University of Louisville

ThinkIR: The University of Louisville's Institutional Repository

Electronic Theses and Dissertations

12-2017

Design, analysis, and application of a cellular material/structure model for metal based additive manufacturing process.

Shanshan Zhang
University of Louisville

Follow this and additional works at: <https://ir.library.louisville.edu/etd>

Part of the [Computer-Aided Engineering and Design Commons](#), and the [Other Materials Science and Engineering Commons](#)

Recommended Citation

Zhang, Shanshan, "Design, analysis, and application of a cellular material/structure model for metal based additive manufacturing process." (2017). *Electronic Theses and Dissertations*. Paper 2832.
<https://doi.org/10.18297/etd/2832>

This Doctoral Dissertation is brought to you for free and open access by ThinkIR: The University of Louisville's Institutional Repository. It has been accepted for inclusion in Electronic Theses and Dissertations by an authorized administrator of ThinkIR: The University of Louisville's Institutional Repository. This title appears here courtesy of the author, who has retained all other copyrights. For more information, please contact thinkir@louisville.edu.

DESIGN, ANALYSIS, AND APPLICATION OF A CELLULAR
MATERIAL/STRUCTURE MODEL FOR METAL BASED ADDITIVE
MANUFACTURING PROCESS

By

Shanshan Zhang

B.S., Northeastern University, China, 2009

M.S., Northeastern University, China, 2011

A Dissertation

Submitted to the faculty of the

J. B. Speed School of Engineering of the University of Louisville

In Partial Fulfillment of the Requirements for the Degree of

Doctor of Philosophy

In Industrial Engineering

Department of Industrial Engineering

University of Louisville

Louisville, Kentucky

December 2017

Copyright 2017 by Shanshan Zhang

All rights reserved

DESIGN, ANALYSIS AND APPLICATION OF A CELLULAR
MATERIAL/STRUCTURE MODEL FOR METAL BASED ADDITIVE
MANUFACTURING PROCESS

By

Shanshan Zhang

A Dissertation Approved on
November 27, 2017

by the following Dissertation Committee:

Dissertation Director: Dr. Li Yang

Dissertation Co-director: Dr. Brent E. Stucker

Dr. Gail W. DePuy

Dr. Thomas L. Starr

ACKNOWLEDGMENTS

First and foremost, I would like to extend my most sincere gratitude to my advisor Dr. Li Yang. It has been a great honor to work with him. With his unwavering commitment and insightful guidance, he instigated in me his excitement for research and knowledge in the area of additive manufacturing.

I am also very grateful to my previous advisor Dr. Brent Stucker. for giving me this amazing opportunity and for always being there with kind and insatiable generosity.

I want to thank other committee members, Dr. Thomas Starr and Dr. Gail DePuy, for your advice and support during this project. I have great respect and admiration for each one of you and truly appreciate your participation in this project.

To my friends and colleagues, I thank you for your friendship, help and moral support. Your presence was a great source of encouragement during this time.

Most importantly, I will always be eternally grateful to my family for their incessant support, love and encouragement throughout my academic career. I would not be so far without them and for that I will always be indebted.

DEDICATION

Dedicated to the memory of my father (9.15.1954 – 10.14.2017), who regarded me as his little princess, always supported me, whatever path I took. Wish him R.I.P. in heaven.

and

To the rest of my family

ABSTRACT

DESIGN, ANALYSIS AND APPLICATION OF A CELLULAR MATERIAL/STRUCTURE MODEL FOR METAL BASED ADDITIVE MANUFACTURING PROCESS

Shanshan Zhang

November 27, 2017

Powder bed fusion additive manufacturing (PBF-AM) has been broadly utilized to fabricate lightweight cellular structures, which have promising potentials in many engineering applications such as biomedical prosthesis, aerospace, and architectural structures due to their high performance-to-weight ratios and unique property tailorabilities. To date, there is still a lack of adequate understanding of how the cellular materials are influenced by both the geometry designs and process parameters, which significantly hinders the effective design of cellular structures fabricated by PBF-AM for critical applications. This study aims to demonstrate a cellular structure design methodology that integrates geometrical design and process-material property designs. Utilizing both analytical modeling and empirical modeling, this research aims to significantly improve the design flexibility and robustness of the metal PBF-AM cellular structures.

Experimental designs were carried out to establish the process-material property knowledge for the Ti6Al4V using the EOS M270 laser powder bed fusion (LPBF) system. Using optical microscopy, scanning electron microscopy (SEM), micro-tensile testing and micro-hardness testing, the characteristics of thin struts with different strut dimensions and orientations under different process conditions were characterized and compared with those of the bulk materials from LPBF. The results clearly indicated significant effects of strut geometries (dimension and orientation angle) on their qualities. Struts with large orientation angle (i.e. more aligned to the build direction) exhibit lower process robustness and are sensitive to process parameters. Due to the resolution limitation of the LPBF process, the geometrical accuracies of the struts increases drastically when the designed dimension is smaller than 0.2 mm, with minimum achievable dimensions around 0.2 mm for the process parameters investigated in this study. On the other hand, the struts with smaller dimensions tend to exhibit higher mechanical properties, which might be associated with the smaller grain size and lower porosities. There also does not appear to be a single set of process parameters that would result in minimum porosities for struts with various dimensions.

Adopting center-joint based unit cell design, an analytical model for unit cells with designable numbers of struts was established in the attempt to enhance the designability of the geometries. Timoshenko beam theory was verified to be the most accurate modeling method, although for large strut orientations Euler-Bernoulli beam theory might suffice. The analytical model for elastic modulus was verified by finite element analysis (FEA) and experiments. Additionally, predictable size effect was modeled for the cellular structures with the center joint connectivity of 8 (octahedral).

Employing the material property database for the cellular designs, the integrated material performance/structural geometry model was demonstrated for both single struts and small cellular patterns. It was shown that the integrated model is able to provide improved prediction to the properties of cellular structures.

TABLE OF CONTENTS

	PAGE
ACKNOWLEDGMENTS	III
DEDICATION	IV
ABSTRACT	V
LIST OF TABLES	XIII
LIST OF FIGURES	XVI
CHAPTER 1 INTRODUCTION AND LITERATURE REVIEW	1
1.1 Cellular structures	1
1.1.1 Introduction of cellular structures	1
1.1.2 Manufacturing of cellular structures	3
1.1.3 Geometry design for cellular structures	9
1.1.4 Assumption in modeling analysis	14
1.2 Additive manufacturing	16
1.2.1 Laser Powder Bed Fusion (LPBF)	16
1.2.2 Capabilities and limitations	19
1.2.3 Characteristics of LPBF-produced parts	21
1.2.4 Methodologies of material characterization	25
1.2.4.1 Macro-scale characterization of cellular structures	25

1.2.4.2 Micro-scale material characterization	28
1.3 Motivation and objectives	32
CHAPTER 2 MATERIAL PROPERTY CHARACTERISTICS IN LASER POWDER	
BED FUSION (LPBF) PROCESS.....	35
2.1 Introduction.....	35
2.2 Material and Laser powder bed fusion (LPBF) process	36
2.3 Exploration of process parameters in EOSINT M 270	39
2.4 Design of Experiments.....	42
2.5 Results and discussion	45
2.5.1 Single track deposits characteristics	45
2.5.2 Top surface of bulk specimens	49
2.5.3 Porosity	50
2.5.4 Microstructure.....	55
2.6 Mechanical properties.....	58
2.6.1 Tensile property	59
2.6.2 Microhardness.....	68
2.7 Conclusion	70
CHAPTER 3 COUPLED EFFECTS OF PROCESS PARAMETER AND SPECIMEN	
SIZE OF TI-6AL-4V THIN STRUCTURES FABRICATED BY LPBF	72
3.1 Introduction.....	72
3.2 Experimental design	75
3.2.1 Material	75
3.2.2 AM process	76

3.2.3 Geometry accuracy and microstructure observations and mechanical properties measurements.....	81
3.3 Sample fabricatability	81
3.3.1 Fabricatability rate	83
3.3.2 Geometry accuracy and quality evaluation	86
3.4 Material characterization	98
3.4.1 Porosity	98
3.4.2 Microstructure	101
3.4.3 Micro tensile test.....	105
3.4.4 Microhardness and ANOVA.....	123
3.5 Qualitative evaluation of process-dependent properties	128
3.5.1 Porosity	128
5.3.4.2 Grain size	132
5.3.4.3 Microhardness.....	135
3.6 Conclusion	136
CHAPTER 4 GEOMETRY FEA SIMULATION.....	139
4.1 Introduction.....	139
4.2 Structural design and simplification	141
4.3 Modeling of mechanical behavior	146
4.3.1 Deformation of the slender struts ($d/l < 0.2$).....	149
4.3.2 Deformation of the short struts ($d/l \geq 0.2$).....	150
4.3.3 Deformation under three types of force condition	152
4.3.4 Further discussions.....	155

4.4 Finite Element Simulation of the joint-based model	157
4.4.1 FEA on the effective Young's modulus.....	159
4.4.2 Size effect.....	162
4.5 Conclusion	168
CHAPTER 5 INTEGRATED MATERIAL/STRUCTURAL DESIGN	170
5.1 Introduction.....	170
5.2 Methodology.....	171
5.2.1 Ti-6Al-4V fabrication with LPBF	171
5.2.2 Prediction of the effective mechanical properties.....	171
5.2.3 Input of initial CAD design and output of fabrication	173
5.3 Results and discussion	175
5.3.1 Predictability of structural properties.....	175
5.3.2 Estimation of mechanical properties.....	178
5.3.3 Experimental verification.....	180
5.4 Conclusion	188
CHAPTER 6 CONCLUSION AND FUTURE WORK.....	190
6.1 Conclusion	190
6.1.1 The effect of process parameter and build orientation on macroscale Ti-6Al-4V specimens.....	190
6.1.2 The effects of process parameter, geometry parameter and build orientation on thin feature Ti-6Al-4V structures.....	191
6.1.3 Analytical and FEA prediction on the thin feature cellular structures.....	194
6.1.4 Prediction of the integrated material/structural model.....	194

6.2 Future work and perspective.....	195
REFERENCE.....	196
APPENDICES	221
Appendix A: Formation success rate summary	221
Appendix B: Analysis on the formation quality of thin struts.....	233
B.1 Analysis on 15° struts.....	233
B.2 Analysis on 45° struts.....	237
B.3 Analysis on 75° struts.....	242
Appendix C: Tensile stress-time curves for thin strut specimens.....	247
CURRICULUM VITAE.....	256

LIST OF TABLES

	PAGE
Table 2.1 Composition of EOS Ti-6Al-4V [120]	37
Table 2.2 Factors and levels of design of experiment	42
Table 2.3 Results of mechanical properties of Ti-6Al-4V bulk samples under different thermal formation conditions [63, 67, 72, 78, 144-150]	58
Table 2.4 Tensile properties of as-built Ti-6Al-4V fabricated by LPBF process.....	60
Table 2.5 Tensile testing results for 15°-built coupons	62
Table 2.6 Tensile testing results for 45°-built coupons	63
Table 2.7 Tensile testing results for 75°-built coupons	64
Table 3.1 Design of experiment for thin feature struts	82
Table 3.2 ANOVA on geometry accuracy of thin-feature struts	94
Table 3.3 ANOVA on geometry accuracy of struts with nominal diameter below 0.2 mm	94
Table 3.4 ANOVA on geometry accuracy of struts with various process parameters	95
Table 3.5 Strain at initial yield and corresponding information of as-built Ti-6Al-4V produced by LPBF	108
Table 3.6 Ultimate tensile strength (UTS) for different process parameters vs. nominal diameters (15°).....	113

Table 3.7 Young's modulus for different process parameters vs. nominal diameters (15°)	114
Table 3.8 Ultimate tensile strength (UTS) for different process parameters vs. nominal diameters (45°)	117
Table 3.9 Ultimate tensile strength (UTS) for different process parameters vs. nominal diameters (75°)	118
Table 3.10 Young's modulus for different process parameters vs. nominal diameters (45°)	119
Table 3.11 Young's modulus for different process parameters vs. nominal diameters (75°)	120
Table 3.12 ANOVA of full factorial experiment (Micro hardness vs. angle and process)	126
Table 3.13 ANOVA of full factorial experiment (Micro hardness vs. ED and specimen size)	127
Table 3.14 Fitting equations and coefficient values for bulk porosity	129
Table 3.15 The coefficients of the transferred fitting equations	130
Table 3.15 Coefficients of the fitting curves in Fig. 3.38	131
Table 3.16 Fitting equations and coefficient values for bulk grain width	133
Table 3.17 Coefficients of the fitting curves in Fig. 3.41	134
Table 3.18 Fitting function and coefficients for microhardness	136
Table 4.1 Geometry factors and levels of octahedral unit cell	160
Table 4.2 ANOVA of full factorial design (Effective modulus vs. angle and aspect ratio)	162
Table 4.3 Design for size effect evaluation	163

Table 4.4 Effective Young's modulus in FEA.....	165
Table 4.5 Fitting equations and coefficients for size effect	167
Table 5.1 Geometry design of unit cells in Case I.....	174
Table 5.2 Geometry design of tilted-fabricated unit cells in Case II.....	175
Table 5.3 Compression results in Case I.....	183
Table 5.4 Compression test results in Case II.....	186
Table 5.5 Design parameters of the octahedral sandwich structure.....	186
Table 5.6 Compression results for the octahedral cellular structures	188

LIST OF FIGURES

	PAGE
Fig. 1.1 Natural cellular structures: Coral (left); Honeycomb (middle); Hip stem insertion (right)	3
Fig. 1.2 Melt gas injection casting to manufacture Al foam [9]	4
Fig. 1.3 Gas releasing agent casting to manufacture Al foam [10, 14]	5
Fig. 1.4 An example of an investment cast 3D Kagome sandwich panel [19]	6
Fig. 1.5 An illustration of the folding operation used to fabricate the single layer pyramidal truss structures [26]	7
Fig. 1.6 Schematic of the process for the expanded pyramidal lattice truss structure [28]	7
Fig. 1.7 An example of a 3D Kagome structure made by the wire-weaving method [6] ...	8
Fig. 1.8 The unit cell of the hexagonal honeycomb	11
Fig. 1.9 Pin-jointed frameworks, (a) $b = 4, j = 4, M < 0$, (b) $b = 5, j = 4, M = 0$, (c) $b = 6, j = 4, M > 0$	12
Fig. 1.10 Stress-strain characteristics of different deformation mechanisms [33]	13
Fig. 1.11 Polyhedral unit cells [5]	14
Fig. 1.12 Schematic of LPBF process [42]	18
Fig. 1.13 Schematic of 3D phase diagram on Titanium alloys [69]	22
Fig. 1.14 Schematic of cooling diagram for Ti-4Al-4V [70, 71]	22
Fig. 1.15 Relationship between build rate, power, and feature definition	24

Fig. 1.16 (a) Test frame for experimental dynamic impact study and (b) schematic of illustration of the mass assembly with upper fixed support.....	26
Fig. 1.18 Schematic of the 4-point bending test on the sandwich beam and loading rollers [101].....	27
Fig. 1.19 Tensile testing using an electrostatic force grip	29
Fig. 1.20 Schematic illustration of the axisymmetric bending test fixture [109]	30
Fig. 1.21 Schematic illustration of the microbeam bending test.....	31
Fig. 1.22 Schematic illustration of radial cracking at a Vickers indentation [114]	32
Fig. 2.1 Distribution and morphology of EOS Ti-6Al-4V powder.....	37
Fig. 2.2 Particle size analysis of Ti-6Al-4V pre-alloyed powder	38
Fig. 2.3 Schematic of LPBF process [42]	39
Fig. 2.4 Parameter map in EOSINT M270 laser powder bed fusion (LPBF) system.....	40
Fig. 2.5 Two basic overlapping scan strategies by layer	41
Fig. 2.6 Illustration of laser beam scanning for the thin feature	43
Fig. 2.7 Single track deposits on the Ti-6Al-4V plate	43
Fig. 2.8 Fabrication of Ti-6Al-4V specimens, (a) Cubic specimens; (b) Thin struts specimens at various tilt angles and dimensions.....	44
Fig. 2.9 Illustration of dimension and section of cubical specimen (left) and thin strut (right)	44
Fig. 2.10 Top surface morphology of single track deposits under SEM	46
Fig. 2.11 Melt pool of single track deposits produced by LPBF and illustration of the measurement	47
Fig. 2.12 Variation of depth and width of melt pools on process parameter sets.....	49
Fig. 2.13 Solidified top surfaces of bulk specimens under varying process parameters ..	50

Fig. 2.14 Optical micrograph of lateral sections under varying process parameters	51
Fig. 2.15 Porosity distribution of LPBF produced Ti-6Al-4V bulk specimens.....	52
Fig. 2.16 Variation of porosity with energy density changed.....	53
Fig. 2.17 Process window for LPW Ti-6Al-4V powder in LPBF process	54
Fig. 2.18 Schematic illustration of microstructures occurring in Ti-6Al-4V after fast cooling quenching.....	55
Fig. 2.19 Vertical cross-sectional microstructure of LPBF produced Ti-6Al-4V samples, (a) 80 W – 400 mm/s; (b) 100 W – 400 mm/s.....	56
Fig. 2.20 Columnar grains measurement in Ti-6AL-4V bulk sample	56
Fig. 2.21 Grain size distribution by varying process parameters.....	57
Fig. 2.22 (a) Illustration of the setup of tensile test; (b) Rupture of a tensile coupon	61
Fig. 2.23 Stress-strain curves of 15°-built tensile specimens	62
Fig. 2.24 Stress-strain curves of 45°-built tensile specimens	63
Fig. 2.25 Stress-strain curves of 75°-built tensile specimens	64
Fig. 2.26 Schematic of transgranular fracture in metal [156].....	66
Fig. 2.27 Fractographs of 15° tensile specimens under SEM, (a) Process parameter of 80 W and 400 mm/s (Zone I); (b) Magnified view of (a); (c) Process parameter of 100 W and 300 mm/s (Zone II); (d) Magnified view of (c); (e) Process parameter of 60 W and 500 mm/s (Zone III); (f) Magnified view of (e).	67
Fig. 2.28 Micro Vickers hardness of as-built bulk Ti-6Al-4V specimens with various energy density	69
Fig 3.1 EOS Ti-6Al-4V powder morphology and particle size distribution.....	76
Fig. 3.2 Schematic diagram of the beam offset [57].....	77
Fig. 3.3 Illustration of exposure points	78

Fig. 3.4 Illustration of exposure to small triangle tip.....	78
Fig. 3.5 Rule of the EF value taken	79
Fig. 3.6 Illustration of radius factor	80
Fig. 3.7 Thin feature struts produced by LPBF using EOS M 270.....	83
Fig. 3.8 Fabrication of 0.5 mm thin struts at contour exposure	83
Fig. 3.9 Fabrication of thin struts in LPBF process at various build orientations with the exposure type of (a) contour and (b) contour/hatch combination.....	84
Fig. 3.10 Statistical fabricability rate of the thin struts produced in typical process parameter sets.....	85
Fig. 3.11 Fabrication of 15° and 75° struts	86
Fig. 3.12 SEM micrographs of surface topologies of thin struts under various build orientation, (a) 15° (80W, 400 mm/s); (b) 45° (80W, 400 mm/s); (c) 75° (80W, 400 mm/s); (d) 15° (100W, 500 mm/s); (e) 45° (100W, 500 mm/s) and (f) 75° (100W, 500 mm/s)..	87
Fig. 3.13 SEM graphs of the thin strut tips with various nominal dimension under contour and hatch exposure, (a) 0.6 mm; (b) 0.3 mm; (c) 0.1mm and (d) 0.05 mm (Magnification: (a) 150 ×; (b-d) 350 ×)	89
Fig. 3.14 SEM graphs of the thin strut tips with various nominal dimension under contour exposure, (a) 0.4 mm; (b) 0.3 mm and (c) 0.2mm (Magnification: 150 ×)	90
Fig. 3.15 Layer collapse in the thin strut	91
Fig. 3.16 Measurement of the thin strut dimensions.....	92
Fig. 3.17 Measured dimension, absolute error and relative error of 15° struts under laser power of 100 W and scan speed of 500 mm/s	93
Fig. 3.18 Hollow interiors in various oriented thin struts, (a) 15°; (b) 45; (c) 75°	96

Fig. 3.19 Cross-sectional thin struts using contour and hatch combined exposure, (a) 0.4 mm; (b) 0.3 mm; (c) 0.2 mm and (d) 0.1 mm	97
Fig. 3.20 Cross-sectional thin struts using contour exposure, (a) 0.4 mm; (b) 0.3 mm; (c) 0.2 mm and (d) 0.1 mm.....	98
Fig. 3.21 Defects in the thin feature struts, (a) 80W and 400 mm/s, 75°; (b) 80W and 400 mm/s, 45° and (c) 100 W and 500 mm/s, 45°	99
Fig. 3.22 Porosity of thin struts.....	100
Fig. 3.23 Microstructure of thin struts and illustration of their sections, (a) 0.6 mm struts using parameters of 100 W and 500 mm/s; (b) 0.6 mm struts using parameters of 80 W and 400 mm/s.....	101
Fig. 3.24 Tilt grain phenomenon, (a) 0.3 mm strut using parameters of 60 W and 300 mm/s; (b) 0.3 mm strut using parameters of 100 W and 500 mm/s; (c) 0.3 mm strut using parameters of 80 W and 500 mm/s; (d) 0.6 mm strut using parameters of 80 W and 400 mm/s.....	102
Fig. 3.25 Measurement of grain width.....	103
Fig. 3.26 Columnar grain width upon varying process parameters and build orientations, (a) 0.5 mm; (b) 0.3 mm; (c) 0.1 mm	104
Fig. 3.27 Microstructure of the thin-feature material produced by LPBF	105
Fig. 3.28 Micro tensile testing setup.....	106
Fig. 3.29 Stress-time curve for thin strut specimens with varying build orientations and diameters at the process parameter of 60 W and 300 mm/s	111
Fig. 3.30 Tensile fractographs of 15° thin-strut specimens under SEM, (a) Process parameter of 80 W and 400 mm/s (Zone I); (b) Magnified view of (a); (c) Process parameter	

of 100 W and 300 mm/s (Zone II); (d) Magnified view of (c); (e) Process parameter of 60 W and 500 mm/s (Zone III); (f) Magnified view of (e).....	116
Fig. 3.31 Tensile fractographs of 75° thin-strut specimens under SEM, (a) Process parameter of 80 W and 400 mm/s (Zone I); (b) Magnified view of (a); (c) Process parameter of 100 W and 300 mm/s (Zone II); (d) Magnified view of (c); (e) Process parameter of 60 W and 500 mm/s (Zone III); (f) Magnified view of (e).....	122
Fig. 3.32 Schematic of Vickers microhardness test. (a) Schematic of Vickers pyramid diamond indenter; (b) Illustration of micro Vickers hardness indentation	124
Fig. 3.33 Comparison of microhardness values of bulk samples and thin struts.....	125
Fig. 3.34 Microhardness of thin struts at varying process parameters and build orientation	126
Fig. 3.35 Micro hardness by nominal diameter of thin struts	127
Fig. 3.36 Porosity of Ti-6Al-4V solids	128
Fig. 3.37 Fitted curves for the porosity of Ti-6Al-4V solids	129
Fig. 3.38 Trending and fitted curves for the power and energy density vertex values versus scan speed	131
Fig. 3.39 Grain width of Ti-6Al-4V solids	132
Fig. 3.40 Fitted curves for the grain width of Ti6-Al-4V solids.....	133
Fig. 3.41 Trending and fitted curves for the minimum grain width value versus laser power	134
Fig. 3.42 Fit of microhardness for Ti-6Al-4V solids	136
Fig. 4.1 Illustration of hexahedron, octahedron and dodecahedron unit cells	142
Fig. 4.2 Illustration of dimension of octahedral unit cell, (a) original joint-based unit cell; (b) transferred enclosing unit cell; (c) CAD model for (a); (d) CAD model for (b).....	143

Fig. 4.3 Slender struts, (a) Joint connected actual thin struts fabricated by EBM; (b) Type I joint; (c) Type II joint	145
Fig. 4.4 Illustration of the effective length of struts	145
Fig. 4.5 Schematic of the loading and deflection of the struts, (a) loading condition and the moment distribution; (b) cantilever deflection of the half strut.....	147
Fig. 4.6 Circular cross-section showing a differential area	151
Fig. 4.7 Effective moduli resulted from 3 types of deformation in FEA (n=4).....	154
Fig. 4.8 Unit cells of (a) hexahedral structure (n=3) and (b) dodecahedral structure (n=6)	155
Fig. 4.9 Analytical computations for (a) hexahedral and (b) dodecahedral unit cells	157
Fig. 4.10 Constraint setup for the unit cell.....	160
Fig. 4.11 Effect of aspect ratio and gradient angle on the effective modulus E_y^*	161
Fig. 4.12 Load and fixture condition setup of cellular structure in FEA	164
Fig. 4.13 Size effect of cellular structures and the fitting curves	166
Fig. 4.14 Fitted curves for size effect	167
Fig. 4.15 Trending and fitted curves for the coefficient a , b and c	168
Fig. 5.1 Build orientation for unit-cell specimens: Perpendicular orientation (left) and tilted orientation (right)	174
Fig. 5.2 Comparison of Effective moduli by analytical model and FEA (n=4).....	176
Fig. 5.3 Comparison of FEA and analytical solutions among nominal and actual various dimensions	177
Fig. 5.4 CAD model of unit-cell specimens for compression test of (a) Hexahedron; (b) Octahedron; (c) Dodecahedron; and (d) LPBF As-fabricated unit-cell specimens of the three types	181

Fig. 5.5 Illustration of convergent struts of a unit cell during fabrication	182
Fig. 5.6 CAD model of octahedral unit cells with varying tilt angles, (a) 15°; (b) 45°; (c) 75°	184
Fig. 5.7 LPBF produced cellular structures	187
Fig. 5.8 Setup for the compression test for cellular structures.....	187

CHAPTER 1

INTRODUCTION AND LITERATURE REVIEW

1.1 Cellular structures

1.1.1 Introduction of cellular structures

Cellular structures are also known as lattice structures or foam structures, which are usually constructed of ligaments and nodes that connect the ligaments. Cellular structures exist widely in nature, such as the coral, the honeycomb, and the natural bones. Cellular structures are also manufactured and used in various applications. Honeycomb bumper structures and truss bridge are good examples of cellular structures.

Cellular structures can be treated as structures with a large fraction of air contained within the structures. Cellular structures can be classified in various ways, such as geometric characteristics, functionality, and manufacturing process. According to how the structures are created, cellular structures can be classified as natural structures and artificial structures. On the other hand, according to the geometrical characteristics, cellular structures can be categorized as stochastic, periodic, or even a mixture of both. Natural cellular structures are in generally more stochastic. Some man-made cellular structures exhibit regular geometry patterns, such as the periodic 2D honeycombs made by crimping and stamping. On the other hand, many man-made cellular structures are not repeatable but

somewhat stochastic, such as metal foam [1, 2]. The most significant limitation of stochastic cellular structures is the lack of freedom given to the designer for the control of the placements of voids in cellular structures [3, 4]. Cellular structures can also be categorized into open cell structures and closed cell structures depending on the morphology of the porosities. The open cell structures have pores that are interconnected with each other and usually also connected to the surrounding environment. On the other hand, the closed cell structures do not have interconnected pores, and these pores are usually enclosed in the cell walls. Due to the presence of continuous cell walls, the closed cell structures exhibit a stretch dominated failure mechanism and therefore are generally stronger than the open cell structures, which exhibit bending dominated failure mechanism [5]. There have been plenty of works related to the modeling, analysis, and manufacturing of the open cell foam structures, which has been mainly focused on stochastic foams [6].

Periodic cellular structures exhibit periodicity in either in two or three dimensions. Two-dimensional periodic cellular structures have 2D periodic features extruded in the third dimension, while three-dimensional periodic cellular structures have 3D features (i.e., unit cells) that are patterned in all three principal directions. 3D periodic cellular structures usually have more complex geometry and could possess more isotropic properties depending on the actual design. Currently, natural cellular structures exhibit much higher degrees of complexity than the man-made cellular structures. For example, the human bone structure is a periodic cellular structure with rather complex patterns of cellular size and shape throughout the structure, as shown in Fig. 1.1.

Cellular materials are potentially well suited to multifunctional applications that demand not only structural strength but also some other attributes such as energy

absorption, heat transfer, thermal protection, or thermal insulation [7]. Stochastic cellular solids, such as foams, have excellent thermal and acoustic insulation properties [5, 7]. On the other hand, periodic cellular solids, such as honeycombs and lattices, can be designed to have tailored mechanical properties, including energy absorption, strength, and stiffness [5], as well as lower pressure drop and high surface area densities, which are important for heat transfer performance [8].

Fig. 1.1 Natural cellular structures: Coral (left); Honeycomb (middle); Hip stem insertion (right)

1.1.2 Manufacturing of cellular structures

Many techniques have been developed for fabricating cellular and foam structures, which exhibit different capabilities and productivities. These processes could be categorized as casting, forming, powder metallurgy, deposition, sheet metal forming/bonding, and wire weaving. Most of the processes are lack of the ability to produce periodic cellular structures or require a large amount of pre- and post-labor and skill investment, which make them unfavorable for the purpose of this project.

Melt gas injection is a conventional method to produce foam structures. Some variations of the process exist, however, the primary procedures are demonstrated in Fig. 1.2. During this process, the premixed materials are heated up to the liquids temperature

and melted. Gas is injected into the melt to form pores directly [9]. The temperature of the liquid is carefully controlled to be close to the melting point, so the viscosity is adequate. In order to further increase the viscosity, stabilizer particles such as aluminum oxide or silicon carbide are added into the melts [9, 10].

This process can produce continuous long foam sheets with tailored width and thickness, and can only produce closed cell foams [11]. Aluminum and its alloys are mostly reported to be suitable by the melt gas injection method because they have relatively low density and do not oxidize much during the process [12, 13]. However, it is difficult to control the distribution of the pores and the exact size and shape of each pore by this method.

Fig. 1.2 Melt gas injection casting to manufacture Al foam [9]

Another similar method is to add forming agents into the melt instead of air bubbles (Fig. 1.3). A widely used foaming agent is titanium hydride (TiH_2), which decomposes into titanium and gaseous H_2 when it is heated above 465°C [9]. During the melting process,

the foam structures gradually dilate under the expansion force of H_2 bubbles and eventually solidifies. By controlling the parameters of foaming agent content, temperature and cooling rate, macroscopically homogeneous foams can be fabricated [10, 14]. This method can produce closed-cell foam structures with significant porosity, which is relatively uniform. However, it is also difficult to control the individual pores, and therefore hard to obtain the precise predictions of the overall properties. Moreover, because of embrittlement effect and high decomposition speed of H_2 in numerous metal alloys, aluminum alloys are currently the primary type of metal processed by this method [15, 16].

Fig. 1.3 Gas releasing agent casting to manufacture Al foam [10, 14]

Investment casting (IC) is a conventional casting process that is capable of producing parts with complex features. The template (or pattern) that are made from a volatile wax or polymer (e.g., polyurethane and ABS) by injection molding are attached on the face sheets at the first step [17, 18]. Then the pattern, together with a system of gating and risers, is coated with a ceramic casting slurry and dried. The wax or polymer is removed by melting or vaporizing completely, and then the empty mold is filled with liquid metal. After solidification, the ceramic mold is removed by vibrating, and the IC-produced parts are formalized. Multiple cellular structures such as pyramidal, tetrahedral and 3D Kagome

[3] have been made by this method. Fig. 1.4 shows an example of an investment casting 3D Kagome sandwich panel.

