Review Article

Metallurgy principles applied to powder bed fusion 3D printing/additive manufacturing of personalized and optimized metal and alloy biomedical implants: an overview

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ABSTRACT

This overview presents a survey of the state-of-the-art of laser and electron beam powder bed fusion, 3D printing design, development, fabrication and applications of porous, or open-cellular metal and alloy personalized implants; and is particularly directed to materials and biomaterials students and professionals. Of particular importance is the application of metallurgy principles, especially the role played by traditional solidification fundamentals, in predicting and characterizing the microstructures and mechanical properties ofadditively manufactured implants representing a host of human skeletal reconstruction and replacement appliances. In addition to presenting important reviews highlighting very recent metallurgical processing strategies and current trends in the global development of hospital point-of-care, 3D printing centers creating surgical planning models in association with the fabrication of personalized, patient-specific implants are described.

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Lawrence E. Murr has been in metallurgical and materials engineering education, research, teaching, and academic administration for more than 50 years at Penn State University, University of Southern California, New Mexico Tech, Oregon Graduate Center, and The University of Texas at El Paso, where he currently is Emeritus Professor. He has also been a member of the ATIG-Phoenix technical staff since 2014. The recipient of numerous distinguished research and teaching awards including The TMS 2007 Educator Award, The 2007 John S. Rinehart Award for global leadership in shock and high-strain-rate phenomena, the 2008 Henry Clifton Sorby Award for lifetime achievement in metallography, the 2009 Albert Easton White Distinguished Teaching Award of ASM International, the 2010 Piper Professor Award, the 2014 Alpha Sigma Mu Materials Science Distinguished Life Member Award, and Lee Hsun Distinguished Lecture Awards of the Shenyang National Laboratory for Materials Science, Institute of Metal Research, Shenyang, China (2010 and 2016). Dr. Murr has also published over 950 technical papers and 23 books, the latest being “Handbook of Materials Structures, Properties, Processing and Performance” published by Springer in 2 volumes in 2015. His 2019 h-index (Google Scholar) is 82. He is the Co-Editor-in-Chief of the Journal of Materials Science and Technology, and Editor of the Journal of Materials Research and Technology. The holder of 6 patents, Dr. Murr also holds BS degrees from Albright College, and Penn State University as well as MS and
1. Introduction

1.1. Overview of layer-building technologies

3D printing/additive manufacturing involving layer-by-layer building of metal and alloy components is now nearly a half century old [1]. However, modern or contemporary layer-building using metal or alloy wire or powder layer fabrication has grown rapidly over the past two decades, primarily because of commercially available systems which are illustrated schematically in Fig. 1 [1–3]. In Fig. 1(a), a wire feed in an inert gas shroud or in a vacuum is melted by a laser or electron beam. The building component is moved in 3 dimensions (x–y and then z). Correspondingly, in Fig. 1(b), the powder feed and beam-melting system scans in the x–y plane to selectively build a layer segment, while the built layer then drops down along the z-axis to repeat the layer building. In both Fig. 1(a) and (b) a computer-aided design (CAD) program selectively controls the layer building process.

In Fig. 1(c), a powder layer is rolled or raked to a thickness of several mean powder particle diameters and then selectively melted by a CAD-driven laser or electron beam. The selectively melted powder layer then drops down (along the z-axis) and a new powder layer created, and the process repeated layer-by-layer to print a 3D product. Selective laser melting (SLM) of the powder bed in Fig. 1(c) is accomplished in an inert environment using either purified nitrogen or argon while the selective electron beam melting (EBM) process is accomplished in a vacuum. Both processes (SLM or EBM) are currently achieved in commercial systems having somewhat restrictive build volumes of roughly 0.09 m³, although some specialized system designs now allow building large components in volumes approaching 1 m³. Fig. 1(d) illustrates a variance of powder-bed-layer fabrication where a metal or alloy powder is rolled or raked into a layer where a CAD-driven nozzle or printer head (jet) selectively applies a binder. This process is repeated layer-by-layer and the unbound powder is then carefully removed from the selectively built product which is sintered or hot isostatically pressed (HIPed) at high temperature. While this process involves some shrinkage and often produces products with some intrinsic porosity, the resolution of features is in the sub-millimeter range for higher temperature metal or alloy product or component fabrication.

As of 2019, there were roughly 40 different commercial metal 3D printing/additive manufacturing systems, with total sales annually and globally in the billions of U.S. dollars [1]. By far the most amenable and popular systems for the production of commercial medical devices, particularly porous biomedical implant devices, involve laser and electron beam powder bed fusion technologies illustrated in Fig. 1(c). Even though the build volumes are relatively modest as indicated above (~0.09 m³), many biomedical devices such as porous cranial facial inserts, various spinal insert, and many other skeletal or orthopaedic implants can be conveniently fabricated in these commercial machines. In addition, while many classical, commercial metal and alloy implants (especially total hip and knee implants) have been fabricated from forged or cast-ingsots and then CNC machined to specific, patient available sizes and surface finished, laser and electron beam 3D printing/AM systems can allow complex and strategically placed porosities to be directly fabricated specific to particular patient skeletal dimensions, and often only final polishing of critical components is required. Excess powders can be conveniently reused and there are no tooling requirements since patient-specific micro-CT data can be embedded in CAD software to create complex and porous implants. Even the difference in cost or forged precursor costs, such as $300/kg for Ta, $20–$35/kg for Ti-6Al-4 V in contrast to commercial powder costs of $2000–$3000/kg for Ta and $250/kg for Ti-6Al-4 V, is especially cost effective since multiple components having specific porosity and dimensional design features can be built in a single process; and each can be personalized or patient-specific. However, in some parts of the world, powder bed fusion technologies (Fig. 1c) provide mass production of prescribed implant designs, rendering considerably lower appliance costs. For example, total hip replacements in China and India averaged ~$2000 in 2018; with knee implants costing ~$1500. In addition, price reduction of 20–50% was imposed in China in 2019, while India imposed price controls on orthopaedic (hip and knee) implants in 2015.

