obtained results.

The optical micrograph of the specimen annealed at 750°C after rolling at 1000°C and air cooling is shown in Fig. 5. A comparison of the microstructure of Fig. 2(b) with Fig. 5 shows that annealing after rolling at 1000°C (Fig. 5) led to an increase in the grain boundary alpha width from 1.2 to 2.0 μm and lamellar alpha width from 0.8 to 1.3 μm.

The purpose of Fig. 5 is to compare the microstructure and mechanical properties of the alloy in the constant heat-treatment cycle with different thermomechanical processing histories by rolling in the two-phase and single beta-phase re-

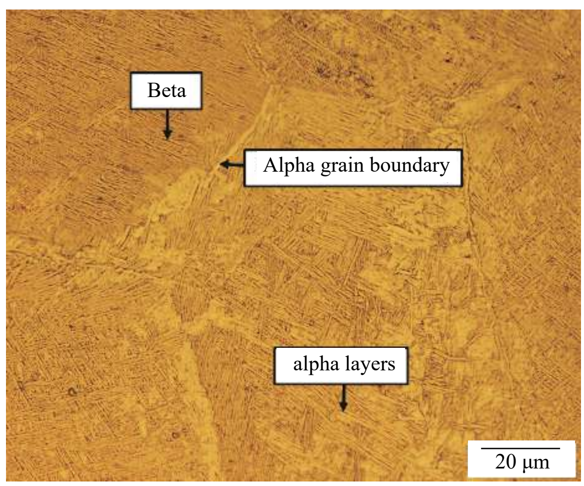

Fig. 5. Optical micrograph of the specimen annealed at 750°C after rolling at 1000°C and air cooling.

gion. An alpha phase with globular morphology and a size of 3.0 μm was formed in the specimen annealed at 750°C after rolling at 850°C (Fig. 3(b)), while alpha layers are observed in the beta grains in the specimen rolled at 1000°C after being annealed in the same condition (Fig. 5). The volume fraction of the alpha phase after annealing and rolling at 1000°C was reduced up to 44% compared with the same annealing condition after rolling at 850°C. The beta grain size in the specimen rolled at 1000°C and annealed at 750°C reached 274 μm, which is significantly larger than that of the specimen rolled at 850°C and subsequently annealed. The beta grain boundaries decreased with an increase in the beta grain size, leading to an increase in the alpha phase nucleation in the beta grains and formation of α/β layers.

The SEM micrographs of the specimens rolled at 850°C and 1000°C and subsequently annealed at 750°C and aged at 570°C for 3 h are shown in Fig. 6. The microstructure of the specimen annealed and aged at 570°C after rolling at 850°C (Fig. 6(a)) consisted of primary alpha in the beta matrix and secondary alpha in the beta grains. In this microstructure, a primary alpha phase with globular morphology and uniform distribution is observed. The secondary alpha width and its volume fraction reached 0.3 μm and 15% in annealing and aging conditions after being rolled in the two-phase region. However, the beta grains grew in annealing and aging conditions after rolling at 1000°C due to the absence of a primary alpha phase. In addition, the secondary alpha was formed with a width of 1.3 μm and volume fraction of 30% in the beta matrix (Fig. 6(b)). The presence of primary alpha makes the retained beta phase more stable after rolling at 850°C followed by aging at 570°C, and consequently, the secondary alpha phase nucleation during aging becomes more difficult. Therefore, annealing plus aging after rolling at 850°C led to the formation of a secondary alpha that is finer in size and lower in volume fraction compared with that formed during annealing and aging after rolling at 1000°C. In a work on Ti-5553 alloy [27], it was reported that the secondary alpha formed during aging after annealing in beta region was larger in size and volume fraction than that of the present study, and this is attributed to the absence of the primary alpha.

3.2. Relationship between microstructure and tensile properties

The tensile properties of the Ti–5Al–4V–2Fe–1Mo alloy rolled at 850 and 1000°C are presented in Fig. 7. The specimen rolled in the beta region shows lower strength and ductility compared with that in the two-phase region. The volume fraction of the alpha phase increased from 21% to 68% by decreasing the rolling temperature to that of the two-phase region. The high volume fraction of the strong alpha phase by rolling at 850°C enhanced the yield strength (YS) and ultimate tensile strength (UTS).