Fig. 1.4 An example of an investment cast 3D Kagome sandwich panel [19]

The approach allows the fabrication of more complex and non-planar shapes. Moreover, with the advent of rapid prototyping, casting pattern manufacture can be automated. However, structures with near-optimal, low-relative-density cores are difficult to fabricate because of the tortuous flow paths and the resulting susceptibility to casting defects. The need to infiltrate the tortuous structure also limits the process to alloys that have high fluidity upon melting. The materials that are suitable for IC process include aluminum alloys, stainless steel alloys, titanium alloys, carbon steel and glass [20-24].

The perforated metal sheet method is commonly used to produce core structures for sandwich panels. Different cellular patterns with low relative density can be fabricated from alloys with high formability, such as 304L stainless steel [25], by relatively simple press forming operations. Queheillalt et al. demonstrated the fabrication of pyramidal lattice truss structures by folding the perforated diamond sheet at the nodes using a punch,

and therefore a single layer truss structure can be made (Fig. 1.5) [26, 27]. Similarly, Sypeck et al. used the same method to make periodic tetrahedral truss cores with hexagonal pattern [25].

Fig. 1.5 An illustration of the folding operation used to fabricate the single layer pyramidal truss structures [26]

Since the perforated sheet folding method tends to have low efficiency, Gregory et al. demonstrated that a similar sheet metal is working based method for pyramidal truss lattices based on the in-plane expansion of partially slit metal sheets [28]. The process for the expanded lattice structures is shown in Fig. 1.6, which involves three primary steps: slitting, flattening, and folding of the metal sheet.

Fig. 1.6 Schematic of the process for the expanded pyramidal lattice truss structure [28]

Another method used by various researchers to produce periodic cellular structures is the wire-weaving method. Some structures that are difficult to produce with sheet metal working are made by this method such as the Kagome structures [6, 29, 30]. The individual wires of steel or other materials are twisted together until plastic deformation occurs, and the wires become helical, then the wires are fixed on the frame and woven together to produce the periodic cellular structures. An example of a 3D Kagome structure made by wire-weaving is shown in Fig. 1.7. The wire-weaving process involves extensive manual operations, which make the fabrication of large-scale 3D cellular structures extremely difficult.

Fig. 1.7 An example of a 3D Kagome structure made by the wire-weaving method [6]

Since AM technologies were first introduced around 1986, after decades of development, versatile AM techniques have potential ability to fabricate parts with predetermined properties with almost any geometry directly from the digital CAD model. Therefore, on account of the emergence of AM techniques, the manufacturing of complex

and highly customized products can be realized. Additionally, in combination with multi-controllable parameters, AM processes can make parts with a wide range of materials.

1.1.3 Geometry design for cellular structures

Cellular structures are a kind of network system of the interconnected solid beam or wall elements with voids inside. Classical theories for cellular structures have been well developed by many research groups [2, 3, 7, 13]. For open or closed cellular structures, the geometrical volume ratio of the cellular components can be identified using relative density (RD) and porosity (ρ), and the relationship of both satisfy the following equation.

$$RD + \rho = 1 \quad (1-1)$$

In general, open cell structures exhibit lower mechanical strength compared to closed cell structures. The mechanical properties of the open cell structures can be predicted by the relative density by Gibson-Ashby method [5, 7].

For stochastic foam structures, there exist explicit relationships between the relative density of the structure and their mechanical properties. Eqn. 1-2 ~ 1-5 shows the theoretical predictions of Young's modulus (E) and strength (σ) of open cell foam and closed cell foam structures, where E_s and σ_s are Young's modulus and strength of the solid material used to make the foam, ρ and ρ_s are the density of the foam structure and the solid material, respectively, Φ is the volume fraction of the solid contained in the cell edges, and C_1 , C_2 , C'_1 , and C'_2 are constants that vary with the specific type of foams. For open cell foams, generally $C_1 = 1$, $C_2 = 0.3$, and for closed cell foams, generally $C_1 = C'_1 = 0.32$, $C_2 = 0.33$, and $C'_2 = 0.44$. Due to the cell wall stretching under stress, the closed cell foams

exhibit higher Young's modulus and strength compared to open cell foams, as clearly shown in the equations [31].

$$\frac{E}{E_s} = C_1 \left(\frac{\rho}{\rho_s} \right)^{n_1} = C_1 \left(\frac{\rho}{\rho_s} \right)^2 \quad (1-2)$$

$$\frac{\sigma_{pl}}{\sigma_s} = C_2 \left(\frac{\rho}{\rho_s} \right)^{n_2} = C_2 \left(\frac{\rho}{\rho_s} \right)^{\frac{3}{2}} \quad (1-3)$$

$$\frac{E}{E_s} = C_1 \phi \left(\frac{\rho}{\rho_s} \right)^{n_1} + C_2 (1-\phi) \frac{\rho}{\rho_s} \quad (1-4)$$

$$\frac{\sigma_{pl}}{\sigma_s} = C_2 \phi \left(\frac{\rho}{\rho_s} \right)^{\frac{3}{2}} + C_2 (1-\phi) \frac{\rho}{\rho_s} \quad (1-5)$$

Although Eqn. 1-2 to 1-5 are commonly used for foam structures; the derivation process is based on the standard beam and wall bending theory. Therefore, they are expected to be applicable to 3D periodic cellular structures.

The hexagonal honeycomb structure is one of the few periodic cellular structures that have been analytically modeled and discussed in great details. Gibson et al. demonstrated the modeling of the hexagonal honeycomb structure in detail employing Euler-Bernoulli beam theory [5]. The unit cell of the hexagonal honeycomb with design parameters is shown in Fig. 1.8. The model assumes the thickness of the wall t is relatively small compared to l (i.e., t/l is small). Therefore, the honeycomb network structure could be conveniently treated as a frame structure, and the individual beams could be analyzed using classical beam theory.

Fig. 1.8 The unit cell of the hexagonal honeycomb

The elastic modulus of the structure in X_1 and X_2 directions can be obtained by:

$$\frac{E_1}{E_s} = \left(\frac{t}{l}\right)^3 \frac{\cos \theta}{\left(\frac{h}{l} + \sin \theta\right) \sin^2 \theta} \quad (1-6)$$

$$\frac{E_2}{E_s} = \left(\frac{t}{l}\right)^3 \frac{\left(\frac{h}{l} + \sin \theta\right)}{\cos^3 \theta} \quad (1-7)$$

In modeling of the yield properties, Gibson et al. assumed that the neutral plane shift due to the normal stress was negligible, and obtained relatively simple equations [32]:

$$\frac{\sigma_{pl1}}{\sigma_s} = \left(\frac{t}{l}\right)^2 \frac{1}{2\left(\frac{h}{l} + \sin \theta\right) \sin \theta} \quad (1-8)$$

$$\frac{\sigma_{pl2}}{\sigma_s} = \left(\frac{t}{l}\right)^2 \frac{1}{2\cos^2 \theta} \quad (1-9)$$

In addition, for the cellular structures with low strut connectivity that has low stiffness, the configuration of their cell edges behaves different stabilities and deformation mechanisms [33]. Therefore, the Maxwell stability criterion is often employed design to

help explain the performance of cellular structures and determine the kinematic stability of the pin-jointed frame. The Maxwell criterion can be expressed as follows [34]:

$$\text{2-D lightweight structure: } M = b - 2j + 3 \quad (1-10)$$

$$\text{3-D lightweight structure: } M = b - 3j + 6 \quad (1-11)$$

where M is the Maxwell stability number; b and j are the numbers of struts and joints of the framework, respectively. For example in Fig. 1.9 [33], the frameworks indicate different Maxwell stability numbers. The frame with $M < 0$ in Fig. 1.9 (a) folds up when loaded, and indicates a kinematically unstable framework system. The triangulated frame with $M = 0$ in Fig. (b) is stiff when loaded because the transverse strut carries tension. The frame with $M > 0$ in Fig. 1.9 (c) is over-constrained, and if the horizontal bar is shortened, the vertical one undergoes tension even while no external loads are applied.

Fig. 1.9 Pin-jointed frameworks, (a) $b = 4, j = 4, M < 0$, (b) $b = 5, j = 4, M = 0$, (c) $b = 6, j = 4, M > 0$

Moreover, the Maxwell criterion is shown to be useful in the design of cellular structures [17, 35, 36]. The Maxwell criterion determines the dominant deformation mechanism of the cellular structures. For cellular structures with $M < 0$, the dominant

deformation mechanism is strut/wall bending, and the structures are defined as bending-dominated cellular structures. For cellular structures with $M = 0$, the dominant deformation mechanism is strut/wall stretching, and the structures are defined as stretch-dominated cellular structures. The cellular structures with $M > 0$ give a state of self-stress.

There exist significant discrepancies of mechanical characteristics between bending- and stretch-dominated cellular structures. Bending-dominated structures usually exhibit lower modulus and strength compared to stretch-dominated structures due to the lower resistance of struts/thin walls to the bending deformation. Fig. 1.10 shows the stress-strain characteristics of different cellular structure mechanisms. As is shown in Fig. 1.10 (a), the bending-dominated structures exhibit the classical plateau stress-strain curve. In comparison, the stretch-dominated structures exhibit higher initial failure stress level, is followed by a softening period before the occurrence of the plateau stage.

Fig. 1.10 Stress-strain characteristics of different deformation mechanisms [33]

Another consideration for design for cellular structures is that the unit cell needs to satisfy the space filling requirement. As the description of central joint-based cellular structure design previously, different spatial patterns can lead to different unit cell design

with different geometrical characters. Examples of some unit cell shapes are shown in Fig. 1.11. Among them, only No. 2, 3, 4, 6 and 8 are space-filling geometries. The unit cells could have more complicated internal features. However, the bounding geometries are always one of these polyhedrons or their patterns. On the other hand, 3D cellular structures usually allow geometrical design adjustment in all principal directions and therefore are potentially favorable for the design of multi-axial loading applications.

Fig. 1.11 Polyhedral unit cells [5]

1.1.4 Assumption in modeling analysis

For analysis of cellular structure, unit cell approaches are often used, which employs structural symmetry to simplify large cellular structures into representative unit cells. Based on the requirements of applications, the internal connections (joints) can be designed as rigid, semi-rigid, or hinged. For many cellular designs, assumptions may include that connected nodes act as hinge joints, neglecting bending moments and

considering only axial forces on individual strut components [36-38]. This pin-joint assumption works efficiently for mathematical manipulation of models for mechanical analysis and is relatively accurate for structures with short and medium length struts [36]. However, for structures with slender struts, since the joints are usually stiffer than the struts, bending becomes the dominant mechanism for the deformation of the structures, and the hinge-joint assumption becomes inaccurate. On the other hand, a rigid joint of the structures eliminates the rotation of the strut ends and allows for the transfer of bending moments.

Studies have shown that models based on the bending-dominated mechanism, such as Euler-Bernoulli and Timoshenko models, yield improved accuracy for metal cellular structures fabricated by AM compared to the models that are based on stretch-dominated mechanism [39, 40]. There are a number of beam theories that can be potentially employed to describe the deformation characteristics of struts. Thereinto, the classic beam theories include Euler-Bernoulli beam theory and Timoshenko theory, which are commonly employed for analyzing the deformation of beam elements in cellular structures.

In Euler-Bernoulli beam theory, transverse shear stress in the beam is not taken into account whereas Timoshenko beam theory takes into account shear deformation and rotational bending effects, making it suitable for describing the behavior of thick beams [41]. In Euler-Bernoulli beams, bending is assumed to behave in such a way that cross-section normal to the neutral axis remains normal to the neutral axis after bending. Euler-Bernoulli beams give good results for normal stress because they are capable of capturing bending dominated deformation fields. On the other hand, in case of Timoshenko beams,

the cross-section initially is normal to the neutral axis but does not remain normal after bending. Timoshenko beam theory is higher order than the Euler-Bernoulli theory.

1.2 Additive manufacturing

Additive Manufacturing (AM) includes manufacturing processes in which complex three-dimensional solid shapes based on a CAD model are added layer by layer to the desired geometry. This technology has unique advantages over conventional manufacturing in its capability to manufacture freeform shapes without any use of molds. AM has been called by many other terms, including solid freeform fabrication (SFF), rapid prototyping (RP), layered manufacturing (LM) and digital manufacturing (DM). The various processing technologies as per the ASTM F2792 - 12a that is part of AM include binder jetting [42-45], directed energy deposition [46, 47], material extrusion [48-50], material jetting [51, 52], powder bed fusion [53-55], sheet lamination [56, 57], and vat photopolymerization [58].

In the current study, an EOSINT M 270 LPBF process is employed to fabricate metal lattice truss structures for experimental studies.

1.2.1 Laser Powder Bed Fusion (LPBF)

Laser Powder Bed Fusion (LPBF) is capable of building parts directly by selectively melting successive layers of metal powder with a high power laser beam. Similar to the Selective Laser Sintering (SLS) process, LPBF utilizes contour information of horizontally sliced cross-sections of a CAD model.

The LPBF process starts by preparing a .STL file which is the standard file format for AM technologies. This file is then loaded into a file preparation software package for

slicing the 3D data into layers, usually from 20 μ m to 100 μ m in thickness, creating a 2D image of each layer. After that, parameters, and additional process settings are assigned to allow the file to be interpreted and built by different types of AM machines. In addition, physical supports are generated during the .STL file preparation process if necessary.

Most LPBF machines from the companies, such as Concept Laser (GmbH), SLM Solutions, and EOS (GmbH), provide highly controllable design reign to choose exposure strategies and laser processing parameters as the users' input manually. Besides, some other factors can significantly affect the quality of products, including layer thickness, protective atmosphere, build substrate temperature and oxygen purity.

During the LPBF process, atomized fine metal powder has to melt and wet previously solidified layers to avoid balling or warping. A metal substrate plate is usually fastened to the building platform before the first layer is distributed. Once a layer is distributed, laser energy will be applied for contouring and scanning on the powder surface according to the 2D slice of the part geometry using two high-speed scanning mirrors aligned in the X- and Y-axes respectively, as shown in Figure 1.12. The melted layer will wet and be fused to the substrate. The subsequent layer will be melted and fused on the previously solidified layer. Overhanging structures need to be attached with supports extended from the substrate plate. This process takes place inside a chamber under the inert gas atmosphere, which can be either argon or nitrogen, until the part is complete.

Fig. 1.12 Schematic of LPBF process [59]

LPBF works with a variety of metal alloys, including Ti64, IN625 and IN718 nickel alloys, and many stainless steel alloys. Additionally, there are multiple process control parameters in LPBF, including laser exposure types and their corresponding scanning parameters, such as laser beam power, scanning speed, hatching distance, as well as material feedstock parameters including particle size distribution, layer thickness for a specific material, that influences the fabrication qualities. In most process development studies, most of these parameters are investigated for each material in order to find optimum process settings for part quality.

The highly focused laser energy input in LPBF results in large thermal gradients, which in turn result in residual stress on the part. Deformation, cracking and dimensional inaccuracy is among the most common defects. A bridge structure [60] and a T-shaped cantilever structure [61] were fabricated using LPBF for residual stress evaluations, and the bending deformation was measured in both cases. The bridge structure showed bending of the two ends into the positive vertical direction, and the T-shaped cantilever structure bent in the negative vertical direction after cooling to ambient temperature and removal of

the support structures. A structure with overhanging features on both sides was built, and it was observed that the overhanging structures bent and broke away from the support structure because of the residual stress [62]. Cracks and delamination could occur during fabrication of M2 Tool Steel parts in LPBF process[63]. Applying pre-heating to the base plate and re-melting every layer was investigated in titanium parts by LPBF process, in order to reduce cracks and delamination that were caused by thermal stress [64]. For typical AM processes, there are two types of support materials: (a) material which surrounds the part as a naturally-occurring by-product of the build process (natural supports), e.g. binder jetting printing process, and (b) rigid structures which are designed and built to support, restrain or attach the part being built to a build platform (synthetic support) [65]. Unlike AM processes, such as SLS and sheet lamination processes, which provide adequate support the part by its surrounding material. In LPBF, although the surrounding powder behaves as a natural support, additional synthetic support structures need to be designed and fabricated to reduce deformation caused by residual stress. Most parts need supports for overhanging structures in order to connect the part to the build platform. This type of support structure needs to be removed from the part during LPBF post-processing.

1.2.2 Capabilities and limitations

LPBF possesses the capability to produce fully dense metal parts with complex geometries. It can fabricate parts with intricate internal features and thin features. The minimum layer thickness for most commercial machines is 20 microns, and thickness setups between 20-50 microns are common. Previous work on mechanical property evaluation for thin struts under varying process parameters was reported. Tsopanos et al. stated that the dimensions of the struts varied between 180-250 μm with a range of

mechanical properties using stainless steel 316L [66]. Another study from Yadroitsev et al. shows the relationship of processing parameters (laser power input, scanning speed and thickness of layer) and quality of thin wall features, which were performed using a 3D System Phenix laser system, and they concluded that the minimal possible thickness of an inner wall is 140 μm for stability [67]. However, the systematical study of process optimization on thin feature structures has not been carried out in details.

Currently, a number of studies on the characterization of cellular structures have focused on the evaluation of topological designs [2, 68], while the relationship between process selection and the material characteristics of the thin features has not been addressed as a significant concern. The fabrication of structures that consist of thin features with dimensions of 1mm or smaller is close to the geometric limit for LPBF. Moreover, these thin feature structures exhibit different characteristics due to significantly different thermal histories [69, 70]. During melting, based on the energy inputs under different scanning strategies and laser power, the dimensions of these thin structures can become a significant factor in determining the condition of the melt pool, which in turn affect the geometrical, microstructural and mechanical characteristics of the structures. When small-scale features are made, more significant thermal dissipation occurs through the surrounding powder bed, which often results in the generation of defects, including unmelted powder inclusions, internal voids, cracks and shape irregularities [11, 71]. Surface defects such as redundant powder sintering and surface cracks also result in mechanical property variations and make accurate quality control more difficult [72, 73].

1.2.3 Characteristics of LPBF-produced parts

The LPBF is a point-by-point and layer-by-layer manufacturing process that involves rapid solidification, directional cooling, and phase transformations induced by repeated thermal cycles. Rapid solidification reduces elemental segregation and extends solid solubility and can result in metastable phase formation. Directional heat dissipation may result in preferred directionality in grain growth. Repeated thermal cycles also exert complex effects, i.e., microstructural differences between depositions layers [74]. As a result, the heterogeneous properties of AM materials are closely associated with the entire processing history that includes various process characteristics, such as the microstructure and mechanical properties. LPBF is currently compatible with many materials, such as Ti-6Al-4V, 316L, cobalt-based and nickel-based alloys and more.

The microstructural evolution of the metals made by LPBF has been studied and modeled [70, 75-82]. Ti-6Al-4V is a typical material used in LPBF. Fig. 1.13 shows an outline of a three-dimensional phase diagram of α - and β -stabilizing element. In LPBF process, the melted region of laser beam processed starts from the β nucleus. The pre-existed β grains partially melt and contribute as the nucleation sites for the newly formed β phase to grow epitaxially in the building direction [83], which is opposite to the heat extraction direction [84, 85]. The β phase is bcc lattice structure and is unstable at low temperature. At a fast cooling rate in LPBF, the metastable martensitic α' phase can be formed from the prior β phase (Fig. 1.14).

Optical microscopy is among the most common approaches used to analyze microstructures under $1000\times$ magnification. In situations where specimens can no longer be imaged using optical microscopy, scanning electron microscopy (SEM) and

transmission electron microscopy (TEM) can be used to observe local microstructural features, such as powder particle morphologies of and precipitate phases.

Fig. 1.13 Schematic of 3D phase diagram on Titanium alloys [86]

Fig. 1.14 Schematic of cooling diagram for Ti-4Al-4V [87, 88]

The mechanical properties of the AM-produced parts can be evaluated by various mechanical tests, such as tensile test, compression test, fatigue loading test, hardness test and so on. The mechanical properties for many metals, such as Ti-6Al-4V [54, 89], Nimonic 263 [90], and PH1 steel [91], fabricated by LPBF are found better than that of the standard materials from the ASM handbook [92, 93]. Vilaro et al. reported that LPBF produced as-built Ti-6Al-4V had a higher yield and ultimate strengths [80], and similar results were also presented by Vrancken et al. [94].

Since the mechanical properties of the LPBF as-built specimens can be affected by microstructural defects and material anisotropy, these characteristics must also be evaluated. The characteristics and influence of defects in LPBF were investigated systematically by Gong [53, 95, 96]. The tensile coupons with lower porosity reflected significant compromise on tensile properties. The internal defects could also become the initiations that created cracking and accelerated the growth of fractures. On the other hand, the variation in tensile properties with respect to the build orientation can be attributed to the fabrication orientation, which reflects as a function of the loading axis [80]. Additionally, the defects exhibited different morphologies due to various energy input. Gong made efforts and stated that the voids due to lack of fusion were shaped irregular while those over melted structures using high energy process parameters usually had spherical morphology [96].

The melt pool size and shape also have a direct impact on the mechanical properties, and therefore, should be controlled in order to produce materials of consistent quality. Beuth and Klingbeil developed process maps for predicting melt pool size and related these properties to deposition rate and power [97, 98]. They showed that it is possible to maintain

a constant melt pool size over a range of deposition rates. In the latest study of Beuth's research group, the geometry-based simulation model was reported to be used to predict porosity caused by the insufficient overlap of melt pools, and the comparison results with experimental data were fitted [99].

Moreover, surface finish also affects the mechanical performance. Chan et al. investigated the effect of surface roughness on the fatigue life of EBM and LBM Ti-6Al-4V using a three-point bend test. They proposed equation (Eqn. 1-12) to correlate fatigue life to surface finish.

$$\ln(\text{Fatigue life}) = -0.34 \ln(\text{Surface Roughness}) + C \quad (1-12)$$

Martukanitz et al. [100] and Petrick et al. [101] reported that the build rate and feature definition are closely related to the surface quality. Figure 1.15 illustrates the relationship between build rate, power, and feature quality.

Fig. 1.15 Relationship between build rate, power, and feature definition

In addition, the microstructural analysis revealed anisotropy of specimens depending on the building direction. For both vertical and horizontal building directions,

the microstructures within each layer have similarities and depend on the thermal gradient and its direction at the boundary of the moving fusion zone [54, 90, 102]. These specimens have significantly improved elongation properties in comparison to specimens fabricated by conventional casting processes [103]. The heat conduction direction based on scan patterns is the driving factor for the orientation of grains [104]. Vertically built specimen often exhibit lower strengths compared to horizontally built specimens [105]. Vilaro et al. also investigated, the samples in the transverse direction have a higher ductility than the ones in the longitudinal direction because of the columnar grain growth [90]. The study of Brandl et al. also stated that the samples fabricated by powder-bed processes showed higher strength and lower ductility in the X-direction that is parallel to the build substrate than those in the Z-direction that is perpendicular to the substrate [54].

1.2.4 Methodologies of material characterization

1.2.4.1 Macro-scale characterization of cellular structures

Studies on the mechanical properties of cellular structures have focused not only on the structure level but also on the individual strut/thin wall level.

For slender beams with slenderness ratio between 75 and 300, several analytical and experimental studies have been performed to investigate the elastic [106, 107] and plastic [108] behaviors. Some of these studies adopted slender beams as long as 700 mm, but the testing methods could be referential. For instance, Fig. 1.16 shows a test apparatus and main upper support frame of a dynamic impact apparatus (Fig. 1.16(a)) with a schematic illustration (Fig. 1.16(b)) [109]. It is used to investigate the plastic buckling

response of a slender beam subject to axial impact with respect to a variation in the effective slenderness ratio and material properties.

Fig. 1.16 (a) Test frame for experimental dynamic impact study and (b) schematic of illustration of the mass assembly with upper fixed support

Quasi-static axial crushing tests were performed on circular thin-walled tubes made of 304 stainless steel, aluminum alloy 6063-T6, and mild steel by Hsu et al. The mode transitions during the crushing of the samples and their characteristics, such as strain rate sensitive properties and strain hardening over a range of different axial lengths, were examined [110]. Similarly, axial compressive impact studies were employed to determine the local and global buckling of cylindrical shells [111] and high-strength concrete cylinders [112], which demonstrated a time-dependent size effect phenomenon in both experiments and numerical simulations.

However, to date, works with structures that have $10 \sim 100 \mu\text{m}$ dimensional scale, which is too large for micro/nanoscale testing but too small for traditional testing, are rarely reported.

For cellular structures, sandwich structures are often tested using tensile test [113], compression tests [11, 113] and bending tests [11, 114, 115] to investigate mechanical properties such as Young's modulus (stiffness), yield strength, fracture strength, Poisson's ratio, resilience, etc. Popular bending tests include 3-point bend (Fig.1.17) and 4-point bend (Fig. 1.18) tests.

Fig. 1.17 Schematic of the 3-point bending test [73]

Fig. 1.18 Schematic of the 4-point bending test on the sandwich beam and loading rollers [116]

The main advantage of a 3-point bending test is the ease of the specimen preparation and testing, but the results of testing are sensitive to specimen and loading geometry and strain rate. On the other hand, the 4-point bend geometry has the advantage that it subjects a beam to a constant bending moment and zero shear force between the central locations of force. It is found that failure takes place on the value of the compressive strength which is higher than that measured by direct compression [117, 118]. Furthermore, the 3-point bending test uses one loading anvil, whereas a 4-point bending test uses two loading anvils, which distributes the stress more evenly on the samples. The bend type is

usually selected based on the material. In most cases, non-homogeneous materials tend to be stiffer and more brittle, such as ceramic, so using a 4-point bending test will prevent the material from failing prematurely (Instron Co.).

Quasi-static and dynamic compression test is also commonly employed for cellular structure characterizations, especially when the energy absorption performance is of interest. It is noted that for the honeycomb specimens (3 in thick and 4×4 in cross-section) in the previous compression tests, the deformation mode is local buckling away from the free surfaces, and bond failure near the edges that triggers a global buckling failure mode [119]. Thus, the free surfaces have an edge effect, and the edge effects would be even more evident for smaller specimens.

1.2.4.2 Micro-scale material characterization

Micro-scale material tests are used to test small samples in the micron and nanometer scale ranges. Similar to bulk structures, the evaluation of the micro-scale structures also includes the characterizations of their elastic and plastic behaviors. In some instances, the standard techniques developed for macro-scale mechanical tests, such as those codified by ASTM, can be applied to characterize small-scale structures. However, there are no standard techniques at present to characterize thin structures in micro dimensions, although some results for initial design have been reported [120].

The mechanical testing methods commonly employed for Microelectromechanical systems (MEMS) devices could provide additional information. Although the minimum feature size of typical MEMS is on the order of 1 μm (Srikar and Spearing, 2003), which is much smaller than the typical feature scale of the LPBF cellular structures. some of the testing methods and theories for these devices still prove insightful for this study.

Typical techniques for the characterization of mechanical properties of MEMS materials include the tension test, bending tests (including axisymmetric plate bending, microbeam bending, bulge test, M-test, and wafer curvature tests), dynamic (resonant) tests, and the fabrication of passive strain sensors [121].

The micro-scale tensile test

The micro-scale tensile test is an extension of macro-scale tensile tests and can characterize Young's modulus, yield strength, fracture strength and ductility of non-integrated, free-standing thin and thick structures. When transverse strain is measured, Poisson's ratio can also be computed [122]. A uniaxial load is applied to both ends of a specimen, and the recorded data of the extension enables the characterization of the material from its stress-strain graph [120]. Tsuchiya and coworkers [123] at Toyota developed a tensile tester based on an electrostatic-force gripping system (Fig. 1.19). The tester is constructed within a Scanning Electron Microscope (SEM) chamber for *in situ* observation. In the setup, the free end is fixed to a flat probe by electrostatic force, and a tensile force is applied until fracture.

Fig. 1.19 Tensile testing using an electrostatic force grip

Axisymmetric bending test

Axisymmetric bending tests can be used to characterize the fracture strengths of silicon structures on the mesoscale. Fig 1.20 shows a schematic diagram of the loading method. The specimen with a diameter much larger than the area of load distribution (See Fig 1.20), can be cut from a wafer and is supported on a ring concentric to the load axis. The maximum tensile stress and the initiation region of fracture occur in a small region at the center of the lower surface of the specimen. The fracture stress is given by:

$$\sigma_F = \frac{3P_F}{4\pi h^2} (1+\nu) \left(\frac{3}{8} + 2 \ln \frac{a}{a_1} - 2 \ln 2 \right) \quad (1-17)$$

where P_F is the applied load at fracture, h is the thickness of the wafer, a is the radius of the hole, and a_1 is the radius of the area of contact between the ball and the specimen. This equation is obtained from experiments focused on semiconductor materials. Since these are brittle materials, their fracture strength critically depends on the characteristics of the processing-induced flaws.

Fig. 1.20 Schematic illustration of the axisymmetric bending test fixture [124]

Microbeam bending test

Microbeam bending has been extensively used to characterize Young's modulus, yield strength, and fracture strength of structures. Fig 1.21 shows the schematic of the microbeam bending experiment. The maximum strain occurs at the fixed end of the beam [125]. A bending load (P) is applied to a beam, and the displacement (δ) is recorded as a function of the load. For a cantilever beam of length L , width b , and thickness h , Young's modulus is given by the expression [126]:

$$E = \frac{4(1+\nu^2)L^3}{bh^3} \frac{P}{\delta} \quad (1-18)$$

The fracture strength is given as [126]:

$$\sigma_F = \frac{6LP_F}{bh^2} \quad (1-19)$$

where P_F is the load of fracture. The preceding equations assume perfectly rigid support behavior and are valid for small deformation mechanisms.

Fig. 1.21 Schematic illustration of the microbeam bending test

Nanoindentation

Hardness measurements are routinely used to characterize the mechanical properties of materials. In a typical test, a diamond tip with well-defined geometry is compressed onto the testing surface, and the depth of penetration is measured, as shown schematically in Fig. 1.22. By estimating the area of impress by the indenter, the values of Young's modulus, residual stress, yield strength, and fracture strength can be calculated [127, 128].

The hardness is given as

$$H = \frac{P}{A} \quad (1-20)$$

where P is the applied load and A is the projected area under the indenter.

Fig. 1.22 Schematic illustration of radial cracking at a Vickers indentation [129]

1.3 Motivation and objectives

Cellular structures have shown prospects in many engineering applications, such as biomedical implants and aerospace structures. LPBF is a highly versatile AM technique for producing complex metal components directly from CAD model, without expensive

tooling, and with minimal material waste, particularly for the cellular structures. Therefore, with the development of the AM technologies, the fabrication of cellular structures is no longer a barrier. AM is, therefore, becoming a driving force to accelerate the pace of innovation of cellular structures.

In many previous studies, the unit cell approach was often adopted to simplify the large structure system into representative unit cells, and the current research and development work for cellular structures are focusing on the geometry-dependent structural optimization. An integrated design tool for cellular components is not readily available. In many engineering systems, design information is still disconnected between material properties and structural geometry, which often makes the design work inefficient.

In this study, it is expected to design the joint-based model. The joint-based model designs unit cells with the symmetrical struts growing from the central joint, and the cellular structure is subsequently created by repeating the unit cell in space as requested. This joint-based model provides additional flexibility for the cellular designs and also provides a platform for the fabrication of melting and solidification process in LPBF. Furthermore, an integrated material/structural geometry model is beneficial to allow for simultaneous design considerations of both the material and the geometry, enabling the high-fidelity design of AM cellular structures.