Orthopaedic implant surgeries worldwide increased from around 1.7 million in 2012 to 2.9 million in 2016; with expectations exceeding 5 million by 2021. The orthopaedic implant market ranking since 2015 includes (in decreasing share): U.S., China, Japan, Germany, France, UK, Italy, Brazil, Spain, Canada, Australia and India. Global market implant AM revenues are expected to exceed $65 billion by 2025. The market is currently dominated commercially by companies such as Stryker, DePuy, Aesaleap, Zimmer/BIomet, Smith and Nephew, Synthes and Medtronic. By far, the primary biomedical device alloy is Ti-6Al-4 V, but some manufacturers utilize Co-Cr-Mo alloys (especially for critical components), 316L stainless steel, and tantalum; all available in purified and optimized (spherical and appropriately prealloyed) precursor powders.

1.2. Fundamentals of powder bed fusion AM of complex and porous biomedical implants

Powder bed fusion (SLM and EBM) technologies have proliferated globally over the past 2 years, while commercial systems are increasingly utilized in hospital 3D printing hubs or labora-
Fig. 1 - Principal commercial metal additive manufacturing process comparisons. (a) Laser or electron beam layer building using metal or alloy wire feed; (b) Powder feed. (c) Powder bed fusion layer building using laser beam or electron beam melting: selective laser melting (SLM) or electron beam melting (EBM), respectively. (d) Binder jet powder layer building using a CAD-driven binder jet to stabilize selective powder layer areas. Excess powder is removed and the fabricated 3D product is sintered to full or nearly full density in a high-temperature furnace. After Murr and Johnson [4].

Fig. 2 summarizes the principal features of the powder bed fusion technologies in Fig. 1(c) using EBM. Fig. 2(a) shows the electron gun and beam forming and scanning system (at 1 and 2). Powder cassettes (3) in Fig. 2(a) gravity feed powder (Fig. 2(b)) onto the powder bed where it is raked (r) into a new layer which is preheated by the scanning electron beam and then selectively melted as directed by CAD software characterizing the layer building strategy necessary to build the specific product (at 4 in Fig. 2a). Fig. 2(c) shows an optical (light) micrograph illustrating the typical α-phase platelet microstructure for solid, EBM fabricated Ti-6Al-4V products. Essentially the same microstructural features characterize SLM fabricated Ti-6Al-4V solid products. The main difference between EBM and SLM fabricated metal or alloy products involves the lower SLM powder bed temperatures and more rapid cooling or cooling rate. This produces internal strain in some SLM, products which often requires hot isostatic processing (HIPing) to relieve this intrinsic strain, and product warping or other distortions. SLM product surfaces are also often smoother, and for some applications this can be an important feature.

The CAD model which directs the beam in the layer-by-layer product building process implicit in Fig. 2(a) is embedded in a digital file whose data can be constructed either mathematically or from data from CT or MRI scans with layer resolutions of ~0.5 mm [5–7]. Similar strategies are also used to fabricate complex-shaped objects replicated by laser and mechanical metrology (co-ordinate measuring) systems which reconstruct 3D-co-ordinate data, and fabricate poly-
mer components by successively curing the polymer layers using suitable, scanned optical or UV laser beams: the original stereolithography concept patented by Charles Hull in 1986. Indeed, layer-building polymer systems have been available commercially for more than 2 decades, and capable of building complex products of a wide variety of polymeric materials extending several meters in size [7].

While complex structure models, especially open-cellular mesh and foam structure models emulating human bone can be developed from CT scans of such structures, software packages to create such models are commercially available or available in the open literature. For example, irregular, polyhedral foam cell models can be constructed from software developed decades ago by Bakke [8], while Materialize NV (Louver, Belgium) currently provides a suite of software which utilizes various polyhedral and geometric unit cells to produce open-cellular mesh structure models. Fig. 3(a and b) provide examples of these structure models while Fig. 3(c and d) show model constructs emulating bone-like foam cylinders. In both Fig. 3(a) and (b), the model structure is actually an EBM-fabricated Ti-6Al-4V component. In Fig. 3(c) and (d), the inner foam cells are essentially twice as large as the outer foam cells, whereas in the model shown in Fig. 3(b), the foam is represented by a polyhedral ligament construct [8]. In real human bone, such as the femur, the outer or hard cortical bone is a more dense foam (of lower porosity) than the inner, medullary or trabecular (soft) bone regime. In fact, porosity in open-cellular structures is ideally represented as

\[
\text{Porosity} = (1 - \rho / \rho_s) \times 100\%.
\]

where \(\rho\) and \(\rho_s\) are the open-cellular product density and the fully dense (solid) product density, respectively. Of course this equation (Eq. 1) applies not only to regular or irregular cellular constructs, but also to randomly porous products, i.e., those containing random pores or bubbles, including closed cell structures.