The higher ductility consisting of elongation (El) and reduction of area (RA) parameters of the specimen rolled at 850°C can be attributed to the formation of the globular alpha phase, smaller beta grain size, and the absence of the alpha grain boundary in comparison with the specimen rolled at 1000°C. As known [28] ductility is determined by crack nucleation resistance, depending on the effective slip length. The globularization process decreases the effective slip length and makes the nucleation of crack difficult due to stress concentration at α/β boundaries. The formation of grain boundary alpha creates continuous soft zones, which are preferentially subjected to plastic deformation during loading. Thus, the stress concentration and high localized strain on the grain boundaries led to the separation of the

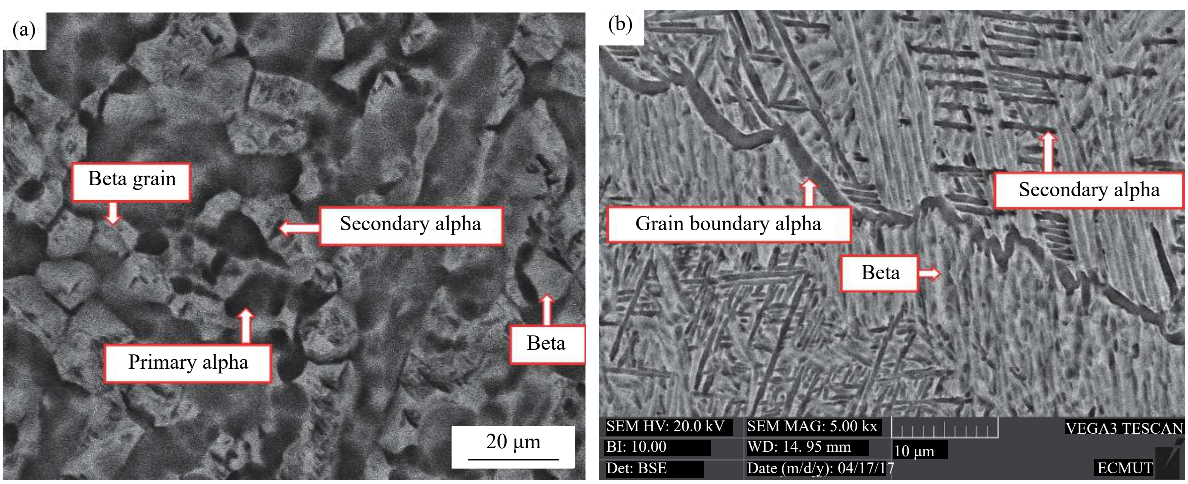

Fig. 6. Microstructure of the specimen rolled at (a) 850°C and (b) 1000°C and subsequently annealed at 750°C and aged at 570°C for 3 h.

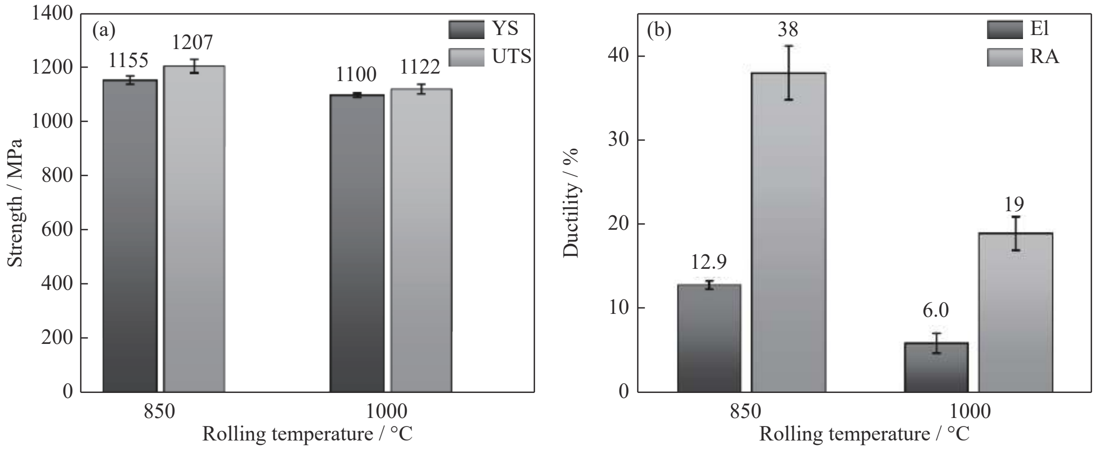

Fig. 7. Variation of the (a) strength and (b) ductility of the alloy after rolling at 850 and 1000°C.

grains, i.e., lower ductility in the specimen rolled at 1000°C. It can be concluded that a suitable balance of tensile properties is obtained through rolling of the Ti–5Al–4V–2Fe–1Mo alloy at 850°C (two-phase region).

