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Damage tolerance performance of high strength and toughness titanium alloys formed by additive manufacturing in aerospace: A review

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ABSTRACT

In aerospace, high strength and toughness titanium alloys (HSTTAs) formed by additive manufacturing (AM) are mostly utilized to create complex shaped structural components. It can fulfill the bespoke design specifications of components, enhance material consumption and production efficiency, and decrease costs and time. As the requirement for safety and stability in structural components rises, damage tolerance performance (DTP) has emerged as the design benchmark for titanium alloys in aviation. This review initially presents the historical evolution of HSTTAs and subsequently discusses related research on the HSTTAs formed by AM. This review covers recent research advancements on the DTP of HSTTAs formed by AM, detailing the deformation behavior, fracture toughness, and fatigue crack propagation characteristics of the alloy. The primary approaches for enhancing the DTP of HSTTAs formed by AM, including process parameter optimization and heat treatment, are examined. Finally, the existing problems on current research and prospective research directions are identified. Investigating the DTP of HSTTAs formed by AM can enhance the overall performance of materials and guarantee structural integrity, while also fostering innovation in AM and propelling technological advancement. And it provides certain reference significance for upgrading AM processes and post-treatment processes, optimizing properties, and developing new high damage tolerant titanium alloys.

1. Introduction

Titanium alloy is the most extensively utilized high-performance structural alloy in the aerospace sector, owing to its superior corrosion resistance, specific strength, and flexibility [1–5]. The operational conditions for aircraft titanium alloys are exceedingly severe, frequently resulting in failures such as wear and fracture [6]. The enhancement of safety standards and service longevity for titanium alloys in aircraft has necessitated elevated safety and performance requirements for these materials. Therefore, the design standards for titanium alloy materials characterized by high strength, high toughness, extended service life, high efficiency, and elevated damage tolerance are presently prominent subjects of discussion. Utilizing AM not only fulfills the performance criteria for alloys in the aviation sector but also expedites the fabrication of intricate components, hence reducing costs and time [7]. Analyzing its DTP can evaluate its service life in the presence of initial cracks or defects, thereby ensuring safety during using [8–12].

In recent years, the rapid advancement of AM has led to a corresponding increase in research on HSTTAs formed by AM. Researchers have transitioned from exclusively focusing on the benefits of high strength or rapid prototyping of materials to attaining optimal service life and durability through processing. However, due to the enormous workload of research on DTP, current research on DTP is often not comprehensive or systematic enough. This review initially presents the development history of HSTTAs, and then introduced the research on HSTTAs formed by AM. Based on the present research status, the prevailing approaches for regulating and improving DTP were summarized. Finally, individual perspectives were presented regarding the primary problems facing contemporary research and prospective avenues for progress. This review aims to provided inspiration for the research on DTP of HSTTAs, which is significant for improving material structural safety, optimizing design and manufacturing processes, and fostering technical innovation and application extension.

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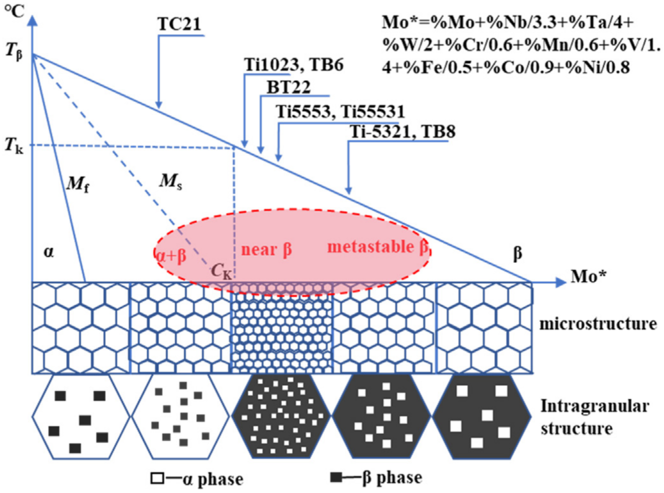

Fig. 1. Distribution and structure of HSTTAs.

2. Development history of HSTTAs

Titanium alloys are extensively utilized in medical equipment, aircraft, petrochemicals and other fields due to their exceptional corrosion resistance, high strength, and significant ductility [1,13]. Titanium alloys in the aerospace sector must perform under extreme conditions while adhering to stringent criteria, including lightweight properties, high strength and toughness, safety and efficiency, and prolonged service life. According to the requirements of different aviation components, titanium alloys have gradually developed into HSTTAs, high temperature titanium alloys [14,15], low temperature titanium alloys [16,17]. HSTTAs were the most prevalent and sought-after alloy among them. The earliest performance standards for HSTTAs were: ultimate tensile strength (UTS) exceed 1000 MPa and fracture toughness (K_{Ic}) over 55 MPa $m^{1/2}$ [18,19]. With the development of a large number of new HSTTAs, the UTS of HSTTAs was generally in the range of 1100–1300 MPa, the elongation (El) was greater than 5 %, and the K_{Ic} was in the range of 55–90 MPa $m^{1/2}$. HSTTAs primarily consist of near α alloy [20,21], near β alloy [22], metastable β alloy [23,24], and α+β alloy [25,26]. The concentration of β-stabilizing components in the majority of alloys range from C_K to C_β, with the Mo equivalent generally exceeding 10. When the Mo equivalent approach C_K, its microstructure become finer, displaying an intragranular microstructure characterized by numerous small α phases interspersed inside the β phase (or small β phases dispersed within the α phase), as shown in Fig. 1. This will endow HSTTAs with complex structures and well mechanical properties, hence enabling regulated characteristics.

The advancement and utilization of HSTTAs have consistently been a hot topic in aerospace materials, with numerous countries vying to develop diverse HSTTAs, as shown in Table 1. As early as 1971, Timet Corporation in the United States developed Ti-1023 alloy [27–29]. This was also one of the often utilized near β titanium alloys. After heat treatment, the alloy exhibited a tensile strength (R_m) ranging from 965 to 1310 MPa, with the K_{Ic} was 99–33 MPa $m^{1/2}$, demonstrating a favorable correlation between strength and toughness. In that year, RIM Corporation also developed the α+β titanium alloy-Ti62222S [30,31]. The design strength of the alloy was greater than 1034 MPa, K_{Ic} was over 77 MPa $m^{1/2}$, and it was utilized in the lower keel chord of F-22 fighter. RIM has also developed a metastable β titanium alloy known as β-C alloy [32]. Studies indicate that following solid solution treatment, the UTS of β-C alloy can reach 1400 MPa, while the K_{Ic} remains at 58 MPa $m^{1/2}$ [33]. In the 1970s, the Soviet Union developed highly alloyed and high-strength near β titanium alloys, specifically BT22 [34,35]. After annealing, the content of α and β phases were each 50 %, resulting in the maximum strength seen among current alloys. Post-annealing, the UTS of large-sized forgings can reach 1300 MPa, and the K_{Ic} remains at 65 MPa $m^{1/2}$. The BT22 alloy possesses great strength, toughness, and favorable plasticity, along with exceptional welding capabilities, making it extensively utilized in big forgings and integral components, including the Su-27, IL-76, Tu-204, and other aircraft structures in Russia [36]. Boeing Corporation of the United States and VSMPO Corporation of Russia have jointly developed the metastable β alloy, Timetal555, also referred to as Ti5553 [37,38]. This alloy was mainly used for large forgings of aircraft, and its strength can reach up to 1517 MPa. To meet the demand of the aerospace industry for titanium based composite materials, including anti-oxidation foils, Timet developed the new metastable β alloy, Ti-15Mo-2.7Nb-3Al-0.2Si, referred to as β-21S for McDonnell Douglas in 1989 [39]. After solid solution aging, the UTS of the alloy remained at 1100–1300 MPa, K_{IC} remained at 60–90 MPa $m^{1/2}$. It possesses remarkable strength and ductility, along with superior high-temperature performance attributed to the use of Si components, rendering it ideal for the production of engine liners and nozzles.

China commenced the study and development of HSTTAs earlier, shown by the TB2 (Ti-3Al-5Mo-5V-8Cr) alloy, developed by the Beijing Nonferrous Metals study Institute in the early 1960s, which is currently extensively utilized in engineering [40,41]. After solid solution and aging, its R_m was about 1100–1200 MPa, and it has good strength plasticity matching. Therefore, it was widely used in aviation components fabricating such as rivets, fasteners, and solid rocket engine casings, and has been used until now. On the basis of TB2 alloy, Ti-3Al-5Mo-5V-2Cr, namely TB10-the nearly β HSTTA-was developed by reducing the Cr content [42,43]. After specific heat treatment, the R_m of TB10 alloy forgings can reach 1110 MPa, and the K_{IC} was 70.5 MPa $m^{1/2}$. TB10 alloy can be used to manufacture high-strength and large-sized rods. It has been used in the connecting rods of the Shenzhou-II orbital module and the Ziyuan-II satellite before 2000. It was an excellent titanium alloy that meets the industrial demands of well structural efficiency and great reliability for structural components using [44]. On the basis of research on HSTTAs in various countries, a large number of new HSTTAs have also been developed domestically, such as TB8 [45–47], TB6 [48–50], TB5 [51,52], Ti-5321 [53], and TB18 [54]. Developed by the Northwest Nonferrous Metals Research Institute in 2004, TC21 alloy was the most typical and widely used alloy [55–62]. This alloy has the characteristics of high strength (R_m over 1100 MPa), high toughness (K_{IC} over 70 MPa $m^{1/2}$), high DTP, as well as excellent fatigue resistance. The performance

Table 1
Partial HSTTAs and their properties.