Figs. 4 and 5 show the mesh strut and foam ligament microstructures for EBM-fabricated Ti-6Al-4V mesh and foam samples, respectively. In each case, there is considerable surface roughness created by incompletely or unmelted powder particles. In addition, Fig. 4(b) shows the Ti-6Al-4V strut microstructure to be noticeably different \(\alpha'\) martensitic phase platelets in contrast to the larger \(\alpha\)-phase platelets shown for bulk, solid, EBM-fabricated Ti-6Al-4V in Fig. 2(c). This is a result of the much smaller strut (volume) features which cool very rapidly and produce the martensitic platelet microstructure. It is notable and somewhat fortuitous that the \(\alpha'\)-martensite in both the mesh strut structure in Fig. 4(b) and the foam ligament microstructure of Fig. 5 is characteristically harder and stronger than the solid Ti-6Al-4V \(\alpha\)-phase microstructure shown in Fig. 2(c): e.g. 4.6 GPa versus 3.4 GPa hardness, respectively. This is due to the very dense and fine \(\alpha'\) martensite features, where the hardness can be approximated by

\[
H \sim H_0 + K / \sqrt{\lambda},
\]

where \(H_0\) is the intrinsic alloy hardness ideally represented by large, equiaxed \(\alpha\) (hcp) grains, \(K\) is a material constant, and \(\lambda\) is the average phase size or platelet width; roughly 3 microns in the case of the \(\alpha\)-phase platelet microstructure in Fig. 3(c) versus roughly 0.3 microns for the \(\alpha'\) martensite platelets in Fig. 4(b); an order of magnitude difference. It might be recalled that as a “rule of thumb”, hardness (especially Vickers hardness) in many metals and alloys is related to the tensile yield stress, \(\sigma_y\), by [10,11]

\[
H \sim 3\sigma_y,
\]

where \(\sigma_y\) can also be represented generally by the Hall–Petch relationship [10,11]:

\[
\sigma_y = \sigma_0 + K' / \sqrt{D},
\]

where \(\sigma_0\) is an intrinsic yield stress of a single crystal of the metal or alloy, \(K'\) is another material constant, and \(D\) is the polycrystalline metal or alloy grain size (or average diameter in the case of an equiaxed grain structure). However, in the case of equiaxed or other grain structures which contain
intragrain microstructures such as phase platelets, stacking faults, twins, etc., which can have a spacing or dimension, d, an additional strengthening term, \( \kappa' / d \), must be added to Eq. (4); where \( \kappa' \) is a related material constant. For large grain size and d very small, a good approximation would allow \( d \sim D \) in Eq. (4).

2. Applications of solidification fundamentals to powder bed fusion fabrication

One of the more fundamental features of metallurgy involves metal and alloy solidification, including the adjustment of thermodynamic and thermo-kinetic parameters to selectively control phase equilibria and associated microstructures. In the case of bulk metal and alloy solidification, such as in traditional casting, the movement of the liquid/solid interface characterizes the initial/solid-state microstructure. Primary parameters defining the solidification process involve the liquidus-solidus temperature variation with time, or the thermal gradient, G, along with the velocity of the liquidus-solidus interface, or solidification rate, \( R \) [12–15]. Similar phenomena are also characteristic of welding solidification [17]. While the solidification process in powder bed fusion fabrication is conceptually different than bulk metal or alloy solidification, the thermo-kinetic phenomena involving G and R are essentially the same. The difference is that each layer heating and melting/solidification sequence represents a liquidus-solidus transition characterized by the layer thickness, t. In powder bed fusion layer building (Fig. 1c), the laser or electron beam
invests energy (P in watts) in the melting of each layer. The corresponding beam power density is given by

\[ Q = \frac{P}{\nu s't} \text{(/cm}^3\text{)}. \]  \hspace{1cm} (5)

where \( \nu \) is the beam travel speed in cm/s, \( s' \) is the beam scanline spacing in cm, and \( t \) is the layer thickness in cm. The invested energy is also often designated as the linear powder density (J/cm):

\[ Q' = \frac{P}{\nu}. \]  \hspace{1cm} (6)

The layer cooling rate is then expressed by

\[ \text{G.R.} = \frac{A}{e^{\alpha Q'}}. \]  \hspace{1cm} (7)

Where \( a \) and \( A \) are constants.

It might be recalled from casting solidification that dendrite formation, and especially the interdendritic spacing, \( d \), is related to the cooling rate by [12–15]:

\[ d = a/(G-R)^n, \]  \hspace{1cm} (8)

where \( a \) is a constant and \( n \) can vary from \( -\frac{1}{4} \) to \( \frac{1}{2} \). For secondary dendritic arm spacing (\( d = d_2 \)), \( n \) in Eq. (8) is 1/3, and this relationship is referred to as the Kurz-Fisher relationship [15]. In powder bed fusion layer building, \( d \) in Eq. (8) can be replaced by \( \lambda \) in Eq. (2). Correspondingly, this describes the microstructure-hardness variations implicit on comparing Figs. 3(c) and 4(b) for the EBM fabrication of Ti-6Al-4 V solid and mesh structures, respectively. It is therefore apparent that increasing G/R in general decreases the size (and spacing) of the residual microstructure.

It is also generally established from solidification fundamentals that G/R is related to the mode of solidification or the type of microstructure. For example, in casting processes, as G/R decreases, the microstructure changes from columnar (or planar) to cellular or dendritic [12–15]. In powder bed fusion layer building, microstructure similarly changes from columnar to equiaxed grain structure as G/R decreases. In this process,

\[ \frac{G}{R} = \frac{2\pi k(\Delta T)^2}{BP
u}. \]  \hspace{1cm} (9)

where \( k \) is the thermal conductivity of the layer, \( \Delta T \) is the liquidus-solidus temperature difference, \( B \) is a constant, and \( P \) and \( \nu \) are as defined above (Eqs. (5) and (6)). Borrowing from traditional solidification principles [12–15] and the observations of residual microstructures for powder bed fusion processes (e.g. SLM, and EBM), which have been summarized in extensive reviews such as the recent review by Deb Roy, et al. [16], including so-called solidification maps for a number of metals and alloys, a generalized plot of G versus R as shown in Fig. 6 can represent the morphology and size of layer-built residual microstructures. As in traditional metal and alloy solidification processing, post heat treatment and treatment cycles can alter both grain structure and intra-grain microstructures:

\[ D-D_0 = c't'^{1/n} e^{Q''/R'T}. \]  \hspace{1cm} (10)

where \( D_0 \) is the original or starting grain diameter, \( D \) is the thermally processed grain diameter, \( c \) is a constant, \( Q'' \) is the activation energy for grain growth, \( R' \) is the gas constant, and \( T \) is the temperature. Here the holding time, \( t' \), at temperature is a critical parameter. It can be noted that \( D \) in Eq. (10) is the same as \( D \) in Eq. (4), and can be manipulated by both the pro-

