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Effects of $\beta$ grain-growth behaviors on lamellar structural evolution and mechanical properties of TC4–DT alloy

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Abstract

Grain-growth behaviors of TC4–DT alloy in a narrow temperature range (990 °C – 1050 °C) were systematically investigated, and the effects of which on the lamellar structural evolution and mechanical properties were quantitatively evaluated. Microstructural observations indicated that prior $\beta$ grain size increased with an increase in heat-treatment temperature and time, which was described by the modified Sellars model. The grain-growth exponent ($n = 2.741$) and activation energy ($Q = 161.0$ kJ mol$^{-1}$) during $\beta$ treatment were confirmed. The $\alpha$ colony size similar to $\beta$ grain varied significantly with the heat-treatment conditions, while $\alpha$ plate thickness changed slightly. The Hall–Petch equation could qualitatively exhibit the relationships between the lamellar microstructure parameters (prior $\beta$ grain size, $\alpha$ colony size, and $\alpha$ plate thickness) and mechanical properties (strength, ductility, and impact toughness). The fine prior $\beta$ grain that contained different orientated $\alpha$ colonies produced more boundaries to hinder dislocation motion and crack propagation, which contributed a more circuitous crack growth path. The results indicated that the control of $\alpha$ colony size was critical to improve the mechanical performance of TC4–DT alloy.

1. Introduction

Titanium and its alloys are used extensively in the fields of aerospace, energy, and medical industries because of their excellent properties, such as a low density, high specific strength, good corrosion resistance, and good biocompatibility [1]. Nowadays, the damage-tolerant design principle has become dominant in structural component manufacture for aerospace applications [2], which leads to a stricter requirement for mechanical properties of titanium alloys. The optimization of process parameters that dominate final microstructures is crucial to improve the mechanical properties. It is reported that a thermomechanical process and its accompanying heat-treatments in $\beta$ phase region (above the $\beta$ transus temperature) can yield a lamellar microstructure, which has a superior crack-propagation resistance compared to those with equiaxed or bimodal structure [3, 4]. However, grain tends to grow rapidly in $\beta$ region because of the relatively open body-centered-cubic (bcc) structure, which makes the control of microstructures difficult in subsequent steps due to the structural heredity effect of titanium alloys [5, 6]. Therefore, it is essential to investigate the grain-growth evolution to improve the microstructure and properties during $\beta$ processing.

Microstructure parameters of the lamellar structure, including prior $\beta$ grain size ($D_\beta$), $\alpha$ colony size ($D_\alpha$), and $\alpha$ plate thickness ($D_p$), are influenced by pre-processing in $\beta$ region [7]. A complicated combination of lamellar morphology and its size is related closed to the final properties of the titanium alloys. Previous researches have been carried out to clarify the effect of lamellar microstructure on the titanium alloy deformation behavior and mechanical properties [8, 9]. Li et al [10] found that a near-$\alpha$ high-temperature alloys with thin initial $\alpha$ lamellar...
embodied a higher flow stress than the one with the thick lamellar during \(\alpha + \beta\) region processing. Wang et al [11] compared the fatigue crack growth behaviors of fully-lamellar and bi-lamellar Ti–6Al–3Nb–2Zr–1Mo alloys and suggested that fully-lamellar occurred more deflection on fatigue cracks. Most previous researchers have considered only one parameter, namely, that of the lamellar structure that affects the properties, and limited literature exists on the interrelationship of all parameters with mechanical properties.

TC4 (i.e. Ti–6Al–4V) alloy, as a low-cost \(\alpha-\beta\) titanium alloy, has been applied extensively to advanced structural materials because of its attractive comprehensive performance [12]. TC4–DT alloy is a new damage-tolerance titanium alloy developed based on conventional TC4 alloy with the permitted maximum content of interstitial elements (C, O and N) in a controlled level to attain a good combination of strength, ductility, and toughness [13–15]. In recent years, \(\beta\) isothermal forging and near-\(\beta\) forging processing have received considerable interest to assure the deformation temperature and improve the damage-tolerance properties in \(\alpha-\beta\) titanium alloys [16, 17], which require hot processing temperature to be confined in a narrow range above the \(\beta\) transus temperature to accurately control microstructure coarsening. However, most of previous studies only investigated the \(\beta\) grain-growth behaviors of titanium alloys in a wide temperature range (much above the \(\beta\) transus temperature) [18–20]. In addition, these studies could not reveal the effects of \(\beta\) grain-growth behaviors on lamellar structural evolution and the relationship between different level lamellar microstructures and mechanical properties. The objective of this work was to evaluate the grain-growth behaviors and establish a kinetics model of TC4–DT alloy in \(\beta\) region with a narrow temperature range (990 °C–1050 °C). The effects of \(\beta\) grain-growth on the lamellar microstructure evolution and mechanical properties were quantitatively investigated, which can deepen the comprehensive understanding of cracking mechanism and provide a basic guidance to evaluate the quantitative relationship between microstructural control and performance improvement in manufacture.