Alloyd	Type	Component	UTS/MPa	K_{Ic}/MPa·$m^{1/2}$	Country
TC21 [54–61]	α+β	Ti-6Al-2Sn-3Mo-1Cr-2Zr-2Nb	≥1100	≥70	China
Timetal555(Ti5553) [37,38]	metastable β	Ti-5Al-5Mo-5V-3Cr-0.6Fe	1100–1450	50–90	USA, Russia
Ti62222S [30,31]	α+β	Ti-6Al-2Sn-2Zr-2Mo-2Cr-0.2Si	≥1034	≥77	USA
TB2 [39,40]	metastable β	Ti-3Al-5Mo-5V-8Cr	1100~1200	≥60	China
β-C [32]	metastable β	Ti-8V-6Cr-4Mo-4Zr-3Al	1200~1400	58~66	USA
β-21S [39]	metastable β	Ti-15Mo-2.7Nb-3Al-0.2Si	1100–1300	60–90	USA
Ti-1023 [27–29]	near β	Ti-10V-2Fe-3Al	965~1310	33~99	USA
TB10 [42,43]	near β	Ti-3Al-5Mo-5V-2Cr	≥1110	≥70.5	China
BT22 [34,35]	near β	Ti-5Al-5Mo-5V-1Fe-1Cr	1080~1280	≥65	Soviet Union

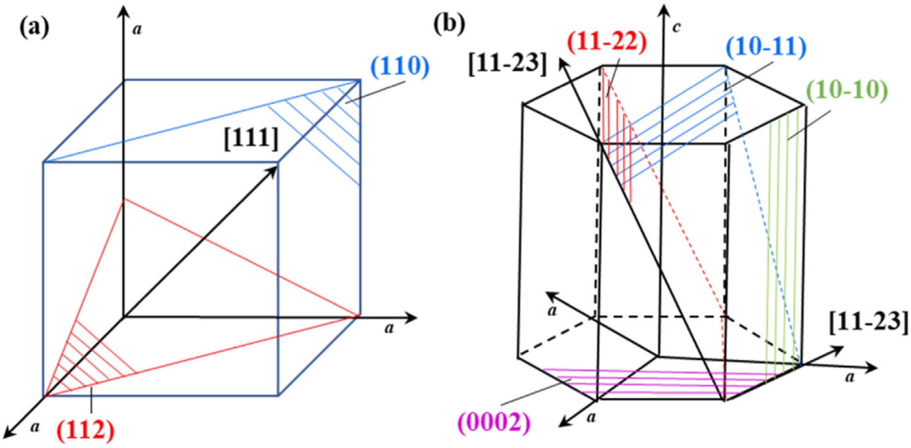

Fig. 2. Slip system of (a) β phase and (b) α phase in titanium alloy.

of this alloy was comparable to that of Ti-6-22-22 alloy used on the fourth generation F-22 fighter in the United States, and its good welding performance also compensates for the poor welding process performance of Ti-6-22-22 alloy [63]. It was suitable for manufacturing important load-bearing components such as large integral frames, beams, and joints.

3. Research on the DTP of HSTTAs formed by AM

3.1. HSTTAs formed by AM

After using AM to form titanium alloys [64–66], it can avoid the shortcomings of traditional preparation processes such as casting and forging, for instance, low material availability and long manufacturing cycle [67]. Especially in the aerospace field [68], the required components have excellent performance and high forming accuracy and complex shapes. With the continuous development and improvement of additive manufacturing technology, HSTTAs, as an important high-performance structural material, have an increasing proportion in the alloys formed by AM. AM not only greatly shortens the forming cycle and reduces costs, but also enables rapid prototyping of complex structures [69–71]. And more diverse microstructures were obtained [72–75] with comparable or better performance [76,77]. However, the microstructures of titanium alloys formed by AM typically exhibit coarse columnar β grains and needle-like martensitic α' [78–81], and thier microstructures were uneven [82–86], which have a bearing on the thermal cycling features of extremely high thermal gradient and extremely fast cooling rate in this process [87,88].

The main types of AM were: wire and arc additive manufacturing (WAAM) [89,90], powder bed fusion (PBF) [91–96], electron beam melting (EBM) [97–101], laser cladding deposition (LCD) [102,103], selective laser melting (SLM) [104–106]. Currently, research on titanium alloys formed by AM shows a positive development trend. For example, Wang et al. [107] developed a new α+β titanium alloy, which was Ti-6.5Al-3.3Mo-1.7Zr-0.15Cr-0.1Ni-1.2Mn-0.3Si-0.6Fe formed using laser directed energy deposition (LDED). It has not only excellent strength (1247 MPa), but also good plasticity (9.2 %). Ding et al. [108] also formed Ti55531 alloy with good strength and excellent ductility through LDED. The Ti5553 alloy prepared by Wu et al. [109] using PBF technology has similar strength to the alloy after annealing.

Currently, there is a lot of research on the forming process parameters of HSTTAs formed by AM. By adjusting the process parameters during forming, such as laser power, scanning speed, and powder feeding speed, the microstructure morphology characteristics can be improved to a certain extent, and a richer variety of microstructure types can be obtained. And many studies have shown that heat treatment can adjust the microstructure of HSTTAs formed by AM, thereby improving the overall performance of the alloy. These methods also have a significant impact on the DTP of alloys, and will be discussed in detail in subsequent chapters. All countries were actively developing HSTTAs to achieve wider applications. At the same time, with the continuous development of new alloys, research on HSTTAs formed by AM was also constantly developing towards high performance, high reliability, well K_{Ic} and low fatigue crack propagation rate [110,111].

3.2. Tensile deformation behavior

The current research on the DTP of HSTTAs formed by AM was relatively limited, lacking an overall overview of fatigue performance [112]. However, its analysis method was similar to the DTP research method of traditional process formed titanium alloys. The research content mainly includes: tensile deformation behavior, fracture toughness (K_{Ic}), and Fatigue crack propagation (FCP).

Due to the diversity of the microstructure of HSTTAs, as well as the complexity of AM forming processes, there were numerous factors that influence the deformation behavior. Not only can the content [113], size [114–117], morphology [118], microstructure [119,120] and crystallographic orientation [121,122] of α phase or β phase affect tensile properties and plastic deformation of the alloy to varying degrees, but there are also multiple deformation mechanisms [123]. The deformation of HSTTAs mainly depends on slip. As shown in Fig. 2(a), there was the body centered cubic structure and a large number of slip systems in β phase. There was closely packed hexagonal structure in α phase, and its slip system includes $<a>$ slip systems along the [11–20] direction and $<a+c>$ slip systems along the $\left[11\overline{2}3\right]$ direction [124,125], as shown in Fig. 2(b). Among them, the $<a>$ slip system can be further divided into basal (0001) $\langle 11\overline{2}0\rangle$ slip, prism $\{10\overline{1}0\}\langle 1\overline{2}01 >$ slip, and pyramidal $\{10\overline{1}1\}\langle 1\overline{2}01 >$ slip. The $<a+c>$ slip system includes first-order pyramidal $\{10\overline{1}1\}\langle\overline{11}23 >$ slip and second-order pyramidal $\{11\overline{2}2\}\langle 11\overline{2}3 >$ slip [126,127].

In the titanium alloy, one or more slip mechanisms of α phase were frequently occurred. For example, June et al. [128] found that in TC4 alloy formed by laser powder bed melting, with temperature increasing, the improvement in strain hardening was result of decrease in critical shear stress and high dislocation density of (0001) $\langle 11\overline{2}0\rangle$ slip and $10\overline{1}1 < 1\overline{2}10 >$ slip systems. In Ti-6.8Zr-2.3Mo-6.9Al-2.2V alloy formed by AM, the activation of prism slip can be promoted by heat treatment, which could increase its ductility [129]. Zhang et al. [129] also found that primary equiaxed α phases have superior capability to coordinate part plastic strain, which brings the sensitization of comprehensive $<a>$ prism slip and improving the ductility. Zhao et al. [130] found that in Ti64 alloy after special heat treatment, $\langle 11\overline{2}0\rangle_{\alpha}$ prism and basal slip

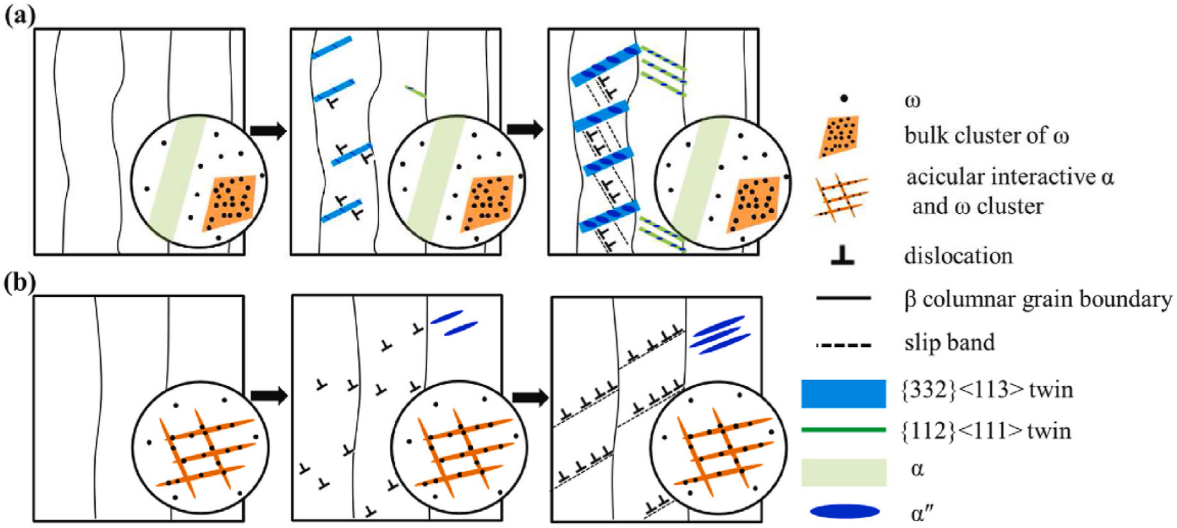

Fig. 3. Schematic diagram of fracture behavior and deformation principle with different laser powers: (a) S1(160 W) and (b) S2(200 W). Taken from Ref. [132] with permission of Elsevier.