Fig. 8 shows the effect of annealing temperature on the tensile properties of the alloy after rolling at 850°C. As seen, the YS initially decreased and then increased with an increase in the annealing temperature from 700 to 850°C. The graph shows the absence of work hardening at an annealing temperature of 700°C, and this caused the amount of YS and UTS become similar. The low strain hardening rate is due to the fact that the low volume fraction of the beta phase at an annealing temperature of 700°C leads to a limited reaction of low density dislocations. By increasing the annealing temperature up to 750°C, the UTS decreased from 1101 to 1062 MPa and then increased. The decrease in the YS can be attributed to 60% reduction in the volume fraction of the strong alpha phase by increasing the annealing temperature (Fig. 3). As a consequence, the ductility increases due to easy slip in beta phase. This causes the yield to decrease as much as the tensile strength. Meanwhile, with the formation of acicular alpha phase at 950°C, the YS increased to 1234 MPa (Fig. 3(e)). By increasing the annealing temperature from 700 to 750°C, the tensile strength decreased due to a reduction in the volume fraction of alpha phase, and the morphology of the alpha phase changed from lamellar to globular. Increasing the annealing temperature up to 850°C caused an decrease in the volume fraction of the globular alpha phase to 43%, and the occurrence of work hardening led to enhanced UTS (dash line in Fig. 8(a)). The maximum yield and tensile strengths achieved at the annealing temperature of 950°C are attributed to precipitation of the acicular alpha phase in beta grains. It has been reported [7,29] that the presence of acicular alpha phase increases the α/β interfaces and creates barriers against the dislocation motion, resulting in an enhanced strength.

According to Fig. 8(b), the ductility initially inreased with an increase in the annealing temperature up to 800°C and then decreased. The globularization of the alpha phase and its small size at 800°C maximize the elongation (14.6%),

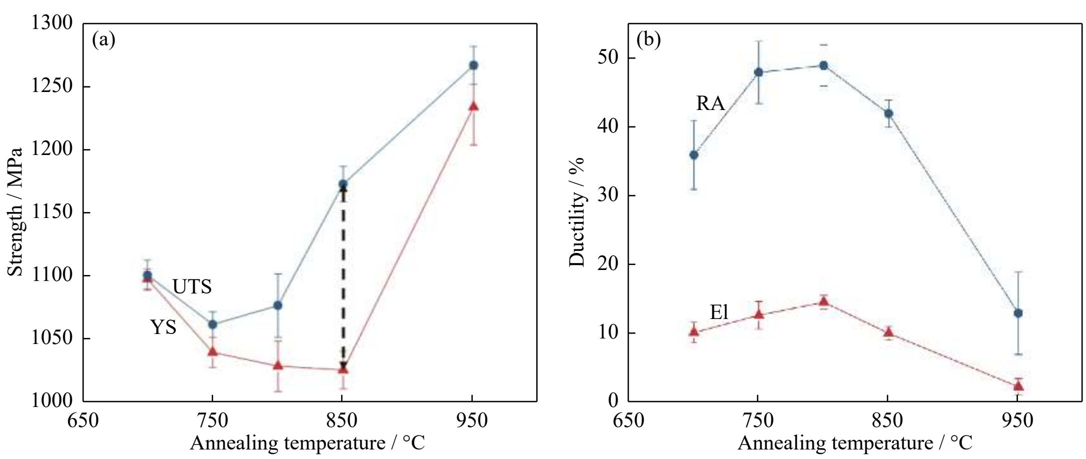

Fig. 8. Effect of annealing temperature on the tensile properties of the alloy rolled at 850°C: (a) strength and (b) ductility of the alloy.