2. Experimental procedures

The material that was used in this study was commercial TC4–DT as-forged alloy with a chemical composition (in wt.%). 5.90 Al, 3.98 V, 0.07 Fe, 0.051 O, 0.002 N, and balance Ti. The \(\beta\) transus temperature was 980 °C as determined by metallographic techniques. Figure 1 shows that the initial microstructure of equiaxed primary \(\alpha\) within a transformed \(\beta\) matrix had a typical two-phase structure characteristic. The average grain size of primary phase was measured to be approximately 15 μm.

Figure 2 shows the heat-treatment procedure and mechanical specimen geometry. Isothermal treatment tests were carried out on a thermal dilatometer (DIL 805), which provided an accurate control of temperature, time, and cooling rate. The 6-mm-diameter and 12-mm-high cylindrical specimens were machined by electro-discharge machining. The chamber was evacuated and purged with argon. To investigate \(\beta\) grain-growth behaviors during isothermal treatment, the specimens were subjected to different \(\beta\) treatments at 990 °C–1050 °C for 0–7200 s, and then quenched to room temperature, as shown in figure 2(a). When the titanium alloys were cooled at different rates (above 0.1 °C s\(^{-1}\)) by continuous cooling from \(\beta\) region, the sizes of prior \(\beta\) grain exhibited no apparent changes [21]. Thus, to study the effects of \(\beta\) treatment on the lamellar structure evolution during continuous cooling, several specimens were solution-treated at 990/1030 °C for 20–1200 s, followed by cooling at 1 °C s\(^{-1}\) to room temperature to obtain a full-lamellar structure, as shown in figure 2(b). This cooling rate is typically applied and consulted in the industrial processes for titanium alloys production. To investigate the relationship between lamellar structural
Evolution and mechanical properties, billet treatment tests (990/1030 °C + 20/600 s) were executed on a Gleeble-3500D simulator under a protective argon atmosphere. The heating and cooling rates were the same as the values specified in figure 2(b) to obtain the same lamellar microstructure. Then tensile and impact toughness specimens were machined from billets. The geometries relationship between billets and specimens are shown in figure 2(c). The tensile properties and Charpy impact energy of the specimens were evaluated using an electronic universal testing machine (CMT 5105) and an impact tester (JBDW 300D) at room temperature, respectively. Three specimens were tested for each heat-treatment condition to ensure accuracy.

All heat-treated specimens were prepared using a standard metallurgical procedure, and the polished surface was etched using Kroll’s reagent (2 ml HF/4 ml HNO3/80 ml H2O). The microstructure observation, electron backscattered diffraction (EBSD) analysis, and fracture surfaces were investigated with optical microscopy (OM, Zeiss Axiovert 40) and scanning electron microscopy (SEM, FEG 450) equipped with an EBSD system, respectively. The quantitative values of average β grain size ($D_{\beta}$), width of the α colony ($D_{c}$), and α plate thickness ($D_{p}$) were measured by using Image Pro-Plus® (IPP) analysis software. The EBSD specimens were polished mechanically and electropolished further with a solution of 10% perchloric acid in methanol at 25 °C for 35 s at 20 V.

3. Results and discussion

3.1. β grain-growth during isothermal treatment

3.1.1. Grain size evolution

Figure 3 presents the β grain morphologies with martensitic α′ microstructure of TC4–DT alloy after heat treatments at 990 and 1030 °C for various holding times. The typical acicular α′ within the prior β grain was obtained by quenching from β region and observed in all the micrographs. Prior β grains were less distinct because of no precipitation of grain boundary α (αGB) under the rapid cooling. The microstructural characteristics show a mixture structure of inhomogeneous equiaxed grains under all heat-treatment conditions. During the high-temperature heat-treatment of titanium alloys, grain-growth of the β phase is believed to include recrystallization-nucleation and coarsening stages [22]. This phenomenon of mixed grains is attributed to the presence of many
vacancies, dislocations, and other defects in the as-forged microstructure of TC4–DT alloy, resulting in different driving forces of grain nucleation and coarsening.