In order to establish the integrated material/geometry model, the works are introduced in detail in the following chapters:

Chapter 2 focuses on the investigation of the material properties of the large-dimension Ti-6Al-4V fabricated via LPBF in order to establish the baseline for the studies of the processing of thin-feature cellular materials.

Chapter 3 explores the relationships between the process and the characteristics of the Ti-6Al-4V cellular materials based on the baseline established in Chapter 2, and compare the material characteristics with the bulk materials. This Chapter establishes the material property model for the integrated cellular design model.

Chapter 4 demonstrates the joint-based geometry design model using strain energy method. The validity of the model was verified by FEA simulations and experiments. In addition, size effect is studied and fitted into the geometry design model to account for finite-size boundary effects of the cellular geometries.

Chapter 5 demonstrates the approach of establishing the integrated material/process design model by integrating the material functions into the geometry model. The effectiveness of the integrated model was verified by experimentation using Ti-6Al-4V cellular structures with the various process and geometry design conditions.

Chapter 6 discusses conclusions from this work and summarizes the future works.

CHAPTER 2

MATERIAL PROPERTY CHARACTERISTICS IN LASER POWDER BED FUSION (LPBF) PROCESS

2.1 Introduction

Laser powder bed fusion (LPBF) is a broadly used additive manufacturing (AM) technique for producing complex metal components direct from CAD with minimal waste. In LPBF, the quality and property of the parts are absorbing considerable attention, because the AM parts are designed from the CAD model and built directly, which processes are lack of additional post manufacturing plans [57, 130]. For this reason, the quality and property of the parts produced by AM processes are of high importance regard to the functional application of the materials.

Generally, the quality and property of the AM fabricated parts are affected by the defects, which can be roughly categorized into two major types: porosity and cracking. Porosity is one of the common types of defects that occur in the LPBF process. The defects in the parts produced by LPBF process can be significantly derived from the character of single track deposits [74]. In general, keyhole and balling contribute primarily to the pores in the LPBF process [57, 96]. Therefore, it is essential to understand the mechanism of defect generation during the single track scan under various process parameters, such as

laser power, scan speed, hatching space and layer thickness. Numerous studies reported research findings from the single track scans on various process parameters and materials [131-133].

The choice of process parameters used to melt and solidify the metal powder is fundamental and critical, as the thermal history of the material influences both of the integrality and the mechanical properties. Selecting the process parameters that suit the materials with desirable full dense or specific production with porosity are critical to success. Currently, the studies on the material of Ti-6Al-4V focus on the high laser power and scan speed, but few comprehensive studies were reported on the low process parameter regions (laser power and scan speed) that are appropriate to thin feature structures.

Therefore, in this and next chapters, studies were carried out using commercial Ti-6Al-4V powder. Single track scans, bulk specimens and thin feature specimens of the Ti-6Al-4V alloy were produced by LPBF based on a 4×4 process parameter matrix. In the process window, the optimal parameter combinations were distinguished from those caused over melting and incomplete melting for distinctive discussion, to explore the influence of process parameters on the evolution of LPBF produced Ti-6Al-4V materials, and also to compare the characteristics of thin structures and bulky specimens using LPBF process.

2.2 Material and Laser powder bed fusion (LPBF) process

Ti-6Al-4V powder

The material used for this study is EOS Ti-6Al-4V alloy, which is a typical lightweight alloy and has excellent mechanical properties and corrosion resistance. The

laser-sintered parts fulfill requirements of ASTM regarding the maximum concentration of impurities [134].

Table 2.1 Composition of EOS Ti-6Al-4V [134]

Composition	Ti	Al	V	O	N	C	H	Fe
Content (wt %)	Balance	5.5 - 6.75	3.5 - 4.5	< 0.2	< 0.05	< 0.08	< 0.015	< 0.3

Fig. 2.1 shows the distribution and morphology of the Ti-6Al-4V powder particles under a Scan Electron Microscope (SEM). Fig. 2.1 (a) shows that the original EOS Ti-6Al-4V powder has a spherical morphology. Fig. 2.1 (b) shows that the characteristic dendritic features can be observed on the surface of the Ti-6Al-4V powder particles. This phenomenon is contributed by the low solidification rate of the particles during the atomization process [135-137].

Fig. 2.1 Distribution and morphology of EOS Ti-6Al-4V powder

The particle size analysis was carried out by Microtrac S3000 laser diffraction analyzer, which is capable of particle sizes of 0.02 to 2800 microns. Fig. 2.2 shows the particle size volume distribution results of the EOS Ti-6Al-4V powder. From the result,

the mean particle size is 33.86 μm under the volume distribution calculation with approximately 60% particle passing. This distribution is unimodal but not monodisperse. It has a width. The distribution width of the powder could then be represented by the span from $D_{v0.9}$, $D_{v0.5}$, and $D_{v0.1}$ values as following Eqn. 2-2 below [138]:

$$Span = \frac{D_{v0.9} - D_{v0.1}}{D_{v0.5}} \quad (2-2)$$

where, $D_{v0.1}$ and $D_{v0.9}$ define the diameters of the particles where 10% and 90% of the volume distribution lie below these values, respectively, and $D_{v0.5}$ is the median for the volume distribution. According to Eqn. 2-2, the span of EOS Ti-6Al-4V powder is 0.81. The larger span value indicates that the distribution is narrower, and that the particle sizes have smaller standard deviation.

Fig. 2.2 Particle size analysis of Ti-6Al-4V pre-alloyed powder

EOSINT M 270 LPBF system

The EOSINT M270 utilizes a Yb-fibre laser with beam diameter as small as 100 μm . Inside the EOS machine building chamber, there is a material dispenser platform along

with a recoating unit used for feeding fresh powder over the building platform, as shown in Fig. 2.3.

Fig. 2.3 Schematic of LPBF process [59]

2.3 Exploration of process parameters in EOSINT M 270

The manufacturer provides a package of standard processing parameters for the specified Ti-6Al-4V powder. These standard parameters usually result in high densities for bulk geometries [134]. In order to enable the parameter optimization by the users, EOSINT M270 also provides users with access to choose numerous laser processing parameters manually such as laser power, scan speed, beam offset and exposure types. There also exist other process-related factors not controlled by the operations that can be of significant effect to the quality of the printed materials, such as the protective atmosphere, build substrate temperature and oxygen purity. Fig. 2.4 summarizes these factors that are related to the setup sequence of the EOS system. For Ti-6Al-4V, the selection of standard layer thickness of 30 μm and the protective inert gas of argon is dependent on the material setting in EOS. Subsequently, based on the geometry of the parts and the exposure features, the exposure strategies for the specific purpose are decided before setting up the process parameters.

Fig. 2.4 Parameter map in EOSINT M270 laser powder bed fusion (LPBF) system

The process control of the EOS M270 system is quite complicated. Various processing parameters, including laser power, scanning speed and hatch spacing, the exposure strategies, including skin/core, contour/hatch/edge scans and the scan pattern, could potentially have extensive interactions in influencing the characteristics of the process. In the development of LPBF processes, the “Energy Density (E_d)” is often used as an important processing characteristic [139], which can be expressed as Eqn. 2.1. Energy density indicates the energy supplied by the laser beam to a volumetric unit of powder material.

$$E_d = \frac{P_{laser}}{v_{scan} \cdot S_{hatch} \cdot t_{layer}} \quad (2-1)$$

where E_d is the energy density, P_{laser} is the laser power, v_{scan} is the scan speed, S_{hatch} is the hatch spacing distance, and t_{layer} is the layer thickness. The ratio of P_{laser}/v_{scan} is also often used as an energy density indicator when the layer thickness and the hatch spacing

remain constant. During the process development and evaluations, the energy density is often used along with other process parameters such as scanning strategies, geometries of parts, building directions, etc.

The EOS INT M 270 system employs two layer-to-layer scanning pattern rotation strategies for the interior hatching stripes, which are shown in Fig 2.5. For each layer, the interiors of the scanning area surrounded by the contour are divided into a series of parallel stripes. Within the same layer, the stripes will be scanned with the same scanning pattern orientation (shown in Fig.2.5 as lines with arrows) regardless of the complexity of the scan area geometries. In the next layer, the scanning pattern orientation will rotate by $\sim 67^\circ$ (Fig. 2.5 (a)) or 90° (Fig. 2.5 (b)) counterclockwise in relation to the direction of the pervious layer. This is principally used to reduce the thermal stresses in X_1 and X_2 directions during scanning, and sequentially improve the metallurgical aspect of the parts by reducing the number of defects [104, 132, 140, 141] . The distance between two adjacent scanning lines is defined as hatching space, which is usually defined as a fixed value.

In this study, the scanning pattern rotation strategy of 90° was used.

Fig. 2.5 Two basic overlapping scan strategies by layer

2.4 Design of Experiments

In this study, process parameters including laser power and scan speed, build orientation and feature diameters were taken into account in order to evaluate their effects on the quality of the thin feature structures. The default process parameter for Ti-6Al-4V support structure is 80 W and 400 mm/s in the EOS M270 system, which was used as the base benchmark parameter setting for the study. The experimental factors and levels are given in Table 2.2. Besides, the hatching space of 100 μm and layer thickness of 30 μm are pre-set in the system.

Table 2.2 Factors and levels of design of experiment

Factor	Level
Laser Power (W)	60, 80, 100, 120
Scan Speed (mm/s)	300, 400, 500, 700

In LPBF, the consolidation of metal powder for a bulky specimen is obtained by melting and solidifying a small amount of material arranged track by track in each layer. Upon the geometry of the CAD model, the next layer solidification subsequently takes place afterward, and thus a 3D object is created. For thin struts in the following study, the dimension of the features is highly dependent on the beam curing size, which reflects on the melt pool (Fig. 2.6). Additionally, the variation of process parameters contributes to the deposition and solidification [83, 131, 132, 142, 143]. However, very little is understood about how the low-level laser power and scan speed, and environment interact with each other and to impact the part qualities of the thin feature structures. Therefore, the single track deposits will be the threshold window to provide a well understanding of LPBF process.

Fig. 2.6 Illustration of laser beam scanning for the thin feature

In order to achieve the single track deposits, cubical blocks with the dimension of 30 mm × 7 mm × 10 mm were used as the starting geometries, which were created in Magics software and imported into the EOS software. During the parameter setup, the hatch and edge scans were disabled, and only contour scan was enabled. In order to realize the single track scanning along the outlines of the blocks. Additionally, the beam offsets were also set to be zero. During the fabrication, a Ti-6Al-4V alloy plate with a thickness of 3.3 mm was placed on the top of EOS build platform as the substrate. One layer of Ti-6Al-4V powder was spread on the top and scanned. The fabricated single track deposits were shown as light, bright contour lines on the plate in Fig. 2.7. The single tracks were sectioned at 1/4, 1/2 and 3/4 of total length (Cross sections 1, 2 and 3) for microscopic studies. In addition, the top surface was investigated under a Zeiss SUPRA 35 scanning electron microscope (SEM).

Fig. 2.7 Single track deposits on the Ti-6Al-4V plate

Fig. 2.8 Fabrication of Ti-6Al-4V specimens, (a) Cubic specimens; (b) Thin struts specimens at various tilt angles and dimensions

Fig. 2.9 Illustration of dimension and section of cubical specimen (left) and thin strut (right)

Beside the single track study, 3D samples were also fabricated. For each set of parameter combination, $5\text{ mm} \times 5\text{ mm} \times 10\text{ mm}$ cubical Ti-6Al-4V specimens and thin struts with various dimensions and build angles were fabricated (shown in Fig. 2.8) in order to investigate the porosity, microstructure, and hardness of the bulk and lightweight geometries. The thin struts were fabricated on the top of solid bases that were also fabricated, in order to reduce the potential damage from the cut-off during post-processing.

The fabricated specimens were sectioned horizontally and vertically, respectively (Shown in Fig. 2.9). The sectioned parts were cleaned in ethanol bath using an ultrasonic cleaner for 3 minutes, and then mounted in groups and polished for microscopy observations.

2.5 Results and discussion

2.5.1 Single track deposits characteristics

Pores or voids in the parts during the fabrication in LPBF are generally attributed to the keyhole and balling phenomena. Keyhole is caused by excessive high energy input over a small region. The resulting geometry of the melt pool exhibits extended depth and relatively narrow width. During the short cooling time, the metal vapor from the bottom of the melt pool cannot escape in time, and thus upon solidification becomes a pore defect. On the other hand, balling phenomenon usually appears when the energy input level is low. It is reported that the balling also happens due to very high energy level because of liquid splash upon to the cohesive powder particles [144]. The size of balling pores is apparently more significant than those caused by keyhole, and this type of pores have a substantial impact on the mechanical property of the fabricated parts.

The top surfaces of the single track deposits were observed under a Zeiss SUPRA 35 scanning electron microscope (SEM), and the surface morphological characters are displayed in Fig. 2.10. It was seen that at a low energy density level, such as at the laser power of 60 W and the scan speeds higher than 300 mm/s, and at the laser power of 80 W and the scan speed of 700 mm/s, the single tracks became discontinuous, and the scan lines showed apparent balling phenomenon, which exhibits spherical in geometry at the fragmentation of the tracks. Additionally, as the scan speed increased at the same laser

intensity level, the height of the balling seemed to reduce. On the other hand, at higher energy density level, the single tracks exhibited more continuous and uniform weld deposits. As the energy intensity increases, the single tracks became noticeably wider.

Fig. 2.10 Top surface morphology of single track deposits under SEM

All the single track deposits were sectioned in the middle of the scan path perpendicularly to the scan direction. The width and depth of melt pools were directly measured from the cross-sectional microscopic graphs. The measurement method of the melt pools is illustrated in Fig. 2.11. It can be identified from Fig. 2.11 that the depth and

width of the melt pools at reduced laser power levels or increased scan speed levels exhibit reduced values, while at higher laser energy (the parameter sets of 120W – 300 mm/s and 100 W – 300 mm/s) the melt pool exhibits typical keyhole-shaped geometries with deep penetrations into the base plate [83, 145].

Fig. 2.11 Melt pool of single track deposits produced by LPBF and illustration of the measurement

The single tracks with significant balling phenomenon exhibit shallow melting depth in the substrate and “mushroom head” shape above the substrate that does not fuse to the substrate. It was believed that the low energy intensity reduced the temperature of the melt pool and caused high viscosity and poor flowability of the melted material. Consequently, the surface tension tended to minimize surfaces by creating liquid spheres [146, 147].

At the same level of laser power, faster laser scan would cause reduced depth and width, while at the same level of scan speed, higher laser power resulted in larger dimensions in depth and width of melt pools. The depth and width of the melt pool of the single tracks were measured and plotted in Fig. 2.12. It was apparent that both laser power and scan speed have significant effects on the dimensions of the melt pools. It was clear that the melt pool would grow bigger with a higher energy density. At the process parameter set of 120 W and 300 mm/s, which was of the energy density of 133 J/mm^3 , the melt pool dimension could reach up to a width of $178.2 \text{ }\mu\text{m}$ and a depth of $150.4 \text{ }\mu\text{m}$. The depth of melt pool was equivalent to the 5-layer thickness and could be regarded that the same region would be melt five times layer by layer. At a lower energy density, such as 28.57 J/mm^3 at the process parameter of 60 W and 700 mm/s, the depth was only $6.7 \text{ }\mu\text{m}$ and much less than the layer thickness of $30 \text{ }\mu\text{m}$ for Ti-6Al-4V in EOS system, which would cause the incomplete melt region between two continuous layers in the LPBF process. Due to the repeated heating/melting, significant defects might be resulted from the process [83, 148].

Fig. 2.12 Variation of depth and width of melt pools on process parameter sets

2.5.2 Top surface of bulk specimens

As shown in Fig 2.13, the solidified top surfaces of bulk specimens were observed under an optical microscope.

It can be seen clearly that the hatch lines were arranged parallel. At the same laser power level, the morphology of melt lines became narrower as the scan speed increased. At the laser power of 60 W, there were gaps between the lines, and the width of those gaps increased as the scan speed further increased. Such discontinuous melt surfaces give rise to potential defects during the fabrication. On the other hand, higher laser power could result in wider melting lines because of high intensity of energy input, which could also lead to the generation of defects or porosity.

Fig. 2.13 Solidified top surfaces of bulk specimens under varying process parameters

2.5.3 Porosity

The as-built bulk specimens were sectioned vertically in relation to the build plate and prepared for microstructural observation. The polished specimens were etched using Keller's reagent. The etched cross sections were observed under the optical microscope, and the micrographs were shown in Fig. 2.14.

Fig. 2.14 Optical micrograph of lateral sections under varying process parameters

The effect of process parameters could be observed apparently in Fig. 2.15. Two parameter sets at the left-bottom (120 W and 300 mm/s) and right-top (60 W and 700 mm/s) place on the diagonal line of the process matrix revealed relatively obvious defects. However, the morphological differences of these defects could be observed. At the process parameters of 120 W and 300 mm/s, spherical pores were generated, which resulted from the keyhole effect under high energy input levels. In general, when the high intensity of laser beam energy melts at a small region, the metal powder vaporizes and creates voids in the melted region. When the vaporization region is open, it can be filled up by the powder from the newly added layer and remelted the region again. But in the case of deep and close vaporization voids, the powder of the top layer cannot fill in the voids, and the vapor is not able to escape during the fast cooling process and thus resulted in round shape pores [149-

151]. On the other hand, irregular-shaped defects were generated under the process parameters of 60 W and 700 mm/s largely due to the insufficient melting of the Ti-6Al-4V powder [85].

Fig. 2.15 Porosity distribution of LPBF produced Ti-6Al-4V bulk specimens

In general, along the other diagonal line of the process matrix relatively good quality cross sections were observed. On the other hand, the porosity increases towards the upper-right or lower-left directions of the matrix in the process map. These results had a good agreement with the observations from the single track deposit.

During the layer-wise laser melting process, even if the pores are generated at the processed layer, the re-melting occurs during the subsequent layer deposition, which could potentially introduce mechanisms for the metal vapor to escape. However, the pores formed at the very bottom of melt pool are difficult to eliminate even with re-melting and could cause permanent porosities [84, 152]. Therefore, the process parameter sets that result in too large or small energy density in the process map are supposed to be avoided.

The porosity was estimated quantitatively using ASTM standard [153] and presented in Fig. 2.15. The significantly high porosity of the specimen that was made at 120 W and 300 mm/s is up to 8% while that at 60 W and 700 mm/s is 2.2%, which shows that between the largest and smallest energy densities in the process map, significant porosity variability could be resulted in.

The relationship between the porosity and the energy density is plotted in Fig. 2.16. A similar trend was also reported for Ti-6Al-4V and some other materials, such as ALSi10Mg and EOS GP1 stainless steel, by other researchers [145, 154, 155]. It was concluded that the desired density ranged with the energy density of 44 to 89 J/mm³, which exhibit porosity levels smaller than 0.2% and could be considered as “high density” solid.

Fig. 2.16 Variation of porosity with energy density changed

Through the observation and analysis on the porosity distributions of bulk specimens under various process parameters, a process map could be determined to

indicate the effect of laser power and scan speed on the porosity defects for the material of Ti-6Al-4V by LPBF process as shown in Fig. 2.17. From the process map, three distinct zones could be defined as shown in Fig. 2.17, which are full dense zone (Zone I), over melting zone (Zone II) and incomplete melting zone (Zone III).

The parameters in Zone I generally result in high-quality bulk specimens with low porosities using the LPW Ti-6Al-4V powder. It was worth noting that the default process parameter set for the support structure in EOS system is also located in Zone I. On the other hand, the parameters in Zone II and Zone III could result in porosities. The Zone II parameters represent high energy density inputs that give rise to keyhole type porosity defects, while the Zone III parameters represent low energy density inputs that led to lack of fusion type of defects. This process window provides an explicit guidance for the following study with the thin strut features.

Fig. 2.17 Process window for LPW Ti-6Al-4V powder in LPBF process

2.5.4 Microstructure

The melting-solidification process in LPBF process has been extensively demonstrated. The solidification of the Ti-6Al-4V starts from the nucleation of the prior β phase. The pre-existing β grains could also potentially act as the nucleation sites for the newly formed β phase to grow epitaxially along the building direction [83], which is opposite to the heat flow direction [84, 85]. The bcc- β phase transforms to metastable martensitic hcp- α' phase by a diffusion mode at the high cooling rate at temperatures below the martensite start temperature M_s [156]. The final microstructure consists of columnar grains that include martensitic α' phase due to the fast cooling rate and grows towards the build direction. The schematic illustrates the phase transformations upon fast cooling in Fig. 2.18. The optical micrographs of the LPBF Ti-6Al-4V materials are shown in Fig. 2.19. The microstructure presents coarse columnar grains with martensite α' phase. The prior β phase exhibits a length across through the width of the columnar grains.

Fig. 2.18 Schematic illustration of microstructures occurring in Ti-6Al-4V after fast cooling quenching

Fig. 2.19 Vertical cross-sectional microstructure of LPBF produced Ti-6Al-4V samples, (a) 80 W – 400 mm/s; (b) 100 W – 400 mm/s

The measurement of columnar grain size for the bulk Ti-6Al-4V materials was carried out by using the linear intercept method following ASTM [95] (Fig. 2.20). Generally, the grain size increases with increased energy density [80, 83, 157]. It was reported by Vilaro et al [80] that the large columnar grains are approximately 150 μm wide at a fully dense parameter setting of 160 W and 600 mm/s, a hatching space of 200 μm and the layer thickness of 40 μm , which corresponds to an energy density of 66.67 J/mm³. The samples are produced on a Trumpf LF250 machine at Poly-Shape (Villecresnes, France).

Fig. 2.20 Columnar grains measurement in Ti-6AL-4V bulk sample

Fig. 2.21 shows the relationship between the grain size and the process parameters significantly. At the same laser power level, the size of the grains increases as the scan speed decreases, while at the same scan speed level, the grains size decreases as the laser power increases. Also, it was noticed that at each laser power level, when the energy density was no more than 60 J/mm³, the grain size appears to exhibit consistent grain size of ~80-100 μm regardless of the power and scan speed levels. Furthermore, at the laser power of 60 W and 80 W, the grains sizes even exhibit a slightly decreasing trend with the increase of energy density, although such effect is not significant. The large grain size of approximately 192 μm occurs at the process parameter of 120 W and 300 mm/s, although at this level the standard deviation of the grain size also becomes significant.

Fig. 2.21 Grain size distribution by varying process parameters

2.6 Mechanical properties

Extensive research works have been done by a number of researchers on the mechanical properties for Ti-4Al-4V, which is summarized in Table 2.3. Among these results, Thijs et al. [84] and Rafi et al. [105] reported the anisotropic behavior of AM produced Ti-6Al-4V tensile coupons on transverse and lateral direction, respectively. The mechanical properties of the LPBF produced Ti-6Al-4V materials were shown to be significantly dependent on the build direction of the specimens. However, there has not presented sufficient evidence on the relationship between the mechanical properties and the building orientation for the low-level parameters of scan speed and laser power.

Table 2.3 Results of mechanical properties of Ti-6Al-4V bulk samples under different thermal formation conditions [80, 84, 89, 94, 158-164]

	SLM	EBM		Wrought	Annealed
Macro hardness (HV)	400	327 ± 2		402.5 ± 15.5	350
Micro hardness (HV)	354 ~ 410	358 ± 11	403 ± 15		
Young's modulus (E, GPa)	109.7 ± 12.5	118 ± 5		104 ± 2	112 ± 2
Yield strength ($\sigma_{0.2}$, MPa)	1112 ± 72.1	830 ± 5	1125 ± 25	790 ± 20	900 ± 20
Ultimate tensile strength (σ, MPa)	1229.5 ± 36.7	915 ± 10	1175 ± 25	870 ± 20	975 ± 25
Elongation (ϵ, percent)	5.4 ± 2.0	13.1 ± 0.4	20.5 ± 4.5	18.1 ± 0.8	13 ± 1

The hardness test is a simple way of determining the resistance of a specific material deformation [165]. From Table 2.3, it can be seen that the macro hardness values are generally larger than the microhardness. This could be attributed to the different loading and indentation range for the hardness testing. The macro hardness is used to evaluate the overall resistance ability of plastic deformation for the bulk materials on a macro scale. On the other hand, the microhardness is based on the measurement of indents made on the

pore-free surface of the testing specimen [162]. Therefore, the microhardness results can be used to observe changes in hardness on the microscopic scale, particularly specific to the small feature specimens.

Also, Table 2.3 includes different fabrication methods, which result in quite different property values. Additionally, even using the same method of fabrication, such as the EBM process, the mechanical properties of as-built EBM samples are different due to the thermal conditions during the fabrication. In the case shown in the left column, spherical Ti-6Al-4V gas atomized powder with the particle size of 45-100 μm was applied. The process was carried out in a vacuum with the beam diameter of 0.5 mm and the power gun of 4 kW [161]. In the case shown in the right column, the commercial Grade 5 Ti-6Al-4V powder was utilized. The as-built samples were fabricated by ARCAM EBM S400, which is equipped with the electron gun of 60 kV and develops an energy density in excess of 10^2 kW/cm^2 . The applied process parameters allowed for the microstructure-property variation from top to bottom [158-160].

In this study, tensile tests on the bulk Ti-6Al-4V solids with various process parameters at various build orientations and the microhardness were investigated, in order to explore the effect of low-level process parameters, such as the laser power and scan speed, and the influence of build orientations on the mechanical behaviors of the LPBF fabricated Ti-6Al-4V production.

2.6.1 Tensile property

The effect of process parameters in LPBF on the mechanical properties of the fabricated metal had been reported in many studies [53, 89, 161]. The tensile properties of

LPBF as-built Ti-6Al-4V materials in different build orientations are exhibited in Table 2.4.

Table 2.4 Tensile properties of as-built Ti-6Al-4V fabricated by LPBF process

Source	Build orientation	Young's modulus (E, GPa)	Yield strength ($\sigma_{0.2}$, MPa)	Ultimate tensile strength (σ , MPa)	Elongation (ϵ , percent)
[91]	LD	134 ± 20.49	1195 ± 19.89	1269 ± 9.57	5 ± 0.52
	TD	114 ± 3.77	1143 ± 38.34	1219 ± 20.15	4.89 ± 0.65
[80]	LD	105 ± 5	1137 ± 20	1206 ± 8	7.6 ± 2
	TD	102 ± 7	962 ± 47	1166 ± 25	1.7 ± 0.3
[94]		109.2 ± 3.1	1110 ± 9	1267 ± 5	7.3 ± 1.12
[162]		94	1125	1250	6

An anisotropy of microstructure in the longitudinal direction (LD) and transverse direction (TD) is attributed to the manufacturing defect and orientation as a function of the build direction. As it was reported in previous literature [80, 84], the ultimate tensile strength, yield strength and Young's modulus in LD are slightly higher than those in TD. Also, the ductility in LD is better than that in TD.

The tensile test was carried out at room temperature for tensile coupons with different build orientations using an Instron 5569A tensile tester machine as shown in Fig. 2.22. During the testing, the crosshead moved with a speed rate of 1.0 mm/min following the testing speed recommendation of the ASTM E8/E8M-11 [166]. In addition, a 50 kN load cell was used for all the tensile testing experiments.

Fig. 2.22 (a) Illustration of the setup of tensile test; (b) Rupture of a tensile coupon

The UTS, yield stress, elastic Young's modulus and the elongation of the tensile coupons with the build orientations of 15° , 45° , and 75° were listed in Table 2.5, 2.6 and 2.7, respectively, and the stress-strain curves of the three build orientations are shown in Fig. 2.23, 2.24 and 2.25, respectively. From the results clearly showed the effects of process parameters on mechanical properties of the materials. Notably, the parameters of Zone I in the process map correspond to materials with comparable UTS, yield strength with Table 2.3. On the other hand, the parameters of Zone II and Zone III correspond to relatively low mechanical properties. Referring to the porosity distribution map shown in Fig. 2.15, it could be reasonably assumed that the porosities are primarily responsible for the low mechanical properties by introducing stress concentrations and failure sites.

Fig. 2.23 Stress-strain curves of 15°-built tensile specimens

Table 2.5 Tensile testing results for 15°-built coupons

15°	UTS (MPa)	Stress at yield (MPa)	Young's modulus (GPa)	Strain at break (%)
60W, 300mm/s	1206.67 ± 17.00	1152.06 ± 35.45	111.1 ± 2.10	7.46 ± 1.06
60W, 400mm/s	1166.67 ± 47.84	1131.55 ± 46.52	108.9 ± 2.06	8.64 ± 1.49
60W, 500mm/s	1043.33 ± 24.94	996.26 ± 25.54	100.41 ± 1.44	3.73 ± 0.40
80W, 300mm/s	1193.33 ± 12.47	1150.65 ± 17.33	107.47 ± 0.69	12.40 ± 0.62
80W, 400mm/s	1373.33 ± 97.41	1336.07 ± 95.03	105.8 ± 1.67	9.38 ± 0.26
80W, 500mm/s	1130 ± 24.49	1091.97 ± 24.25	107.9 ± 2.47	4.85 ± 0.21
100W, 300mm/s	1120 ± 29.44	1067.81 ± 26.04	99.63 ± 2.26	3.48 ± 1.68
100W, 400mm/s	1203.33 ± 36.82	1162.36 ± 39.85	111.09 ± 0.88	2.83 ± 1.11
100W, 500mm/s	1163.33 ± 38.59	1116.34 ± 37.45	117.79 ± 0.82	3.03 ± 1.31

Table 2.6 Tensile testing results for 45°-built coupons

45°	UTS (MPa)	Stress at yield (MPa)	Young's modulus (GPa)	Strain at break (%)
60W, 300mm/s	1070 ± 66.50	1026.3 ± 34.62	109.79 ± 1.10	2.86 ± 0.61
60W, 400mm/s	1110 ± 28.67	1070.19 ± 47.80	102.62 ± 3.69	2.3 ± 1.10
60W, 500mm/s	1045 ± 38.59	1050.16 ± 31.73	99.52 ± 1.38	3.38 ± 0.19
80W, 300mm/s	1105 ± 17.00	1075.72 ± 38.49	104.9 ± 0.96	4.98 ± 0.34
80W, 400mm/s	1140 ± 37.42	1091.9 ± 42.02	104.61 ± 4.12	5.38 ± 0.82
80W, 500mm/s	1280 ± 28.67	1250.06 ± 56.84	106.23 ± 2.07	4.39 ± 0.21
100W, 300mm/s	1020 ± 20.55	1014.27 ± 17.7	96.07 ± 1.41	2 ± 0.28
100W, 400mm/s	1110 ± 16.33	1068.94 ± 20.10	108.33 ± 0.49	4.62 ± 1.64
100W, 500mm/s	1155 ± 49.22	1126.03 ± 61.43	110.83 ± 1.05	3.6 ± 0.64

Fig. 2.24 Stress-strain curves of 45°-built tensile specimens

Table 2.7 Tensile testing results for 75°-built coupons

75°	UTS (MPa)	Stress at yield (MPa)	Young's modulus (GPa)	Strain at break (%)
60W, 300mm/s	1205 ± 25	1103.03 ± 35.59	103.95 ± 1.67	8.29 ± 2.67
60W, 400mm/s	1030 ± 20	941.64 ± 55.50	106.87 ± 3.71	4.26 ± 2.98
60W, 500mm/s	655 ± 125	457.44 ± 42.56	84.93 ± 5.68	0.82 ± 0.21
80W, 300mm/s	1135 ± 45	1108.86 ± 65.87	107.5 ± 3.15	1.11 ± 0.10
80W, 400mm/s	1135 ± 65	1070.43 ± 62.11	106.09 ± 3.23	7.52 ± 1.20
80W, 500mm/s	1095 ± 15	969.24 ± 48.89	107.16 ± 0.16	1.34 ± 0.62
100W, 300mm/s	1010 ± 20	928.74 ± 67.21	90.4 ± 2.71	1.31 ± 0.23
100W, 400mm/s	1140 ± 20	1030.59 ± 58.53	106.6 ± 3.72	6.49 ± 1.73
100W, 500mm/s	1120 ± 30	1063.25 ± 59.59	107.43 ± 1.19	1.96 ± 0.17

Fig. 2.25 Stress-strain curves of 75°-built tensile specimens

Additionally, the results of the tensile testing also show that the Ti-6Al-4V tensile coupons with different build orientation exhibited anisotropic mechanical strength. The microstructure of the as-built Ti-6Al-4V specimens exhibits distinct anisotropy with the columnar grains orienting along the building direction. When the tensile stress is applied along the direction of 15°-oriented along the columnar grains, which corresponds to the tensile testing of 75°-built coupons, the resistance effect from the continuous boundary of grains through multi-layers reveals inconspicuously, and thus gives rise to a low strength. On the other hand, when the direction of tensile stress is perpendicular to the grain growing direction, the grain boundaries of coarse columnar grains result in substantial resistance effect, and thereby reflect the high strength. During the martensite transformation due to the fast cooling rate, the linear coefficients of expansion of hcp α' and bcc prior β phases are different, and therefore it is easy for the microcracking to initiate at the phase boundaries. Subsequently, the microcracking starts expanding by interior necking, and dimples are nucleated at multiple locations. As the deformation increases, the dimple walls become thinner and connect to the neighbors by tearing mode [167]. Furthermore, dislocations near the phase boundaries and grain boundaries are activated by the stress, and micropores are created, which reduce the repulse for dislocations. A flow of dislocations is moving to and accumulates at the micropores along the sliding surfaces, and eventually forms a fresh cracking initiation [88, 168, 169],

Moreover, the elongation results reveal the ductility of the materials in tensile testing. It was noticed that the strain values at the break in Zone I were generally larger than those in Zone II and Zone III. This phenomenon could be attributed to the defects in the testing specimens. On the other hand, the primary strengthen mechanism is contributed

by the martensite and the dislocation density inside the fine martensitic laths [91], which gives rise to high mechanical strength. Since the width of the martensitic lath is about 1 ~ 2 μm and the length can be up to the width of the columnar grains, this type of needle-like phase could also cause significant reduction of the ductility of the Ti-6Al4V material [105].