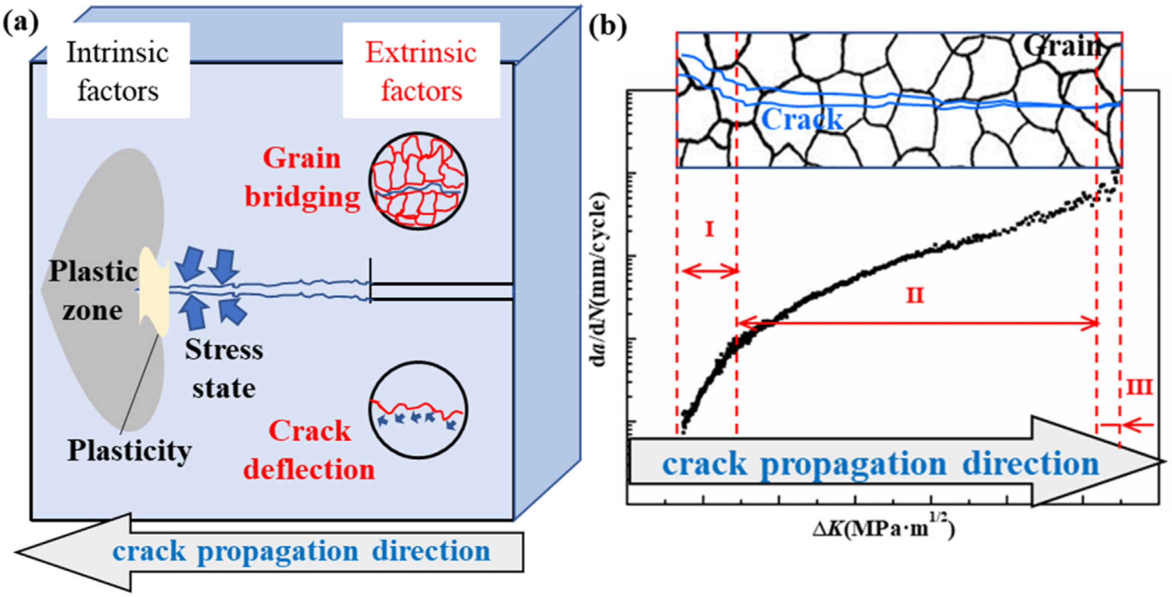

Fig. 4. Crack propagation in titanium alloy (a) factors affecting K_{Ic}; (b) FCP stage.

modes were the easiest to activate slip modes in α variant. In the TA15 alloy, Ti-6.5Al-2Zr-Mo-V, prepared by laser powder bed melting, the deformation mechanism at room temperature was $<a>$ slip [131].

There are still one or more deformation mechanisms in HSTTAs formed by AM. In the study of the deformation mechanism of Ti1023 alloy formed by L-PBF, Ma et al. [118] and Wang et al. [132] found the presence of α" phase and {332}<113> and {112}<111> deformation twins, confirming the existence of multiple deformation systems consisting of twin induced plasticity (TWIP), transformation induced plasticity (TRIP), and dislocation slip, as shown in Fig. 3. Shen et al. [133] identified that the fracture behavior of Ti-4Mo-4Cr-2Sn-2Zr-5Al samples prepared by LDED was caused by stress concentration at the α/β phase interface and compatible deformation between grains. Schaal et al. [134] found the formation of lath α′ and stress induced martensite transformation in L-PBF formed Ti-22Zr-9Nb-2Sn metastable β alloy, which will make it difficult for dislocation slip to occur and improve strain hardening rate. These studies demonstrated that the influencing factors of deformation mechanisms were diverse, including α phase morphology, interface, and stress.

3.3. Fracture toughness

Fracture toughness (K_{Ic}) [135] refers to the capacity of the alloy to withstand certain amount of stress without causing rapid crack propagation and eventual fracture, assuming it has existing defects. Its influencing factors were divided into extrinsic factors for example temperature or strain rate, and intrinsic factors for example element composition or microstructures. From the perspective of fracture mechanics, it can be divided into intrinsic and extrinsic factors, as shown in Fig. 4(a). The intrinsic factors mainly include the state of plastic deformation, the plastic zone size of the crack tip, and the distribution of strain force in the plastic zone, all of which are related to the microstructure characteristics of titanium alloys, while the extrinsic factors mainly include the degree of crack deflection, crack bridging direction, etc. [136].

There was currently limited research on the K_{Ic} of HSTTAs formed by

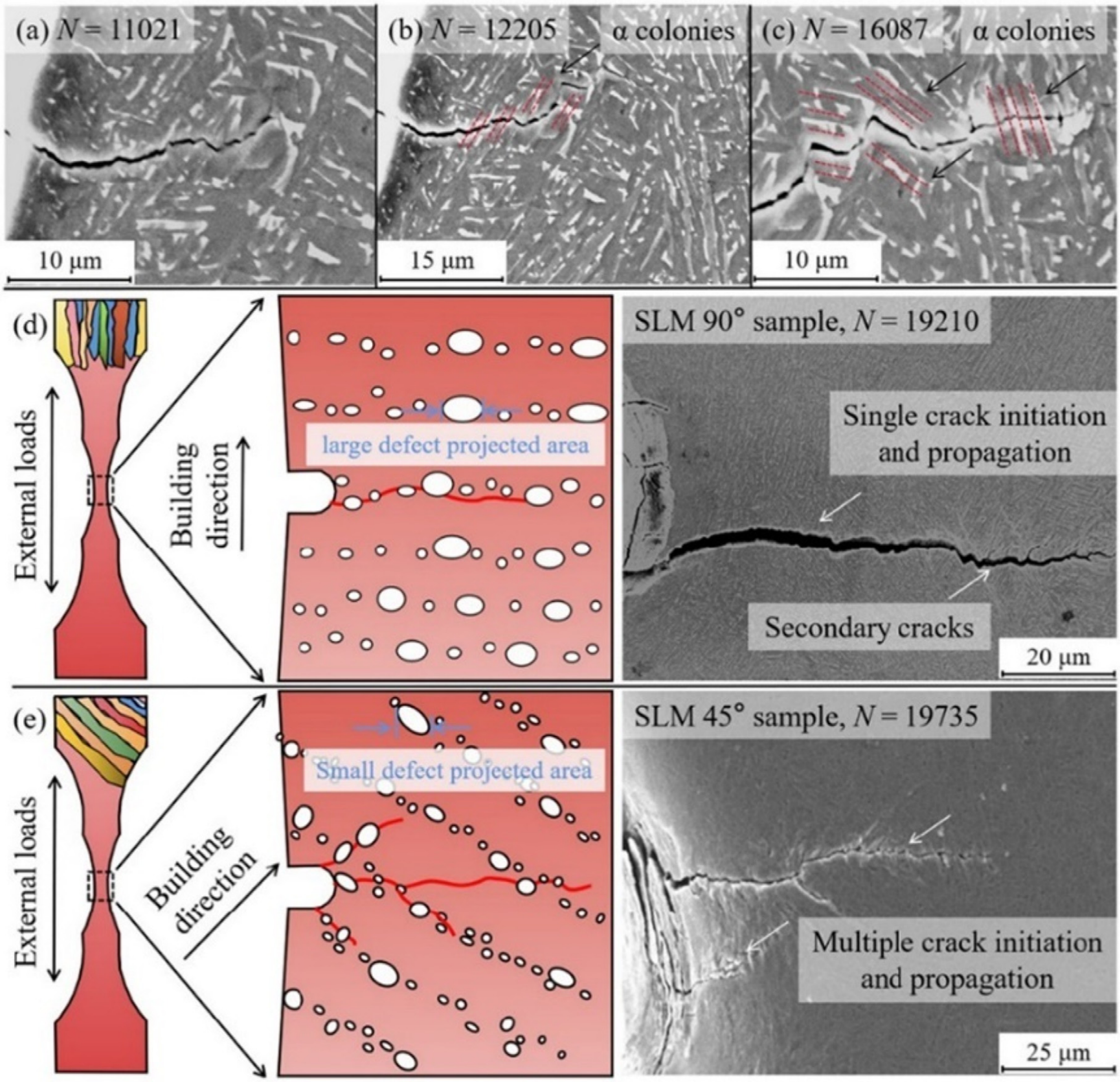

Fig. 5. At the building direction of 90°, fatigue crack propagation behavior (a–c) and single fatigue crack initiation (d), along with at the building direction of 45°, multiple fatigue crack initiation (e). Taken from Ref. [159] with permission of Elsevier.