Fig. 4 – EBM-fabricated Ti-6Al-4 V rhombic dodecahedral mesh (Fig. 4a) section showing rough, particle-sinter strut surfaces (a) and corresponding, rapidly cooled \( \alpha \)-phase (martensitic) microstructure (b). From Murr [42].

Fig. 5 – EBM-fabricated Ti-6Al-4 V foam (Fig. 4b) section showing rough particle-sinter ligament surfaces as in Fig. 4(a).
cessing powder bed fusion temperature or by post-processing HIP or heat treatment.

It can be observed in Fig. 6 that by choosing appropriate layer-building parameters (pre-heat scan, beam scan speed, beam focus or scan line width or spacing) and layer thickness as well as other beam scan, strategy features; specific and performance optimized microstructures and their associated mechanical properties (hardness: Eqs. (2), (4) and (8)) can be achieved. Of course Fig. 6 simply provides a general overview of achievable microstructures, especially grain features including columnar or equiaxed grains, or mixtures; as well as sub-grain microstructures such as martensitic phase structures. In addition, related phase or precipitation phenomena can also occur either during initial layer solidification and growth as well as post-processing heat treat schedules. CAD selectivity can provide for the development of specific microstructures and related properties in various regions of complex implants by altering the local scan strategies appropriately. Deb Roy et al. [16] have recently reviewed these features for laser and electron beam powder bed fusion processing, while Murr [3,11,18] has illustrated extensive microstructure comparisons for numerous powder bed fusion fabricated metals and alloys.

3. Design strategies for fabricating porous, optimized metal and alloy biomedical implants by powder bed fusion technologies

Metal and alloy implants have been used in a variety of orthopaedic applications for nearly 70 years [19], including stainless steels (especially 316L), Co-Cr alloys, Ti, Ti alloys and Ta. Throughout the first half-century of these biomaterials device applications, issues of biocompatibility were notable concerns. These involved metal leaching toxicities (such as Ni in stainless steels, and Ti), loosening of cemented fixtures due to debonding and micromotion which was often recognized as resulting from bone remodeling because of stress shielding where the bone is “shielded” from its normal load-carrying ability because of load sharing by the implant. This was especially observable in hip arthroplasties where the metal or alloy implant stem replaced the intramedullary or cancellous (soft) inner bone. J. Wolff recognized in the latter part of the 19th Century that bone will adapt to loads (or stress) under which it is placed by “remodeling” to become stronger to resist the load usually by increasing its density. This became known as Wolff’s Law, and for bone implanted with a solid insert, the remodeling involved the reverse response: the bone density was reduced, sometimes by as much as 50 percent 4–8 years following surgery. This response, referred to as mechanotransduction, occurs by special bone molecular signaling.

While implant-induced bone stress shielding is a complex issue, it became identified mainly with the difference in bone stiffness versus that of the implant metal or alloy: a variance in the elastic or Young’s modulus. For example, the elastic (Young’s) modulus for bone can vary from <1 GPa for cancellous (soft or trabecular) bone to roughly 20 GPa for the hardest cortical bone which composes the outer bone regime [21,22]. Structurally, soft or cancellous bone such as the

![Fig. 6 - Generalized temperature gradient (G) versus growth rate (R) plot characteristic for microstructures for laser or electron beam powder-bed-fusion-fabricated metal and alloy products. Note microstructure morphology/feature sizes decrease (D2 < D1) with increasing cooling rate (G-R). Intragrain or intraphase microstructure size or spacing is indicated by d.](image)

intramedullary (marrow) regime in the femur center is composed of large, open-cellular (porous) foam cells in contrast to a less porous and smaller foam cell structure in the outer (cortical) bone structure. Correspondingly, as shown in Eq. (1), the more porous bone is less dense. Even more notable is the fact that nearly 3 decades ago, it was observed that the elastic modulus (E) and density (ρ) of essentially any material were related, and Gibson and Ashby [20] definitively showed that almost universally

$$\frac{E}{E_0} = \left(\frac{\rho}{\rho_0}\right)^2.$$

where E0 and ρ0 are the elastic modulus and density of the solid (fully dense) material. Many researchers and orthopaedic clinicians soon recognized that by creating porosity in metal or alloy implants, especially at their surface regions, could provide not only a more compatible match in the elastic modulus of the implant and bone, but also encourage bone cell in growth and bonding which could ideally suppress loosening of the appliance. For example, in Ti and Ti-alloy implants such as Ti-6Al-4 V, the elastic modulus is ~110 GPa. Correspondingly a femoral stem would normally replace the intramedullary central bone regime and contact the inner cortical bone, where the elastic modulus might be ~10 GPa; an order of magnitude lower modulus than the implant stem.
Recognizing the advantages of developing porous implant strategies, many attempts have been made to manufacture effective, porous materials. While excellent foams such as aluminum have been manufactured for a half century, high temperature metal and alloy foams are nearly impossible to manufacture. Recognizing these shortcomings, medical device/implant manufacturers have developed novel processes to produce components having varying but sometimes limited degrees of porosity. Zimmer-Biomet continues to manufacture implant components from tantalum: Trabecular Metal—a trademark. These components are made by pyrolyzing a thermosetting polymer foam to create a low-density carbon strut skeleton onto which Ta is chemical vapor deposited to create an interconnected scaffold structure. This product is \(~80\%\) dense and has an elastic modulus ranging from \(\sim 2.5\) to \(4\ \text{GPa}\) [23]. Solid Ta has an elastic modulus of \(\sim 190\ \text{GPa}\), a density of \(16.7\ \text{g/cm}^3\), and it melts at \(\sim 3000\ ^\circ\text{C}\). While the high density and temperature for Ta present problems in powder bed fusion fabrication of components, Wauthle, et al. [24] have recently described SLM fabrication of porous Ta implants.