The fracture surfaces of the Ti-6Al-4V tensile samples are shown in Fig. 2.20. Significant differences can be observed among the specimens. In Fig. 2.20 (a) and (b), the fracture surface of the tensile coupon by 80 W and 400 mm/s (Zone I) is characterized by microscopic void and shallow dimples along with some transgranular fractures, indicating a typical ductile rupture failure mechanism mixed with brittle cleavage fracture. Fig. 2.26 presents the transgranular fracture mechanism.

Fig. 2.26 Schematic of transgranular fracture in metal [170]

Fig. 2.27 Fractographs of 15° tensile specimens under SEM, (a) Process parameter of 80 W and 400 mm/s (Zone I); (b) Magnified view of (a); (c) Process parameter of 100 W and 300 mm/s (Zone II); (d) Magnified view of (c); (e) Process parameter of 60 W and 500 mm/s (Zone III); (f) Magnified view of (e).

An increase of defect density in the specimens degrade the tensile behavior, especially for the ductility. Some specimens cracked immediately after initial yielding in tensile testing. Fig. 2.27 (c-f) displays fracture surfaces with an increased amount of defects from parameters in Zone II and Zone III. For samples with Zone II process parameters

(100W, 300mm/s), many spherical defects could be clearly observed in the specimen (Fig. 2.27 (c) and (d)) that likely contributed to both the cracking initiation and propagation. The extensive existence of dimple fracture features from the close view of the fracture surface from the samples in Zone II (Fig. 2.27 (d)) suggests that the failure mode is also a ductile + brittle mixed mode. Fig. 2.27 (e) and (f) shows the tensile coupon built by 60 W and 500 mm/s (Zone III). The unmelt regions can be identified visually. Additionally, the necking phenomenon was not observed in the tensile coupons in Zone II and Zone III.

2.6.2 Microhardness

A micro-hardness test is a means of determining the local material properties and is sometimes employed to obtain a quick estimation of the mechanical strength variabilities within the samples. The microhardness of LPBF samples is typically higher due to the rapid cooling during the fabrication process. The microhardness values of the bulk Ti-6Al-4V fabricated by LPBF were previously reported and are listed in Table 2.3. The microhardness values of bulk Ti-6Al-4V material by LPBF from literature vary 354 to 410 HV (Table. 2.3).

In this study, the microhardness test was carried out in a Shimadzu HMV-G21 micro indentation (Vickers) hardness tester. The hardness number is calculated by the applied force divided by the surface area of the indentation. In this study, a force of 4.903 N and a hold time of 15 seconds were utilized on the polished cross-section surface of the as-built Ti-6Al-4V specimens by LPBF process. For each specimen, a minimum of 8 indentations was made at random and isolated locations on the finished surface of the samples in order to obtain statistically significant results.

Generally, microhardness exhibits a significant dependency on the local porosity levels of the microstructure. Therefore, it was expected that the processing parameters have significant influence to the microhardness of these Ti-6Al-4V bulk samples. From the “micro hardness-energy density” chart in Fig. 2.28, the microhardness value of bulk Ti-6Al-4V samples exhibits an increasing trend with decreasing energy density below the energy density level of 80 J/mm^3 . When the energy density is higher than 80 J/mm^3 , the microhardness decreases with the increasing energy density levels. For samples with large and small energy density levels, it could be seen that the standard deviation of the microhardness increased, which revealed the defects start to play more significant roles in introducing quality variability in these specimens. The highest microhardness value of 380 HV was obtained at the process parameters of 120W and 500 mm/s.

Fig. 2.28 Micro Vickers hardness of as-built bulk Ti-6Al-4V specimens with various energy density

2.7 Conclusion

This chapter discussed the characterization of Ti-6Al-4V solid material under various process parameters by LPBF process. The range of process parameters, including laser power and scan speed, were decided according to the default parameter set for support structures in the EOS system, which applied lower laser power and scan speed levels than the standard process parameters for bulk Ti-6Al-4V materials.

The single track study was performed for all designed the laser power and scan speed combination. It was apparent that laser power and scan speed had a significant effect on the geometry and quality of single track melt pool. Increased laser power or decreased scan speed would result in an increased melt pool depth and width, as well as their ratio (depth/width). In order word, larger energy density caused broader and deeper melt pool. At higher energy density level, in some cases, keyhole effect was observed. The investigation of the single tracks establishes a good preliminary understanding about the features fabricated under low energy level process parameters, which could be used to guide the further study of fabrication of thin-feature structures.

The effect of process parameters on the porosities of Ti-6Al-4V solids was significant. The variation of process parameters was directly related to the energy change to the Ti-4Al-4V defect generation during the LPBF process. Fully dense zone (Zone I), over melting zone (Zone II) and incomplete melting zone (Zone III) in the parameter window were responsible for the porosity distribution and the corresponding properties.

Microstructural evolution is primarily a function of the cooling rate. The material in LPBF is processed at high cooling rates. Martensitic α' phase in Ti-6Al-4V parts resulted

in high strength. The tensile properties of Ti-6Al-4V coupons in Zone I exhibited better mechanical performance than those in Zone II and Zone III. The mechanical strength exhibits strong anisotropy, with the build orientation exhibiting lowest mechanical strength and highest ductility. For example, the strengths of low angle (15°) tensile coupons exhibit an average of ~50 MPa increase compared to that of high angle (75°).

In addition, the mechanical properties of the LPBF Ti-6Al-4V bulk materials also exhibit a significant dependency on the process parameters, which appeared to be mainly contributed by different defect porosities. A few high-porosity points occurred at large and small energy densities, which was considered to be related to the porosity level. The highest microhardness value of 380 HV was observed at the process parameters of 120W and 500 mm/s.

CHAPTER 3

COUPLED EFFECTS OF PROCESS PARAMETER AND SPECIMEN SIZE OF TI-6AL-4V THIN STRUCTURES FABRICATED BY LPBF

3.1 Introduction

“When modern man builds large load-bearing structures, he uses dense solids; steel, concrete, glass. When nature does the same, she generally uses cellular materials; wood, bone, coral. There must be a reason for it.” -- Metal Foams: A Design Guide [5].

In general, natural cellular structures are more stochastic existing commonly, such as stem piths of plants, honeycombs, bones, and so forth, due to their high performance-to-weight ratios. Man has made use of these natural cellular structures for centuries. As historical records, the cork was used for bungs in wine bottles in Roman times Hooke [171], and was observed under a microscope and found out “cells” structures by Robert Hooke in 1665. In modern engineering times, artificial cellular structures are highly demanded in many applications, such as light-weight aerospace and aircraft products, which are experienced from natural cellular structures, created and developed in order to be easy to fabricate. Especially additive manufacturing overcomes the limits of conventional manufacturing to realize the fabrication of cellular structures with higher degrees of flexibility.

Cellular structures are highly desired in many applications such as aerospace and aircraft products and artificial biomedical prosthesis due to their high performance-to-weight ratio, lightweight, and high energy absorption. Cellular structures usually contain small-dimensional features and complex geometries, therefore are often infeasible for the conventional manufacturing technologies. Additive manufacturing (AM) overcomes the limits of conventional manufacturing and is capable of producing structures that exhibit a high level of geometrical complexities.

Laser Powder Bed Fusion (LPBF) is a type of metal powder bed fusion AM process, which utilizes laser beam to fuse powder into predetermined geometries layer by layer selectively. In the LPBF, the characteristics of the processes are largely determined by the thermal behavior during the fabrication [69, 172]. Commercial LPBF systems usually have standard process parameters for typical metal AM materials that could produce high-density bulk parts (>1 mm). There exists extensive literature about the process investigations on the LPBF process. Numerous studies were reported on the optimization of process parameters for material properties and part qualities, such as surface morphology [67, 173], microstructural evolution [82, 84, 94, 131, 132, 145, 174], defect detection [53, 95, 96, 154, 175] and mechanical characterizations [80, 162]. However, the standard parameter sets might give rise to low build rate for cellular structures with small-scale individual thin struts, owing to the significant amount of “jumps” in the laser movements. Since the cross sections of the thin struts are small and thermally insulated by the surrounding powder, the input energy from standard parameters could potentially result in excessive energy densities. Additionally, the exposure strategy also has a significant influence on the part qualities. For bulk part fabrication, the contour scan helps to improve

the surface roughness, while the hatch scan plays a most significant role in determining the mechanical properties of the bulk materials. On the other hand, for thin strut features the influence of contour scan becomes more significant due to the small cross-sectional dimensions and consequently large peripheral-to-area ratios. For the EOS LPBF system used in this study, additional scan theme termed edge scan that could facilitate the fabrication of certain thin features further increases the complexity of the exposure setup.

Current there exist quite a few studies on the investigation of process characteristics and mechanical properties of thin-feature structures by AM. Lots of work has contributed to the understanding of the AM manufactured thin-feature structures on the mechanical behaviors at the structural level [176-179]. Mechanical tests, such as tensile, compressive and bending testing for cellular structures, were conducted by many research groups. Maskery et al. designed body-centered-cubic (BCC) lattice structures into the gauge of the tensile coupons and investigated the relative lattice cell size and the effective modulus [176]. Lots of works [33, 180-182] have focused on the compressive strength of LPBF produced foams and lattice structures. In applications such as biomedical bone implants, in the benefits of the porous designs have been demonstrating in many cases [183-185]. The surface morphology and internal microstructure of the cellular structures were investigated to evaluate the fabrication in selective laser melting (SLM) and electron beam melting (EBM) [78, 153, 174, 186]. However, LPBF is subject to specific process limits, such as fabricatable feature size and build angle for the specific materials and machines, and thus these limits bring constraints on the fabricatable geometry. Mazur et al. did detailed work on single struts and reported a desirable set of minimum manufacturable strut angle and diameter using Ti-4Al-4V in SLM250 HL machine (SLM Solutions) [177]. The

geometrical shape and the properties of the structure composed of thin-feature components can be affected directly by the process parameters [187, 188]. Although prior work has contributed to the understanding of characteristics of LPBF produced cellular structures, insufficient works have been reported on the systematic study of the material characteristics of the thin-feature struts typical to the cellular structures, especially with low-level process parameters (scan speed and laser power) in LPBF. Therefore, in the current study, the effect of process parameters and geometry parameters on the material characteristics of the thin struts fabricated via LPBF were investigated in detail. Quantitative analysis was carried out to establish a model for the prediction the optimum process parameters of cellular structures by fitting the evaluated trending results.

3.2 Experimental design

3.2.1 Material

EOS pre-alloyed Ti-6Al-4V powder supplied by LPW was used in this study to fabricate the thin feature structures. Fig. 3.1 shows the EOS Ti-6Al-4V powder morphology and its particle size distribution. The particle size analysis presented that the EOS Ti-6Al-4V powder has a size distribution with a range between 22.47 μm ($D_{v0.1}$) and 48.71 μm ($D_{v0.9}$) and a mean diameter of 34.26 μm . The particle morphology is spherical.

Fig 3.1 EOS Ti-6Al-4V powder morphology and particle size distribution

3.2.2 AM process

In EOS INT M 270 system, there are three basic scanning operations, which are contour, hatch and edge scans. Contour scan occurs along the outer skin of the scanned region. Hatch operation fills the interior area inside the contour using specific filling methods. Edge scan is used to realize single-line scan for specific features such as sharp corners in a small-dimension structure. These scanning operations must be pre-defined during the process setup and can be used either individually or in combinations.

Another process control parameter that could be potentially important for the thin feature fabrication is the beam offset (BO) of the laser beam. The beam offset is defined as the spacing between the center of the laser beam and the nominal contour of the scanned area. Generally, beam offset reflects the effective laser beam size and is largely determined by the laser optics. The value of laser beam size offset is usually pre-set and fixed for each type of material used by the system. In addition, in the EOS system, there is another type of beam offset that is defined as “in-process” beam offset (Fig. 3.2). In-process beam offset

is able to adjust further the offset of the contour scan path, which allows for additional flexibility with the process optimization.

Fig. 3.2 Schematic diagram of the beam offset [74]

Contour Scan Strategy

The contour scan applies the laser beam along the contours of the scanned areas in each layer. In EOSINT M 270 system, The laser beam focus is automatically switched between two pre-defined values during the exposure using a dual focus system. A fine focus is typically used for accurate exposure of contours and a coarse focus for fast exposure of hatching scan in a larger area.

During the setup of the contour scan parameters, the control software (PSW3.5 M270D) also provides additional control flexibility such as the exposure sequence for skin/core scan and contour scan as well as numbers of repeated scans (up to 4 times). One of the primary purposes of the contour scan for bulk geometry is to ensure proper part surface qualities.

Edge Scan Strategy

Edge scan operation cannot be applied individually and is instead always set up with contour scan. For the setup of the edge scan, several unique parameters are employed.

As is shown in Fig. 3.3, edge factor (EF) defines the stopping points of the edge scan from the outermost points on the nominal contour. The exposure of points (EOP) is defined as the product of edge factor (EF) and beam offset (BO) by the following equation:

$$EOP = EF \times BO. \quad (3-1)$$

Fig. 3.3 Illustration of exposure points

Another factor that determines the exposure of points for the edge scan is a threshold (TH). The value of threshold multiplying beam offset is defined as a critical point to edge factor. At the triangular contour tip point P (shown in Fig. 3.4), the exposure of point is the distance between the point P and the center of the laser beam. Considering the effective radius of the laser beam (r), which is dependent on the materials type, the distance L between the actual boundary of the scan and the nominal contour tip point P is equal to the exposure of point deducts the radius of the laser beam, a.k.a.

$$L = EOP - r. \quad (3-2)$$

Fig. 3.4 Illustration of exposure to small triangle tip

If the distance (L) exceeds the value of $TH \times BO$, the edge factor is automatically set as $EF=1.45$. Otherwise, the edge factor is taken as the actual value set by the user. Therefore, the EF value is a variable that is dependent on TH, BO and r. It can be expressed as

$$\left\{ \begin{array}{ll} EF = 1.45, & \text{when } EF > TH + r \times BO \\ EF = \text{the value being set,} & \text{when } EF \leq TH + r \times BO \end{array} \right. \quad (3-3)$$

However, during the process design, the distance (L) is a more functional parameter from process evaluation perspective. Therefore Eq.(3-4) can be used for the setup of the edge factor as

$$\left\{ \begin{array}{ll} EF = 1.45, & \text{when } TH \times BO < L \\ EF = \text{the value being set.} & \text{when } TH \times BO \geq L \end{array} \right. \quad (3-4)$$

For example, given that the distance (L) from the actual scan stop point to the nominal contour is 0.1mm. If the value of $TH \times BO$ is larger than 0.1mm, the EF value will be set as 1.45 automatically by the software regardless of the user's inputs. The distance of the exposure point will be set as 1.45 times of the total beam offset. On the other hand, when $TH \times BO$ is smaller than 0.1mm, the EF value takes the number set by the user. Fig. 3.5 shows both of the cases.

Fig. 3.5 Rule of the EF value taken

However, through the preliminary experimental investigation, it is noticed that the distance of the exposure of point L is always 1.45 times of beam offset regardless of the values of TH and BO. Therefore, it was decided that the distance of exposure of point L cannot be assumed as a fixed number. Without further information into the setup of L available, it was assumed that for all the thin feature builds the EF was set as 1.45.

Additionally, it should be noted that the level of exposure of the points can also be defined either by the minimum radius factor (mRF). The minimum radius factor is expressed as a function of the radius of the laser beam (r), shown as Eqn. 3-5.

$$EOP = f(mRF) \quad (3-5)$$

Fig. 3.6 Illustration of radius factor

When the minimum radius factor is 1, the edge scan can ensure that the laser beam does not directly heat any areas outside of the model contour. On the other hand, when the minimum radius factor is 0, the effect of the diameter of the laser beam is completely ignored, and the center of the laser beam would move along the part contour of the layer, which would result in an oversized heating area. The minimum radius factor is a value between 0 and 1.

Hatch scan strategy

In the EOS INT M270 system, the minimum threshold dimension for the hatch scan to actually take place is 250 μm [59]. In the fabrication of the thin struts with the lateral thickness smaller than 250 μm , the hatch scan exposure will not occur regardless of the setup option. When the lateral thickness of the struts is larger than the threshold dimension, the hatch exposure follows the customized setup by the operator.

3.2.3 Geometry accuracy and microstructure observations and mechanical properties measurements

The surface topology of the strut specimens was observed using Olympus MX51 Optical Microscope (OM) and Zeiss SUPRA 35 scanning electron microscope (SEM). The specimens were consequently polished using silicon carbide sandpapers and etched with Keller's reagent. The microstructure of the sectioned specimens was observed using Olympus MX51 OM and Zeiss SUPRA 35 SEM as well. The micro tensile test was conducted under a Shimadzu EZ-Test material testing apparatus. The microhardness test used a Shimadzu HMV-G21 micro-indentation (Vickers) hardness tester.

3.3 Sample fabricatability

Following the experimental design mentioned in Section 2.4, thin feature struts using the material of Ti-6Al-4V (LPW) were fabricated in an EOS LPBF system under various process parameter combinations.

In the evaluation of the fabrication quality and accuracy of thin feature structures under different processing parameters, the process parameter was selected carefully. Since in the EOSINT M270 system the default process parameter for the Ti-6Al-4V support structure is 80 W and 400 mm/s [59], in this study $\pm 25\%$ of laser power and scan speed

based on the default values were selected for the fabrication of thin struts. The exposure type and beam offset were also considered in the design of experiments. The diameters of the struts were pre-defined between 0.05 ~ 0.6 mm. Besides, the hatching space of 100 μm and layer thickness of 30 μm were set in the study. The experiments were carried out with and without beam size offset of 40 μm , while the in-process beam offset was 0 for all the samples.

A total of 4536 thin feature struts that followed full factorial design in Table 3.1 with six replicates was fabricated in LPBF. The thin-strut specimens were designed with varying diameters, and orientations using SolidWorks and the corresponding .stl files were converted and loaded to the EOS M270 machine for fabrication purpose. Afterwards, the strut specimens were evaluated for quality by measuring the as-built dimensions quantitatively and summarizing the building capacity statistically. The results were consequently stated in Chapter 3 and would be indexed as a database by distinct process conditions.

Table 3.1 Design of experiment for thin feature struts

Factor	Level
Laser Power (W)	60, 80, 100
Scan Speed (mm/s)	300, 400, 500
Build Orientation ($^{\circ}$)	15, 45, 75
Scanning Strategy	Contour + Hatch, Contour only
Beam Offset (μm)	0, 40
Diameter (mm)	Varies between 0.6 ~0.05 mm

Fig. 3.7 shows a build of the thin struts on the substrate. Fig. 3.8 shows a representative strut tip exposed by contour scan only, which results in a hollow interior structure.

Fig. 3.7 Thin feature struts produced by LPBF using EOS M 270

Fig. 3.8 Fabrication of 0.5 mm thin struts at contour exposure

3.3.1 Fabricatability rate

The thin struts were fabricated in different process parameters (laser power and scan speed levels) in LPBF process. It was found that the fabricatability rate of the thin feature struts varied significantly with the build orientation. The fabricatability of the thin

struts at various build orientations can be identified visually in the two exposure types from Fig. 3.9 (a) and (b), respectively.

Fig. 3.9 Fabrication of thin struts in LPBF process at various build orientations with the exposure type of (a) contour and (b) contour/hatch combination

With larger build orientation, fabrication failure occurred more frequently for all levels of process conditions. Most of the struts at the build angle of 15° were built successfully. As the build orientation increases, the fabricatability rate of the struts decreases.

Fig. 3.10 Statistical fabricability rate of the thin struts produced in typical process parameter sets

The fabricatability rate of the thin struts is summarised by build orientation statistically. The corresponding data in each process are shown as histograms in the graphs in Appendix A. Fig. 3.10 shows 4 typical fabricatability rates to illustrate the orientation effect in the different zones of the process map. The processes that weren't built successfully are marked as “×”. It is also noted that the half-built struts were also

considered as failed. It was found the fabricatability in Zone I and Zone III was higher than that in Zone II. It was believed that a high energy input would result in residual stress. This might cause the deformation of the scanned parts that might extrude above the powder bed level and broken away by the recoater blade, which eventually cause destructive failure.

3.3.2 Geometry accuracy and quality evaluation

The geometry accuracy and quality of the struts were significantly influenced by the build orientation. Fig. 3.11 showed the fabricated samples of 15° and 75° struts. The down-facing side of the 15° struts exhibits significant defects, although most samples with different strut dimensions were built successfully (shown in Fig. 3.11 (a)). The 75° struts exhibit relatively small down-facing surface defects, but the fabricatability rate was lower as stated previously (shown in Fig. 3.11 (b)).

Fig. 3.11 Fabrication of 15° and 75° struts

3.3.2.1 Exterior characteristics of thin struts

One important aspect of the evaluation of the strut quality is the exterior characteristics, which is significant due to the large surface area ratios of these structures. Thus, a close check at the surface of the thin struts was carried out under an SEM. As shown in Fig 3.12 (a), the rough down-facing surface could be clearly observed on the 15° struts due to the thermally-induced adhesion of unmelt powder, while the upper surface

was relatively smooth. Hence, it was identified that in the LPBF process the surface quality is limited due to powder attachment on the surface of struts. It was reported that unmelted spherical particles were sintered and visible on the strut surface by LPBF due to inefficient heat input on the edge borders of the samples [189, 190]. On the other hand, small splashed droplets of the melted metal were also presented to attribute to the attached sintered particles because of melting instability and the reduction of surface energy in a short length scale [144, 191]. Therefore, the geometry accuracy of 15° struts was highly compromised due to the rough down-facing surface. Similar characteristics were also observed for the 45° struts, which is shown in Fig. 3.12 (b), but the number of unmelted particles was much reduced and not affected the geometry accuracy of the struts. On the other hand, for 75° struts, the roughness of the upper and lower surfaces was not significantly different, but the powder adhesion on the upper surface becomes slightly more significant.

Fig. 3.12 SEM micrographs of surface topologies of thin struts under various build orientation, (a) 15° (80W, 400 mm/s); (b) 45° (80W, 400 mm/s); (c) 75° (80W, 400 mm/s); (d) 15° (100W, 500 mm/s); (e) 45° (100W, 500 mm/s) and (f) 75° (100W, 500 mm/s)

In order to better understand how the scanning exposure influences the thin feature struts, the tips of the as-built struts were observed under SEM. Fig. 3.13 shows the SEM micrographs of the tips of the thin struts fabricated using the combined scan exposure of contour and hatch and with the beam offset of 40 μm . As previously discussed, when the nominal diameter of the strut was smaller than 0.25 mm, the hatch scan would not be enabled during the scanning. From the samples of 0.6 mm diameter, the pattern of the hatch scan line is clearly discernable across the top layer as shown in Fig. 3.13 (a). Below the threshold dimension of hatch scan activation, such as the samples with diameters of 0.2 mm, 0.1 mm and 0.05 mm, the hollow tubular feature from only the contour exposure can be clearly seen from the tips of the struts as shown in Fig. 3.13 (b-d). In addition, it was visually observed that the actual diameters of the struts with nominal dimensions of 0.1 mm and 0.05 mm were close. This was attributed to the resolution limit of the laser beam focus dimensions in the EOS system and was investigated quantitatively in further discussions.

Fig. 3.14 further illustrates the transition of feature characteristics near the hatch scan threshold values. The struts with diameters of 0.4 mm and 0.3 mm exhibited as hollow tubes, which did not occur in the 0.2 mm diameter strut.

Moreover, for the thin-feature struts, it was expected that the staircase effect is more obvious, which was clearly verified in the microscopic graphs of this study. However, it was also noticed that for the low build orientation (such as 15° in this study) struts, the successive layers were not aligned horizontally. There exists a slight collapse of each layer towards the direction of strut extension as shown in Fig. 3.15. This can be explained qualitatively by the combination of the rapid movement of the high-intensity energy source and the consequent rapid moving thermal gradient during the melting and solidification

processes, as well as the lack of a support structure, which results in the penetration of melt pool into the metal powder bed.

Fig. 3.13 SEM graphs of the thin strut tips with various nominal dimension under contour and hatch exposure, (a) 0.6 mm; (b) 0.3 mm; (c) 0.1mm and (d) 0.05 mm (Magnification: (a) 150 $\times$; (b-d) 350 $\times$)

Fig. 3.14 SEM graphs of the thin strut tips with various nominal dimension under contour exposure, (a) 0.4 mm; (b) 0.3 mm and (c) 0.2mm (Magnification: 150 ×)

Fig. 3.15 Layer collapse in the thin strut

3.3.2.2 Geometry accuracy analysis of thin struts

The dimensions of the thin feature struts were measured with an optical microscope (Olympus MX51). Due to the surface sintering effect, the dimension of the thin feature was not completely uniform. The smallest section became a critical dimension for the strength of the struts. This smallest dimension was measured at five randomly chosen locations on a strut sample as shown in Fig. 3.16, and three struts (if built successfully) were measured for each process/geometry conditions. The measurement locations on the struts were selected largely randomly, however, care was taken to ensure that the smallest and largest dimension in the field of view in the photograph were also included in the selections (Shown in Fig. 3.16). The averaged dimension measurement results were used to evaluate the fabrication accuracy under varying process conditions. The standard deviations, relative errors and absolute errors were used to quantify the difference between the measured and nominal dimensions for the struts.

Fig. 3.16 Measurement of the thin strut dimensions

The analyzed results are plotted in Fig.3.17. In general, the measured dimensions were larger than the nominal dimensions, and the difference between measured and nominal values are shown as deviation bars in Fig. 3.17. The relative errors exhibit increasing trend with the decrease of the strut diameters, and the errors become much more significant at diameters smaller than 0.2 mm. The beam offset also has significant effects on the dimensional accuracies. The struts that were produced with beam offset of 0 were generally larger than those with beam offset of 40 μm . On the other hand, the effect of exposure type was not significant. Also, the hatch exposure did not appear to introduce significant differences on the actual dimensions of the struts. The similar trend with the geometrical accuracies was observed for all the other process/geometry combinations. The additional figures of similar results from the full set of experimental design are attached in Appendix B.

Fig. 3.17 Measured dimension, absolute error and relative error of 15° struts under laser power of 100 W and scan speed of 500 mm/s

ANOVA was performed to further evaluate the influence of all factors on the geometry accuracy of the thin-feature struts, and the results are shown in Table 3.2. It can be seen that the F values of diameter, beam offset, build orientation, and process parameter set are larger than the threshold ($F_{0.05, 1, 513}$, $F_{0.05, 6, 513}$, and $F_{0.05, 8, 513}$), which indicate a significance with 95% confidence level; while the F value of exposure type ($F_{0.05, 1, 513}$) indicates low significance. The most significant effect appears to be the diameter, which is followed by the beam offset, the build orientation, and process parameter set. Besides, the P values also reveals that diameter, beam offset, build orientation, and process parameter are significant factors while the exposure type is insignificant within 95% confidence level.

Moreover, it was also noticed that for the struts with the diameter smaller than 0.2 mm, the relative error increased significantly in Fig. 3.17. Thus, ANOVA was conducted by focusing on the population with small-scaled dimensions (0.05 ~0.1 mm). The results are summarized in Table 3.3. Similarly, from the F values and P values, the experimental

factors of beam offset, build orientation, and process parameter set could be considered to have a significant influence on the dimension of the small-scale struts, and the beam offset is the most significant process factor. On the other hand, it was noticed that below 0.2 mm nominal strut diameter the actual diameters of the struts become largely independent to the designs, which is attributed to the resolution limit of in the LPBF EOS system from the melt pool dimension at the certain process parameter combinations.

Table 3.2 ANOVA on geometry accuracy of thin-feature struts

Source	Degree of Freedom	Adj. Sums of Squares (SS)	Adj. Mean Squares (MS)	F-Value	P-Value
Orientation	2	0.07	0.04	26.24	0
Process Set	8	0.10	0.01	8.62	0
Beam Offset	1	0.12	0.12	84.67	0
Exposure	1	0.00	0.00	0.85	0.36
Diameter	6	8.02	1.34	951.43	0
Error	513	0.72	0.00		
Total	531	9.22			
$F_{0.05, 1, 513}=3.8601$, $F_{0.05, 6, 513}=2.1299$, $F_{0.05, 8, 513}=1.9684$					

Table 3.3 ANOVA on geometry accuracy of struts with nominal diameter below 0.2 mm

Source	Degree of Freedom	Adj. Sums of Squares (SS)	Adj. Mean Squares (MS)	F-Value	P-Value
Orientation	2	0.00	0.00	4.09	0.02
Process Set	8	0.01	0.00	6.15	0
Beam Offset	1	0.00	0.00	19.09	0
Exposure	1	0.00	0.00	0.46	0.50
Diameter	1	0.00	0.00	1.11	0.29
Error	141	0.03	0.00		
Total	154	0.05			
$F_{0.05, 1, 141}=3.9201$, $F_{0.05, 6, 141}=2.1750$, $F_{0.05, 8, 141}=2.0164$					

From the geometrical accuracy results it was observed that at a low energy density within Zone I of the process map, such as 80 W and 400 mm/s, the build quality was generally desirable, which agrees with the observations from the preliminary study [192]. Moreover, when fabricating the same strut designs, the actual dimensions of struts using a process parameter set in Zone II were significantly larger than that in Zone III, which is likely due to the different sizes of the melt pool. ANOVA was carried out with measurements on struts of 0.4 mm and the beam offset of 0 in order to investigate the effect of fabrication processes. The quantitative results in Table 3.4 show the significance of the process parameters, including nine process sets and two types of laser beam exposure.