AM. This was partly due to the fact that its K_{Ic} was not particularly outstanding compared to traditional formed alloys, and partly due to the difficulty of microscopic research, which lacks corresponding mechanisms and analytical methods. The presently investigate mainly center at the effect of microstructure on the K_{Ic} of alloys. In the study of Zhou et al. [137], the size of layer has significant impact on the K_{Ic}. Qin et al. [138] found that the priority plastic deformation of β phase during crack passivation process caused low K_{Ic} in Ti-5553 alloy. Liu et al. [139] found on LCD formed Ti-5321 alloy, the size, microstructure, and orientation relationship of α phase could cause crack deflection and path changes at the process of fracture. The key influencing factors of K_{Ic} was the microstructure of the alloy. Ding et al. [108] demonstrated that through aging and supercritical β annealing (SBCA-A and SBA-A), different grain boundaries were obtained in the Ti55531 alloy constructed by LDED, Widmanstätten grain boundaries and serrated grain boundaries, respectively. Due to the effective suppression of crack propagation by these grain boundaries, the elongation and K_{Ic} were significantly improved. In the study of Liu et al. on Ti-5Mo-5V-5Al-1Fe-1Cr alloy [140], it was found that the β grain size was negatively correlated with K_{Ic}, and the serrated morphology of α_{GB} could improve ductility and K_{Ic}. Moreover, the orientation of titanium alloys formed by AM will also affect the K_{Ic} of the alloy [141]. As found by Ojo et al. [142], the K_{Ic} of samples perpendicular to the building layer was comparable or slightly superior than that of TC4 with annealing. It can be seen that compared to the study of factors affecting the K_{Ic} of titanium alloys formed by traditional processes, there were more and more complex factors affecting the K_{Ic} of titanium alloys formed by AM, and the correlation between these factors was lack.

3.4. Fatigue crack propagation (FCP) behavior

FCP behavior was commonly characterized by fatigue crack propagation rate (da/dN) and stress intensity factor (ΔK) [143–145]. As shown in Fig. 4(b), the FCP curve could be divided into three zones: (i) the fatigue crack non propagation zone or near threshold zone [146,147]; (ii) the linear crack propagation stage, as subcritical FCP stage or Paris zone; (iii) the unstable propagation of fatigue cracks, as the instantaneous fracture zone. The fracture surface in near threshold zone has a rock sugar-like morphology similar to the cleavage plane [148]. Researchers have shown that the factors affecting the FCP in the near threshold region mainly include microstructure, load ratio, and external experimental factors [149–152]. In the Paris zone, fatigue cracks were in the steady-state propagation stage, which was mainly affected by factors

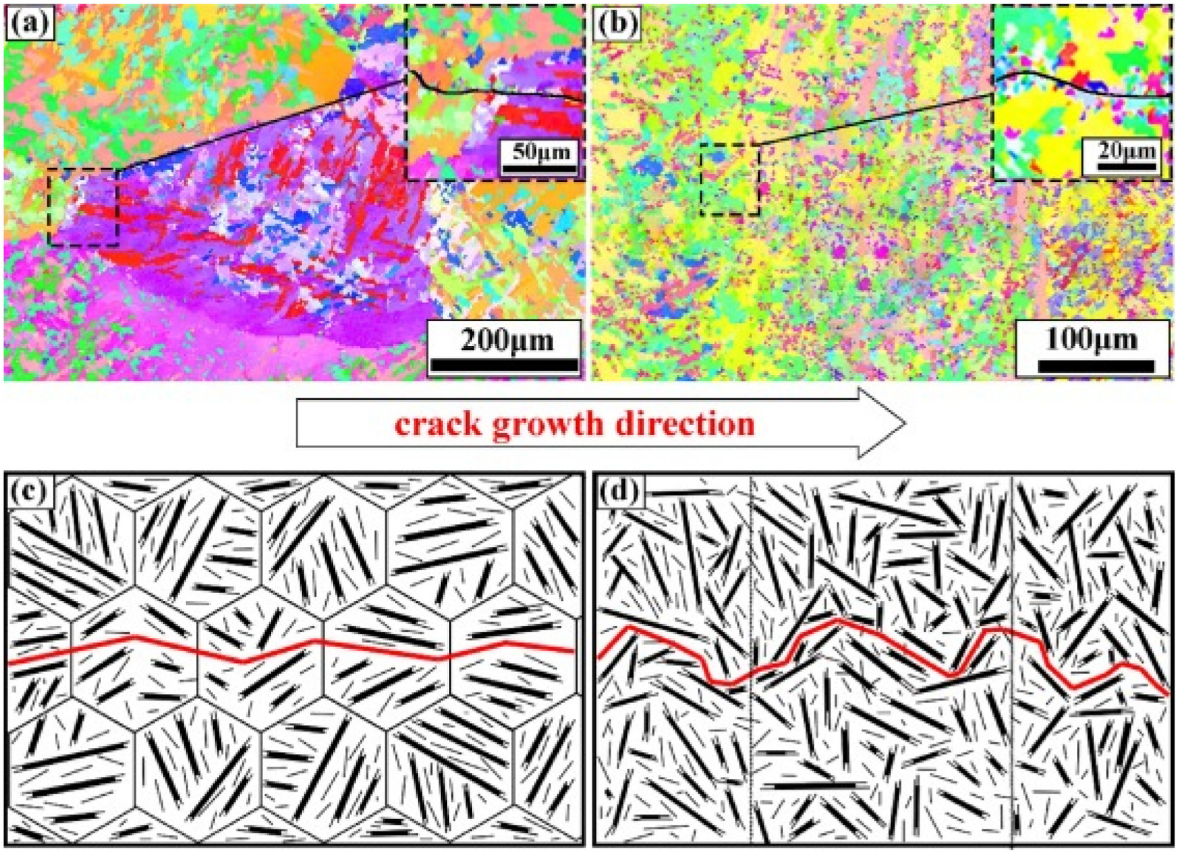

Fig. 6. The FCP along the grain boundary in the equiaxed (a) and columnar grains (b) showing and the schematic of the FCP in the E specimen (c) and the E-C specimen (d). Taken from Ref. [166] with permission of Elsevier.

such as residual stress [153,154]. The fracture morphology of its instantaneous fracture zone was consistent with tensile fracture, mainly influenced by factors such as microstructure and load ratio [155]. Due to the particularity of AM, its process characteristics and resulting defects, such as anisotropy [156–159] as shown in Fig. 5, non-uniformity [160], pores [161], unmelted powder [162], sample size [163], etc. can also affect the FCP of alloys.

Numerous studies have shown that the dimension of α phase, the microstructure and the distribution of HSTTAs has significant impact on the FCP behavior [164,165]. As shown in Fig. 6, Wang et al. [166] identified that the differences in FCP behavior (crack initiation stage) between specimens were mainly owed to the various sizes and shapes of α_{layers} and α_{bundles} within columnar or equiaxed grain regions instead of grain boundaries. As found by Liu et al. [167], in Ti-6Al-2Zr-Mo-V alloy formed by LDED, the FCP behavior was likely interrelated to the size and the crystallographic spatial orientation of the α lamellar. Fatigue cracks with shorter lengths grew along the α/β boundary, while those with longer lengths were mainly affected by the deflection of α/β boundary in printed specimen. With studying the FCP behavior of TC4 alloy formed by WAAM, Wang et al. [168] detected that the initiation and propagation of cracks were influenced by microstructure, and the crack propagates along the sliding line due to the presence of continuous slip bands, but result from the blocking effect of dislocation slip, the grain boundaries will hinder crack growth. In studying the FCP of TC11 alloy formed by laser AM, Wang et al. [144] found that was higher at low ΔK values, mainly due to the morphology and the size of α phase. In the study of Naab et al. [169], it was found that the increase of ΔK_{th} was due to the increase of lamellar α thickness and β phase content result from heat treatment. When studying the FCP behavior in Ti-55511 alloy formed by LMD, Wang et al. [170] realized that the ordered arrangement of α_p phase will deflect the crack path and consume a large amount of energy, thereby prolonging the fatigue life. The microstructure affects the propagation and initiation of cracks, and its essence was that it affects the deformation mechanism. When studying the FCP behavior of WAAM formed TC4 alloy, Wang et al. [168] found that the synergistic and competitive effects between numerous factors for instance grain boundaries, sustained slip bands, and crack branches. Ueki et al. [171] found that at the crack tip, the active slip system was determined to the crystal orientation of α phase, as shown in Fig. 7. There were three main mechanisms for crack propagation: (1) accumulated damage caused by dislocation-dislocation interaction from out-plane slip; (2) alternating shear caused by in-plane prism slip and (3) interlayer debonding force caused by damage accumulation caused by interlayer boundary dislocation interaction.