as it reduces the effective slip length in alpha phase. Therefore, crack nucleation would be difficult, and consequently, the elongation is improved. It has been reported [13] that for the metastable beta titanium alloy, Ti–10V–2Fe–3Al, the microstructure with globular alpha phase shows a better ductility compared with that with an elongated alpha one. Decreasing the volume fraction of the alpha phase and changing the morphology from elongated to globular increases the ductility. Hence, the elongated alpha phase in the microstructure of the specimen annealed at 700°C caused a lower ductility compared with the specimen annealed at 750 and 800°C containing the globular alpha phase. In addition, decreasing the volume fraction of the elongated alpha phase from 10% to 5% increased the ductility from 700 to 750°C. When the deformation bands cut across the elongated alpha phase, interpractice spacing decreased and void density increased [13]. The drop in ductility at temperatures above 800°C (Fig. 8(b)) (El of 2.3% and RA of 13% at 950°C) can be attributed to grain growth and the formation of the acicular alpha phase in beta grains. The results of previous studies on Ti–4.5Al–3V–2Mo–2Fe alloy [29] show that ductility decreased to 8% by increasing the volume fraction of acicular alpha phase at higher annealing temperatures in air cooling condition. The low ductility after annealing at 950°C can be as a result of the formation of 27% acicular alpha phase and a significant increase in the beta grain size from 3.0 to 311.0 μm. Since the acicular alpha is a source of low ductility, the presence of acicular alpha in beta grains creates more α/β interface acting as a source of void nucleation. As a result, the stress concentration in acicular alpha phase can lead to the formation of microcracks, ductility decline, and failure of the sample during the tensile test at room temperature. Based on the results, the optimum annealing temperature of the Ti–5Al–4V–2Fe–1Mo alloy after rolling in the two-phase region was obtained to be 750°C.

The results of the tensile test of the specimen annealed at 750°C for 1 h after rolling at 850 and 1000°C are presented in Fig. 9. The specimen annealed after rolling in the beta region showed an elongation of 10.4% and YS of 1021 MPa, which are lower than those observed after rolling in the two-phase region and subsequent annealing. In both conditions, there was no significant change in the tensile strength.

As previously mentioned, the primary alpha phase morphology is an effective parameter in determining the ductility of beta titanium alloys. The higher ductility (El: 12.7%, RA: 48%) was obtained by annealing and rolling in the two-phase region due to the formation of the globular alpha phase. As the crystallographic planes of the alpha phase are not parallel to those of the beta phase, slip transfer between the globular alpha phase and the beta matrix cannot occur easily. Both thick lamellar alpha phases and large beta grains decrease the ductility of the annealed specimen after rolling in beta region. During deformation of the specimen containing the grain boundary alpha, the stress concentration in the interface results in an early slip and decreased yield stress. The cause of the higher tensile strength of the specimen annealed after rolling in the beta region is the presence of a lamellar alpha phase as it increases the interface area. The value of El × UTS can be used as a criterion for alloy toughness. This parameter was calculated to be 13487.4 and 11138.4 MPa·% for annealing after rolling in two- and single-phase regions, respectively. Therefore, a better balance of the strength and ductility is achieved by the thermomechanical processing of the Ti–5Al–4V–2Fe–1Mo alloy by rolling at 850°C and annealing at 750°C.

The results of the tesile test of the alloy rolled at 850 and 1000°C followed by annealing at 750°C and aging at 570°C are presented in Fig. 10. A comparison of the tensile proper-