Table 3.4 ANOVA on geometry accuracy of struts with various process parameters

Source	Degree of Freedom	Adj. Sums of Squares (SS)	Adj. Mean Squares (MS)	F-Value	P-Value
Process set	8	0.015	0.002	5.35	0.014
Exposure	1	0.000	0.000	1.39	0.272
Error	8	0.003	0.000		
Total	17	0.018			
$F_{0.05, 1, 8}=5.3177$, $F_{0.05, 8, 8}=3.4381$					

3.3.2.3 Cross-sectional feature of thin struts

Among the thin strut specimens, the 75° thin struts were more feasible for the purpose of the observation of the hollow interior than the other strut orientations, because the melt pool is penetrating downwards the previous layer and thermal flow potentially causes the particles sintered that fills the original space. Fig. 3.18 exhibits the difference of the hollow interiors in various oriented thin struts with the diameter of 0.3 mm using the process parameters of 60 W and 300 mm/s. It could be noticed that the hollow part in the

15° and 45° oriented specimens in Fig. 3.18 (a) is nearly submerged, while in the 75° specimen, the melt pool deposits after subsequent layers and affects the hollow morphology least.

Fig. 3.18 Hollow interiors in various oriented thin struts, (a) 15°; (b) 45°; (c) 75°

Therefore, to evaluate the effect of the exposure type on the geometry of thin struts, the 75° thin struts using contour exposure and contour/hatch combined exposure were sectioned perpendicular to the longitudinal direction along the strut axis. Compared to the other two build orientations, 75° struts were easier to observe the laser exposure path after sectioned. Using parameter sets that result in high fabricatability rates, investigated a 2×2 design matrix with the laser power of 60 W and 80 W, and the scan speed of 400 mm/s and 500 mm/s was used for analysis. Fig. 3.19 and Fig. 3.20 show the cross sections of the thin struts fabricated under contour/hatch combined exposure and contour-only exposure, respectively. It can be seen that the width of contour exposure was significantly influenced by the processing energy input. Among the four processes investigated, the process parameter set of 80 W and 400 mm/s resulted in the highest energy density, and the contour scan tracks also exhibited the widest sizes when the strut dimension was not less than 0.4 mm (seen from Fig. 3.20 (a)). For the struts with the diameter of 0.3 mm, in both 80 W – 400 mm/s and 80 W – 500 mm/s process parameter sets, the contour scan tracks exhibit

overlapping towards the center areas of the strut, and thus the interiors of the struts were largely melted. With the other two parameter sets of 60 W – 400 mm/s and 60 W – 500 mm/s, the struts exhibit interior voids towards the centers (shown in Fig. 3.20 (b)). When the dimensions of struts are smaller than 0.2 mm, no interior void was observed for all the process parameters (shown in Fig. 3.20 (c)). On the other hand, in Fig. 3.19, with the inclusion of the hatch exposure, the cross sections of struts with the range of diameters investigated in this study (0.1 mm to 0.4 mm) were largely macro-void free.

Fig. 3.19 Cross-sectional thin struts using contour and hatch combined exposure, (a) 0.4 mm; (b) 0.3 mm; (c) 0.2 mm and (d) 0.1 mm

Moreover, from Fig. 3.19, it was also found that under the contour /hatch combined exposure the shapes of the strut cross-sections became more irregular when the strut diameter was no larger than 0.2 mm.

Fig. 3.20 Cross-sectional thin struts using contour exposure, (a) 0.4 mm; (b) 0.3 mm; (c) 0.2 mm and (d) 0.1 mm

3.4 Material characterization

3.4.1 Porosity

As discussed earlier in Chapter 2, different process parameters generally result in variability with the material characteristics in the LPBF processes. With the thin feature

struts, microstructural defects due to lack of fusion were shaped irregular as shown in Fig. 3.21 (a) and (b). Fig. 3.21 (c) shows the keyhole defects resulted from the high intensity of laser energy, which generally exhibit spherical morphology due to the keyhole evaporation effects.

Fig. 3.21 Defects in the thin feature struts, (a) 80W and 400 mm/s, 75°; (b) 80W and 400 mm/s, 45° and (c) 100 W and 500 mm/s, 45°

The porosity was measured for the thin struts with different process/geometry parameter combinations. The corresponding process-geometry-porosity mapping for struts

are shown in Fig. 3.22. In general, the process parameters in both the over melting zone (Zone II) and the incomplete melting zone (Zone III) tend to result in higher porosities with the struts. The porosity of thin struts exhibits a decreasing trend with the reduction of the diameter of struts and the increase of build orientation. For example, from Fig. 3.22, at the same build orientation, the porosity of struts at 100 W and 300 mm/s reduced from 2% to 0.8% as the dimensional diameter decreased from 0.5 mm to 0.1 mm. The requirement for part density is highly dependent on the engineering applications. In some cases, $0.3 \pm 0.1\%$ is considered acceptable [193].

Fig. 3.22 Porosity of thin struts

Overall, the process parameter sets in Zone I result in lower porosity levels, while the parameter sets in Zone II and Zone III tend to result in higher porosities, which exhibit decreasing trends as the build orientation increases as shown in Fig. 3.22.

3.4.2 Microstructure

Due to the rapid cooling process during the fabrication in the LPBF process, the columnar grains grow along the direction of the thermal gradient, which is largely parallel to the build direction. Fig. 3.23 shows the optical micrographs of the microstructure of the 0.6 mm thin struts, and it could be observed that the columnar grains were largely aligned along the build directions that is similar to the columnar grains of bulk parts.

Fig. 3.23 Microstructure of thin struts and illustration of their sections, (a) 0.6 mm struts using parameters of 100 W and 500 mm/s; (b) 0.6 mm struts using parameters of 80 W and 400 mm/s

It was also noticed that these grains were not perfectly arranged along the build direction, but instead exhibits a slight angle towards the strut axial direction (shown in Fig. 3.24). The grain orientation tilting phenomenon was observed in samples with different process/geometry combinations, which indicated that such phenomenon might be a

common occurrence for LPBF cellular materials. This could be contributed by the direction of the thermal gradient during the strut growing.

Fig. 3.24 Tilt grain phenomenon, (a) 0.3 mm strut using parameters of 60 W and 300 mm/s; (b) 0.3 mm strut using parameters of 100 W and 500 mm/s; (c) 0.3 mm strut using parameters of 80 W and 500 mm/s; (d) 0.6 mm strut using parameters of 80 W and 400 mm/s

The measurement of columnar grain size for the thin feature struts was carried out by following the linear intercept method described in ASTM [194] as illustrated in Fig. 3.25. This method determines the average grain sizes by dividing the intercept length of the sample by the grain counts. The intercept length can be decided at a distance between both ends of the yellow dot lines as shown in Fig. 3.25. The grain intercept count is

determined by counting the number of individual grains or grain boundaries that each line is tangent to. Note that the uncompleted grains are considered as a half grains.

Fig. 3.25 Measurement of grain width

The results of measured average grain size for the struts under different process/geometry parameter conditions are shown as scatter plots in Fig. 3.26.

From the plots, it could be seen that for the thin struts with all the dimensions and orientations investigated the energy density increased due to different thermal history during the laser melting process; the columnar grain size went up with a growing trend. In addition, larger build orientation also appears to result in smaller averaged grain width. This could be explained by the smaller cross-sectional areas with larger build angle, which leads to more significant rapid cooling and consequently smaller grain widths.

On the other hand, the grain widths exhibit different sensitivity to the lateral strut size at different orientation angles. The larger build orientation angles generally correspond to more significant reduction of grain width with the decrease of the strut diameters from

0.5 mm to 0.1 mm. For example, for the build orientation of 75°, the averaged differences of grain widths between the 0.5mm and 0.1 mm samples were ~20 μm . On the other hand, for build orientation angle of 45° to the grain widths exhibit slightly decreasing trend with the decrease of strut dimensions. At 15° angle, for all the sizes of struts, the measured grain widths were quite similar.

Fig. 3.26 Columnar grain width upon varying process parameters and build orientations, (a) 0.5 mm; (b) 0.3 mm; (c) 0.1 mm

Under metallographic observations, the as-built microstructure of the thin-feature specimens by LPBF exhibit a fine acicular morphology as shown in Fig. 3.27. The needle-shaped microstructure is believed to be martensite α' plate due to large cooling rate [94, 156, 195], which is orthogonally oriented within the grains. The existence of martensitic

microstructure in the as-built Ti-6Al-4V attributes to the higher yield and ultimate strengths in LPBF than other processes, such as EBM and wrought.

Fig. 3.27 Microstructure of the thin-feature material produced by LPBF

3.4.3 Micro tensile test

3.4.3.1 Objective

The main objectives for the micro tensile test are to fully characterize the tensile mechanical properties of the thin strut specimens and to establish the process-geometry-mechanical property relationships for these thin feature structures.

3.4.3.2 Equipment setup

The micro-tensile test for the thin strut specimens was carried out under a Shimadzu EZ-Test material testing apparatus, as shown in Fig. 3.28 that is equipped with a 500 N load cell. All the specimens were loaded with a rate of 0.5 mm/min during the tensile test. As ASTM standard [166] requires that for tensile coupons with circular cross sections, the gauge length needs to be at least five times of the diameter of the cylinder and the length of the grip section needs three times of the diameter. Therefore, based on the various

diameters of the thin struts, the safety lengths of 11 times of the diameter was used as the baseline reference. Since the designed length for the struts was 10 mm, all strut specimens satisfied the requirement for testing.

Fig. 3.28 Micro tensile testing setup

For each cylinder specimen, the minimum cross section location establishes the critical conditions for fracture failure during the tensile testing. Therefore, the minimum diameter of each specimen was used as the effective diameter to determine the mechanical properties of the thin strut specimens. Before the measurement of the specimen diameter, each strut was gently rubbed by a #320 silicon carbide sandpaper to remove the partially attached powder on the exteriors. Then three random locations were selected within the gauge section, and two diameter measurements that were perpendicular to each other were taken for each location. The minimum dimensions were selected as the diameter to calculate the area of the cross-section. The cross-sectional area was assumed to be round shape. During the tensile test, the normal stress within the gage area was obtained by $\sigma = F / A$. The force values were obtained directly from the load cell. Besides, the gauge length

was measured using a caliper with a resolution of 0.01 mm when the specimen was set up in the testing apparatus.

3.4.3.2 Results and discussion

Stress-time curves were plotted instead of stress-strain curves since strain was difficult to measure in this study. In theory, the strain values during the tensile testing could be estimated by the crosshead extension when accurate gauge length extension is unmeasurable. However, in the current study the extension cannot be measured accurately. Two critical reasons were believed to affect the inaccuracy significantly. Firstly, there exists a systematic bias in addition to the regular extension. For example, the reported extension value for a 0.5 mm 15°-built strut was 1.54 mm, which would correspond a strain value of 34.45% at break. However, the actual extension for the tensile specimen was considerably below the displayed extension value, since the former could be barely observed using the caliper (resolution of 0.01 mm). Also from previous literatures, for the Ti-6Al-4V alloy produced by metal powder-bed fusion process, the strain upon initial yield ranges from 0.9 ~ 1.1% [89, 94, 105, 162, 193, 196], which is briefly summarized in Table 3.5. Therefore, it was concluded that the reported extension values did not reflect the real displacement of the specimen due to tensile stress. Secondly, there also exist “creep” in the system that cause slack during testing, which is largely attributed to the connectors such as screws. This extra displacement along the tensile direction also introduces errors in the extension measurements and is not negligible compared to the small deflection of the thin feature specimens. As Ti-6Al-4V is a high-strength $\alpha + \beta$ titanium alloy [134, 197, 198], it was noticed that even for the tensile testing of the thin feature specimens, the instant load on the gripping setup could reach up to 480 N. Such a load was sufficient to introduce the

“creep” that affect the accuracy of the extension measurement. Due to the use of fixed displacement rate during the testing, the stress-time curves can be used instead to represent the real-time mechanical performance of the thin struts in the tensile testing. Properties such as UTS and the time that the stress reaches UTS under constant loading speed, are able to be obtained in stress-time curves. As a result, stress-time curves were plotted.

Table 3.5 Strain at initial yield and corresponding information of as-built Ti-6Al-4V produced by LPBF

Ref.	Strain at initial yield (ϵ , %)	Machine/ Process parameters (if indicated)	Stress-strain curve
[94]	~1.1	LM-Q SLM in-house developed machine Laser power: 250 W Scan speed: 1600 mm/s Layer thickness: 30 μ m Hatch spacing: 60 μ m
[53]	~1.1	EOSINT M270(GmbH) Scan speed: 960 mm/s Energy density: 42 J/mm ³
[53]	~1.1	EOSINT M270(GmbH) Scan speed: 540 mm/s Energy density: 74 J/mm ³
[53]	~1.1	EOSINT M270(GmbH) Scan speed: 400 mm/s Energy density: 100 J/mm ³

[53]	1.0~1.1	EOSINT M270(GmbH) Scan speed: 1260 mm/s Energy density: 32 J/mm ³
[53]	~1.1	EOSINT M270(GmbH) Scan speed: 1500 mm/s Energy density: 27 J/mm ³
[162]	~1.1	Concept Laser M3 Linear (GmbH)
[193]	0.9~1.0	EOSINT M270(GmbH) Laser power: 195 W Scan speed: 225 mm/s
[105]	~1.1	EOSINT M270(GmbH) Vertically built
[105]	0.9~1.0	EOSINT M270(GmbH) Horizontally built
[89]	1.0~1.1	SLM Solution (GmbH) Layer thickness: 60 μm and 30 μm Energy density: 52.62 J/mm ³ and 68.47 J/mm ³ Focus offset distance: 2 mm

Iteration 1: Identify material property and cross-sectional effect on the mechanical behavior

The stress-time curves exhibit smooth non-linear curves without apparent yielding periods before tensile rupture (Fig. 3.29 shows the specimens with the process parameter of 60 W and 300 mm/s as. More curves see Appendix C), and the ruptured specimens did not appear to have any distinct local plastic deformation and the cross-sectional area reduction which is known commonly “necking” near the fracture locations. The as-built Ti-4Al-4V in LPBF consist of martensitic α' phase and hence the plastic deformation is mainly restricted to the basal and prismatic slipping systems, namely the $(0002) \langle 11\bar{2}0 \rangle$ and $(10\bar{1}0) \langle 11\bar{2}0 \rangle$. As the α' phase does not form colonies of laths to share the same orientation, the effective slip length is restricted to within single grains. When the external load increases, the increased deformation would eventually lead to transgranular rupture. Therefore, the as-built LPBF Ti-6Al-4V tensile specimens exhibit relatively low ductility.

In addition, the maximum load was able to read from the system. According to the minimum cross section area measurements, the UTS of each specimen could be calculated.

From Fig. 3.29, it could also be noticed that the time for each process parameter and strut diameter to reach UTS varies. Generally, the larger diameter resulted in a longer time for the specimens to reach to UTS under a constant loading speed.

Fig. 3.29 Stress-time curve for thin strut specimens with varying build orientations and diameters at the process parameter of 60 W and 300 mm/s

The results of ultimate tensile strength (UTS) for thin strut specimens with the build orientation of 15° were listed in Table 3.6. It was noticed that the UTS under all the evaluated process parameters were significantly larger compared to the bulk specimens in the previous tensile testing. Besides, the UTS exhibits large variability at various diameters ranging from 0.6 mm to 0.2 mm. Therefore, the cross-sectional dimension is considered to be a significant factor to the UTS of the struts. For a single strut, the minimum cross-section is the critical characteristic that determines the tensile properties. Due to the large roughness of the strut surface, the minimum cross-sectional area is difficult to be identified and measured by a caliper accurately. Therefore, the calculated UTS tends to be underestimated. Besides, the cross-section of the strut specimens is typically not in round shape, which further introduces inaccuracy in the predictions. Moreover, based on the previous discussion on the porosity in various process parameters, various defects tend to act as crack initiations under the tensile stress and eventually cause the reduction of UTS.

The tensile properties of metals are highly dependent on the grain orientation, grain size and microstructural defects[199]. For Ti-6Al-4V, rapid cooling rate generally favors the generation of the acicular martensitic structure as opposed to the stable two-phase α and β structure. The presence of fine acicular microstructure generally improves the tensile property because the existence of the crack path deviations and bifurcations in the microstructure [91].

Table 3.6 Ultimate tensile strength (UTS) for different process parameters vs. nominal diameters (15°)

15°	0.6mm	0.5mm	0.4mm	0.3mm	0.2mm
60W, 300mm/s	1303.35 ± 11.85	1198.95 ± 26.75	804.26 ± 27.66	1266.34 ± 53.19	N/A
60W, 400mm/s	1219.18 ± 45.78	1196.41 ± 60.31	724.52 ± 26.42	1431.9 ± 44.06	789.41 ± 37.49
60W, 500mm/s	1168.35 ± 26.85	N/A	677.14 ± 12.24	1526.47 ± 42.79	N/A
80W, 300mm/s	1700.95 ± 68.65	N/A	602.67 ± 21.17	1672.68 ± 73.98	1229.72 ± 68.56
80W, 400mm/s	N/A	1213.72 ± 20.22	1044.25 ± 18.22	1462.04 ± 67.24	812.24 ± 82.45
80W, 500mm/s	1330.11 ± 41.61	1282.53 ± 21.73	555.94 ± 15.58	1526.27 ± 51.60	N/A
100W, 300mm/s	1144.21 ± 2.51	N/A	1377.3 ± 18.83	N/A	N/A
100W, 400mm/s	1356.32 ± 30.32	1451.93 ± 63.83	747.97 ± 30.25	1713.7 ± 86.74	N/A
100W, 500mm/s	N/A	803.63 ± 55.23	1194.65 ± 58.36	1253.54 ± 60.40	N/A

As previously discussed, as it was difficult to estimate tensile strain reliably, other methods must be considered for the estimation of the elastic modulus. From the previous literature in Table 3.2 about the macro tensile test for the standard tensile coupons, the LPBF Ti-6Al-4V usually exhibits a strain of ~1.1% at the end point of the elastic period. Therefore, this strain value was used to estimate Young's modulus in this study. The calculated Young's modulus under each process parameter set and nominal diameter design are listed in Table 3.7. It could be noticed that the values of Young's modulus exhibited discrepancy for each case, and for some specimens, the discrepancies appear rather large. Young's modulus was very sensitive to the tiny change in the cross-section area, and therefore was not able to be measured accurately in the current testing system.

Table 3.7 Young's modulus for different process parameters vs. nominal diameters (15°)

15°	0.6mm	0.5mm	0.4mm	0.3mm	0.2mm
60W, 300mm/s	118.49 ± 1.08	109 ± 2.43	73.11 ± 2.51	115.12 ± 4.84	N/A
60W, 400mm/s	110.83 ± 4.16	108.76 ± 5.48	65.87 ± 2.40	130.17 ± 4.01	71.76 ± 3.41
60W, 500mm/s	106.21 ± 2.44	N/A	61.56 ± 1.11	138.77 ± 3.89	N/A
80W, 300mm/s	154.63 ± 6.24	N/A	54.79 ± 1.92	152.06 ± 6.73	111.79 ± 6.23
80W, 400mm/s	N/A	110.34 ± 1.84	94.93 ± 1.66	132.91 ± 6.11	73.84 ± 7.50
80W, 500mm/s	120.92 ± 3.78	116.59 ± 1.98	50.54 ± 1.42	138.75 ± 4.69	N/A
100W, 300mm/s	104.02 ± 0.23	N/A	125.21 ± 1.71	N/A	N/A
100W, 400mm/s	123.3 ± 2.76	131.99 ± 5.80	68 ± 2.75	155.79 ± 7.89	N/A
100W, 500mm/s	N/A	123.3 ± 5.02	108.6 ± 5.31	113.96 ± 5.49	N/A

Fig. 3.30 shows the typical SEM fractographs of thin-strut specimens fabricated by LPBF with varying process parameters. Overall, the influence of process parameters on the fracture surfaces was significant. For samples with the process parameter of 80 W and 400 mm/s in Zone I as shown in Fig. 3.30 (a), the fracture surface of the tensile specimens exhibits a rough and layered topology, and it could be speculated that the microcracks initiated at the outer surface [200]. Cleavage planes could be seen clearly in Fig. 3.30 (b), and the fracture surface generally exhibits transgranular fracture which is the typical characteristic of ductile fracture. The dimple network indicated a mixed mode of brittle and ductile failure mechanisms. The fractograph of the specimen fabricated at 100 W and 300 mm/s (Zone II) exhibits less water-flow like the pattern on the fracture surface and

apparently more defects (Shown in Fig. 3.30 (c)). From Fig. 3.30 (d), a number of spherical defects in the specimen are clearly visible, which became the crack initiation regions that caused the premature fracture. The extensive existence of the defects in the specimens contributes to the lower tensile properties and especially the lower ductility. The fractograph of the thin-strut specimen fabricated at 60 W and 500 mm/s (Zone III) exhibited smooth topology as shown in Fig. 3.30 (e). Similar to the second tensile specimens in Zone II, many defects caused by incomplete fusion were visible in the specimen in irregular shapes. From the magnified view in Fig. 3.30 (f), a network of shallow dimples could be observed around the defects, which indicated shear deformation that is coupled with some plastic deformation.

Fig. 3.30 Tensile fractographs of 15° thin-strut specimens under SEM, (a) Process parameter of 80 W and 400 mm/s (Zone I); (b) Magnified view of (a); (c) Process parameter of 100 W and 300 mm/s (Zone II); (d) Magnified view of (c); (e) Process parameter of 60 W and 500 mm/s (Zone III); (f) Magnified view of (e).

Iteration 2: Identify the effect of build orientation on the mechanical behavior

The UTS and Young's moduli for specimens with the build orientations of 45° and 75° are presented in Table 3.8, 3.9, 3.10 and 3.11, respectively.

Table 3.8 Ultimate tensile strength (UTS) for different process parameters vs. nominal diameters (45°)

45°	0.6mm	0.5mm	0.4mm	0.3mm	0.2mm
60W, 300mm/s	1353.93 ± 47.81	1270.19 ± 78.56	1054.53 ± 42.79	998.67	935.39 ± 28.49
60W, 400mm/s	1394.45 ± 59.28	1199.75 ± 51.56	1590.22 ± 76.04	1158.29	1486 ± 79.03
60W, 500mm/s	1252.11 ± 58.00	971.96 ± 73.90	1299.58 ± 71.09	1046.92	1406.42 ± 53.60
80W, 300mm/s	1195.59 ± 75.09	1045.33 ± 5.17	1480.1 ± 29.86	1446.27	1467.72 ± 63.92
80W, 400mm/s	1582.41 ± 60.40	1157.73 ± 66.53	1288.13 ± 73.55	1219.65	1036.08 ± 114.04
80W, 500mm/s	1392.66 ± 63.84	1242.77 ± 70.11	1723.6 ± 57.34	1097.32	1387.63 ± 114.04
100W, 300mm/s	1321.6 ± 61.41	N/A	N/A	1404.84	N/A
100W, 400mm/s	1195.17 ± 69.25	1336.35 ± 77.21	1438.36 ± 61.87	N/A	N/A
100W, 500mm/s	1481.74 ± 28.21	1382.38 ± 85.63	1433.66 ± 30.04	N/A	1238.46 ± 39.42

Table 3.9 Ultimate tensile strength (UTS) for different process parameters vs. nominal diameters (75°)

75°	0.6mm	0.5mm	0.4mm	0.3mm	0.2mm
60W, 300mm/s	123.08 ± 4.35	115.47 ± 7.14	95.87 ± 3.89	90.79	85.04 ± 2.59
60W, 400mm/s	126.77 ± 5.39	109.07 ± 4.69	144.57 ± 6.91	105.3	135.09 ± 7.18
60W, 500mm/s	113.83 ± 5.27	88.36 ± 6.72	118.14 ± 6.46	95.17	127.86 ± 4.87
80W, 300mm/s	108.69 ± 6.83	95.03 ± 0.47	134.55 ± 2.71	131.48	133.43 ± 5.81
80W, 400mm/s	143.86 ± 5.49	105.25 ± 6.05	117.1 ± 6.69	110.88	94.19 ± 6.81
80W, 500mm/s	126.61 ± 5.80	112.98 ± 6.37	156.69 ± 5.21	99.76	126.15 ± 10.37
100W, 300mm/s	120.15 ± 5.58	N/A	N/A	127.71	N/A
100W, 400mm/s	108.65 ± 6.30	121.49 ± 7.02	130.76 ± 5.62	N/A	N/A
100W, 500mm/s	134.7 ± 2.56	125.67 ± 7.78	130.33 ± 2.73	N/A	112.59 ± 3.58

Table 3.10 Young's modulus for different process parameters vs. nominal diameters (45°)

45°	0.6mm	0.5mm	0.4mm	0.3mm	0.2mm
60W, 300mm/s	123.08 ± 4.35	115.47 ± 7.14	95.87 ± 3.89	90.79	85.04 ± 2.59
60W, 400mm/s	126.77 ± 5.39	109.07 ± 4.69	144.57 ± 6.91	105.3	135.09 ± 7.18
60W, 500mm/s	113.83 ± 5.27	88.36 ± 6.72	118.14 ± 6.46	95.17	127.86 ± 4.87
80W, 300mm/s	108.69 ± 6.83	95.03 ± 0.47	134.55 ± 2.71	131.48	133.43 ± 5.81
80W, 400mm/s	143.86 ± 5.49	105.25 ± 6.05	117.1 ± 6.69	110.88	94.19 ± 6.81
80W, 500mm/s	126.61 ± 5.80	112.98 ± 6.37	156.69 ± 5.21	99.76	126.15 ± 10.37
100W, 300mm/s	120.15 ± 5.58	N/A	N/A	127.71	N/A
100W, 400mm/s	108.65 ± 6.30	121.49 ± 7.02	130.76 ± 5.62	N/A	N/A
100W, 500mm/s	134.7 ± 2.56	125.67 ± 7.78	130.33 ± 2.73	N/A	112.59 ± 3.58

Table 3.11 Young's modulus for different process parameters vs. nominal diameters (75°)

75°	0.6mm	0.5mm	0.4mm	0.3mm	0.2mm
60W, 300mm/s	85.6 ± 2.85	N/A	116.65 ± 5.71	60.15 ± 3.29	N/A
60W, 400mm/s	114.82 ± 6.55	52.73 ± 1.48	135.5	54.18 ± 3.42	111.8 ± 8.79
60W, 500mm/s	110.97 ± 3.89	108.3 ± 5.43	96.28 ± 0.90	N/A	98.87 ± 3.52
80W, 300mm/s	72.49 ± 3.54	92.8 ± 0.83	129.54 ± 3.99	N/A	N/A
80W, 400mm/s	62.21 ± 1.33	113.09 ± 0.66	143.96 ± 7.45	88 ± 10.47	N/A
80W, 500mm/s	126.46 ± 1.01	128.38 ± 1.29	148.38 ± 7.25	80.28 ± 1.74	N/A
100W, 300mm/s	108.37 ± 3.58	N/A	N/A	N/A	N/A
100W, 400mm/s	71.06 ± 1.92	107.99 ± 4.15	148.23 ± 3.51	86.07 ± 2.13	N/A
100W, 500mm/s	98.67 ± 5.66	75.94 ± 3.14	154.05 ± 14.21	N/A	N/A

Similar to the 15° thin-strut specimens, from the observation of the rupture of the 45° and 75° thin-strut tensile specimens in the three process parameter zones (I, II and III), and the “necking” phenomenon was not apparent.

However, significant anisotropic material properties were observed in the micro-tensile testing. The epitaxial columnar grains during the fabrication grow through multi-layers under different building orientations gave rise to varying degrees of resistance effect, and therefore results in the different ultimate tensile stress and ductility in the testing. In addition, the variation in tensile properties could also be partially attributed to the defects as a function of loading axis [80]. Since the build direction and the longitudinal axis of the thin strut forms an angle of θ , many defects tend to exhibit layer-wise patterns that also

form an angle of θ with the loading axis. In this case, the low-angled built tensile specimens with defects are stronger than the vertically built specimens. Moreover, the defects cause the microstructural discontinuity, which causes the grain orientation change between the successive layers and therefore makes the interface weaker [91]. In this case, the nucleation and opening of defects in the high-angled built specimens are faster at the same tensile stress level.

Fig. 3.31 shows the representative fractographs of 75° built thin-strut specimens using LPBF. The overall characteristics of the fracture surfaces are similar to the 15° struts. The Zone I specimen in Fig. 3.31 (a) and (b) exhibit obvious features of ductile dimples and a number of fine microvoids occurred at the fracture surface, indicating a ductile fracture failure mode. Meanwhile, the transgranular fracture surfaces were also observed. All of these features indicated a mixed mode of brittle and ductile failure mechanisms. Both Zone II and Zone III specimens show defects in the thin-strut specimens shown in Fig. 3.31 (c-e), but the formation mode of defects are different. A large amount of defects could even result in a macro-scale cross section reduction and causes a drastic decrease in part strength.

The sample setup the testing apparatus poses a challenge in this micro tensile test. It was found that the smallest strut specimens that could be set up and tested successfully using the current testing equipment were the ones with a nominal size of 0.2 mm. However, as discussed previously, the actual dimension of LPBF fabricated thin strut specimen exhibited different extra-dimensional deviations which were at least 20%. Therefore, the actual diameter of the thin-strut specimens that are able to be tested needs to be larger than 0.24 mm. Moreover, because of the low strength of the thin strut, it was challenging to

record an accurate loading value due to the preload and resolution limit of the equipment system.

Fig. 3.31 Tensile fractographs of 75° thin-strut specimens under SEM, (a) Process parameter of 80 W and 400 mm/s (Zone I); (b) Magnified view of (a); (c) Process parameter of 100 W and 300 mm/s (Zone II); (d) Magnified view of (c); (e) Process parameter of 60 W and 500 mm/s (Zone III); (f) Magnified view of (e)

Another challenge in the specimen setup step was the misalignment of the specimens along the direction of tensile stress. As the small-sized specimens are sensitive

to tiny alignment deviations, in this tensile testing, efforts were made by marking the positions on the grips and trying to fix the specimens at the same location for each test in order to avoid the undesirable shearing or twisting effects.

Overall, the micro tensile imposes significant challenges. During the setup of the thin-strut tensile specimens at the current testing apparatus, stability and consistency are often difficult to maintain especially for the very thin strut specimens, such as those with a diameter of 0.1 mm and below. These provided further insights into the development of future testing methods for micro-tensile testing of AM metal thin struts.

3.4.4 Microhardness and ANOVA

In this study, the microhardness test was carried out by a Shimadzu HMV-G21 micro-indentation (Vickers) hardness tester. A force of 500 g force load and 15 seconds were applied in the transverse direction of bulk and thin strut samples. The bulk and thin struts were mounted to support by the mounting medium so that the specimens would not move during the testing.