Nearly a decade ago, Murr et al. [11,25,26] made fairly accurate measurements of both Ti-6Al-4V porous mesh and foam EBM-fabricated components along with Co-Cr-Mo alloy mesh and foam components using acoustic resonance, and essentially confirming the Gibson-Ashby relationship in Eq. (11). Representative samplings of this data are reproduced in the \(E/E_0\) versus \(\rho/\rho_0\) plot in Fig. 7. Since Ti-6Al-4V and Co-Cr-Mo alloy (Co- (26–29) Cr-6Mo) represent range of elastic moduli and solid densities: \(110\ \text{GPa}\) and \(4.43\ \text{g/cm}^3\) for Ti-6Al-4V; \(210\ \text{GPa}\) and \(10\ \text{g/cm}^3\) for Co-Cr-Mo; and melting temperatures of \(1625\ ^\circ\text{C}\) and \(1330\ ^\circ\text{C}\) respectively (with both mesh and foam structure components corresponding to the Gibson-Ashby relationship (Eq. 11)), it would appear that Eq. (11) can in fact reasonably represent any porous metal or alloy component, even those with irregular or non-polyhedral structures; including random pores. In addition, CAD modelling to fabricate a wide variety of articulated mesh and foam metal and alloy products can selectively produce any desirable elastic modulus and corresponding density, including variations in the pore (or opening) dimensions as well as mesh strut or foam ligament dimensions; which can independently influence strength, fatigue, and related mechanical properties. Fig. 7 illustrates these corresponding modulus/density design features where for \(E/E_0 = 0.01\) and \(\rho/\rho_0 = 0.18\), the elastic modulus \(E\) and density \(\rho\) for an implant design are \(1.1\ \text{GPa}\) and \(0.8\ \text{g/cm}^3\) for trabecular mesh and \(2.1\ \text{GPa}\) and \(1.8\ \text{g/cm}^3\) for Ti-6Al-4V and Co-Cr-Mo alloy, respectively; the nominal range for trabecular (cancellous) bone.

While CAD model densities can be selectively adjusted to produce desired elastic moduli, other mechanical properties must be adjusted through geometry and microstructure, which, as noted in Fig. 6, is dependent upon the actual AM process parameters. However, to approach these mechanical design features, the specific implant positioning and purpose must be considered. For example, nominal compressive stress requirements for matching those for trabecular (cancellous) bone would be \(\sim 20–30\ \text{MPa}\) [25], while the maximum compressive strength for the hard, cortical (outer) bone is \(\sim 190\ \text{MPa}\) [25]. For porous metal or alloy constructs, the compressive strength, \(\sigma_c\), is expressed by [26,27];

\[
\sigma_c = \sigma_y \cdot C_0 \left(\frac{\rho}{\rho_0}\right)^{1.5}.
\]

(12)

where \(\sigma_y\) is the optimum compressive yield strength, and \(C_0\) is a constant. \(\sigma_y\) (the compression yield stress) is generally not equal to the tensile yield stress, \(\sigma_y\), as represented in Eq. (4); and is usually larger (often \(\sim 50\%)\) larger, especially for non-isotropic alloys such as Ti-6Al-4V which has a hexagonal close-packed (hcp) crystal structure. Indeed, cortical bone has a tensile strength of \(\sim 130\ \text{MPa}\) and a much lower torsion and shear stress: \(\sim 40–50\ \text{MPa}\) [22].

The most common implants and implant components involve hip and knee arthroplasties where not only are compressive stresses dominant, but also these stress situations are intermittent, such as in resting and walking or running, which ideally involve cyclic stress, or compressive fatigue. Both compressive stress are fatigue stress would depend upon the patient’s weight and gate, and taken together can be a complicated issues. Additionally, active and younger recipients of implants may subject implant components to impact.

Fig. 7 – Relative elastic modulus (\(E/E_0\)) versus relative density (\(\rho/\rho_0\)) plots for EBM-fabricated Ti-6Al-4V mesh and foam and Co-Cr-Mo alloy mesh and foam samples. Adapted from Murr [9]. Note arrows correspond to \(E/E_0 = 0.02\) and \(\rho/\rho_0 = 0.18\) along the fitted line whose slope corresponds to the exponent 2 in Eq. (11).
stresses by jumping, adding additional complications which might affect not only the metal or alloy implant but other components such as polymeric inserts in total knee or hip replacement appliances.