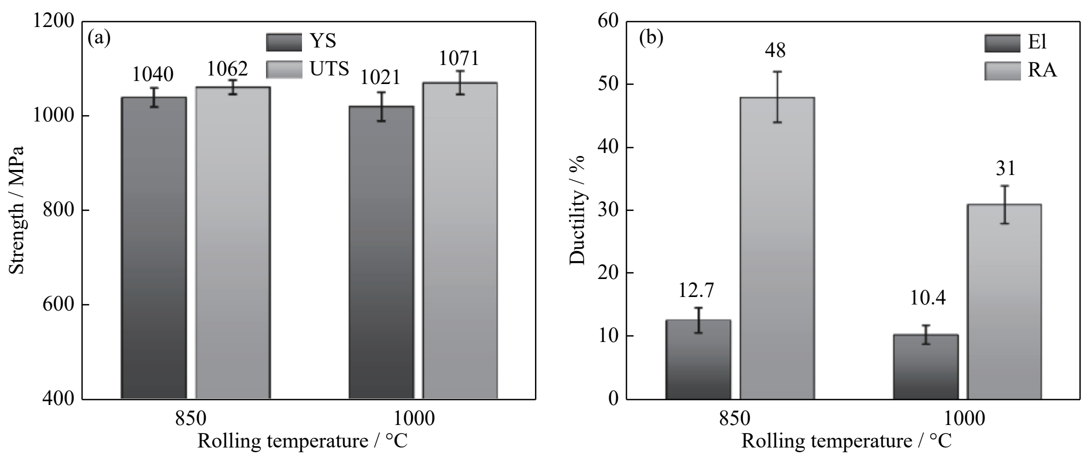

Fig. 9. Variation of (a) strength and (b) ductility of the alloy rolled at 850 and 1000°C and subsequently annealed at 750°C.

ties of the specimen aged at 570°C after annealing and rolling at 850°C and 1000°C indicates that the higher ductility and yield strength were obtained by rolling in the two-phase region. Meanwhile, the tensile strengths in both conditions were approximately similar after aging. The drop in ductility can be attributed to the increase in the beta grain size to 346.0 μm and the formation of the secondary alpha phase during aging, after annealing and rolling in the beta region. The presence of grain boundary alpha and the formation of a secondary alpha phase lead to an early crack nucleation and fast propagation and consequently a lower ductility. The most important and effective factor influencing the mechanical properties of beta titanium alloys is the precipitation of the secondary alpha phase during aging. The distribution of the secondary alpha phase in the beta matrix creates a larger interface area acting as barrier against dislocations movement. The strength of beta titanium alloys is inversely proportional to the distance between secondary alpha phases. As the distance between secondary alpha phases decreases, the strength increases. It has been reported [9] that the high volume fraction of the secondary alpha phase leads to closer distance and higher strength. Thus, the formation of the secondary alpha phase with higher volume fraction (30%) and consequently closer distance of secondary alpha phase by annealing and aging after rolling in the beta region resulted in a higher YS, lower ductility, and tensile strength retention (Fig. 6(b)). Du *et al.* [9] reported that the precipitation of a thick secondary alpha caused a drop in ductilty and an increase in the strength. The values of the El × UTS parameter for annealing and aging after rolling in the two-phase and beta-phase regions were obtained as 14869.4 and 7490.6 MPa·%, respectively, indicating a higher balance of strength and ductility of the alloy in the first condition.

To achieve a suitable balance of strength and ductility and optimize the tensile properties of the Ti–5Al–4V–2Fe–1Mo alloy tested in this work, rolling and solution treatment should be performed in the alpha/beta-phase regions before aging heat treatment.

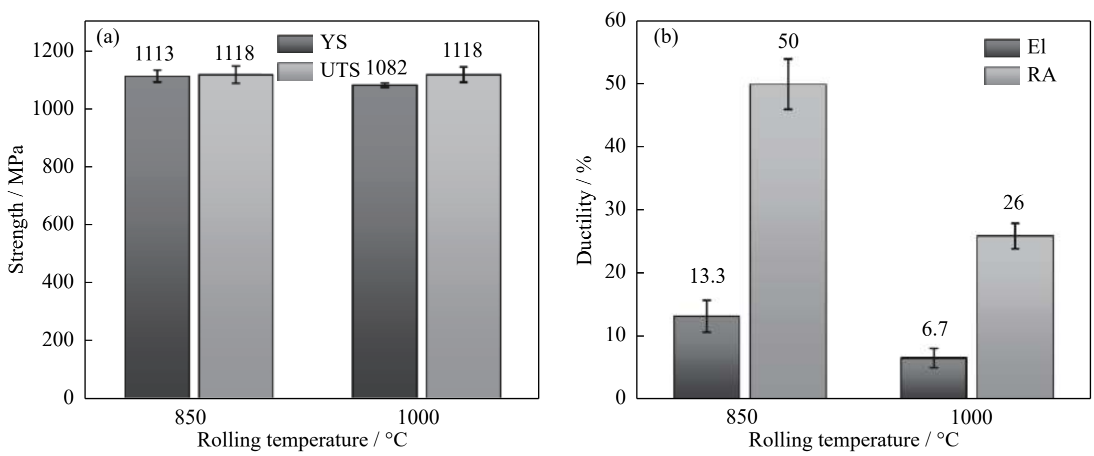

Fig. 10. Variation of (a) strength and (b) ductility of the alloy rolled at 850 and 1000°C followed by annealing at 750°C and aging at 570°C.