4. Methods for improving DTP

Recently research on improving the performance of HSTTAs formed by AM mainly concentrates on improvement of process parameters in the AM process [50,172] and post heat treatment [173,174]. Due to the unique thermal cycling characteristics of AM processes [175,176], where parts were formed layer by layer through the transition of high-temperature liquid metal droplets, coupled with a very rapid cooling rate, microstructures different from traditional forming techniques can be produced. Typical TC4 alloys formed by AM exhibit extremely columnar grains along the construction orientation, and its length range from 100 μm to even 10 mm [177–180]. This was owing to nucleation of β grains on the boundary between the previously deposited layer and the substrate or melt pool. In the melt pool, the unavoidable excessive thermal gradient (over 105 K/m) greatly reduces subcooled quantity, when solidification occurs to bring growth of extension grains and extends to lengths much greater than the thickness of the deposited layer

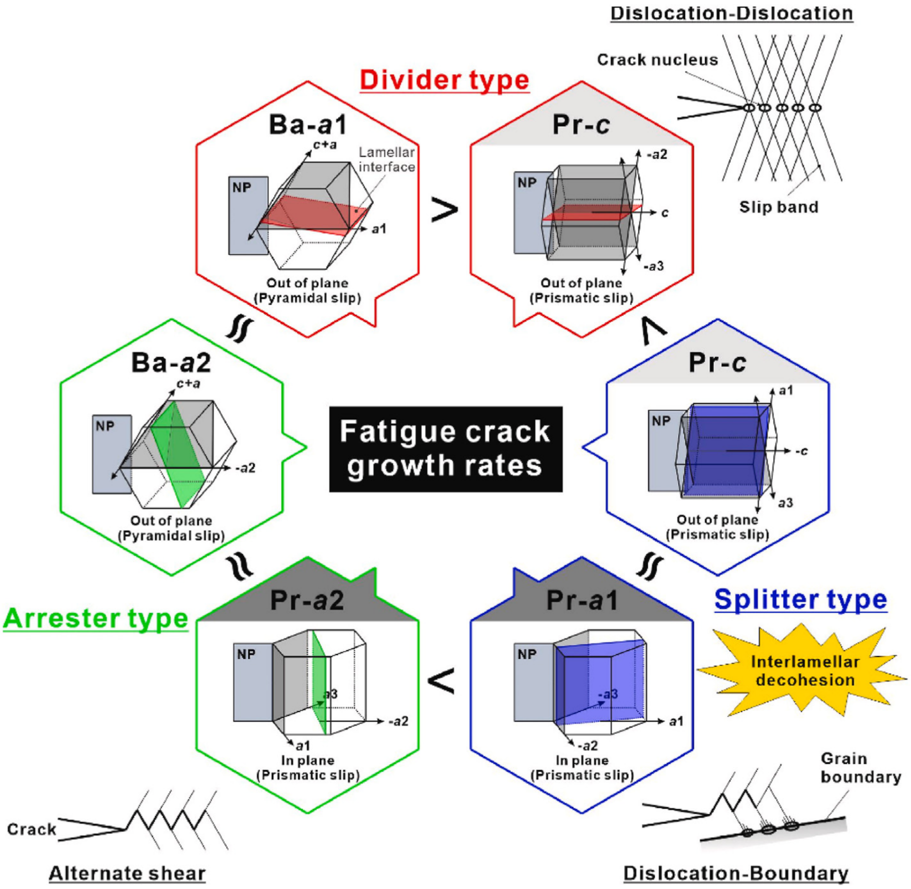

Fig. 7. Schematic diagram of the relationship between the crystallographic α phase orientation, layered structure, activated slip system, FCP and the basic process of crack propagation in TC4 layered structure. Taken from Ref. [171] with permission of Elsevier.

[181]. Meanwhile, there was distinct $<001>_{\beta}$ stripes in deposition orientation [182]. It was precisely the unique process and thermal cycling characteristics of AM that provide the possibility for generating new microstructures and properties.

4.1. Adjustment of process parameters

There have been numerous studies on the process parameters of AM. By adjusting the process parameters during forming, such as energy density [183,184], laser power [185–190], scanning speed [186], powder feeding speed, deposition strategy [191,192], it was possible to improve the microstructure characteristics to a certain extent and to obtain more diverse types microstructure. This was because, for one thing, complex thermal during AM affects the phase transition process of HSTTAs, and for the other, its processing characteristics extend methods and possibility for designing new HSTTAs with special microstructures and superior performance that traditional processing cannot reach [193]. Hua et al. [102] adjusted the proportion of self-lubricating phase CaF_2 in the TC21 coating by controlling the laser power. Higher input power increases the content of CaF_2 phase in TC21, but reduces the proportion of hard phase, leading to a diminish in friction resistance and ductility of the alloy. Yao et al. [194] proposed a refinement splitting (RS) method by optimizing scanning strategies and increasing energy density, as shown in Fig. 8. This method could optimize its microstructure, and refined β grains or α' bundling increase the elongation of TC4 alloy by 1.9 times. Montalbano et al. [195] identified that selecting proper input power can greatly improve plasticity of SLM formed TC4 titanium alloy and change its fracture type, as shown in Fig. 9. Selecting appropriate process parameters can improve the plasticity and toughness of the alloy. However, due to the fixed thermal cycles and microstructural characteristics of AM, this improvement was limited. Therefore, researchers have paid more attention to the subsequent processing of HSTTAs formed by AM.

4.2. Assistive technology and simulation methods

The strengthening mechanism of some auxiliary technologies on alloys was similar to that of adjusting process parameters. These processes did not involve the complex phase transformation process in AM, and were implemented before or during AM forming. Equal channel angular pressing (ECAP) [196] could increase the K_{Ic}. The grain refinement, plastic deformation, and strain hardening generated by laser shock peening (LSP) [197–199] help enhance mechanical properties. Hybrid additive manufacturing (HAM) generates high-density dislocations and nano twins to improve plasticity [200]. For example, Kang et al. [175] proposed that introducing a transverse magnetic field can improve the ductility of Ti60 alloy formed by AM. Simultaneously applying *in situ* rolling during the WAAM forming process can improve the resistance to crack propagation, as shown in Fig. 10. This was because *in situ* rolling can reduce the size of defects, promote arising of the secondary cracks, as well as make FCP path more tortuous [201]. As shown in Fig. 11, Jin et al. [202] found that by adding different proportions of 316L to TA15 deposited by LMD, its microstructure can undergo Widmanstädter→basket weave→bimodal→equiaxed transformation. During this process, the mating relationship between strength and plasticity can be increased, while effectively maintaining a certain level of toughness. These technologies were achieved by assisting other technical means in the forming process to improve the ductility of alloys, and related research was also increasing, e.g., Zhang et al. [203] proposed a fresh AM with flat-top laser-aided cold metal transfer (F-LCAM) for HSTTAs, which can reduce the weld morphology defects of Ti6321 alloy, refine the grain

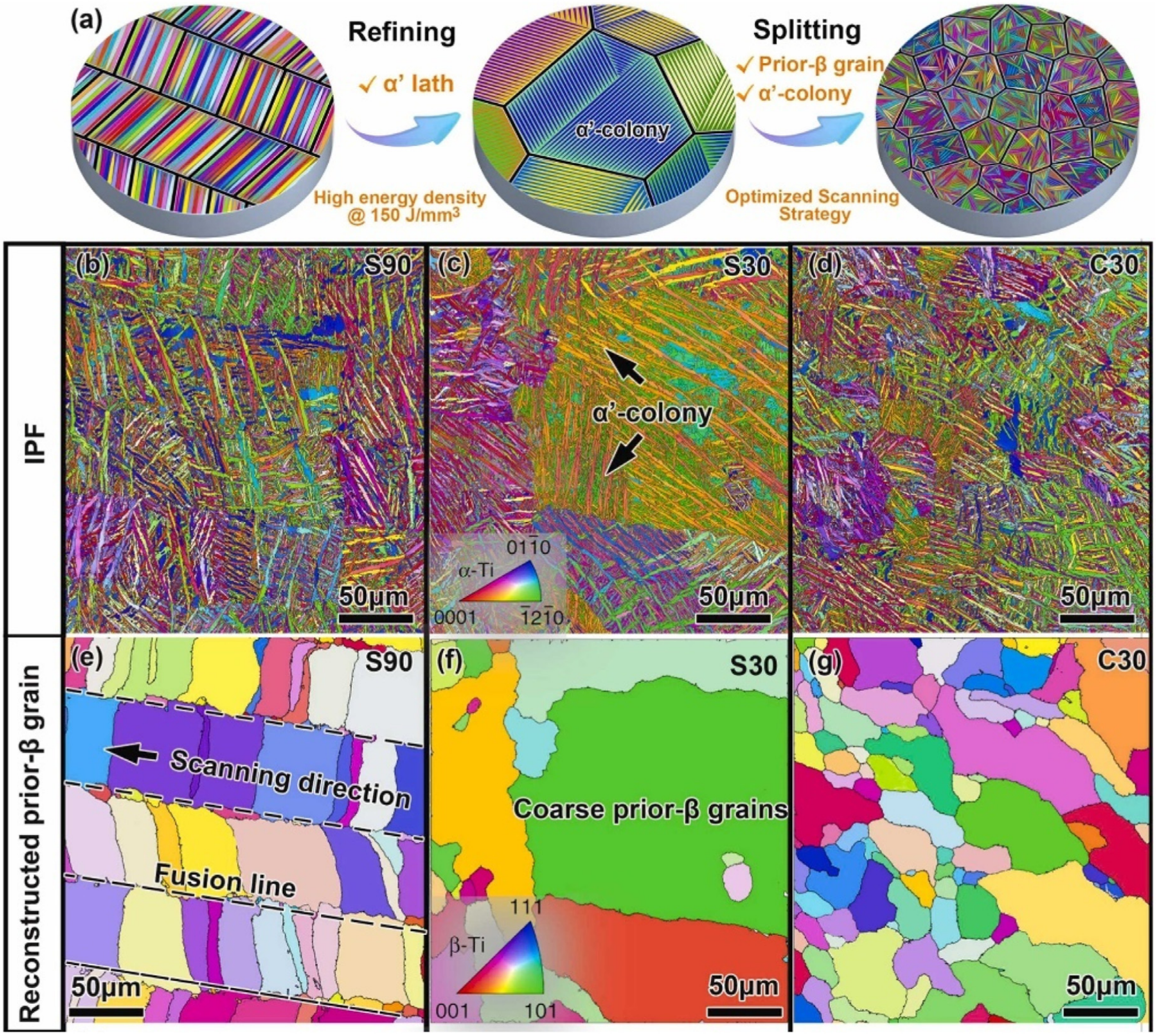

Fig. 8. Using R&S methods, microstructure transformation in TC4 alloy formed by L-PBF. Taken from Ref. [194] with permission of Elsevier.