There is no standard shape or size for a Vickers test specimen. The tested surface should be free of any visible defects. In many instances of specific small samples, special mounting is required [201].

For the micro Vickers hardness test, loads are typically in grams-force (gf), and indentation diagonals are in micrometer (μm). The Vickers hardness number, using gf and μm as units, is calculated as follows:

$$HV = 1.000 \times 10^3 \times P / A_s \quad (3-7)$$

and

$$A_s = \sin(\alpha / 2) / d^2 \quad (3-8)$$

where:

P = force, gf,

A_s = surface area of the indentation, μm^2 ,

d = mean diagonal length of the indentation, μm , and

α = face angle of the indenter, 136° (See Fig. 3.32 (a))

Fig. 3.32 Schematic of Vickers microhardness test. (a) Schematic of Vickers pyramid diamond indenter; (b) Illustration of micro Vickers hardness indentation

In order to investigate the influence of process parameters, build orientation and specimen size respectively, thin strut specimens with all process/geometry design combinations were evaluated by the microhardness test. Fig. 3.33 shows the comparison of microhardness values of bulk specimens and thin strut specimens. It was found that the microhardness for the thin struts exhibits a similar trend with that of bulk samples, and the microhardness curves exhibit opposite trends compared with the porosity (shown in Fig.

2.14). Besides, the microhardness values for the thin struts were generally higher than those of bulk samples, and the difference was about 10 HV.

Fig. 3.33 Comparison of microhardness values of bulk samples and thin struts

Differences of micro harden among varying building orientations of 0.5 mm struts are exhibited in Fig. 3.34. The samples generally exhibit the similar dependency on energy densities, while the effect of build orientation was not significant.

ANOVA was conducted to investigate the influence of process parameters and build orientations, and the results of ANOVA was shown in Table 3.12. Both of the F values and P values for the factors and their interactions show that that laser power and scan speed were both significant factors to the microhardness, and the effect of laser power was more significant. However, the microhardness does not show significant dependency on the build orientation, which is also shown in Fig. 3.33. Additionally, any interaction of the three factors was not significant to the microhardness.

Fig. 3.34 Microhardness of thin struts at varying process parameters and build orientation

Table 3.12 ANOVA of full factorial experiment (Micro hardness vs. angle and process)

Source	Degree of Freedom	Adj. Sums of Squares (SS)	Adj. Mean Squares (MS)	F-Value	P-Value
Power	2	0.00	0.00	19.08	0.00
Speed	2	0.00	0.00	4.28	0.02
Angle	2	0.00	0.00	2.58	0.09
Power*Speed	4	0.00	0.00	2.09	0.09
Power*Angle	4	0.00	0.00	0.21	0.93
Speed*Angle	4	0.00	0.00	0.67	0.61
Power*Speed*Angle	8	0.00	0.00	0.36	0.94
Error	54	0.00	0.00		
Total	80	0.00			

$F_{0.05,2,54}=3.17$, $F_{0.05,4,54}=2.54$, $F_{0.05,8,54}=2.12$

Furthermore, the influence of specimen size on microhardness was investigated. Four process parameter sets were selected with energy density from the three zones in the process map, including the parameters of 80 W – 400 mm/s (66.67 J/mm^3 , Zone I), 80 W

– 300 mm/s, (88.89 J/mm³, Zone II), 80 W – 300 mm/s (111.11 J/mm³, Zone II), and 60 W – 500 mm/s (40.00 J/mm³, Zone III), as shown in Fig. 3.35. It clearly shows that the microhardness increases as the size of the thin-strut specimens increases. Meanwhile, the group with 40.00 J/mm³ exhibit the lowest microhardness values, while other three groups yield similar results.

Fig. 3.35 Micro hardness by nominal diameter of thin struts

Table 3.13 ANOVA of full factorial experiment (Micro hardness vs. ED and specimen size)

Source	Degree of Freedom	Adj. Sums of Squares (SS)	Adj. Mean Squares (MS)	F-Value	P-Value
ED	3	2.40	0.88	92.80	0.00
Size	6	0.90	0.15	17.40	0.00
Error	13	0.11	0.01		
Total	22	3.30			

$F_{0.05,3,13}=3.4105$, $F_{0.05,6,13}=2.9153$

The effects of energy density and specimen size were analyzed in ANOVA, which was shown in Table 3.13. The F values and P values in the results of ANOVA presented that the energy density and specimen size were both affecting the microhardness significantly, and the former factor was more sensitive to the microhardness.

The tensile properties of the Ti-6Al-4V alloy are quite microstructurally sensitive. The grain size, grain orientation and defects inlayed in the material structure are highly dependent on the solidification of the thermal conditions, which is controllable primarily by the process parameters during the LPBF fabrication. The previous investigation on Ti-6Al-4V solid material indicated the process parameters had a significant effect on the microstructure and mechanical properties.

3.5 Qualitative evaluation of process-dependent properties

3.5.1 Porosity

Fig. 3.36 Porosity of Ti-6Al-4V solids

The results of the porosity are shown in Fig. 3.36. As could be observed clearly, the porosity of Ti-6Al-4V bulk materials exhibited a significant relationship with the laser power. By fitting the Power-Porosity plots, the porosity of Ti-6Al-4V bulk materials showed a parabola trend as the experimental results. The fitted curves are shown in Fig. 3.37 and the fitting formulas and the corresponding coefficients are listed in Table 3.14.

Fig. 3.37 Fitted curves for the porosity of Ti-6Al-4V solids

Table 3.14 Fitting equations and coefficient values for bulk porosity

Equation: $y(\%) = A + B \cdot x + C \cdot x^2$							
	A	Std. Error	B	Std. Error	C	Std. Error	Adj. R-square
300 mm/s	17.123	0.415	-0.478	9.657E-03	3.321E-03	5.339E-05	99.985%
400 mm/s	13.159	0.013	-0.317	3.130E-02	1.929E-03	1.730E-04	92.579%
500 mm/s	7.579	0.715	-0.165	1.663E-02	9.069E-04	9.196E-05	96.969%
700 mm/s	13.225	0.352	-0.242	8.190E-03	1.101E-03	4.528E-05	99.914%

The inherent nature of second-degree parabola reveals that a vertex point exists on the parabola curve and the minimum value of the formula occurs at the vertex point, due to the positive coefficient C of the second-degree term. Worthy noted, the vertex locates on the axis of symmetry of the parabola and indicates the corresponding laser power of minimum porosity at the same speed level.

By quadratic polynomial expansion, the equations are equivalent to the square quadratic formula format: $y = C * (x + \frac{B}{2C})^2 + D$, where $D = \frac{4AC - B^2}{4C}$. The vertex value of on the polynomial equation can be read apparently from the transferred quadratic equations. It turns out that the lowest y value will be obtained at $x = -\frac{B}{2C}$. The relationship of energy density and scan speed can be obtained in the same way. Therefore, the coefficients of the transferred fitting equations for the laser power and energy density versus scan speed respectively are shown in Table 3.15.

Table 3.15 The coefficients of the transferred fitting equations

Speed (mm/s)	Power vs. Speed			ED vs. Speed		
	C	$-\frac{B}{2C}$	D	C	$-\frac{B}{2C}$	D
300	3.321E-03	71.975	0.5146	2.690E-03	79.92	17.127
400	1.929E-03	82.153	0.1382	3.011E-03	69.08	14.382
500	9.069E-04	90.969	0.0740	2.040E-03	60.79	7.576
700	1.101E-03	109.87	-0.0692	4.856E-03	52.30	13.225

By plotting the locations of the symmetry axis of the fitting parabolas, a linear trend in Power-Speed plot and an exponential trend in the Energy density-Speed plot are

obviously seen from Fig. 3.38, which provide guidance to estimate the process parameters of the high-density solid material.

Fig. 3.38 Trending and fitted curves for the power and energy density vertex values versus scan speed

Table 3.15 Coefficients of the fitting curves in Fig. 3.38

	Power vs. Speed			Energy density vs. Speed		
Fitting Equation	$y = a + b \cdot x$ (Linear)			$y = a \cdot x^b$ (Exponential)		
Adj. R-Square	99.94%			99.72%		
		Value	Std. Error		Value	Std. Error
Coefficient	a (Intercept)	44.020	0.677	a	1451.966	140.399
	b (Slope)	0.094	0.001	b	-0.509	0.016

In the EOSINT M270 system, because of the maximum laser power of 200 W and scan speed of 7000 mm/s, the fitting equations have constraints to limit the dependent output.

5.3.4.2 Grain size

The columnar grains in the Ti-6Al-4V bulk specimens grow parallel to the build direction, and the measured grain width is plotted in Fig. 3.39. The thermal history during the laser melting gives rise to the columnar grains with varying sizes even in a certain process parameter set. The range of the grain width can be up to times of the average values. As the previously described, a high laser power level and a low scan speed level would lead wider columnar grains. For all power levels, an increased scan speed results in a reduced grain size, and an increased laser power results in an increasing grain width, due to the increased energy density.

Fig 3.39 Grain width of Ti-6Al-4V solids

By fitting the plots of grain width vs. scan speed, it turns out that the plots follow an exponential trend that fits the experimental measurement as shown in Fig. 3.40. The fitting equation and corresponding coefficients are listed in Table 3.16.

Fig. 3.40 Fitted curves for the grain width of Ti6-Al-4V solids

Table 3.16 Fitting equations and coefficient values for bulk grain width

Equation: $y = a + b \cdot c^x$							
#	a	Std. Error	b	Std. Error	c	Std. Error	Adj. R-square
60 W	83.702	3.273	8.264E+09	1.415E+13	0.933	5.326	68.20%
80 W	95.737	2.076	6.635E+05	1.715E+06	0.970	0.008	99.47%
100 W	104.622	1.561	3.197E+05	4.706E+05	0.973	0.005	99.77%
120 W	135.344	4.267	7.048E+09	9.607E+11	0.940	0.427	94.21%

It is noticed that the constant bases (value c in Table 3.16) in the fitting formulas for tested power levels maintain the same at an acceptable error range. Besides, the constant-based exponential function has the nature that when the base c is a positive value

less than 1, the curve infinitely approaches to a linear level parallel to the X-axis but never intersects or coincident with it, which reveals from the constant a . As read from Table 3.16, it is apparently indicated that the values of a increases as the laser power increases.

By plotting the value a in various power level and fitting the curve, a linear trend is found for the distribution of the value a , shown in Fig. 3.41, and the corresponding coefficients are listed in Table 3.17.

Fig. 3.41 Trending and fitted curves for the minimum grain width value versus laser power

Table 3.17 Coefficients of the fitting curves in Fig. 3.41

Equation	$y = a + b \cdot x$		
Adj. R-Square	90.11%		
		Value	Standard Error
Coefficient	a (Intercept)	36.16	6.56314
	b (Slope)	0.744	0.07121

It is worthy noted that even though the fitting curves lead a constant columnar grain width value mathematically, according to the experimental results of porosity for Ti-6Al-4V solids, the grain size would not be obtained infinitely. Either when the porosity is too

large to identify the columnar grains, or when the nominal maximum scan speed in the AM fabrication system is reached, the computation will be interrupted by the limitation.

5.3.4.3 Microhardness

Since the microhardness of metallic solids is significantly related to the porosity level as described earlier, the microhardness plots had an opposite trend compared with the porosity. However, consider the case that the micro-indentation for high property specimens located at pore-free positions, these parameter sets resulted in relative higher microhardness than their capacity, and therefore might exist large deviation of prediction by the revised curves from the result of porosity.

In order to investigate the fitting issue for microhardness of Ti-6Al-4V solids, efforts were made by referring the fitted curves of porosity. Firstly, it was found that vertexes matched those in porosity curves by neglecting the deviating points temporarily. Secondly, according to the inherent property of parabola functions, the plot will become flatter and wider by magnifying the coefficient of a quadratic independent variable with the vertex at the same location in the X-Y coordinates. For this reason, refer to the scale range of “microhardness” axis, a series of parabola curves (shown in Fig. 3.42) were plotted using 12 times magnification and reversed to negative value of the coefficient C , the same vertex laser power values of the parabola functions, and various values at the vertexes based on the experimental results of microhardness. The corresponding coefficients are listed in Table 3.18. Of this fitting, it was noted that the experimental microhardness values of Ti-6Al-4V solid parts with desirable dense located at the regions of fitted curves within acceptable standard error. As for the deviating points of parts with significant porosity, higher values revealed in the plot.

Fig. 3.42 Fit of microhardness for Ti-6Al-4V solids

Table 3.18 Fitting function and coefficients for microhardness

Equation: $y=A + B*x + C*x^2$		
A	B	C
163.576	5.736	-3.985E-02
213.759	3.804	-2.315E-02
279.941	1.980	-1.088E-02
218.469	2.904	-1.322E-02

3.6 Conclusion

In this chapter, thin feature struts were fabricated using LPBF EOSINT M270 with various design conditions, including process parameters, exposure types, build orientations, beam offset, and diameters. The fabrication of thin feature structures is sensitive to the scanning exposure. In the EOSINT M270 system, the hatch exposure was confirmed to be turned on when the diameter of the thin strut was above 0.25 mm. The contour exposure contributed to the roughness of the lateral surface. It was found that the edge exposure did

not dedicate in the scanning procedure of Ti-6Al-4V thin structures with circular cross-section regardless of their dimensions.

The fabricatability of the thin-feature struts was investigated. Generally, lower build orientation led to higher success rate. In the Zone I and Zone III of process window, the fabricatability was higher than that in Zone II due to residual stress caused by high energy input.

The geometry accuracy and quality of the thin-feature struts were evaluated through exterior characteristics and internal microstructure examination. The exterior observation indicated that the geometry accuracy was highly dependent on the build orientation. Due to the attachment of the unmelted particles, 15°-built struts had a rough down-facing surface that reduced the surface quality, while 45° and 75° struts had a smoother surface. Without support, the low-orientation-built struts exhibited “collapsing” down-facing surface morphology. The actual dimensions of the struts were generally more extensive than the designs. The relative errors exhibited increasing trend with the decrease of the strut diameters, and the errors become much more significant at a diameter below 0.2 mm. Results of ANOVA indicated that the factors of diameter, beam offset, build orientation, process parameters significantly influence the geometry accuracy successively. At the diameter smaller than 0.2 mm, the resolution limit of LPBF EOS system caused the actual strut diameter to become independent of the nominal size. Process parameters in Zone II would result in larger dimension than those in the other zones. Exposure type was not a significant factor in the geometry accuracy of the thin feature struts.

Zone II and Zone III tend to result in higher porosities than Zone I. The porosity of thin struts exhibits a decreasing trend with the reduction of the diameter of struts and the

increase of build orientation. Due to rapid cooling in LPBF, columnar grains grew perpendicular to the build substrate, but with a slight angle towards to the strut axial direction. The grain size had a growing trend as the energy density increased and building orientation decreased. Martensitic α' was observed in the sectioned samples.

The mechanical properties of the thin feature struts were characterized by micro tensile test and microhardness test. In the micro tensile test, due to the inaccuracy of extension of the current testing equipment, stress-time curves were plotted instead of stress-strain curves. The UTS was calculated under the assumption of round shape cross sections, and Young's moduli were estimated using the 1.1% strain as the initial yield point. Compared to the large-scale bulk tensile specimens, the UTS exhibited generally larger values, while Young's moduli were comparable. Similar to the large-scale tensile specimens, the process parameters in Zone II and Zone III resulted in defects via different mechanisms, which lead to reduced mechanical behaviors. In a particular process parameter, the low build orientation tended to increase the UTS and modulus. At a specific orientation, smaller strut diameter results in larger UTS in some parameters due to the cross-sectional effect. However, the results of UTS and Young's moduli exhibited large variabilities at various diameters and process parameters. The ductility of the thin-feature struts was low, and no necking phenomenon was observed. Additionally, the results of microhardness were generally higher than those of bulk samples, and the difference was about 10 HV. ANOVA results showed the laser power was the most significant factors to the microhardness, followed by scan speed and build angle. The microhardness results also verified the specimen size effect that smaller dimension led to higher hardness values.

CHAPTER 4

GEOMETRY FEA SIMULATION

4.1 Introduction

The development of AM technologies has enabled the fabrication of metal structures, which are difficult for the conventional manufacturing. Many metal based processes, such as electron beam melting (EBM) and selective laser melting (SLM) have the capability to produce complex shaped cellular structures. With the fabricatability of the cellular structures greatly overcome various by AM, more research and development efforts have shifted to the design the cellular structures with controlled geometric parameters [5].

The nature of laser powder bed fusion (LPBF) decides that only a single type of materials can be used for one batch of fabrication. Thus, the LPBF parts are expected to exhibit homogenous material characteristics. In the design of LPBF cellular structures, there exist a number of studies dealing with homogenous lattice structures and related mechanical properties for optimizing the cellular characteristics, which then ensures appropriate deflection and a target masses while ensuring the structural quality and functionality. Structural optimization is the minimization of mass by varying or determining dimensional variables, such as geometry and shape parameters, under the

constraints of performance requirements, such as stiffness and compliance. Structural optimization requires the explicit definition of objective-related variables in the mesh [202]. A classical problem in the structural design is the maximization of the static stiffness subject to a volume constraint. The objective function of such problems can be evaluated by an analytical computation or finite element method (FEM) regardless of changing the mesh. On the other hand, topology optimization doesn't prescribe any structural-property relationships. Instead, the material distribution within the design domain is permuted as the means of optimization[202]. Various topology optimization in AM has been reported that includes penalization (SIMP), bidirectional evolutionary structural optimization (BESO), homogenization, and proportional topology optimization (PTO), genetic algorithms [203-206]. Currently, the main challenges include multiple aspects. Firstly, the existing software packages that offer topology optimization provide little or limited capabilities with multi-physics problems and dynamic loading problems [207]. Secondly, the optimized geometries from the topology optimization often contain small-dimension features that are beyond the fabrication capacity of AM processes or include features with abrupt dimensional changes that give rise to significant thermal stress concentration. Besides, most topology optimization software still cannot deal efficiently with the anisotropic materials. All of these challenges result in a lack of efficiency with the design method in meeting the expected requirement.

In this study, beam elements based modeling were employed to evaluate the mechanical properties of the cellular structures. Theoretical model, finite element analysis (FEA) and experimental study were employed to estimate the mechanical properties of cellular structures, such as the effective Young's modulus.

4.2 Structural design and simplification

The unit cell approach was commonly used on the design of periodic cellular structures to simplify the design problem since the unit cells could be readily decomposed into individual beam components, which could be analyzed by the beam theory. Many researchers used the unit cell approach for their study of cellular structures [2, 208].

The joint-based method in the cellular structure design constructs cellular structures based on the central joint, from which the struts will be created on account of specific structural requirements. This approach can potentially provide higher flexibility of the cellular structural design and is therefore investigated in this study. Using the strategy of the joint-based method, the establishment of a cellular structure starts with the joint node. In the selected space-filling bounding volumes, the node located spatially at the geometrical center of the volume. Then the struts are created from this node with axial symmetry. The struts are created with one end at the central joint and the other end located at the boundaries of the bounding volume. In order to satisfy the spatial periodicity and integrity of the unit cell in the three principal directions, the end of each strut at the edge of the bounding volume must be coincident with the corresponding end of the strut in the neighboring unit cell. Fig. 4.1 shows three unit cells of the octahedron, hexahedron, and dodecahedron designed using the joint-based method.

Fig. 4.1 Illustration of hexahedron, octahedron and dodecahedron unit cells

The joint-based method can be used to generate many periodic cellular structures with one center joint in the bounding volumes. For example, a cubic prism can be used to generate an octahedron; a hexahedral prism can be used to create a typical diamond structure or a hexahedron, which are shown in Fig. 4.1. Therefore, this model can provide a generalized geometry design approach that is applicable to any joint-based structure with different variables of geometric parameters.

There are 4 primary geometry parameters in the design of central-joint structures: the length of the strut (l), the diameter of the strut (d), the gradient angle of each symmetrical strut relative to the horizontal plane (θ), and $1/2$ of the total number of struts from the same joint in the unit cell (n). Fig. 4.2 shows the octahedral unit cell ($n=4$). Note the dimension of the strut is not given in the figure. A_{cell} is defined as the effective cross-section area of the unit cell of the structure, which is shown as the region surrounded by dot lines in Fig. 4.2. A_{cell} can be calculated as a function of l , θ , and n . Due to the instinct of structural symmetry, the mechanical properties of the structure along the X_1 and X_2 directions are identical.

Fig. 4.2 Illustration of dimension of octahedral unit cell, (a) original joint-based unit cell; (b) transferred enclosing unit cell; (c) CAD model for (a); (d) CAD model for (b)

The forces in each of struts from the same central joint are considered to be equally distributed when a normal force is applied onto the structure since the struts from the same central joint are designed to exhibit axial symmetry. It must be noted, however, that the anisotropy caused by the fabrication in LPBF is not considered here. Several assumptions are considered in this study. First, the unit cell is assumed to be located in an infinite structure, which eliminates the boundary effects and allows the structure to retain a high degree of symmetry. As a result, any arbitrary strut can be treated equivalently to the others. However, the actual structures always have finite dimensions, so the size effect needs to

be taken into account, which is addressed in this work too. Second, all the joints in the structure are considered to be rigid. The deformation of the struts would be contributed by the bending primarily. For metal or other high strength materials by AM, the joints of cellular structures are welded to the structure and therefore cannot rotate. Third, the strut components are treated as beam structure subjected to symmetrically uniaxial loading, and thus loading for each strut element can be considered as in-plane stress. Therefore, a torsional effect that is caused by out-of-plane force can be neglected. Fourth, small deflection upon loading will be employed in this study, since the Ti-6Al-4V exhibits high elastic modulus, the geometry modeling will take the linear-elastic deformation into consideration.

Design for joints

The actual cellular structures are different from the analytical model with features such as overlapped geometries between struts, chamfers, and edges at the intersections. These geometric features could also potentially influence the accuracy of the analytical model.

Fig. 4.3 (a) shows the actual joint in a cellular structure fabricated from the model shown in Fig. 4.3 (b). Fig. 4.3 (b) and (c) show two types of joint designs on the cellular unit structure with circular cross-sections. The joint model (shown in Fig. 4.3 (c)) better represents the real geometrical condition after fabrication. Therefore, the fabrication of cellular structures in this study was based on the type I joint in Fig. 4.3 (b) to keep the model consistency, while FEA model was adopted using the type II joint in Fig. 4.3 (c).

Fig. 4.3 Slender struts, (a) Joint connected actual thin struts fabricated by EBM; (b) Type I joint; (c) Type II joint

Effective length of struts

The mechanical properties of the cellular structures are sensitive to the strut lengths. It is noticed that the length of the actual struts subject to bending is shorter than the designed values due to the thickness of the struts (shown in Fig. 4.4). Therefore, in this study, the length of struts in the structure used for the theoretical predictions was set as the distance “b” as marked in Fig. 4.4.

Fig. 4.4 Illustration of the effective length of struts

4.3 Modeling of mechanical behavior

The octahedron is a typical central joint-based structure. Unlike other cellular structures with tetrahedral or triangular configurations, the unit cell of the octahedral structure is comprised of eight symmetrical struts from the central joint and follows the same geometries in the three principal directions. (See Fig. 4.5). Therefore, the individual strut could be treated as a representative structure in the description of the mechanical behavior of the entire structure. The octahedral structure has been previously reported by the researchers experimentally [66, 176, 209, 210] and analytically [176, 211], which were focused on the mechanical properties of the structures based on the geometrical parameters of the selected unit cell.

Consider a uniaxial force loaded along the Y direction, the architecture of the octahedral structure implies that the struts from the same central joint are under the same loading mechanism. Therefore, an arbitrary strut with both ends constrained by rigid joints was analyzed. Under the assumption of Euler-Bernoulli beam theory, the neutral plane of the strut remains perpendicular to the bending line after the deformation takes place. Such an assumption is generally valid for slender beam structures. The loading condition and the moment distribution on the strut are shown in Fig. 4.5. It is noted that the effective moment at the middle point of the strut is zero. Therefore, the half strut is adopted as a cantilever beam with one end fixed that is located in the middle of the strut and the other free to move in-plane at either end of the strut (shown in Fig. 4.5 (b)). u is a curvilinear coordinate along the deflected strut in Fig.4.5 (b), and du is a differential component along the u direction.

Fig. 4.5 Schematic of the loading and deflection of the struts, (a) loading condition and the moment distribution; (b) cantilever deflection of the half strut

In the deformation of relatively short or thick beams, the effect of shear stresses is more significant, which causes a rotation between the cross-section and the bending line in the structure. This shear-induced rotation is not accounted for in the Euler-Bernoulli beam. It has been shown that in order to predict the mechanical behavior of shorter elements more accurately, the Timoshenko beam theory should be employed. Under the same loading conditions, the Euler-Bernoulli beams are generally stiffer than the Timoshenko beams, and the discrepancies increase with decreasing slenderness ratio (SR). The slenderness ratio is defined as l/R_g where l is the length of the column and R_g is the radius of gyration. The latter term estimates the relative significance of bending compared to axial force effects and could be defined by $R_g^2 = I/A$, where A is the area of the strut's cross section, and I is the second moment of area of the cross-section. The reciprocal of the slenderness ratio $\alpha = 1/SR = I/(Al^2)$ can be used to describe the dimensional characteristics of the struts [212]. For the rectangle cross-section, $\alpha = (1/12)(b/l)^2$; for the circular cross-section, $\alpha = (1/$

$8)(d/l)^2$. Generally larger α corresponds to smaller slenderness ratio, which also indicates that the bending of the strut is more significant. In this study, in order to evaluate the effects of the slenderness of the struts, the aspect ratio (d/l) is used as a design parameter, where d is the diameter of the strut and l is the length of the strut.

To demonstrate the modeling process, the strain energy method was adopted to model and analyze the deformation behaviors of octahedral structures under uniaxial loading. In general, the beam deformation is determined by the combination of bending, axial compression, shear force and torsion. When uniaxial loading is applied to the unit cell, the torsion is negligible and therefore does not contribute to the strain energy.

The various components of strain energies for a half-length strut can be expressed as:

$$\text{Bending strain energy: } U_b = \int_L \frac{M^2(u)}{2EI} du \quad (4-1)$$

$$\text{Axial compression strain energy: } U_c = \int_L \frac{R^2}{2EA} du \quad (4-2)$$

$$\text{Shear strain energy: } U_s = \frac{1}{2} \int_L \int_A \frac{Q^2 S^2}{I^2 b^2 G} dA du \quad (4-3)$$

where,

$$\text{The bending moment: } M(u) = \int W u \cos \theta du \quad (4-4)$$

$$\text{The axial force: } R = W \sin \theta \quad (4-5)$$

$$\text{The shear force: } Q = W \cos \theta \quad (4-6)$$

Therefore, the total strain energy for a strut of length l due to the effect of bending moment M , axial force R and shear force P could be given as:

$$U_1 = 2U_b + 2U_c + 2U_s = \int_0^{l/2} \frac{M^2(u)}{EI} du + \int_0^{l/2} \frac{R^2}{EA} du + \int_0^{l/2} \int_A \frac{Q^2 S^2}{I^2 b^2 G} du \quad (4-7)$$

Consider a remote load F applied along the Y direction applied to the unit cell of the octahedral cellular structure (shown in Fig. 4.6). Due to the structural symmetry, the load on each strut is $W = \frac{1}{4}F$. The loading condition is shown in Fig. 4.6. The

compressive stress σ_y loaded on the unit cell can be obtained as:

$$\sigma_y = \frac{F}{A_{cell}} \quad (4-8)$$

4.3.1 Deformation of the slender struts ($d/l < 0.2$)

Based on the previous discussion, the slender strut could be treated as Euler-Bernoulli beams with only bending deformation considered. The strain energy of the half-length strut caused by the bending is expressed as Eqn. 4-1. The bending moment is $M(u) = \int W u \cos \theta du$. The displacement in the Y direction is:

$$\delta_{ly} = \frac{\partial U_1}{\partial W} = \frac{W l^3 \cos^2 \theta}{12 E_s I} \quad (4-9)$$

where E_s exhibits Young's modulus of the solid material. Of this, the strain of the strut along the Y direction can, therefore, be calculated as:

$$\varepsilon_y = \frac{\delta_{ly}}{l \sin \theta} = \frac{\sigma_y A_{cell} l^2 \cos^2 \theta}{48 E_s I \sin \theta} \quad (4-10)$$

Due to the symmetry of the struts in the octahedral unit cell, the effective Young's modulus of the unit structure can be obtained as:

$$E_y^* = \frac{\sigma_y}{\varepsilon_y} = \frac{48E_s I \sin \theta}{A_{cell} l^2 \cos^2 \theta} \quad (4-11)$$

In this study, the struts were designed to have circular cross sections to represent the actual structures better. Therefore, some of the terms in Eqn. 4-11 can be replaced by the effective area of the unit cell $A_{cell} = \frac{1}{4} \pi d^2$ and the second moment of area of the cross-section $I = \frac{\pi d^4}{64}$. Therefore, the effective Young's modulus of the octahedral structure can be expressed as:

$$E_y^* = \left(\frac{d}{l}\right)^4 \frac{3\pi \sin \theta}{8 \cos^4 \theta} E_s \quad (4-12)$$

4.3.2 Deformation of the short struts ($d/l \geq 0.2$)

For short or thick struts, the elastic strain energy of shear stresses becomes significant in the total elastic work. From Timoshenko beam theory, the mechanical behaviors of the structures under the loading in the y direction can be described by the combined deformation of bending/shearing.