It is apparent on comparing Eqs. (11) and (12) that both the relative modulus ($E/E_o$) and relative compressive stress ($σ_c/σ_y$) are related to the relative density ($ρ/ρ_s$). Li, et al. [27] and Zhao, et al. [28] have recently measured the relationship between the compressive stress as well as the fatigue strength versus Young’s modulus ($E$) for Ti-6Al-4 V mesh components fabricated by EBM. This data is reproduced in Fig. 8. It can be noticed in Fig. 8 that the three shaded regimes represent a range of elastic moduli ($E$) from 1 to 10 GPa, with corresponding compressive strengths ranging from $\sim$23 MPa to 190 MPa, and fatigue strengths ranging from $\sim$0.9–18 MPa for Ti-6Al-4 V mesh components. While the more cortical (higher modulus) mesh design would nearly match the upper compressive strength, the fatigue strength would be notably less than the

Fig. 8 − Fatigue and compression strengths versus elastic (Young’s) modulus for EBM-fabricated Ti-6Al-4 V rhombic dodecahedron mesh samples (Fig. 3a). C and T designate corresponding cortical (C) and trabecular (T) bone regimes. Trabecular (soft) bone modulus reference is indicated by arrow at 1 GPa while cortical (hard) bone modulus reference is shown by arrow at 10 GPa. Corresponding fatigue and compression strengths are indicated at the elastic modulus intersection with the respective, shaded data plots. From data in Zhao, et al. [28]. After Murr [29].

Fig. 9 − EBM-fabricated Ti-6Al-4 V rhombic dodecahedron mesh cranial insert in patient specific, CAD-generated, polymer skull model. Form Murr [9].

Fig. 10 − SLM-3D printed, personalized, Ti-6Al-4 V spinal cage (mesh) implant (MM) in 3D (CT-CAD) custom printed, personalized fifth vertebra and adjacent disc polymer vertebral section (PM) model. Adapted from collaborative work by Anatomics RX and RMIT University, Melbourne, Australia in 2015.

20–30 MPa fatigue strengths for cortical bone, especially the femur [22,30] at $\sim$10$^5$ stress cycles to failure. But experimental fatigue tests relate continuous failure stress to number of cycles, and bone is never continuous stressed in any particular mode. For example, of the 126 bones composing the appendicular skeleton, the femur and tibia bone are probably subjected to the more extreme stresses and a variety of intermittent and cyclic axial stress modes, including compression, bending and torsion. However, unlike more traditional materials such as the metal or alloy implant materials, bone will not accumulate fracture features such as cracks because sig-
naling molecules in bone structure continuously repairs such damage (by signaling bone cell regeneration), rendering bone self-healing [31,32]. In addition, porous metal or alloy implants can ideally be considered scaffolds because it is now clinically recognized that if there is efficient bone cell ingrowth, not only is the implant stabilized to eliminate any loosening from the intrinsic bone but the bone cell ingrowth (or osseointegration) and osteoinduction within the implant porosity can progressively assume a role in responding to complex stress regimes. Strategies are also increasingly optimistic for extending intrinsic bone vascular structures into the implant or inducing additional vascular structures (blood vessels) within the open-cellular implant collagen matrix; rendering the implant a “living” scaffold [9,33–36]. In this respect, it has been shown that the irregular surface features of powder-bed fusion fabricated, open-cellular (mesh and foam) structures (Figs. 4 and 5) optimize bone cell attachment migration; creating necessary collagen/hydroxyapatite matrix configurations which facilitate osteoinduction and angiogenesis.

4. Examples of porous, powder-bed fabricated implants

4.1. Custom-built cranial/maxillofacial/implants and surgical, pre-operative models

For nearly 3 decades, acrylate biomodels using liquid resins have been fabricated using rapid prototyping/stereolithography systems to create preoperative guides in surgical planning involving both soft tissue prototypes such as blood vessels and other organs, as well as more rigid polymers for various skeletal models [7,37–40]. Skeletal, anatomical models facilitate the actual placement of personalized and off-the-shelf (commercial) implants. While, as noted previously, there are millions of knee and hip arthroplasties annually around the world, there are also similarly increasing implant surgeries for many other craniofacial, cranio-maxillofacial procedures: cranial plate inserts and dental facial deformities that effect both children and adults, where, the design and manufacturing of patient-specific/customized implants are essential.

Fig. 9 illustrates an example of both a pre-operative polymer skull model and a customized porous (mesh) cranial Ti-6Al-4 V insert. Since the nominal skull bone Young’s modulus can vary from roughly 1–7 GPa for children and adults respectively [12,41], the design features outlined in Figs. 7 and 8 ideally represent the requisite fabrication strategies for optimized implants. Porosity in cranial plate implants (Fig. 9) is especially important in providing bone cell ingrowth and fastening to the skull.

4.2. 3D printed, open-cellular structure spinal implants

Like personalized, cranio-maxillofacial reconstruction implant requirements, spinal implants have been facilitated by 3D printing spinal inserts utilizing customized, 3D
polyurethane and other polymer biomodels of patient spines which, like cranial models represented in Fig. 9, can facilitate successful spine surgery management and planning. An example of a Ti-6Al-4V 3D-printed spinal cage insert is illustrated in Fig. 10, and represents a wide range of personalized spinal inserts manufactured primarily by SLM either in point-of-care, in-hospital 3D printing facilities, or by familiar, commercial biomedical implant device manufacturers (e.g. Stryker, Zimmer-BioMet, etc.) where FDA approval has been received for a host of innovative spinal implants. Fig. 11 shows an example of such spinal cage implants following experimental surgery in 2015. Xiu et al. [43] in China were among the first in the world to develop open-cellular spinal implants (circa 2009) using EBM fabrication.

4.3. 3D implant designs for total hip arthroplasty

Fig. 12 illustrates the components for total hip arthroplasty involving an acetabular cup which fastens the appliance to the pelvis, a special polymer ball joint attached to the top of the femoral stem, and a polymer liner within the cup, and invisible in the X-ray image. The acetabular cup is manufactured with porous (often mesh) structure and often is screwed to the pelvic bone which provides for attachment of the appliance while it is stabilized by bone cell ingrowth. The polymer cup insert and ball head are usually highly cross-linked, high molecular weight polyethylene, sometimes coated with a lubricious layer of other polymeric materials such as poly(2-methacryloxyethyl) phosphorocholine (PMPC) which mimics
articulating cartilage, and can eliminate contact wear by a factor of 10².