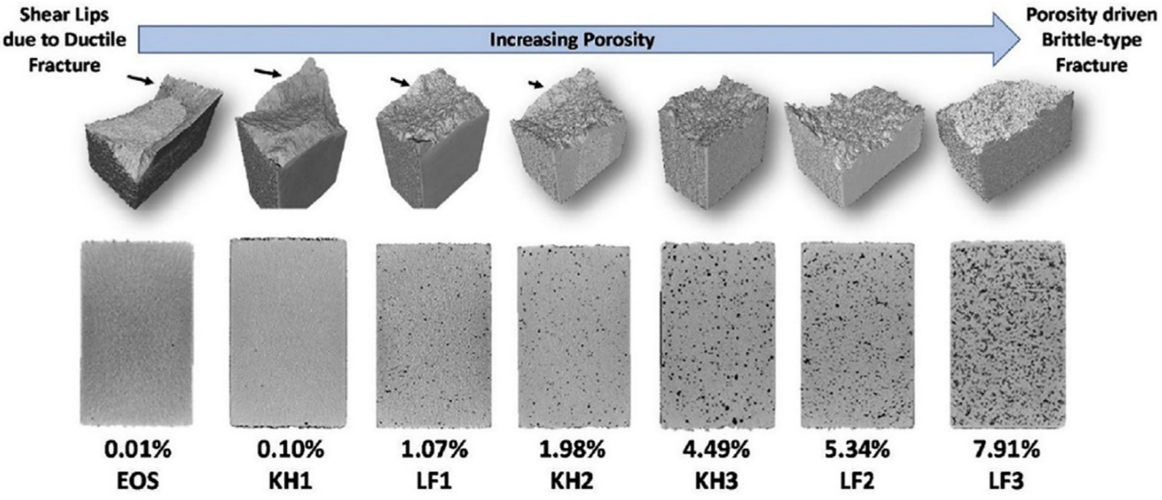

Fig. 9. In TC4 alloy formed by SLM, as porosity increases, tensile fracture morphology and tomography morphology. Taken from Ref. [195] with permission of Elsevier.

size, achieve higher elongation and lower anisotropy. In addition, adding other elements such as copper to titanium alloys for alloying could similarly upgrade the plasticity and strength of HSTTAs [204]. The improvement and enhancement of K_{Ic} can also be achieved through

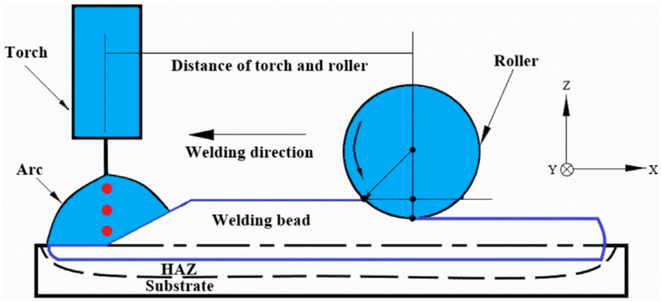

Fig. 10. Diagrammatic sketch of *in situ* rolling WAAM technique. Taken from Ref. [201] with permission of Elsevier.

special technical means, for instance, equal channel angular pressing technology (ECAP), which can significantly improve the K_{Ic} of TC4 alloy formed by DED [196].

Mechanical learning [205], numerical calculation, or simulation [206] can be used to explore the impression of certain parameters on DTP of HSTTAs formed by AM. For example: using the improved Hartman-Schijve equation, to calculate the small crack propagation rate in TC4 alloy produced by AM [207]; simulate FCP behavior to verify its interface effects [208]; simulate the effect of residual stress on the FCP of HSTTAs [209]; model evaluation of effective variable intensity factor to obtain the trend of FCP [210]. For example, Karpenko et al. [211] put forward a dynamic fatigue former to evaluate the influence of hole size or pore site on the fatigue life of Ti6Al4V formed by AM. According to variational mode decomposition (VMD), Yang et al. [212] proposed an adaptive denoising imaging defect detection system to ultrasound array detection. From this, it can be seen that computation and simulation can not only provide prediction and optimization functions, but also promote microstructure control and multi-scale mechanical research, providing forceful support for improving the DTP performance of HSTTAs.

4.3. Heat treatment

Many studies have shown that microstructures of alloys can be adjusted through heat treatment [213,214] and their mechanical properties can be further improved [215,216]. Researchers have developed various heat treatment methods, including annealing [217,218], solid solution [219,220], aging [221,222], slow cooling in the β phase region [139], and multiple heat treatments [223–225], to regulate the

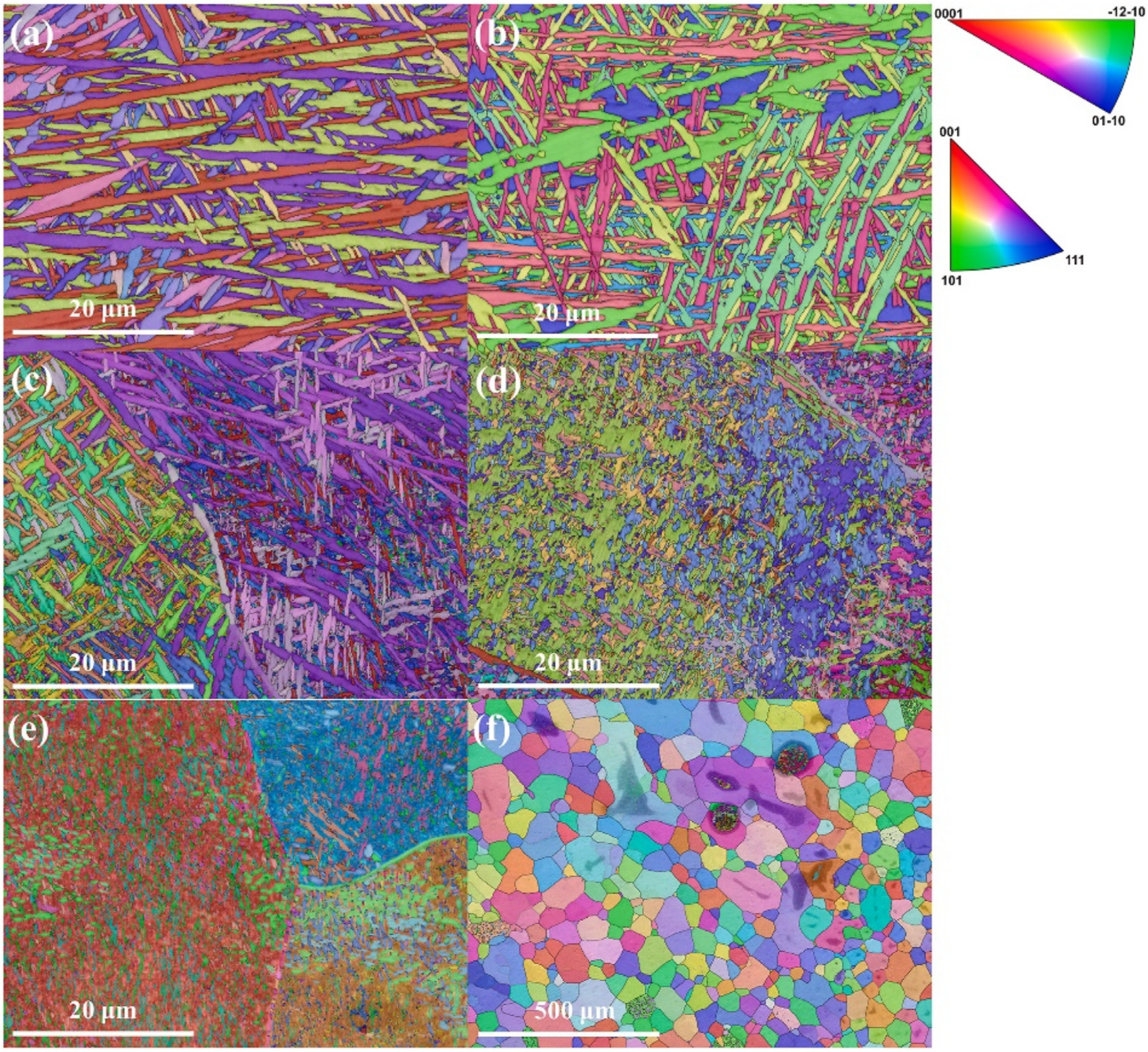

Fig. 11. EBSD images of TA15 with 316L (0, 1, 3, 5, 7, 9 wt%). IPF maps of (a–f) TA15-0, 1, 3, 5, 7, 9. The evolution of microstructure and morphology rule was developed from near α phase to α+β phases and ultimately transformed into near β phase. The composition of α and β phases undergone corresponding varied. Taken from Ref. [202] with permission of Elsevier.