From the shear stress formula, the strain energy in a beam of length L due to shear can be written as:

$$U_s = \frac{1}{2} \int_L \int_A \frac{Q^2 S^2}{I^2 b^2 G} dA du \quad (4-13)$$

where S is the static moment of the cross-sectional area above the point where the shear stress is being calculated. b is the width at the point where the shear stress is being calculated. G is the shear modulus, which satisfies the Hooke's Law, $G = \frac{E}{2(\nu+1)}$.

$$\text{Let} \quad k = \frac{A}{I^2} \int_A \frac{S^2}{b^2} dA \quad (4-14)$$

k can be derived by substituting the static moment, which is expressed as a function of v and the width at the location of v , the area of the cross-section, and the moment of inertia in the Eqn. 4-14 (Fig. 4.6). k is a constant that depends only on the shape of the cross-section of the strut. For the rectangle cross section, $k = 6/5$; for the circular cross-section, $k = 10/9$; for the thin-wall circular tube cross section, $k = 2$. Eqn. 4-14 becomes:

$$U_s = \frac{1}{2} \int_L \frac{kQ^2}{GA} du \quad (4-15)$$

Fig. 4.6 Circular cross-section showing a differential area

Under the bending/shearing combined loading condition, the total strain energy of the single strut could be obtained as:

$$U_1 = \int_0^{l/2} \frac{M^2(u)}{EI} du + \int_0^{l/2} \frac{kQ^2}{GA} du \quad (4-16)$$

The displacement in the Y direction is, therefore:

$$\delta_{1y} = \frac{Wl^3 \cos^2 \theta}{12E_s I} + \frac{kWl \cos^2 \theta}{G_s A} \quad (4-17)$$

Considering the expression of the structural compressive stress $\sigma_y = \frac{4W}{A_{cell}}$, and the

strain in the Y direction becomes:

$$\varepsilon_y = \frac{48\sigma_y E_s I \sin \theta}{A_{cell} l^2 \cos^2 \theta} + \frac{k\sigma_y A_{cell} \cos^2 \theta}{4G_s A \sin \theta} \quad (4-18)$$

Therefore, the effective modulus of the octahedral unit cell due to the combined loading condition can be expressed as:

$$E_y^* = \frac{\sin \theta}{A_{cell} \cos^2 \theta \left(\frac{l^2}{48E_s I} + \frac{k}{4G_s A} \right)} \quad (4-19)$$

For struts with circular cross-sections, Eqn. 4-19 becomes:

$$E_y^* = \left(\frac{d}{l} \right)^4 \frac{\pi \sin \theta}{\cos^4 \theta \left(\frac{8}{3E_s} + \frac{20}{9G_s} \left(\frac{d}{l} \right)^2 \right)} \quad (4-20)$$

4.3.3 Deformation under three types of force condition

Consider the total strain energy due to the axial compressive force within the effective Young's modulus model, the expression of the structural strain and effective Young's modulus parallel to the Y direction could be presented as follows:

$$\varepsilon_y = \sigma_y A_{cell} \left(\frac{l^2 \cos^2 \theta}{48 E_s I \sin \theta} + \frac{k \cos \theta}{4 G_s A \sin \theta} + \frac{\sin \theta}{4 E_s A} \right) \quad (4-21)$$

$$\text{and} \quad E_y^* = \left(\frac{A_{cell} \cos^2 \theta}{\sin \theta} \left(\frac{l^2}{48 E_s I} + \frac{k}{4 G_s A} \right) + \frac{A_{cell} \sin \theta}{4 E_s A} \right)^{-1} \quad (4-22)$$

For the circular cross-sectional struts, the structural strain and the effective modulus in the Y direction are:

$$\varepsilon_y = \frac{1}{\left(\frac{d}{l}\right)^2} \frac{\sigma_y \cos^4 \theta}{\pi \sin \theta E_s} \left[\frac{8}{3} \frac{1}{\left(\frac{d}{l}\right)^2} + \frac{40(\nu+1)}{9} + 2 \tan^2 \theta \right] \quad (4-23)$$

$$\text{and} \quad E_y^* = \left(\frac{d}{l}\right)^4 \frac{\pi \sin \theta}{\cos^4 \theta \left[\frac{8}{3} + \frac{40(\nu+1)}{9} \left(\frac{d}{l}\right)^2 + 2 \tan \theta \left(\frac{d}{l}\right)^2 \right]} E_s \quad (4-24)$$

From the three equation of effective Young's moduli due to three types of deformation, it could be noticed that the effective Young's moduli are dependent on the dimensional parameters, including the aspect ratio (d / l), the tilt angle (θ), and the solid bulk material properties, including the Young's modulus E_s and the Poisson ratio ν (or the shear modulus G_s).

Fig. 4.7 Effective moduli resulted from 3 types of deformation in FEA (n=4)

Gibson-Ashby theory assumes that the axial deformation due to the bending could be negligible for slender beams [7]. The effective modulus results that took into account the three types of deformation in various strut orientations are shown in Fig. 4.7. It can be noticed that the effective moduli agree with Gibson-Ashby theory when the aspect ratio (d/l) is below 0.2, and the elastic modulus of the struts with lower orientation angles exhibit good agreement with all three types of deformation. However, for struts with the aspect ratio of 0.2, the three types of deformation result in significant difference on the effective modulus. At the strut orientation of 45°, the effective moduli calculated with shear force effect also exhibits significant differences from the ones without it. The corresponding differences for $E_{y^* (b+s)}$ and $E_{y^* (b+s+c)}$ away from $E_{y^* (b)}$ are 9.2% and 2.5%, respectively. It

indicates that Timoshenko beam theory is more suitable in this case. At the strut orientation of 75° , the three type of deformation exhibit significant differences. It is considered that the stiffness exhibits stretch-dominated behavior, which can be effective to the axial compression on each individual strut.

4.3.4 Further discussions

Without losing generality, Eqn. 4-25 can be extended to more spatially symmetrical polyhedral structures, such as hexahedron (triangular bipyramid) and dodecahedron (shown in Fig. 4.8). Because of the structural symmetry of either hexahedral or dodecahedral structure, each of the struts from the same central joint in the unit cell shares the load with the adjacent struts. So for an arbitrary strut, the load that it undergoes can be determined as $W = \frac{1}{n} F$, where n represents the half number of struts from the same central

joint in the unit cell. The effective stress on the unit cell can be expressed as $\sigma_y = \frac{nW}{A_{cell}}$.

Fig. 4.8 Unit cells of (a) hexahedral structure (n=3) and (b) dodecahedral structure (n=6)

Therefore, the structural strain of the arbitrary strut and the effective Young's modulus parallel to Y direction are provided with universality as follows:

$$\varepsilon_y = \frac{\sigma_y A_{cell}}{n} \left(\frac{l^2 \cos^2 \theta}{12 E_s I \sin \theta} + \frac{k \cos \theta}{G_s A \sin \theta} + \frac{\sin \theta}{E_s A} \right) \quad (4-26)$$

$$\text{and} \quad E_y^* = \frac{n}{A_{cell}} \left(\frac{l^2 \cos \theta}{12 E_s I \sin \theta} + \frac{k \cos^2 \theta}{G_s A \sin \theta} + \frac{\sin \theta}{E_s A} \right)^{-1} \quad (4-27)$$

The effective moduli of hexahedral and dodecahedral unit cells in FEA were calculated and compared with those of octahedron. The three types of unit cells at the same geometry parameters exhibit the same order of magnitude of effective modulus values, but the dodecahedral unit cells behave slightly higher effective moduli, and the octahedron shows the smallest. The analytical computations under three types of forces using the for hexahedral and dodecahedral unit cells are shown in Fig. 4.9. It is noticed that the analytical results of effective moduli are comparable to the octahedral structures since the hexahedral and dodecahedral structures are similar to the octahedron that does not have the additional substructures. Likely to the octahedron, the individual struts in hexahedral and dodecahedral unit cells are constructed around the central joints symmetrically, and the Maxwell stability criterion numbers are $M < 0$ [33].

Fig. 4.9 Analytical computations for (a) hexahedral and (b) dodecahedral unit cells

4.4 Finite Element Simulation of the joint-based model

In order to provide the accurate predictions of mechanical behaviors and optimal designs of cellular structures, the modeling of these structures is a critical breakthrough. It is often time-consuming work to apply finite element discrete analysis on cellular structures directly. On the other hand, there exists an abundance of literature about the

modeling and analysis of periodic cellular structures. Gibson and Ashby demonstrated the modeling of triangular, square and hexagonal honeycomb structures, employing standard beam theories. The honeycomb unit cell could be treated as a frame structure, and the elastic properties of individual components could be analyzed conveniently [5]. Hassani and Hinton described a method for the estimations of the effective constitutive parameters of complex materials with a periodic structure using homogenization method and provided the analytical formulation for the effective modulus of rectangular cellular structure [213]. Many other researchers [214-218] reported the structural topology and material optimization problems using microcell asymptotic expansion, which assumes periodicity for differential equations, and obtains limit solutions asymptotically approach the original solutions by replacing the more spatially defined model with a simplified equivalent model. Wang [2] presented a hybrid geometric modeling method for the design, analysis, optimization, and manufacture of conformal truss structures. The hybrid method creates the .stl model of each unit truss using solid modeling and surface modeling techniques, and then stacks all the tessellated surface models together without complex operations to generate the .stl model of the entire structure and saves significant computational resources.

However, the size effect of the cellular structures remained a significant issue in the modeling of a finite structure. With many modeling approaches, the effective modulus is independent of the size of structural size. On the other hand, the dimensions of the actual structures in engineering applications are always finite, especially for the design of sandwich structures with insufficient space for a few layers of periodic cellular cores [211]. To verify the mechanical behavior in a finite structural system, and to provide additional reference information to both the analytical solution and experimental observations, the

FEA method is still feasible despite the excessive computational costs. With the current theory for cellular structures, about 8-10 unit cells are required for the structures to minimize the boundary effect [31]. The understanding of the size effect for cellular structures with small overall dimensions must, therefore, be developed.

4.4.1 FEA on the effective Young's modulus

In order to verify the analytical modeling, FEA was employed on the octahedral structures using SolidWorks Simulation. Unit cells were modeled for the corresponding cellular structures, and the effective Young's modulus of the cellular parts was then calculated. The effective mechanical properties of these unit cells were considered for the estimations of the overall cellular structure performance.

The selected unit cells were constrained at the bottom, and uniaxial stress was applied on the top as shown in Fig. 4.10. The effective stiffness could be calculated as:

$$E_y^* = \frac{\sigma_y}{\varepsilon_y} = \frac{P}{\delta_y / H_y} \quad (4-28)$$

where P is the applied uniaxial pressure on the top surface, δ_y is the displacement that is caused by the compression loading and H_y is the initial height of the unit cell in the Y direction and is dependent on the geometrical design of the cellular part. δ_y is obtained from the strain measurement in FEA. In the simulation, the curvature-based solid mesh with dimensions of 0.24 mm ~ 1.2 mm was employed for all the analysis. The material was defined as Ti-6Al-4V with an elastic modulus of 104.8 GPa and Poisson's ratio of 0.31.

Fig. 4.10 Constraint setup for the unit cell

The length of the struts was set to be constant of 8 mm, and the diameter of the struts varied from 0.05 to 2 mm based on the capacity of the LPBF fabrication. Consequently, the aspect ratio (d/l) ranged from 0.005 to 0.2. The tilt angle θ was set at 5 levels from 15° to 75° . The geometry factors and levels are shown in Table 5.1.

Table 4.1 Geometry factors and levels of octahedral unit cell

Factors	Levels
d/l	0.005, 0.01, 0.02, 0.03, 0.04, 0.05, 0.1, 0.2
$\theta (^\circ)$	15, 30, 45, 60, 75

The effective Young's modulus of the structure E_y^* under uniaxial loading in the Y direction for each geometry design was calculated. The effect of the design parameters on the effective modulus E_y^* is shown in Fig. 4.11. From Fig. 4.11, it is clearly shown that with the increase of d/l ratio and the increase of the tilt angle θ of the struts, the effective modulus of the structure increases. This result is intuitive, since the reduced bending

moment on the struts and increased height dimension in the Y direction would contribute to higher modulus and reduced strain. It was also noticed that E_y^* roughly increased an order of magnitude as the strut orientation increases by about 15° , and increased exponentially with the increasing aspect ratio. The effective modulus value for an octahedral unit cell with 75° strut orientation and aspect ratio of 0.2 could be up to 28.3 GPa, which is almost 1/5 of the elastic modulus of the solid Ti-6Al-4V material.

Fig. 4.11 Effect of aspect ratio and gradient angle on the effective modulus E_y^*

The significances of various design factors were analyzed using ANOVA, which is summarized in Table 4.2. From Table 4.2, it can be seen that the F values of angle and aspect ratio are both larger than the threshold F values ($F_{0.05,4,28}=2.71$, $F_{0.05,7,28}=2.36$), which indicate a significance within 95% confidence interval, which clearly suggests that both factors are significant in determining the effective Young's modulus of the structures. the aspect ratio appears to exert a more significant effect on the responses. This can be verified by Fig. 4.12, in which the slope in the 'aspect ratio - E_y^{eff} ' coordinate plane is steep.

Therefore, the effective modulus as a function of the structural geometry using Eqn. 4-25 has a guideline on the octahedral structural design.

Table 4.2 ANOVA of full factorial design (Effective modulus vs. angle and aspect ratio)

Source	DF	Adj SS	Adj MS	F-Value
Angle	4	0.002	0.001	37451
Aspect ratio	7	0.009	0.001	80410
Error	28	0.000	0.000	
Total	39	0.011		
$F_{0.05,4,28}=2.71$, $F_{0.05,7,28}=2.36$				

4.4.2 Size effect

Andrews et al. [31] experimentally evaluated the hexahedral honeycomb structures and found out that the effective Young's modulus was significantly reduced when the ratio of macrostructure dimension (L) and unit cell size (α) was relatively small ($L / \alpha < 6$). Tantikom et al. [219] investigated quasi-static compressive properties of hexagonal closed-pack arrayed tubes and verified the existence of size effect in the regularly cell-structured materials. They stated that the specimen with a small structure-to-cell dimensional ratio ($L / \alpha < 5$) exhibit an unusual stress-strain relationship. Lestari et al. employed a measurement approach to estimate the elastic bending and transverse shear moduli of the sandwich structures from the dynamic response from the experimental results. This method offered a simple and relatively reliable way to predict the stiffness properties by creating functional computation of effective moduli and intrinsic vibration frequency; and by elongating the beam, which is equal to reducing the size of unit cells in effect. The results showed that the effective moduli of the structures were dependent on the overall structural sizes, and when

the length of the beam and cell size satisfies $L / \alpha > 7$, the moduli trend off to constant values. Zhang and Sun [220] proposed the design element concept on the study of the stiffness of 2D layered structures with the cellular core and found that the optimal solution was strongly related to the overall scale of the structures.

As size effect exists in the finite structure and results in altered mechanical behaviors, FEA was carried out with the octahedral structure to investigate the size effect. In this study, the unit structural parameters were kept constant, and the structures were designed to have the same numbers of unit cells in X_1 and X_2 directions. The size effects were investigated by varying both the numbers of unit cells in X_1/X_2 directions and in the Y direction (loading direction) with designs. An arbitrary octahedral unit cell was designed for the study: $l = 8$ mm, $d = 0.45$ mm and $\theta = 40^\circ$. The cell repetition experimental design is shown in Table 4.3.

Table 4.3 Design for size effect evaluation

Layer	Lateral units					
	1x1	2x2	3x3	4x4	6x6	8x8
1	1x1x1	2x2x1	3x3x1	4x4x1	6x6x1	8x8x1
2	1x1x2	2x2x2	3x3x2	4x4x2	6x6x2	8x8x2
3	1x1x3	2x2x3	3x3x3	4x4x3	6x6x3	8x8x3
4	1x1x4	2x2x4	3x3x4	4x4x4	6x6x4	8x8x4
6		2x2x6	3x3x6	4x4x6	6x6x6	8x8x6
8		2x2x8	3x3x8	4x4x8	6x6x8	8x8x8
10		2x2x10	3x3x10	4x4x10	6x6x10	8x8x10
12		2x2x12	3x3x12	4x4x12	6x6x12	8x8x12
14		2x2x14	3x3x14	4x4x14	6x6x14	8x8x14
20		2x2x20	3x3x20	4x4x20	6x6x18*	
30		2x2x30	3x3x30	4x4x40	6x6x30	

*Note: 18 layers of unit cells were applied to the 6×6 base.

The cellular structures were sandwiched between two rigid plates on the top and bottom sides in order to apply the boundary constraints and loads. The bottom platen was fixed, and the uniaxial compressive pressure was applied to the top platen surface. The loading and boundary condition setup is shown in Fig. 4.12.

Fig. 4.12 Load and fixture condition setup of cellular structure in FEA

According to the computation of Eqn. 4-25, the effective Young's modulus of this octahedral structure is 2.30 MPa. The simulated results in FEA are shown in Table 4.4 and Fig. 4.13.

From the results, the cellular structures with the layer numbers smaller than the lateral number of repetitions exhibited significant size effect. The size effect of elastic modulus appears to exhibit a logarithm dependency on the number of unit cells. The fitting curves were plotted for the size effect of structures with different lateral numbers of unit cells and are shown in Fig. 4.14. The fitting formulas and the corresponding coefficients are also listed in Table 4.5.

Table 4.4 Effective Young's modulus in FEA

Unit repetition	FEA modulus (MPa)	Unit repetition	FEA modulus (MPa)	Unit repetition	FEA modulus (MPa)
1x1x1	2.289	3x3x6	3.310	6x6x2	111.49
1x1x2	2.293	3x3x8	3.086	6x6x3	72.90
1x1x3	2.285	3x3x10	2.992	6x6x4	45.12
1x1x4	2.296	3x3x12	2.947	6x6x6	11.26
2x2x1	61.22	3x3x14	2.895	6x6x8	7.356
2x2x2	3.610	3x3x20	2.816	6x6x10	6.593
2x2x3	3.291	3x3x30	2.821	6x6x12	6.377
2x2x4	2.877	4x4x1	135.93	6x6x14	6.107
2x2x6	2.888	4x4x2	71.54	6x6x18	6.029
2x2x8	2.817	4x4x3	30.29	6x6x30	5.802
2x2x10	2.750	4x4x4	6.721	8x8x1	186.75
2x2x12	2.719	4x4x6	6.664	8x8x2	139.30
2x2x14	2.731	4x4x8	5.772	8x8x3	105.62
2x2x20	2.756	4x4x10	5.503	8x8x4	78.16
2x2x30	2.666	4x4x12	5.523	8x8x6	38.84
3x3x1	100.05	4x4x14	5.548	8x8x8	12.99
3x3x2	42.88	4x4x20	5.286	8x8x10	8.830
3x3x3	4.815	4x4x40	5.243	8x8x12	7.768
3x3x4	3.660	6x6x1	167.49	8x8x14	7.423

Fig. 4.13 Size effect of cellular structures and the fitting curves

Fig. 4.14 Fitted curves for size effect

Table 4.5 Fitting equations and coefficients for size effect

Equation: $y = a - b \ln(x + c)$							
Base repetition	a	Std. Error	b	Std. Error	c	Std. Error	Adj. R-square
3 × 3	98.01055	NA	93.69128	NA	0.28814	NA	NA
4 × 4	131.8443	23.03793	91.78339	13.5846	0.04528	0.25596	0.99778
6 × 6	184.9536	15.75924	95.98559	7.83071	0.19384	0.1924	0.9984
8 × 8	254.5432	11.19867	111.4591	4.6443	0.83091	0.15065	0.99953

As the coefficients exhibit a dependency on the layer number, such relationship was plotted in Fig. 4. 15. The fitting coefficients a and c generally follow a linear trend while the fitting coefficient b exhibits an exponential trend as the function of the layer number.

The fitted formulas for the three coefficients a , b and c are listed on the graphs. The fitting of size effect curves provides statistical evidence to further study the finite structures with various geometry parameters.

Fig. 4.15 Trending and fitted curves for the coefficient a , b and c

4.5 Conclusion

The unit cell approach was adopted to simplify the design of the cellular structure. The join-based method was employed to create the unit cell from the central node and the struts of the unit cells. Joints and strut length were designed as fabrication feature in LPBF.

Various cellular designs including hexagonal, octahedral, and dodecahedral cells can be constructed when taking structural periodicity and symmetry into consideration.

By applying the strain energy method and beam theories, the model for the effective Young's modulus in the building direction was established analytically for the unit cells under three types of deformation models. For strut with large orientations of 45° and 75° , large aspect ratio (0.2 in this study) gave rise to disagreement with Gibson-Ashby theory. On the other hand, at the aspect ratio of 0.2, deformation caused by bending and shearing should be taken account at the strut orientation of 45° ; while deformations caused by all the three types of forces were significant to the effective modulus for the struts with the orientation of 75° . Moreover, the analytical computation was also available to hexahedral and dodecahedral unit cells, and the analytical solutions for hexahedron and dodecahedron exhibit significantly dependence on build orientation and strut diameter.

FEA modeling work was carried out with the octahedron to further analyze the size effect of the octahedral cellular structures. The octahedral structures exhibited strong size effect, which exhibits a logarithm dependency on the number of unit cells. The logarithm curves fitted to the formula of $y = a - b \cdot \ln(x+c)$.

CHAPTER 5

INTEGRATED MATERIAL/STRUCTURAL DESIGN

5.1 Introduction

As the additive manufacturing (AM) technologies become increasingly adopted, the industries have started to pay more interest into the designs and applications of AM cellular lightweight structures. One of the objectives of cellular structure design is to minimize the material consumption and maximize the performance efficiency of materials while ensuring the desired quality by appropriate fabrication approaches.

One of the challenges of realizing cellular structures for applications is the control of the qualities of the fabricated structures. Currently, there exist a large number of research works that investigate the geometrical design and performance evaluations of AM cellular structures, but there is a lack of understanding with the relationships between optimal process selection and the performance of the structures. From the previous studies on both the material-process characterization on in Chapter 2 ~ 3 and the geometry design in Chapter 4, the LPBF products with thin features exhibit heterogeneous material properties, which make them more difficult for the cellular designs. This chapter aims to demonstrate the integrated design of cellular structures that combine the process design and geometry

design to provide a guideline for the optimal processing selection and quality prediction for the cellular structures using LPBF.

5.2 Methodology

5.2.1 Ti-6Al-4V fabrication with LPBF

Ti-6Al-4V pre-alloyed powder material supplied by LPW Technology Inc. (Pittsburg, US) and the EOSINT M 270 LPBF system with a Yb-fiber laser (nominal maximum power 200W) were used to fabricate the parts in this study. For the material of Ti-6Al-4V, a layer thickness of 30 μm is set as default by the EOS system, which determines the slicing thickness. Also, the hatching space of 100 μm (if employed) was set for all the sample fabrication. From Chapter 3, it was shown that the process parameters and geometry parameters have a significant influence on the porosity and the microstructures of the Ti-6Al-4V thin feature structures, as well as their mechanical properties. As the cellular structures are composed of the thin feature struts, the mechanical properties are consequently significantly related to the process parameters and geometry parameters.

5.2.2 Prediction of the effective mechanical properties

The central-joint based structural design was introduced previously to model cellular structures using analytical modeling approach. The simplicity of the beam elements made it possible to utilize beam theories to predict the effective mechanical properties. For the central joint-based cellular units, four primary geometrical parameters (l , d , θ , and n) are designable. The analytical prediction of effective Young's modulus was presented as Eqn. 4-27, which is re-listed here for convenience.

$$E_y^* = \frac{n}{A_{cell}} \left(\frac{l^2 \cos \theta}{12E_s I \sin \theta} + \frac{k \cos^2 \theta}{G_s A \sin \theta} + \frac{\sin \theta}{E_s A} \right)^{-1} \quad (4-27)$$

For the octahedral cellular parts, the expression could be given as Eqn. 4.25, which is also re-listed here.

$$E_y^* = \left(\frac{d}{l} \right)^4 \frac{\pi \sin \theta}{\cos^4 \theta \left[\frac{8}{3} + \frac{40(v+1)}{9} \left(\frac{d}{l} \right)^2 + 2 \tan \theta \left(\frac{d}{l} \right)^2 \right]} E_s \quad (4-25)$$

The analytical prediction of mechanical properties clearly shows the dimensional-dependency of central-jointed cellular structures.

The strength of the LPBF cellular structures could be estimated by the initiation of the yield, which primarily treats these structures as brittle materials. Since the LPBF-produced thin-feature struts exhibit significant surface defects and therefore low ductility [69, 192, 221], it is reasonable to estimate the strength conservatively. By applying von Mises Stress Criterion, the maximum allowable stress can be estimated as follows [211]:

$$\sigma_m = \frac{4}{\frac{A_{cell} l^2 \cos \theta}{\sqrt{2} I} u + \frac{2 \sin^2 \theta}{A^2} + 6 \frac{\cos^2 \theta D^2}{I^2 b^2}} \sigma_y \quad (5-1)$$

where I present the second moment of inertia of the cross-section; A is the cross-sectional area of the strut; u is the distance from the geometrical center of the cross-section to the location of interest; D is the moment of area of the cross-section; b is the width of the cross-section at the location of interest.

Moreover, the geometrical complexity of the structural cellular parts often results in increased computational costs for the FEA analysis. On the other hand, unit cell design approach has been utilized to simplify the periodic structure and to facilitate efficient

designs. Unit cells that represent the geometrical periodicity of the cellular structures are modeled for the properties of interest such as the effective Young's modulus, which can be obtained by Eqn. 4-28. The prediction of effective modulus was performed for the design in Table 4.1, and the relationship between the effective elastic modulus and geometry designs was shown in Fig. 4.10.

In the FEA simulation studies, the unit cells were meshed using the curvature-based method with a tetrahedron mesh size of 0.24 ~ 1.2 mm throughout the FEA work, and the material type was set as Ti-6Al-4V with the elastic modulus of 104.8 GPa and Poisson's ratio of 0.31.

5.2.3 Input of initial CAD design and output of fabrication

A total of 72 unit cells were designed, which can be categorized into two groups. 15 unit-cell designs were fabricated with “perpendicular” orientation, while an additional 9 unit-cell designs were fabricated with tilted orientations. For all the designs, 3 sample replicates were fabricated. The samples were fabricated by EOS M270. The unit cell experimental evaluation consisted of two cases. In Case I, the unit cells were designed with various aspect ratios, building orientations and strut numbers n from the central joint in a half unit cell. Taguchi method was employed for the experimental design for the unit cell geometries as shown in Table 5.1.

Fig. 5.1 Build orientation for unit-cell specimens: Perpendicular orientation (left) and tilted orientation (right)

Table 5.1 Geometry design of unit cells in Case I

#	Aspect ratio l/d	Diameter d (mm)	Build orientation θ ($^{\circ}$)	n	Mark
1	50	0.2	15	3	3-1
2	50	0.2	45	4	3-2
3	50	0.2	75	6	3-3
4	25	0.4	15	4	3-4
5	25	0.4	45	3	3-5
6	25	0.4	75	6	4-1
7	20	0.5	15	4	4-2
8	20	0.5	45	6	4-3
9	20	0.5	75	3	4-4
10	10	0.5	15	6	4-5
11	10	0.5	45	3	6-1
12	10	0.5	75	4	6-2
13	5	1	15	4	6-3
14	5	1	45	6	6-4
15	5	1	75	3	6-5

In Case II, the fabrication qualities of the cellular structures with anisotropic material properties were investigated. An additional variable of “Tilted orientation” was added to be considered. The tilted orientation represents the angle between the symmetry axis and the build direction that is perpendicular to the substrate. Unit cells with three tilted

orientations were fabricated to evaluate the effect of overall strut orientation on part qualities. The design of unit cells in Case II is shown in Table 5.2.

Table 5.2 Geometry design of tilted-fabricated unit cells in Case II

#	Aspect ratio l/d	Build orientation θ (°)	Process combination*	Tilted orientation (°)	Mark
1	5	15	A	15	A1.6
2	10	45	C	30	B1.6
3	16	75	B	45	C1.6
4	5	15	B	45	B0.8
5	10	45	A	15	C0.8
6	16	75	C	30	A0.8
7	5	15	C	30	C0.5
8	10	45	B	45	A0.5
9	16	75	A	15	B0.5

* Process parameter annotation: A: 80 W and 400 mm/s, B: 80 W and 500 mm/s, and C: 60 W and 400 mm/s.

5.3 Results and discussion

5.3.1 Predictability of structural properties

5.3.1.1 FEA model for ideal cellular designs

Generally, the FEA simulations deal with designs with exact features and properties, which means that the material is homogenized and the geometries are in fully defined. Octahedral unit cells ($n = 4$) were taken as an example for FEA evaluations as shown in Fig. 5.2. The effective moduli of the octahedral units with different strut aspect ratio (d/l) were compared with the results from the analytical model. As expected, the analytical results agree to the FEA solutions at different orientation level. In addition, the same agreements were also found for the hexahedral and dodecahedral unit cells.

Fig. 5.2 Comparison of Effective moduli by analytical model and FEA (n=4)

5.3.1.2 FEA model with consideration of dimensional deviations

The quality of LPBF produced thin feature struts were significantly dependent on the dimension as previously discussed in Chapter 3. Firstly, the actual dimension of the thin struts is not very consistent due to the intrinsic process variability. Secondly, the resolution limit of the laser beam in LPBF EOS system causes large dimensional deviation when the nominal diameter is below 0.3 mm, which could exceed 400% with smaller features. Thirdly, the beam offset could also contribute to the dimensional errors. It was found that a beam offset of 0 could generally lead to larger dimensional errors of the thin struts compared to those with beam offset of 40 μm . Under a certain processing condition, the actual dimensions of a strut are as a function of its nominal geometry parameters, which can be expressed as:

$$G = G(i) \quad (5-1)$$

where i represents the input nominal geometry parameter set. In this section, the diameter of the thin struts was considered as a critical parameter, and therefore the function could be defined as:

$$d = d(i) \quad (5-2)$$

where d represents the diameter of the thin struts.

Fig. 5.3 Comparison of FEA and analytical solutions among nominal and actual various dimensions

Consider the case of struts with a specific process parameter set in the designed process window. The actual diameters of the as-fabricated thin struts with varying exposure types and dimensions could be found in Appendix B. The effective moduli were adjusted by applying the actual diameter values on the calculations. Fig. 5.3 shows that the effective

Young's moduli of the as-fabricated octahedral unit cells via both analytical and FEA solutions are larger compared to those with nominal sizes. As expected, as the strut diameter increases, the predicted effective moduli exhibit improved accuracies. Since the exposure of contour and hatch combination tend to result in oversized struts, it is expected that the modeled effective moduli would exhibit larger values.

5.3.2 Estimation of mechanical properties

Based on the results from Chapter 3, the anisotropic material property model was integrated with a geometrical model in the design. As a result, the mechanical properties could be described as:

$$M = F(i, j, \theta) \quad (5-3)$$

where M represents the mechanical property, F is the function of mechanical properties of the Ti-6Al-4V thin-feature struts, j represents the input process parameters in the current LPBF EOS M270 machine, and θ is defined as the build orientation. The mechanical properties of Ti-6Al-4V materials were obtained by tensile testing for macroscale and mesoscale specimens and Eqn. 5-3 was derived via fitting formulations.

5.3.2.1 Tensile properties for bulk specimens

The tensile results for as-built Ti-6Al-4V fabricated in LPBF were discussed in Section 2.6.1, and the corresponding stress-strain curves could be found in Appendix A.

For the tensile test of bulk samples, the geometrical dimension effect becomes less significant. Thus, the variable of i could be treated as a constant in the expression, and Eqn. 5-3 can be simplified as:

$$F(i = bulk) = F(j, \theta) \quad (5-4)$$

For the design of structures with bulk dimension, the mechanical properties can be determined from the existing material property database from previous literature as well as the current studies.

For example, the elastic Young's modulus of 15°-built specimens using 60W and 300 mm/s from experimental results is 106 GPa, and the corresponding UTS exhibits 1190 MPa. In this case, the function of F can be replaced with the particular functions of E and S , and the expressions are:

$$E(60W - 300mm / s, 15^\circ) = 106GPa \quad (5-5)$$

$$U(60W - 300mm / s, 15^\circ) = 1190MPa \quad (5-6)$$

5.3.2.2 Tensile properties for thin-strut specimens

Similarly, for the design of the structures with struts, the applied mechanical parameters can also be determined from the database for the thin struts. In this case, the expression in Eqn. 5-3 is applicable for the thin-strut materials due to their dependency on dimensions.

For example, the elastic Young's modulus of 45° thin-strut specimens with a diameter of 0.5 mm using 60W and 300 mm/s is 115 GPa, and the corresponding UTS exhibits 1270 MPa. The expressions of E and S are, therefore:

$$E(0.5mm, 60W - 300mm / s, 45^\circ) = 115GPa \quad (5-7)$$

$$U(0.5mm, 60W - 300mm / s, 45^\circ) = 1270MPa \quad (5-8)$$

5.3.3 Experimental verification

The experimental- and simulated-based results can provide a database of Ti-6Al-4V thin feature structures for a specific machine, even though in the current studies only a small subset of such database was demonstrated. Based on the database established from the current studies, the experiments were carried out to verify the integrated material performance/structural model.