4. Conclusions

(1) The alloy rolled at 850°C exhibited a combination of more suitable tensile properties than the alloy rolled at 1000°C due to the globularization and increase in the volume fraction of the alpha phase.

(2) At lower annealing temperatures (700 to 800°C), for the alloy rolled in the two-phase region, the strength decreased and the ductility changed inversely, but the UTS increased with an increase in annealing temperature from 750 to 800°C. However, the formation of acicular alpha at 950°C caused an increase in the strength and a decrease in the ductility.

(3) The specimen annealed at 750°C after rolling in the two-phase region showed a suitable balance of YS (1040 MPa) and elongation (12.7%).

(4) The highest tensile strength and ductility of the studied Ti–5Al–4V–2Fe–1Mo alloy was achieved by rolling at 850°C followed by annealing and aging at 750°C and 570°C, respectively. This is attributed to the small beta grain size and thick secondary alpha phase with a relatively low volume fraction.

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References

[1] W.C. Chung, L.W. Tsay, and C. Chen, Microstructure and notch properties of heat-treated Ti–4.5Al–3V–2Mo–2Fe laser welds, *Mater. Trans.*, 50(2009), No. 3, p. 544.

[2] P. Blenkinsop, Titanium alloys. Advances in alloys, processes, products and applications, *J. Phys. IV*, 3(1993), p. 161.

[3] Y. Oshida, *Bioscience and Bioengineering of Titanium Materials*, Elsevier, Oxford, 2010.

[4] L.M. Lei, X. Huang, M.M. Wang, L.Q. Wang, J.N. Qin, H.P. Li, and SQ.. Lu, Effect of hot compressive deformation on the martensite transformation of Ti–10V–2Fe–3Al titanium alloy, *Mater. Sci. Eng. A*, 530(2011), p. 591.

[5] L.Q. Wang, J. Qu, L.Y. Chen, Q. Meng, L.C. Zhang, J.N. Qin, D. Zhang, and W.J. Lu, Investigation of deformation mechanisms in β-type Ti–35Nb–2Ta–3Zr alloy via FSP leading to surface strengthening, *Metall. Mater. Trans. A*, 46(2015), No. 11, p. 4813.

[6] Z.X. Du, S.L. Xiao, L.J. Xu, J. Tian, F.T. Kong, and Y.Y. Chen, Effect of heat treatment on microstructure and mechanical properties of a new β high strength titanium alloy, *Mater. Des.*, 55(2014), p. 183.

[7] B. Gunawarman, M. Niinomi, T. Akahori, J. Takeda, and H. Toda, Mechanical properties of Ti–4.5Al–3V–2Mo–2Fe and possibility for healthcare applications, *Mater. Sci. Eng. C*, 25(2005), No. 3, p. 296.

[8] Z.F. Shi, H.Z. Guo, J.Y. Han, and Z.K. Yao, Microstructure and mechanical properties of TC_{21} titanium alloy after heat treatment, *Trans. Nonferrous Met. Soc. China*, 23(2013), No. 10, p. 2882.

[9] Z.X. Du, S.L. Xiao, Y.P. Shen, J.S. Liu, J. Liu, L.J. Xu, F.T. Kong, and Y.Y. Chen, Effect of hot rolling and heat treatment on microstructure and tensile properties of high strength beta titanium alloy sheets, *Mater. Sci. Eng. A*, 631(2015), p. 67.

[10] G. Lütjering and J.C. Williams, *Titanium*, Springer, Berlin, Heidelberg, New York, 2003, p. 13.

[11] A.K.M.N. Amin, *Titanium Alloys: Towards Achieving Enhanced Properties for Diversified Applications*, InTech, Croatia, 2012, p. 111.

[12] S. Sadeghpour, S.M. Abbasi, M. Morakabati, and S. Bruschi, Correlation between alpha phase morphology and tensile properties of a new beta titanium alloy, *Mater. Des.*, 121(2017), p. 24.