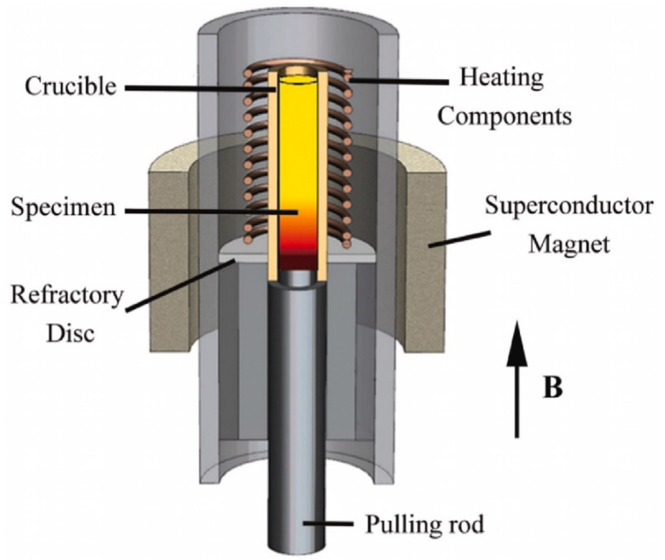

Fig. 12. Based on Bridgman furnace, modified with a ring of superconductor magnet, designed HMF-HT setup. Taken from Ref. [130] with permission of Elsevier.

microstructure, improve ductility or DTP [226,227]. As found by Schwab et al. [228], Ti-5553 alloy (Ti-5Mo-5V-5Al-3Cr) formed by SLM has the microstructure of uniformly distributed sheet-like α phase and β phase matrix after annealing at 500 °C, and its tensile strength would be enhanced to 1600 MPa. Similarly, using certain heat treatment, Zuo et al. [229] found that it was expected to weakened the anisotropy of Ti-55531 alloy formed by SLM and improved its corrosion resistance, resulting in bimodal and bilayer structures. Raising aging temperature lead to α phase coarsening and decreasing in its content, causing cracks to deflect and consuming additional energy, thereby improving K_{Ic}. In Ti-55511 alloy, Liu et al. [230] demonstrated that its microstructure evolution and K_{Ic} were highly sensitive to multiple heat treatment mechanisms. Multiple heat treatment was an extension of single annealing and one of primary means for adjusting the microstructure morphology of HSTTAs. For instance, Furuhara et al. [231] enhanced the strength-toughness matching relationship of β-Ti-15-3 alloy through specifically customized two steps aging heat treatment at high or low temperatures. As shown in Fig. 12, Zhao et al. [130] combined high magnetic field with heat treatment, to significantly improve the ductility of Ti64 alloy prepared by L-PBF, which was related to its easy activation of $<11\text{–}20>_\alpha$ prism and basal slip. This was because heat treatment can eliminate the inhomogeneity and internal stress caused by AM process to a great extent, and further optimize the microstructure, so as to improve the plasticity and toughness of HSTTAs.

Similarly, Liu et al. [232] found that, formed by EBRM, after solution aging treatment, the tensile strength of TC4 alloy was significantly improve. Due to the refinement of α lath, with solution temperature increasing from 930 °C, its strength could be improved. Ghosh et al. [233] found that prolonged annealing has an adverse effect on its fatigue life of TC4 alloy formed by AM. In the laser deposited TC4 alloy, Qin et al. [234] also found that there were extensive fine phases after solid solution aging, which can hinder dislocation slip and cause dislocation accumulation, resulting in decrease in plasticity and improvement of its strength. Therefore, single solution treatment will mostly refine the α phase and greatly improve the strength of the alloy, which was not conducive to the improvement of the plasticity and toughness of the alloy. Annealing treatment has been proved to be effective in improving toughness. An et al. [235] detected that after annealing Ti-5111 alloy formed by WAAM, more secondary twins and deformation twins were sensitized, which can significantly enhance its toughness. After critical annealing and aging of Ti55531 alloy formed by LDED, Ding et al. [108] found that the K_{Ic} not only reached 81.7 MPa m$^{1/2}$, but also had good strength-plasticity matching. Wang et al. [107] also found primary α phase (α_p) in which 1.4 ± 0.2 μm thickness and fibrous root morphology can be obtained after annealing, which was the principal basis for its improved ductility. On the basis of annealing treatment, many new heat treatment methods which can improve the plasticity and toughness of the alloy have been developed. Zhang et al. [129] further indicated that cyclic heat treatment of AM-ed Ti-2.3Mo-6.8Zr-2.2V-6.9Al alloy can effectively reduce length-width ratio of layer α, improving spheroidization of primary α, and enhancing plasticity and toughness of the alloy. Heat treatment can also increase its dislocation density, which leads to improvement of its ΔK_{th} [169]. Zeiler et al. [236] found that heat treatment leads to the generation of layered structures in HSTTAs formed by EBM. Large α bundles result in higher K_{Ic}, but have little effect on the ΔK_{th} for long FCP, while small α bundles are beneficial for increasing the ΔK_{th} for long FCP. After heat treatment, Zhang et al. [237] also found that the TC4 alloy obtained basket weave structure, and lamellar α thicken at 800 °C. Its crack tip will growth and deflect along lamellar α back and forth, which lead to the increase of energy dissipation in FCP and the generation of tortuous propagation paths, as shown in Fig. 13. After performing quasi β heat treatment on TC4, Wang et al. [238] eliminated the interlayer HAB, generating clustered microstructure and uniformly equiaxed grains, which improved the FCP anisotropy. Therefore, a strategy was proposed to obtain lower and isotropic FCP, which was also based on optimizing additive manufacturing processes and post heat treatment designs, as shown in Fig. 14. Su et al. [239] found that in the Ti3662 alloy formed by LPBF, a good synergistic effect can simultaneously improve the strength and plasticity by adjusting annealing time or temperature, while maintaining tensile strength over 1100 MPa and elongation over 15 %. In conclusion, heat treatment can effectively improve the microstructure and maintain a good matching relationship between strength and ductility of HSTTAs.

5. Summary and prospects

5.1. Problems on the DTP of HSTTAs formed by AM

Upon examining the existing research on the DTP of HSTTAs formed by AM, it was determined that numerous studies still exhibit various issues.

Since its initial development in the United States in 1954, the Ti-6Al-4V or TC4 alloy has been extensively researched and employed due to its superior mechanical and processing characteristics. Owing to its established processing system and extensive commercial application, HSTTAs formed by AM have consistently been a focal point of research. Fig. 15 illustrates that a statistical examination of more than 270 research articles on titanium alloys, sourced from *Addit. Manuf.* journals in recent years, indicates that studies on TC4 alloy [65,240–245] comprise over 70 %, whereas research on other HSTTAs is comparatively limited. This clearly illustrates the insufficient study on novel HSTTAs, thus constraining their development and applications.

In the previous research, titanium alloy showed high strength due to AM process, so the research mainly focused on how to significantly improve its static load strength, or how to achieve a better strength plasticity matching relationship. Nevertheless, the existing research on the DTP of HSTTAs remains limited, mostly owing to the variability of research topics and the constraints of characterization techniques. The DTP encompasses numerous performance indicators, including plastic deformation, K_{Ic}, and FCP behavior, indicating a wide array of research topics and various contributing elements, such as temperature, stress state, loading rate, and microstructural features. Secondly, the current experimental techniques and methodologies frequently fall short in effectively reproducing the intricate damage states, making it challenging to precisely represent the damage behavior under real operational situations. In practical implementation, the components of HSTTAs may sustain damage in various forms, including cracks and holes.