Figure 4 shows the variation of the average $\beta$ grain size with temperature and time. Combined with the above results, although the existing grains were inhomogeneous, the average grain size increased with an increase of
heat-treatment temperature and time. For a given temperature (figure 4(a)), the \(\beta\) grain grew rapidly for a treatment time of less than 1200 s, then slowed with time. As shown in figure 4(b), the growth rate gradually became rapid and the grain size increased obviously with an increase in temperature. Grain-growth is a diffusion-controlled thermal activation process that driven by grain-boundary energy \([23]\). With an extended treatment time, the increase of grain size reduces the total grain boundary area, and the migration velocity of the grain boundary decreases.

3.1.2. Grain-growth kinetics

Two traditional empirical models, namely, the Sellars model and the Anelli model are used extensively to evaluate the grain-growth behavior in metals \([24, 25]\), and is described by equations (1) and (2):

\[
D^n - D_0^n = A t \exp\left(\frac{-Q}{RT}\right) \quad (1)
\]

\[
D = Bt^m \exp\left(\frac{-Q}{RT}\right) \quad (2)
\]

where \(D\) is the final average \(\beta\) grain size (\(\mu\)m), \(D_0\) is the initial \(\beta\) grain size (\(\mu\)m), \(t\) is the holding time (s), \(T\) is the absolute temperature (K), \(n\) is the grain-growth exponent, \(m\) is the time index, \(Q\) is the activation energy of grain growth (kJ mol\(^{-1}\)), \(R\) is the gas constant (8.3145 kJ/(mol \cdot K)), and \(A\) and \(B\) are constants that depend on the experimental parameters and materials.

The initial grain size \(D_0\) is much smaller compared with \(D\) after the holding time, which is assumed to be zero and negligible, especially at high temperatures \([26]\). Thus, the Sellars model can be simplified as the following relation:

\[
D^n = A_0 t \quad (3)
\]

where \(A_0\) is a constant, which is expressed as an Arrhenius-type equation of \(A_0 = A\exp(-Q/RT)\).

The values of \(n = 2.516\) and \(A_0\) were derived from the slope and the intercept of \(\ln D - \ln t\) plots by linear fitting (figure 5(a)). The activation energy \(Q = 102.5\) kJ mol\(^{-1}\) and \(A = 10.55 \times 10^7\) were determined from \(\ln A_0 - 1/T\) (figure 5(b)). As a result, an expression for the simplified grain-growth model of Ti6Al4V alloy was written as:

\[
D^{2.516} = 10.55 \times 10^7 t \exp\left(\frac{-102.5 \times 10^3}{RT}\right) \quad (4)
\]

Although this approach of ignoring \(D_0\) simplified the process of experiment and operation, it undoubtedly reduced the credibility of the prediction model. Based on equations (1) and (2), a modified Sellars model was proposed to depict the \(\beta\) grain-growth kinetics, which considered the grain-growth exponent \(n\) and the time index \(m\) \([27]\), as follows:

\[
D^n - D_0^n = C t^m \exp\left(\frac{-Q}{RT}\right) \quad (5)
\]

To calculate the unknown parameters \((n, m, Q,\) and \(C)\), the natural logarithm of both sides of equation (5) were taken.
When the heat-treatment temperature and time are constants, the average values of $m$ and $Q$ was calculated at a given value of $n$ by linear regressions expressed as equations (7) and (8), respectively. Then the corresponding value of $C$ was calculated using the method of averaging the results according to equation (6).

\[
\ln(D^n - D_0^n) = \ln C + m \ln t - \frac{Q}{RT}
\]  

(6)

When the heat-treatment temperature and time are constants, the average values of $m$ and $Q$ was calculated at a given value of $n$ by linear regressions expressed as equations (7) and (8), respectively. Then the corresponding value of $C$ was calculated using the method of averaging the results according to equation (6).

\[
m = \left. \frac{\partial[\ln (D^n - D_0^n)]}{\partial (\ln t)} \right|_T
\]  

(7)

\[
Q = -R \left. \frac{\partial[\ln (D^n - D_0^n)]}{\partial (1/T)} \right|_T
\]  

(8)

The values of the material parameters ($m$, $Q$, and $C$) were determined for different values of $n$ from 1 to 5 with an interval of 0.5 according to the above method, and the corresponding modified model was defined to obtain the predicted value of the prior $\beta$ grain size. The square sum of errors $\text{SSE}(n)$ between the experimental and the predicted values was introduced to obtain the optimum value of $n$. Figure 6 presents plots of $\text{SSE}(n)$−$n$ and the corresponding polynomial fitting curve. A seventh-order polynomial $f_{\text{SSE}(n)}$ was fitted the relationship between $\text{SSE}(n)$ and $n$ well, and $f_{\text{SSE}(n)}$ reached a minimum when $n$ was 2.741.