[13] G.T. Terlinde, T.W. Duerig, and J.C. Williams, Microstructure, tensile deformation, and fracture in aged Ti–10V–2Fe–3Al, *Metall. Trans. A*, 14(1983), No. 10, p. 2101.

[14] K.N. Kumar, P. Muneshwar, S.K. Singh, A.K. Jha, B. Pant, and K.M. George, Effect of grain boundary alpha on mechanical properties of Ti–5.4Al–3Mo–1V alloy, *JOM*, 67(2015), No. 6, p. 1265.

[15] M. Ninomi, D. Eylon, S. Fujishiro, C. Ouchi, and T. Kazino, Improvement in fracture toughness of Ti–4.5Al–3V–2Mo–2Fe through microstructural optimization, *Metall. Mater. Trans. A*, 34(2003), No. 2, p. 267.

[16] J.K. Nieh, C.T. Wu, Y.L. Chen, C.N. Wei, and S.L. Lee, Effect of cooling rate during solution heat treatment on the microstructure and mechanical properties of SP-700 titanium alloys, *J. Mar. Sci. Technol*, 24(2016), No. 2, p. 99.

[17] A.H. Sheikhali, M. Morakkabati, and S.M. Abbasi, Hot torsion behavior of SP-700 near beta titanium alloy in single and dual phase regions, *Int. J. Mater. Res.*, 109(2018), No. 12, p. 1136.

[18] R. Boyer, E.W. Collings, and G. Welsch, *Materials Properties Handbook: Titanium Alloys*, ASM International, Materials Park, 1994, p. 685.

[19] SAE International, ASE Internationla Standard AMS 4964A: *Titanium Alloy, Sheet, Strip, and Plate 4.5Al–3V–2Fe –2Mo Annealed*, Warrendale, 2011.

[20] ASTM Internationalm, ASTM Standard E1181-02: *Standard Test Methods for Characterizing Duplex Grain Size*, West Conshohocken, 2015.

[21] S.L. Semiatin, V. Seetharaman, and I. Weiss, The thermomechanical processing of alpha/beta titanium alloys, *JOM*, 49(1997), No. 6, p. 33.

[22] I. Weiss, F.H. Froes, D. Eylon, and G.E. Welsch, Modification of alpha morphology in Ti–6Al–4V by thermomechanical processing, *Metall. Mater. Trans. A*, 17(1986), No. 11, p. 1935.

[23] C.H. Park, J.W. Won, J.W. Park, S. Semiatin, and C.S. Lee, Mechanisms and kinetics of static spheroidization of hot-worked Ti–6Al–2Sn–4Zr–2Mo–0.1Si with a lamellar microstructure, *Metall. Mater. Trans. A*, 43(2012), No. 3, p. 977.

[24] S.L. Semiatin, N. Stefansson, and R.D. Doherty, Prediction of the kinetics of static globularization of Ti–6Al–4V, *Metall. Mater. Trans. A*, 36(2005), No. 5, p. 1372.

[25] S. Banerjee and P. Mukhopadhyay, Phase Transformations: Examples From Titanium and Zirconium Alloys, Elsevier, Oxford, 2010, p. 148.

[26] A. Gunawarman, M. Niinomi, K.I.. Fukunaga, D. Eylon, S. Fujishiro, and C. Ouchi, Fracture characteristics and microstructural factors in single and duplex annealed Ti–4.5Al–3V–2Mo–2Fe, *Mater. Sci. Eng. A*, 308(2001), No. 1-2, p. 216.

[27] S. Shekhar, R. Sarkar, S.K. Kar, and A. Bhattacharjee, Effect of solution treatment and aging on microstructure and tensile properties of high strength β titanium alloy Ti–5Al–5V–5Mo–3Cr, *Mater. Des*, 66(2015), p. 596.

[28] J.W. Xu, W.D. Zeng, Y.W. Zhao, Z.Q. Jia, and X. Sun, Effect of globularization behavior of the lamellar alpha on tensile properties of Ti-17 alloy, *J. Alloys Compd*, 673(2016), p. 86.

[29] M. Niinomi, K. i. Fukunaga, D. Eylon, S. Fujishiro, and C. Ouchi, Effect of cooling rate on microstructure and fracture characteristics of β-rich α+β type Ti–4.5Al–3V–2Mo–2Fe alloy, *Mater. Trans.*, 42(2001), No. 7, p. 1339.