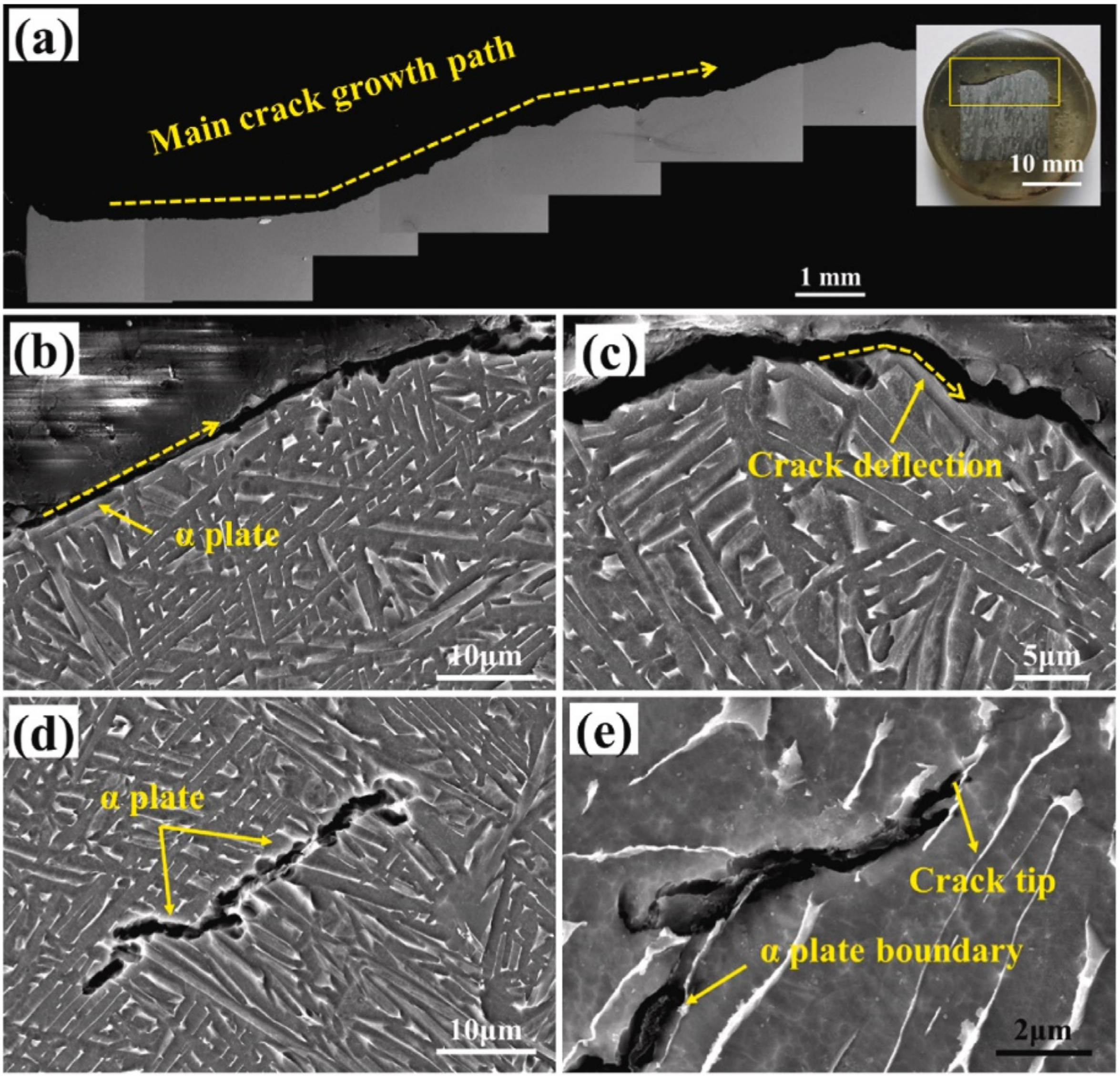

Fig. 13. FCP in HT2 specimen: (a) overall growth path; (b) the main crack; (c) path at boundary α; (d) overall secondary crack; (e) secondary crack tip at boundary α. Taken from Ref. [237] with permission of Elsevier.

Nonetheless, effectively simulating these damage states and assessing their impact under laboratory circumstances presents significant challenges. The existing assessment methodology for the DTP of HSTTAs formed by AM lacks standardized criteria and specifications, complicating direct comparisons of various investigations and undermining the repeatability and generalizability of the findings. For instance, several research teams may employ divergent experimental circumstances and methodologies when assessing K_{Ic} or FCP rate, leading to inconsistent test outcomes. These considerations collectively result in the deficiency of study on the DTP of HSTTAs formed by AM.

5.2. Prospect of research on the DTP of HSTTAs formed by AM

Upon reviewing the research on the DTP of HSTTAs formed by AM, it is evident that future research topics may concentrate on:

1). Investigate the multi-scale determinants of DTP, including micro scale (grain size, phase distribution, etc.), meso scale (voids, fractures, etc.), and macro scale (overall mechanical characteristics). Investigating the factors regulating these various scales can elucidate the link between microstructure and DTP.
2). Investigate the interrelation among various damage processes: the damage process of HSTTAs may encompass multiple damage mechanisms, including FCP, loss in fracture toughness, and plastic deformation, among others. These various degradation mechanisms may simultaneously occur and interact, complicating the link between microstructure and characteristics.
3). Novel testing procedures and characterization methods: sophisticated experimental apparatus and methodologies are essential for the precise measurement and characterization of the DTP of HSTTAs formed by AM. Furthermore, to effectively simulate the AM process and the design-to-production pipeline, it is essential to develop a sophisticated mathematical model and account for the impact of numerous variables.
4). Establish a cohesive evaluation system for DTP: the paramount objective is to provide a standardized and quantitative criterion for performance assessment. The variability in testing equipment and methodologies has resulted in disparate evaluation systems for DTP, and there is an absence of a robust and broadly applicable assessment standard.

In summary, the existing study on the DTP of HSTTAs formed by AM

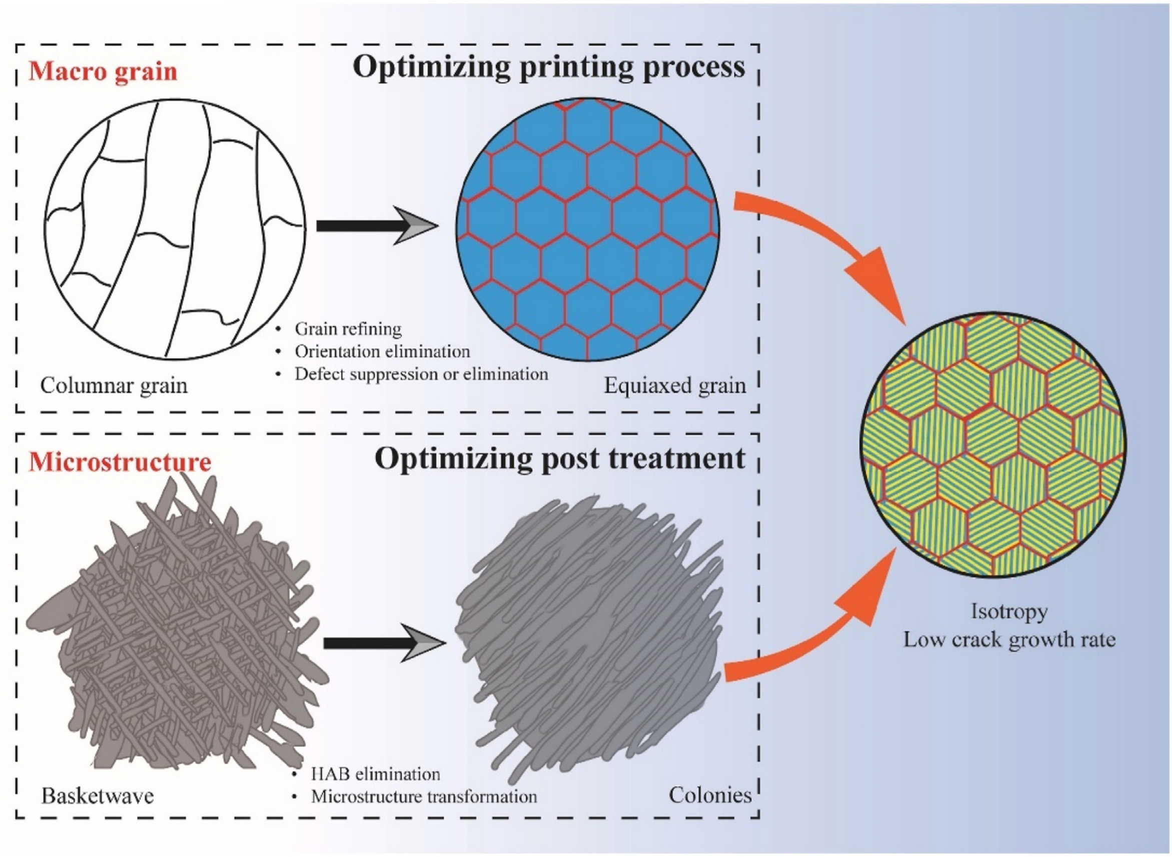

Fig. 14. Upgradation of FCP and inspiration for related AM technology. Taken from Ref. [238] with permission of Elsevier.

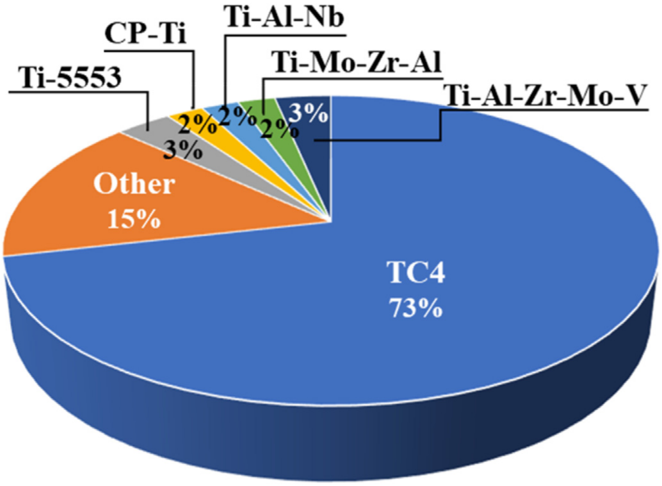

Fig. 15. Proportion chart of various titanium alloys in AM.

continues to encounter numerous challenges. Future research should focus on the comprehensive examination of novel alloys, the advancement of multi-scale performance, experimental and simulation methodologies, and the establishment of standardized evaluation criteria, to elucidate the intrinsic relationship between microstructure and DTP of new HSTTAs, thereby offering guidance for quantifying the DTP of HSTTAs formed by AM.

CRediT authorship contribution statement

Guozheng Liu: Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Data curation. **Qinyang Zhao:** Writing – review & editing, Supervision, Conceptualization. **Weiju Jia:** Writing – review & editing, Supervision, Methodology. **Yan Zhang:** Writing – review & editing, Supervision. **Shuo Song:** Writing – review & editing, Supervision. **Chengliang Mao:** Validation, Software. **Wei Zhou:** Visualization, Conceptualization. **Siyuan Zhang:** Funding acquisition. **Yongqing Zhao:** Writing – review & editing, Supervision, Methodology, Investigation.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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