The femoral stem insert shown in Fig. 12 often contains various porous, open-cellular structure features along the stem exterior. In addition, long rods implanted in severely broken femur bones can also benefit from porous, foam structure designs illustrated in the models provided in Fig. 3(c) and (d). Commercial manufacturers have also been providing 3D printed acetabular cup components for nearly 10 years [44] with FDA approval (in the US), and many tens of thousands have been surgically implanted over the past decade worldwide. While appliance loosening is dramatically reduced in implant designs involving open-cellular structures, polymer liner erosion damage can occur due to galling of the polymer surfaces and the intrusion of arthritic deposits in some patients. Infection is also an issue in 1–2 percent of patients, and while effective antibiotics are often mixed with cements to secure the implant component, the open-cellular structure can allow antibiotics to be invested in these regions before surgical insertion [45] to provide an efficient and long term infection deterrent.

Because total hip implants experience a complex variation of multiple stress modes and often cyclic high stresses, especially in a wide range of patient weights and body motions in walking (gate variations), personalized, 3D appliance fabrication is particularly useful in assuring optimum mechanical compatibility designs (Figs. 7 and 8). Xia, et al. [46] have recently reviewed roughly 50 papers written since 2015 regarding clinical applications of 3D printing in hip arthroplasty devices, including patient-specific instrument fabrication in hospital point-of-care, 3D printing centers and surgical service centers.

4.4 Total knee arthroplasty and open-cellular implant components

Fig. 13 shows a total knee arthroplasty with a femoral Co-Cr-Mo component fabricated by EBM, and having a mesh structure for femoral bone attachment and a semi-polished contact surface. The meniscus or articular cartilage replacement of highly cross-linked, high molecular weight polyethylene insert is not visible in the X-ray image. Fig. 14 shows the total knee arthroplasty appliance in detail, where the highly-cross linked HMWPE insert is shown on the tibial table (HCPE in Fig. 14). In this rendering, both the femoral and tibial components have porous bone stems to facilitate implant placement and fixation. Fig. 15 illustrates some conceptual and design details for stem-component fabrication of Ti-6Al-4V knee appliances (Fig. 14), including open-cellular features strengthened by a solid, internal stem rod (Fig. 15d). These features (Fig. 15(c) and (d)) can also include the foam models or variations shown in Fig. 3(c and d).

While the use of Co-Cr-Mo tibial knee implant components by commercial biomedical device manufacturers has somewhat routine, Ti-6Al-4V components can also be utilized with some advantage. For example, cast Co-Cr-Mo tibial knee inserts have required HIPing using ASTM F-75 standard (1200 °C, 4 h in 0.1 GPa Ar; homogenization at 1200 °C, 4 h in 0.1 GPa Ar; quench to 760 °C in 8 min.) which dissolves Cr₂₃C₆ carbide precipitates, and eliminates corrosion sensitization due to Cr depletion by precipitate formation [47,48]. This HIP treatment produces a relatively equiaxed grain structure often with a high density of stacking faults in the fcc Co-Cr grain matrix. This microstructure produces a residual, Vickers microindentation hardness ranging from ~3.2–3.6 GPa. Similar, but somewhat unique Cr₂₃C₆ carbide production can occur for EBM fabrication of Co-Cr-Mo (Co-26 Cr-6 Mo (0.03–0.5 C)) alloy biomedical products as illustrated in Fig. 16 which shows an optical micrograph composition section view of columnar grains and regular columns of Cr₂₃C₆ carbides which result from particular powder bed fusion processing parameters implicit in Fig. 6. The TEM insert in Fig. 16 shows the cubic (a = 1.6 nm) carbides along with stacking faults. Fig. 17 shows
Fig. 15 – Features of powder-bed fusion (3D) fabricated porous implant stems. (a) Upper section view of femur showing soft, intramedullary (trabecular bone) regime into which porous implant stems are inserted (Fig. 14). (b–d) Show mesh CAD models while (e) shows an EBM-fabricated, Ti-6Al-4 V tapered mesh stem having a solid core as represented in (d). (e) Is specific to the tibial insert appliance (TS) in Fig. 14.

5. **Complex skeletal reconstruction implants: personalization of implant fabrication through hospital point-of-care, 3D printing centers**

It is apparent that there are many advantages of powder bed fusion 3D printing of metal and alloy implants. Designed porosity in various biomedical devices can eliminate loosening through bone cell ingrowth, and osseointegration and osteoinduction within the porous implant can render it a simple bone scaffold. Vascularization could render the implant “living” or an extension of the living bone structure [9].
More importantly, and as illustrated in the examples of Figs. 10-17, CT scanning and embedding personalized layer data in processing CAD software can allow for a wide range of personalized, patient-specific implant fabrication, often at low cost. In addition, many patients requiring total hip or knee arthroplasties may not require specially fabricated devices, and can be reasonably accommodated by a range of standard component sizes, thereby allowing for the manufacture of a large number of optimized biomedical appliances at low cost using 3D printing by powder bed fusion.