Figure 7 shows the linear relationships of $\ln(D^{2.741} - D_0^{2.741})$–$\ln t$ and $\ln(D^{2.741} - D_0^{2.741})$–$1/T$. When we relied on the experimental data and used the same solving process, the values of $m$, $Q$, and $C$ were obtained as 1.098, 161.0 kJ mol$^{-1}$, and $50.87 \times 10^9$, respectively. Then the modified model of $\beta$ grain-growth kinetics was described by:
and are the mean values of \( E \) and \( E_i \). The average absolute relative error (AARE) where \( E_i \) is the experimental value of the grain size, \( \bar{E} \) is the predicted value, and \( N \) is the number of data points. \( \bar{E} \) and \( \bar{P} \) are the mean values of \( E_i \) and \( P_i \), respectively.

3.2. Lamellar structural evolution during continuous cooling

3.2.1. Effect of \( \beta \) grain size on lamellar microstructure

Figure 8 shows the microstructures and EBSD orientation maps of TC4–DT alloy that were obtained at a continuous cooling rate (1 °C s \(^{-1}\)). As slowly cooling from the \( \beta \) region, a lamellar widmanstätten microstructure was observed under all heat-treated conditions (figures 8(a)–(c)). The prior \( \beta \) grains were easily identified by continuous \( \alpha_W \) plates formed at \( \beta/\beta \) grain boundaries, and \( \alpha \) colonies within \( \beta \) grains were composed of \( \alpha \) plates (\( \alpha_W \)) and retained \( \beta \) phase. The \( \alpha \) colony size significantly increased as the heat-treatment temperature and time, and the \( \alpha_W \) plate thickness showed slight change. It is because the \( \alpha \) phase will be easier to nucleate and grow at boundaries, and larger lamellar structure parameters can be produced at a higher treatment temperature and for a longer holding time. According to the orientation maps, the \( \alpha_W \) plates in a single \( \alpha \) colony had the same orientation, and the neighboring \( \alpha \) colonies showed a different orientation relationship within a prior \( \beta \) grain. The number of different oriented \( \alpha \) colonies inside the fine \( \beta \) grain (figures 8(d) and (e)) exceeded those inside the coarse \( \beta \) grain (figure 8(f)), which indicates the importance of the prior \( \beta \) grain boundaries in...
providing nucleation sites and α phase precipitation. Figure 10 presents a qualitative analysis for lamellar structure parameters ($D_c$, $D_c'$, and $D_p$) of TC4–DT alloy. Under the current continuous cooling condition, prior β grain size agreed approximately with that of quenched samples at the same treatment temperature and time. In addition, the relationships of $D_c$–$D$ and $D_p$–$D$ all conformed to the linear correlation, while a strong linear relationship existed between $D$ and $D_p$. The reason for this is that the αW plate morphology mainly depends on the chemical composition and cooling rate during continuous cooling from the β region, and the α plate thickness is inversely related to the cooling rate [31, 32]. Therefore, the αW plate thickness is not only affected by the cooling rate, but also controlled by prior β grain size during continuous cooling. A previous study revealed a similar microstructure feature in a two-phase titanium alloy [33].

3.2.2. Relationship between lamellar structure and mechanical properties

Figure 11 shows the relationship between tensile strength (yield strength/YS and ultimate tensile strength/UTS) and lamellar structure parameters, which is indicated that strength of lamellar structure conformed to the Hall–Petch equation [34–36]. This is attributed mainly to a rapid coarsening of the Widmanstätten structure, which increases the slip distance and reduces the stress required for dislocation glide during plastic deformation. Figure 12 shows the variations of tensile elongation (EL) and impact energy ($A_k$) with lamellar structure parameters. Moreover, EL and $A_k$ also increased with decreasing lamellar structure parameters and obeyed the
Figure 11. Variation of yield strength (YS) and tensile strength (UTS) with reciprocal square root of (a) prior β grain size ($D_p$); (b) α colony size ($D_c$); (c) αW plate thickness ($D_p$).

Figure 12. Variations of tensile elongation (EL) and impact energy ($A_k$) with reciprocal square root of (a) prior β grain size ($D_p$); (b) α colony size ($D_c$); (c) αW plate thickness ($D_p$).
Hall–Petch relationship. During deformation process, pile-up of dislocation at the boundaries promotes the formation of micro-cracks and micro-voids, which propagate along these boundaries. The grain and lamellar structure refinement of titanium alloys increases the areas of grain boundary, colony boundary and $\alpha/\beta$ phase boundary [17]. The ability of fine lamellar microstructures to spread the deformation can cause less stress concentration near the boundaries, such that a cavitated boundary is hard to be debonded [37, 38], which improves effectively elongation. Furthermore, it is concluded that $D_c$ had the best fitting correlation coefficient with YS, UTS, EL, and $A_k$ as 0.9947, 0.9898, 0.9983, and 0.9869, respectively (figures 11(b) and 12(b)). Thus, among lamellar structure parameters, $\alpha$ colony size is the key factor in adjusting the strength, ductility and impact toughness.

The impact energy is related closely to the fracture mode, which is determined by the microstructure. Figure 13 shows the impact fracture surfaces of TC4–DT alloy for different heat-treatments. With an increase in heat-treatment temperature and time, the fracture characteristics showed a significant difference. From the macro-fractographies, the fracture surface of the sample at 1030 °C/600 s was tearing mode. The micro-fractographies of samples (figures 13(d) and (e)) were characterized by dimples and small cleavage facets, but the dimples in figure 13(d) were deeper and denser. Figure 13(f) shows that the fracture characteristics were composed mainly of shallow dimples, river patterns and large flat cleavage facets, resulting in a reduction of toughness. The cleavage facet size increased with the increasing of temperature and time, which was close to the $\alpha$ colony size and much less than the prior $\beta$ grain size. It can be inferred that the $\alpha$ colony has an obvious effect on the fracture mode of TC4–DT alloy with a full-lamellar microstructure.

Figure 14 shows the crack propagation process in various sizes of the prior $\beta$ grain. Fine prior $\beta$ grains (figure 14(a)) containing small $\alpha$ colonies of different orientations had much colony boundaries and $\alpha/\beta$ phase...
boundaries, by which the crack propagation could absorb more energy. Cracks tend to grow along boundaries and interfaces [39, 40]. Hence, the crack deflection is more significant in the refined microstructures and contributes to a tortuous crack path. This result is consistent with the fracture surface in figure 13(d). In contrast, coarse prior β grains (figure 14(b)) had a lower volume fraction of boundaries in front of the crack tip, resulting in a relatively smooth crack path, which exhibited lower impact toughness.

4. Conclusions

(1) The β grain-growth behaviors of TC4–DT alloy during isothermal treatment was sensitive to the heat-treatment temperature and time. The modified Sellsars model for grain-growth kinetics, simultaneously incorporating the grain-growth exponent n and the time index m, was determined as $D_\beta = 50.87 \times 10^{1.998 \times (-161.0 \times 10^{-\beta D})}$ and provided a more reliable prediction.

(2) The main characteristic of the lamellar microstructure during continuous cooling is that the variation tendency of the α colony size and the α plate thickness was positively correlated with the prior β grain size, which followed a strong linear dependence ($D_\alpha = 4.919 + 0.1302D$) and a weak linear dependence ($D_p = 0.8767 + 1.701 \times 10^{-3}D_p$), respectively.

(3) Lamellar structure parameters significantly affected strength, ductility and impact toughness of TC4–DT alloy, which obeyed the Hall–Petch relationship. The cleavage facet size of fractured surface had a strong correspondence with α colony size. The crack propagation path of the impact toughness was ascribed to the effects of boundaries and crack deflection, which were determined mainly by the α colony morphology.