While personalization of total knee and hip arthroplasties by powder bed fusion technologies can be somewhat routine, more complex implants and 3D-produced reconstructions involving cranial-dental or related cranial-maxillofacial reconstructions become very specialized. Similarly, many other complex revision or reconstruction surgeries requiring metal implant fabrication along with appropriate, polymeric surgical planning models are truly patient-specific and one-of-a-kind. Fig. 18 illustrates an example of such specialty, skeletal 3D printing involving reconstructing a diseased hip section which has been inserted in the surgical planning model for the patient. In this medical protocol, it is apparent that the patient lost the hip connection, and was unable to walk. A corresponding hip acetabular cup could also be fabricated and a total hip arthroplasty connected to the pelvic insert as shown by the arrow in Fig. 18 [34,42]. There have been many such surgical successes of personalized 3D-printed implants and porous metal skeletal replacement devices in many parts of the world in the last decade, especially China, which began the development of point-or-care, in hospital 3D printing centers with commercial powder bed fusion EBM and SLM systems to fabricate patient-specific biomedical devices, such as that shown as an example in Fig. 18, around 2009 [43,49].

As of 2019, there were hundreds of so-called point-of-care 3D printing centers or hubs in hospitals world-wide; some in orthopedic units and other serving as service units for a variety of surgical departments, including developing 3D printed surgical planning models. These centers utilize CT scan systems to create initial layer or slice data which is then embedded in digital imaging or CAD software which drives the 3D fabrication process. Software such as Materialise Interactive Medical Image Control System (MIMICS), or 3D Slicer are common commercial packages for converting CT scans into CAD models [46]. Materialise (MIMICS) technology was employed worldwide in hospital 3D printing centers in 2019 as follows: US/Canada, 113; UK, 31; EU, 48; South America (mainly Brazil), 9; China, 24; Japan, 34; Australia, 9. While this is only one example, there are many other similar facilities globally. Materialise acquired a 75% stake in Engimplan, a Brazil-based medical device manufacturer since 1992, to develop personalized (especially orthopedic) implants in Brazil.
Because of the tens-of-thousands of patient-specific and individualized 3D implant fabrications and reconstruction surgeries performed worldwide, most of these clinical successes are becoming routine medical practice, and are unreported in the literature. In addition, Ti-6Al-4V powder has become the 3D powder bed fusion fabrication precursor of choice not only for a wide variety of commercially manufactured appliances, including personalized devices, but for the vast majority of point-of-care 3D printing centers or service laboratories. As a consequence, there are few incentives to introduce new alloy powders, even those having advantages in producing products having some design advantages regarding mechanical compatibility as implicit in Figs. 7 and 8.

6. Summary and conclusions

Aside from key references describing fundamental and historical issues of significance in developing an understanding of metal and alloy implant development and the contributions of 3D printing to modern implant fabrication, this overview has focused primarily on contemporary and especially customized, porous implants [3,9,18,28,29,34,37,38,42,46]. These references describe implant design, development, fabrication, applications, and examples of clinical outcomes (or performance) over the past 4 years. In addition, Tofail et al. [50] and Dutta et al. [51] have recently provided overviews of additive manufacturing challenges and opportunities, many related to biomedical applications of metals and alloys, while more timely reviews of customized 3D printed metal and alloy implants and implant components have been written by Ahangar, et al. [52] and Dall’Avà, et al. [53], respectively. Liu, et al. [54] have also recently reviewed aspects of Ti-6Al-4V AM, which, as evident in the present overview, is the global choice for implant fabrication, especially using commercial, laser and electron beam power bed fusion systems. The progressive and continuing use of Ti-6Al-4V for a broad variety of implants such as those illustrated in the examples in Figs. 9–15 and 18 continues to assure optimum biocompatibility, safety, and post-surgical quality of life. Indeed, although there have been some instances of titanium and vanadium toxicity, recent research concludes these issues are rare [57].

While considerable research on the development of a wide range of Ti-alloy systems has been conducted over the past 2 decades [55], including alloys such as Ti-24Nb-4Zr-7.9 Sn, having a Young’s modulus less than half that for Ti-6Al-4V [56], there is little incentive, either biomedical or economical, to adopt these alloys for implant fabrication, as noted above.

As illustrated in this overview, the application of solidification principles defining powder bed fusion processing parameters in terms of the thermal gradient (G), solidification rate (R) and cooling rate (G-R) allows AM metal and alloy microstructures to be fairly well defined by constructing a solidification processing map as illustrated in Fig. 6. For Ti-6Al-4V, the principal microstructures, consisting of various α-phase dimensions or α′-(martensite) phase dimensions (Figs. 2(c) and 4(b)), allow the hardness (Vickers) to be adjusted over a range of roughly 3.5–4.5 GPa, as well as the corresponding manipulation of strength and ductility. While this can assure a high degree of biomechanical compatibility with bone for porous, open-cellular implant designs, optimal bone ingrowth not only assures effective fastening of the implant but also renders the implant a bone scaffold, or ideal bone replacement.

It is also notable, as emphasized in this overview, that within the past decade, 3D printing of metal and alloy implants using powder bed fusion technologies, especially with commercial laser and electron beam systems, has rapidly emerged worldwide. More than two hundred point-of-care 3D printing centers existed in hospital surgical departments around the world employing multidetector CT scanning systems and specialized software to create CAD models to fabricate personalized, porous, and often complex skeletal replacement implants ranging from cranial-maxillofacial reconstructions, skull plates (Fig. 9), spinal inserts (Figs. 10 and 11), total hip and knee implants (Figs. 12–15) and a host of other complex, patient-specific implants (Fig. 18). While customized, patient-specific implant fabrication continues to proliferate, commercial biomedical device companies are also incorporating powder bed fusion
technologies into their manufacturing arena, allowing cost reductions for a range of standard orthopedic and related appliances, especially using Ti-6Al-4V powders, which are increasingly available worldwide. And while metallurgical issues are now well established for Ti-6Al-4V biomedical device processing and performance, continuing research and development of customized prostheses following amputation surgeries and other exoskeletal reconstructions is especially promising.

Conflicts of interest

The author declares no conflicts of interest.

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