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References

[1] Gao X, Zeng W, Wang Y, Long Y, Zhang S and Wang Q 2017 J. Alloy. Compd 725 536–43
[2] Wen X, Wen M, Huang C, Tan Y, Lei M, Liang Y and Cai X 2019 Mater. Des. 180 107898
[3] Guo P, Zhao Y and Zeng W 2015 Rare Metal Mat. Eng. 4 277–81
[4] Xiao Y, Liu H, Yi D, Le I, Zhou H, Jiang Y, Chen Z, Wang J and Gao Q 2018 J. Mater. Eng. Perform. 27 4941–54
[5] Balasundar I, Ravi R K and Raghu T 2017 Mater. Sci. Eng. A 684 135–45
[6] Wang W L, Wang X L, Mei W and Sun J 2016 Mater. Charact. 120 263–7
[7] Wen X, Wen M, Huang C and Lei M 2019 Mater. Sci. Eng. A 740 741 121–9
[8] Liang Y and Wang H 2015 Mater. Sci. Eng. A 622 16–20
[9] Xu J, Zeng W, Zhao Y and Jia Z 2016 Mater. Sci. Eng. A 676 434–40
[10] Li H, Zhao Z, Guo H, Yao Z, Yin Y, Miao X and Ge M 2017 Mater. Sci. Eng. A 682 345–53
[11] Wang Q, Ren J Q, Wu Y K, Jiang P, Li J Q, Sun Z J and Liu X T 2019 J. Alloy. Compd 789 249–55
[12] Seshacharyulu T, Medeiros S C, Frazier W G and Prasad Y 2002 Mater. Sci. Eng. A 325 112–25
[13] Liu J, Zeng W, Shu Y, Xie Y and Yang J 2016 Rare Metal Mat. Eng. 45 1647–53
[14] Fu P, Mao Z, Zuo C, Wang Y and Wang C 2014 Chinese J. Aeronaut 27 1015–21
[15] He S, Zeng W, Xu J and Chen W 2019 Mater. Sci. Eng. A 745 203–11
[16] Shi Z, Guo H, Liu R, Wang X and Yao Z 2015 T. Nonferr. Metal. Soc. 25 72–9
[17] Yang X, Guo H, Zhao Z, Yin Y, Yuan S and Xin S 2017 Procedia Eng. 207 2167–72
[18] Gil F J and Planell J A 2000 J. Mater. Sci. Lett. 19 2023–4
[19] Javashin O M, Shevechenko S V and Semiatin S I 2002 Mater. Sci. Eng. A 332 343–50
[20] Gil F J and Planell J A 2000 Mater. Sci. Eng. A 283 17–24
[21] Feng S, Jinshan L, Hongchao K, Wenzhong L, Xianghong L and Yong F 2015 Rare Metal Mat. Eng. 44 848–53
[22] Wang T, Guo H, Tan L, Yao Z, Zhao Y and Liu P 2011 Mater. Sci. Eng. A 528 6375–80
[23] Liang X, Yin L X, Zheng L Y, Ma M Z and Liu R P 2016 Mater. Des. 99 396–402
[24] Anelli E 1992 IISI Int. 32 440–9
[25] Sellsars C M and Whiteman J A 1979 Met. Sci. 13 187–94
[26] Liu W H, Wu Y, He Y J, Nieh T G and Lu Z P 2013 Scripta Mater. 68 526–9
[27] Chen R C, Hong C, Li J J, Zheng Z Z and Li P C 2017 Procedia Eng. 207 663–8
[28] Chevrikiu B, Srinivasan R, Tamirisakandala S and Miracle D B 2009 Scripta Mater. 60 496–9
[29] Fei Y, Wang X N, Zhu Z S, Li J, Shang G Q and Zhu H L 2013 Mater. Sci. Forum 747–748 844–9
[30] Lu J, Zhao Y, Ge P and Niu H 2013 Mater. Charact. 84 105–11
[31] Ahmed T and Rack H J 1998 Mater. Sci. Eng. A 243 206–11
[32] Sen I, Tamirisakandala S, Miracle D and Ramamurty U 2007 Acta Mater. 55 4983–93
[33] Xu J, Zeng W, Zhao Y, Sun X and Du Z 2016 J. Alloy. Compd 688 301–9
[34] Hall E O 1951 Phys. Soc. Sect. B 64 747–53
[35] Petch N J 1953 J. Iron Steel Inst. 174 25–8
[36] Hansen N 2004 Scripta Mater. 51 601–6
[37] Huang H E and Koo C H 2004 Mater. Trans. 45 562–8
[38] Dang N, Liu L, Adrien J, Cazottes S, Xiao W, Ma C, Zhou L and Maire E 2019 Mater. Sci. Eng. A 747 154–60
[39] Tian X J, Zhang S Q, Li A and Wang H M 2010 Mater. Sci. Eng. A 527 1821–7
[40] Shi X, Zeng W and Zhao Q 2015 Mater. Sci. Eng. A 645 82–7