5.3.3.1 Case study 1: Geometrical and mechanical quality evaluation of unit cells

To evaluate the quality of the fabrication of 3D structures in LPBF and also verify the analytical and FEA computations that were presented in Chapter 4, the unit-cell specimens were designed into two cases (I and II). Case I would focus on the evaluation of an integrated model for symmetric material properties, whereas Case 2 focuses on the evaluation of an integrated model for non-symmetric/heterogeneous material properties. The primary purpose of these cases was to investigate the actual geometry parameters in the as-built unit-cell specimens and compare the experimental mechanical results with the analytical and FEA solutions.

Fig. 5.4 shows the CAD models and samples of the three types of unit cell cellular structures of Case I. The unit cells were built perpendicular to the building plate using the process parameters of 80W and 500 mm/s. Therefore, all the struts in the unit cell structures were oriented along the identical angles. Uniaxial compression tests were performed using a Shimadzu micro-tensile tester, which is equipped with a 500 N load cell. The crosshead moved with a speed rate of 0.5 mm/min during the compression test.

Fig. 5.4 CAD model of unit-cell specimens for compression test of (a) Hexahedron; (b) Octahedron; (c) Dodecahedron; and (d) LPBF As-fabricated unit-cell specimens of the three types

The compression testing results are shown in Table 5.3, and the sample codes are matched to Table 5.1. It was found that the build success rates for the 3D unit cells were lower than the 2D strut specimens, which might be attributed by the damage of the struts in the unit cell structures that face towards the moving direction of the recoater. Due to the shear effect, the strut with an orientation direction opposite to the powder feeding direction is more accessible to break by the recoater, especially if thermal stress causes additional deformation with these struts. For this reason, it is also generally recommended that for LPBF processes the samples should be placed with the orientation angle along the powder feeding direction. In addition, it was observed that the unit-cell specimens were natural to fail while depositing close to the upper joint, especially for the dodecahedral units. It was

believed that at the location when the struts converged to the joint, the increased cross sections would change the thermal situation in the convergent region and reduce the cooling rate, and hence result in additional residual stress (as shown in Fig. 5.5). Since the thin feature structures are quite sensitive to the thermal condition, the increasing energy would result in the failure of the entire structure.

Fig. 5.5 Illustration of convergent struts of a unit cell during fabrication

Moreover, the predictions from the integrated model are consistently higher than the original solutions. Besides, it could also be observed that the experimental modulus values of the unit cells for #3-1 and #3-2 are larger than those from the integrated FEA model, while the others are generally lower than the integrated predicted values. Since the designed diameters for the samples of #3-1 and #3-2 are 0.2 mm, the actual experimental diameters were significantly larger than the design due to the resolution of the fabrication process as discussed previously. Also, due to the aforementioned mechanical testing setup issues, the actual modulus for the struts might not be reliable for the single struts of 0.2 mm. On the other hand, the lower experimental moduli of the samples with larger struts could be contributed by the defects during the fabrication.

Table 5.3 Compression results in Case I

#	Original		Integrated		Experiment		Success rate
	Analytical E_y^* (MPa)	FEA E_y^* (MPa)	Analytical E_y^* (MPa)	FEA E_y^* (MPa)	Exp E_y^* (MPa)	σ_{exp} (MPa)	
3-1	0.00587	0.00604	0.00618	0.00694	0.105	0.078	2 of 3
3-2	0.094	0.095	0.497	0.503	0.864	0.385	1 of 3
3-3	165.19	169.02	171.80	175.56	NA	N/A	0 of 3
3-4	34.39	34.85	68.28	69.65	59.173	9.733	1 of 3
3-5	39063.47	39542.44	30742.78	31023.59	NA	N/A	0 of 3
4-1	0.056	0.059	0.072	0.078	NA	N/A	0 of 3
4-2	0.094	0.110	0.059	0.060	NA	N/A	0 of 3
4-3	0.229	0.231	1.860	1.937	1.784	0.846	2 of 3
4-4	2600.07	2629.52	1188.02	1194.58	NA	N/A	0 of 3
4-5	57.43	58.98	99.31	103.53	39.330	29.781	1 of 3
6-1	4.248	4.262	4.514	4.539	NA	N/A	0 of 3
6-2	67.80	69.58	86.23	89.31	NA	N/A	0 of 3
6-3	2.174	2.205	2.180	2.249	NA	N/A	0 of 3
6-4	3.650	3.796	10.12	11.984	9.528	5.412	1 of 3
6-5	526.09	530.645	840.41	874.91	78.696	26.533	1 of 3

For Case II, the octahedral unit cells were investigated. The additional variable of tilt angle ϕ was added into the design to include heterogeneous material effects. The tile angle ϕ indicates the angle between the symmetry axis and the build direction of the specimens, which is illustrated in Fig. 5.6 (a). Fig. 5.6 exhibits the CAD models of the octahedral unit cells with three different tilt angles of 15°, 45°, and 75°. The experimental design was presented in Table 5.2.

During the fabrication of these samples, in addition to the recoater shearing issue, due to the tilting, some of the unit cell struts were also oriented in a near-horizontal orientation, which could aggravate the strut fabrication failure rates.

Fig. 5.6 CAD model of octahedral unit cells with varying tilt angles, (a) 15°; (b) 45°; (c) 75°

Back to the analytical prediction, since the struts in the unit cells were built with different angles, the strain energy of each strut must be individually calculated due to the now heterogeneous material properties. Treating the struts as Euler-Bernoulli beams, the strain energy for the octahedral unit cell U_i can be expressed as follows:

$$U_{unit} = \sum_i^{2n} U_i \quad (5-9)$$

The strain energy for each strut U_i satisfies the condition in Eqn. 4-9. Similarly, for Timoshenko beams, The U_i will follow Eqn. 4-17. The displacement δ_y of the entire unit cell in the Y direction can be obtained as:

$$\delta_{y-unit} = \frac{\partial U_{unit}}{\partial F} \quad (5-10)$$

and the corresponding strain of the unit cell in Y direction is:

$$\epsilon_{y-unit} = \frac{\delta_{y-unit}}{H_{y-unit}} \quad (5-11)$$

where H_{y-unit} represents the height of the unit cell and can be obtained from the geometry parameters:

$$H_{y-unit} = 2l \sin \theta \quad (5-12)$$

The effective stress applied to the unit cell is contributed by the uniaxial load and the effective area of the unit cell, which can be expressed as:

$$\sigma_{y-unit} = \frac{F}{A_{cell}} \quad (5-13)$$

Therefore, the effective modulus in Y direction for the octahedral unit cell is expressed as follows:

$$E_{y-unit}^* = \frac{\sigma_y}{\varepsilon_y} = \frac{FH_y}{A_{cell}\delta_y} \quad (5-14)$$

The results of the compression test are listed in Table 5.4. The analytical computation for the effective modulus considered heterogeneous material properties by Eqn. 5-14. And the FEA solutions were conducted in SolidWorks by disassembling the octahedral unit cell by eight strut components and applying the corresponding mechanical properties that were achieved previously based on specific geometry parameters and build orientations. The comparison of the modulus values from experiment and FEA modeling is shown. From the results, the differences of elastic modulus among the struts with different build orientations seem to be insignificant. On the other hand, the mechanical strength of the samples appears to be affected more significantly by the quality variations of the strut surfaces and the defects in the specimens due to the differences of orientations.

Table 5.4 Compression test results in Case II

#	E _{Ana} (MPa)	E _{FEA} (MPa)	E _{exp} (MPa)	Difference	σ_{exp} (MPa)	Success rate
A1.6	79.065	81.348	78.49 ± 1.34	3.51%	2.68 ± 0.15	1 of 3
B1.6	584.13	593.20	473.59 ± 0.68	20.16%	10.10 ± 0.32	2 of 3
C1.6	45367.12	45892.38	NA	NA	NA	NA
B0.8	4.590	4.682	2.47 ± 0.45	47.28%	1.04 ± 0.02	1 of 3
C0.8	41.178	42.829	NA	NA	NA	0 of 3
A0.8	3111.23	3193.39	NA	NA	NA	0 of 3
C0.5	0.705	0.724	NA	NA	NA	0 of 3
A0.5	7.198	7.268	NA	NA	NA	0 of 3
B0.5	507.04	510.40	NA	NA	NA	0 of 3

5.3.3.2 Case study 2: compression of sandwich structures

The theoretical effective moduli are expressed as Eqn. 4-13, Eqn. 4-21 or Eqn. 4-25 for an infinite structure. On the other hand, from the size effect for the cellular structures can become significant and therefore should also be taken into consideration. The fitting curves for the size effect are listed in Table 4.5.

Sandwich structures with two skin plates and arbitrary geometrical parameters and strut angle were fabricated in LPBF EOSINT M270 system in order to evaluate the mechanical property size effects of the octahedral structures. The design of parameters is shown in Table 5.5, and the sandwich structure specimens are displayed in Fig. 5.7.

Table 5.5 Design parameters of the octahedral sandwich structure

Design #	Diameter (mm)	Length (mm)	Strut orientation (°)	Repetition of unit cells
1	1.0	8	30	4x4x4
2	1.0	8	30	4x4x2
3	1.3	8	45	2x2x2

Fig. 5.7 LPBF produced cellular structures

Compression test was performed for the cellular samples with an Instron 5569A universal tester. The compressive testing setup is shown in Fig. 5.8. 50 kN load was utilized for the compression test. The actual dimension of the as-built struts was measured using the previously mentioned method.

Fig. 5.8 Setup for the compression test for cellular structures

The specimens were built perpendicularly to the build plate, and therefore, it was assumed that the mechanical properties of the specimens were isotropic, and the experimental results could be compared with the prediction directly. The results of compression tests are shown in Table 5.6. The size effect for the structures with the same number of lateral unit cells was observed. The experimental results also verified that the effective modulus of the octahedral structures exhibited best agreements with the FEA when the layer number was equal to the number of base units. Besides, the experimental values were lower than the predicted values. This could be contributed to the large variability of cross-sectional areas with the struts, which creates an unpredicted critical point of the entire structure and eventually caused the crush.

Table 5.6 Compression results for the octahedral cellular structures

Design #	Dimension D ₁ (mm)	Dimension D ₂ (mm)	Height (mm)	σ_{Exp} (MPa)	E _{Exp} (MPa)	E _{Ana} (MPa)	E _{FEA} (MPa)	E _{FEA-Adj} (MPa)
1	45.88 ± 0.02	45.92 ± 0.05	36.32 ± 0.06	2.07 ± 0.02	40.75 ± 1.53	26.34	69.32	73.57
2	45.62 ± 0.07	45.53 ± 0.05	18.05 ± 0.05	2.95 ± 0.08	175.45 ± 9.81	26.34	224.07	239.17
3	18.76 ± 0.02	18.80 ± 0.03	24.13 ± 0.02	9.94 ± 0.02	239.46 ± 4.98	233.92	528.53	637.68

5.4 Conclusion

To create the material/structural model for Ti-6Al-4V cellular structures, the material process optimization and structural optimization were literature reviewed. The AM fabricated cellular structures behave highly heterogeneous. However, to date, there has not been much literature that discusses this issue. The establishment of the central-joint model provided a simplified approach to integrating material performance and structural

design. The integrated material/structural model was then verified experimentally by selecting the geometry parameter, process parameter and build orientation from the database of the Ti-6Al-4V thin feature structures and calculating the effective mechanical properties. The unit-cell specimens and sandwich cellular structures were tested in compression, and the results were verified with the analytical and FEA solutions.

CHAPTER 6

CONCLUSION AND FUTURE WORK

6.1 Conclusion

6.1.1 The effect of process parameter and build orientation on macroscale Ti-6Al-4V specimens

This study discussed the characterization of Ti-6Al-4V solid material under various process parameters by LPBF process. The range of process parameters, including laser power and scan speed, were decided according to the default parameter set for support structures in the EOS system, which applied lower laser power and scan speed levels than the standard process parameters for bulk Ti-6Al-4V materials.

The single track study was performed for all designed the laser power and scan speed combination. It was apparent that laser power and scan speed had a significant effect on the dimensions and quality of single track melt pool. Increased laser power or decreased scan speed would result in an increased melt pool depth and width, as well as their ratio (depth/width), which is a direct result of the wider and deeper melt pool. At higher energy density level, in some cases, keyhole effect was observed. The investigation of the single tracks establishes a good preliminary understanding about the features fabricated under low energy level process parameters, which could be used to guide the further study of fabrication of thin-feature structures.

The effect of process parameters on the porosities of Ti-6Al-4V solids was significant. The variation of process parameters was directly related to the energy change to the Ti-6Al-4V defect generation during the LPBF process. Fully dense zone (Zone I), over melting zone (Zone II) and incomplete melting zone (Zone III) in the parameter window were responsible for the porosity distribution and the corresponding properties.

Microstructural evolution is primarily a function of the cooling rate. The material in LPBF is processed at high cooling rates. Martensitic α' phase in Ti-6Al-4V parts resulted in high strength. The tensile properties of Ti-6Al-4V coupons in Zone I exhibited better mechanical performance than those in Zone II and Zone III. The mechanical strength exhibit strong anisotropy, with the build orientation exhibiting lowest mechanical strength and highest ductility. For example, the strengths of low angle (15°) tensile coupons exhibit an average of ~ 50 MPa increase compared to that of high angle (75°).

In addition, the mechanical properties of the LPBF Ti-6Al-4V bulk materials also exhibit a significant dependency on the process parameters, which appeared to be mainly contributed by different defect porosities. A few deviating points occurred at large and small energy densities, which was considered to be related to the porosity level. The highest microhardness value of 380 HV was observed at the process parameters of 120W and 500 mm/s.

6.1.2 The effects of process parameter, geometry parameter and build orientation on thin feature Ti-6Al-4V structures

The thin feature struts were fabricated using LPBF EOSINT M270 with various design conditions, including process parameters, exposure types, build orientations, beam offset, and diameters. The fabrication of thin feature structures is sensitive to the scanning

exposure. In the EOSINT M270 system, the hatch exposure was confirmed to be turned on when the diameter of the thin strut was above 0.25 mm. The contour exposure contributed to the roughness of the lateral surface. It was found that the edge exposure was not effective in the scanning procedure of Ti-6Al-4V thin structures with circular cross-section regardless of their dimensions.

The fabricatability of the thin-feature struts was investigated. Generally, lower build orientation led to higher formation success rate. In the Zone I and Zone III of process window, the fabricatability was higher than that in Zone II due to residual stress caused by high energy input.

The geometry accuracy and quality of the thin-feature struts were evaluated through exterior characterization and internal microstructure examination. The exterior observation indicated that the geometry accuracy was highly dependent on the build orientation. Due to the attachment of the unmelted particles, 15°-built struts had a rough down-facing surface that reduced the surface quality, while 45° and 75° struts had a relatively smooth surface. Without support, the low-orientation-built struts exhibited “collapsing” down-facing surface characteristics. The actual dimensions of the struts were generally larger than the designs. The relative errors exhibited increasing trend with the decrease of the strut diameters, and the errors become much more significant at a diameter below 0.2 mm. Results of ANOVA indicated that the factors of diameter, beam offset, build orientation, process parameters significantly influenced the geometry accuracy successively. When the diameter is smaller than 0.2 mm, the resolution limit of LPBF EOS system caused the actual strut diameter to become independent of the nominal size. Process parameters in

Zone II would result in larger actual dimension. Exposure type was not a significant factor in the geometry accuracy of the thin feature struts.

Zone II and Zone III tend to result in higher porosities than Zone I. The porosity of thin struts exhibits a decreasing trend with the reduction of the diameter of struts and the increase of build orientation. Due to rapid cooling in LPBF, columnar grains grew perpendicular to the build substrate, but with a slight angle towards to the strut axial direction. The grain size had a growing trend as the energy density increased and building orientation decreased. Martensitic α' was observed in the sectioned samples.

The mechanical properties of the thin feature struts were characterized by micro tensile test and microhardness test. In the micro tensile test, due to the inaccuracy of extension of the current testing equipment, stress-time curves were plotted instead of stress-strain curves. The UTS was calculated under the assumption of round shape cross sections, and Young's moduli were estimated using the 1.1% strain as the initial yield point. Compared to the large-scale bulk tensile specimens, the UTS exhibited generally larger values, while Young's moduli were comparable. Similar to the large-scale tensile specimens, the process parameters in Zone II and Zone III resulted in defects via different mechanisms, which lead to reduced mechanical properties. the UTS and modulus tend to increase with lower build orientation. Also, at a specific orientation, smaller strut diameter results in larger UTS in some parameters due to the cross-sectional effect. However, the results of UTS and Young's moduli exhibited large variabilities at various diameters and process parameters. The ductility of the thin-feature struts was low, and no necking phenomenon was observed. Additionally, the results of microhardness were generally higher than those of bulk samples, and the difference was about 10 HV. ANOVA results

showed the laser power was the most significant factors to the microhardness, followed by scan speed and build angle. The microhardness results also verified the specimen size effect that smaller dimension led to higher hardness values.

6.1.3 Analytical and FEA prediction on the thin feature cellular structures

Unit cell approach was adopted to simplify the design of the cellular structure. The joint-based method was demonstrated for the creation of the unit cell from the central node with the struts extending to the edge or boundary of the unit cell symmetrically. The connection to the boundaries with periodicity is needed to ensure each strut connected to the neighboring unit cell in the structure.

By applying the strain energy method and beam theorems, the effective Young's modulus in the building direction was established theoretically. FEA modeling work took the octahedron as an example and revealed that the octahedral cellular structure exhibited predictable size effect.

6.1.4 Prediction of the integrated material/structural model

To create the material/structural model for Ti-6Al-4V cellular structures, the material process optimization and structural optimization were integrated. And was utilized to evaluate the performance of the AM fabricated cellular structures with heterogeneous material properties. The integrated material/structural model was verified experimentally by selecting the geometry parameter, process parameter and build orientation from the database of the Ti-6Al-4V thin feature structures and calculating the effective mechanical properties. The unit-cell specimens and sandwich cellular structures were tested in compression testing, and the results were verified with the analytical and FEA solutions.

6.2 Future work and perspective

Currently, the database for the mechanical properties of thin strut specimens is still incomplete with varying fabrication processes, geometry dimensions and build orientations. It is time-consuming to create the database for a specific material with different geometrical and dimensional inputs coupled with different fabrication processes, and it is expected that extensive continuous efforts are needed to create a complete and reliable library by testing large population of specimens. However, in this work, the methodology to create the database for Ti-6Al-4V thin feature structures by a specific machine was demonstrated, even though the current studies contribute to a relatively small set of information compared with the expected comprehensive data library. This study offers a valuable guideline to make a decision of process parameters for a cellular structure. In the future work, more options of geometry design, such as the symmetrical attributes regarding of mutual angles between the neighboring struts, can be taken account in.

Additionally, the current micro tensile test was shown to pose some challenges to the test result reliability. Positively, the difficulties and challenges during this testing provided guidance for the future design of an advanced micro-tensile testing apparatus.

REFERENCE

1. Wang, H.V., *A unit cell approach for lightweight structure and compliant mechanism*. 2005, Citeseer.
2. Wang, H., Y. Chen, and D.W. Rosen. *A hybrid geometric modeling method for large scale conformal cellular structures*. in *ASME Computers and Information in Engineering Conference, Long Beach, CA, Sept.* 2005.
3. Chiras, S., et al., *The structural performance of near-optimized truss core panels*. International Journal of Solids and Structures, 2002. **39**(15): p. 4093-4115.
4. Chu, C., G. Graf, and D.W. Rosen, *Design for additive manufacturing of cellular structures*. Computer-Aided Design and Applications, 2008. **5**(5): p. 686-696.
5. Gibson, L.J. and M.F. Ashby, *Cellular solids: structure and properties*. 1999: Cambridge university press.
6. Lee, Y.-H., et al., *Wire-woven bulk Kagome truss cores*. Acta Materialia, 2007. **55**(18): p. 6084-6094.
7. Gibson, L.J., *Cellular solids*. Mrs Bulletin, 2003. **28**(4): p. 270-274.
8. Hayes, A.M., et al., *Mechanics of linear cellular alloys*. Mechanics of Materials, 2004. **36**(8): p. 691-713.
9. Ashby, M., et al., *Metal foams: a design guide* Butterworth Heinemann. Boston, 2000.
10. Banhart, J., *Manufacturing routes for metallic foams*. Jom, 2000. **52**(12): p. 22-27.

11. Yang, L., *Structural design, optimization and application of 3D re-entrant auxetic structures*. 2011: North Carolina State University.
12. Leitlmeier, D., H.P. Degischer, and H.J. Flankl, *Development of a foaming process for particulate reinforced aluminum melts*. *Advanced Engineering Materials*, 2002. **4**(10): p. 735-740.
13. Ashby, M.F., et al., *Metal foams: a design guide*. 2000: Elsevier.
14. Banhart, J., *Metallic foams: challenges and opportunities*. Eurofoam, 2000. **2000**: p. 13-20.
15. Reed-Hill, R.E., R. Abbaschian, and R. Abbaschian, *Physical metallurgy principles*. Vol. 17. 1973: Van Nostrand New York.
16. Pearson, W.B., *A handbook of lattice spacings and structures of metals and alloys: International series of monographs on metal physics and physical metallurgy*. Vol. 4. 2013: Elsevier.
17. Deshpande, V., M. Ashby, and N. Fleck, *Foam topology: bending versus stretching dominated architectures*. *Acta Materialia*, 2001. **49**(6): p. 1035-1040.
18. Cheah, C., et al., *Rapid prototyping and tooling techniques: a review of applications for rapid investment casting*. *The International Journal of Advanced Manufacturing Technology*, 2005. **25**(3-4): p. 308-320.
19. Wadley, H.N., N.A. Fleck, and A.G. Evans, *Fabrication and structural performance of periodic cellular metal sandwich structures*. *Composites Science and Technology*, 2003. **63**(16): p. 2331-2343.
20. Lee, C., et al., *Rapid investment casting: direct and indirect approaches via fused deposition modelling*. *The International Journal of Advanced Manufacturing Technology*, 2004. **23**(1-2): p. 93-101.

21. Saha, R., et al., *On the evaluation of stability of rare earth oxides as face coats for investment casting of titanium*. Metallurgical and Materials Transactions B, 1990. **21**(3): p. 559-566.
22. Pattnaik, S., D.B. Karunakar, and P. Jha, *Developments in investment casting process—a review*. Journal of Materials Processing Technology, 2012. **212**(11): p. 2332-2348.
23. Jones, S. and P. Marquis, *Role of silica binders in investment casting*. British ceramic transactions, 1995. **94**(2): p. 68-73.
24. Jiang, W., et al., *A new shell casting process based on expendable pattern with vacuum and low-pressure casting for aluminum and magnesium alloys*. The International Journal of Advanced Manufacturing Technology, 2010. **51**(1): p. 25-34.
25. Sypeck, D. and H. Wadley, *Multifunctional microtruss laminates: textile synthesis and properties*. Journal of Materials Research, 2001. **16**(03): p. 890-897.
26. Queheillalt, D.T. and H.N. Wadley, *Pyramidal lattice truss structures with hollow trusses*. Materials Science and Engineering: A, 2005. **397**(1): p. 132-137.
27. Queheillalt, D.T., Y. Murty, and H.N. Wadley, *Mechanical properties of an extruded pyramidal lattice truss sandwich structure*. Scripta Materialia, 2008. **58**(1): p. 76-79.
28. Kooistra, G.W. and H.N. Wadley, *Lattice truss structures from expanded metal sheet*. Materials & Design, 2007. **28**(2): p. 507-514.
29. Kang, K.-J., *A wire-woven cellular metal of ultrahigh strength*. Acta Materialia, 2009. **57**(6): p. 1865-1874.

30. Lee, Y.-H. and K.-J. Kang, *A wire-woven cellular metal: Part-I, optimal design for applications as sandwich core*. Materials & Design, 2009. **30**(10): p. 4434-4443.
31. Andrews, E., et al., *Size effects in ductile cellular solids. Part II: experimental results*. International Journal of Mechanical Sciences, 2001. **43**(3): p. 701-713.
32. Gibson, L., et al. *The mechanics of two-dimensional cellular materials*. in *Proceedings of the Royal Society of London A: Mathematical, Physical and Engineering Sciences*. 1982. The Royal Society.
33. Ashby, M., *The properties of foams and lattices*. Philosophical Transactions of the Royal Society of London A: Mathematical, Physical and Engineering Sciences, 2006. **364**(1838): p. 15-30.
34. Maxwell, J.C., *L. on the calculation of the equilibrium and stiffness of frames*. The London, Edinburgh, and Dublin Philosophical Magazine and Journal of Science, 1864. **27**(182): p. 294-299.
35. Deshpande, V. and N. Fleck, *Collapse of truss core sandwich beams in 3-point bending*. International Journal of Solids and Structures, 2001. **38**(36): p. 6275-6305.
36. Deshpande, V.S., N.A. Fleck, and M.F. Ashby, *Effective properties of the octet-truss lattice material*. Journal of the Mechanics and Physics of Solids, 2001. **49**(8): p. 1747-1769.
37. Wang, H.V., S.R. Johnston, and D.W. Rosen. *Design of a graded cellular structure for an acetabular hip replacement component*. in *International Solid Freeform Fabrication Symposium, Austin, TX*. 2006.
38. Hutchinson, R. and N. Fleck, *The structural performance of the periodic truss*. Journal of the Mechanics and Physics of Solids, 2006. **54**(4): p. 756-782.

39. Yang, L., et al. *Design and characterization of orthotropic re-entrant auxetic structures made via EBM using Ti6Al4V and pure copper*. in *International Solid Freeform Fabrication Symposium*. 2011.
40. Chen, H., et al., *Mechanical property of lattice truss material in sandwich panel including strut flexural deformation*. *Composite Structures*, 2012. **94**(12): p. 3448-3456.
41. Timoshenko, S., *History of Strength of Materials* McGraw-Hill. New York, 1953.
42. Miyanaji, H., S. Zhang, and L. Yang, *A new physics-based model for equilibrium saturation determination in binder jetting additive manufacturing process*. *International Journal of Machine Tools and Manufacture*, 2018. **124**: p. 1-11.
43. Miyanaji, H., et al., *Optimal Process Parameters for 3D Printing of Porcelain Structures*. *Procedia Manufacturing*, 2016. **5**: p. 870-887.
44. Zhang, S., et al. *An experimental study of ceramic dental porcelain materials using a 3D print (3DP) process*. in *Proceeding of Solid Freeform Fabrication (SFF) Symposium*. 2014.
45. Miyanajia, H., et al. *A preliminary study of the graded dental porcelain ceramic structures fabricated via binder jetting 3D printing*. in *Proceedings of the 24th International Solid Freeform Fabrication Symposium. Austin, TX, USA*. 2014.
46. Heigel, J., P. Michaleris, and E. Reutzel, *Thermo-mechanical model development and validation of directed energy deposition additive manufacturing of Ti-6Al-4V*. *Additive manufacturing*, 2015. **5**: p. 9-19.
47. Carroll, B.E., T.A. Palmer, and A.M. Beese, *Anisotropic tensile behavior of Ti-6Al-4V components fabricated with directed energy deposition additive manufacturing*. *Acta Materialia*, 2015. **87**: p. 309-320.

48. Li, W., et al., *Extrusion-on-demand methods for high solids loading ceramic paste in freeform extrusion fabrication*. Virtual and Physical Prototyping, 2017: p. 1-13.
49. Hong, S., et al., *Fabrication of 3D printed metal structures by use of high-viscosity cu paste and a screw extruder*. Journal of Electronic Materials, 2015. **44**(3): p. 836-841.
50. Li, W., et al., *Fabricating ceramic components with water dissolvable support structures by the ceramic on-demand extrusion process*. CIRP Annals-Manufacturing Technology, 2017.
51. Baumer, M., et al. *How can material jetting systems be upgraded for more efficient multi-material additive manufacturing*. in *Proceedings of the Solid Freeform Fabrication (SFF) Symposium*. Texas: The University of Texas at Austin. 2014.
52. He, Y., et al., *An investigation of the behavior of solvent based polycaprolactone ink for material jetting*. Scientific reports, 2016. **6**: p. 20852.
53. Gong, H., et al., *Influence of defects on mechanical properties of Ti-6Al-4V components produced by selective laser melting and electron beam melting*. Materials & Design, 2015. **86**: p. 545-554.
54. Brandl, E., C. Leyens, and F. Palm. *Mechanical properties of additive manufactured Ti-6Al-4V using wire and powder based processes*. in *IOP Conference Series: Materials Science and Engineering*. 2011. IOP Publishing.
55. Yang, L., et al., *Mechanical properties of 3D re-entrant honeycomb auxetic structures realized via additive manufacturing*. International Journal of Solids and Structures, 2015. **69**: p. 475-490.

56. Bourell, D.L., et al., *A brief history of additive manufacturing and the 2009 roadmap for additive manufacturing: looking back and looking ahead*. Proceedings of RapidTech, 2009: p. 24-25.
57. Gibson, I., D.W. Rosen, and B. Stucker, *Additive manufacturing technologies*. Vol. 238. 2010: Springer.
58. Li, T., et al., *3d printing—a review of technologies, markets, and opportunities for the forest industry*. J. Sci. Technol. For. Prod. Process, 2016. **5**(2): p. 30.
59. *EOSINT M270 Basic Training Manual. Section 9*.
60. Kruth, J.-P., et al. *Part and material properties in selective laser melting of metals*. in *Proceedings of the 16th international symposium on electromachining*. 2010.
61. Zaeh, M.F. and G. Branner, *Investigations on residual stresses and deformations in selective laser melting*. Production Engineering, 2010. **4**(1): p. 35-45.
62. Wang, D., et al., *Research on the fabricating quality optimization of the overhanging surface in SLM process*. The International Journal of Advanced Manufacturing Technology, 2013. **65**(9-12): p. 1471-1484.
63. Kempen, K., et al. *Producing crack-free, high density M2 Hss parts by selective laser melting: pre-heating the baseplate*. in *Proceedings of the 24th international solid freeform fabrication symposium. Laboratory for freeform fabrication, Austin, TX*. 2013.
64. Kruth, J.-P., et al., *Benchmarking of different SLS/SLM processes as rapid manufacturing techniques*. 2005.
65. Gibson, I., D. Rosen, and B. Stucker, *Additive Manufacturing Technologies: Rapid Prototyping to Direct Digital Manufacturing*. 2010, Springer.

66. Tsopanos, S., et al., *The influence of processing parameters on the mechanical properties of selectively laser melted stainless steel microlattice structures*. Journal of Manufacturing Science and Engineering, 2010. **132**(4): p. 041011.
67. Yadroitsev, I., P. Bertrand, and I. Smurov, *Parametric analysis of the selective laser melting process*. Applied surface science, 2007. **253**(19): p. 8064-8069.
68. Shen, Y., et al., *The mechanical properties of sandwich structures based on metal lattice architectures*. Journal of Sandwich Structures and Materials, 2009.
69. Cormier, D., et al. *Characterization of thin walled Ti-6Al-4V components produced via electron beam melting*. in *Solid freeform fabrication symposium*. 2004. Citeseer.
70. Murr, L., et al., *Characterization of Ti-6Al-4V open cellular foams fabricated by additive manufacturing using electron beam melting*. Materials Science and Engineering: A, 2010. **527**(7): p. 1861-1868.
71. Gümrük, R. and R. Mines, *Compressive behaviour of stainless steel micro-lattice structures*. International Journal of Mechanical Sciences, 2013. **68**: p. 125-139.
72. Yan, C., et al., *Evaluations of cellular lattice structures manufactured using selective laser melting*. International Journal of Machine Tools and Manufacture, 2012. **62**: p. 32-38.
73. Yang, L., et al., *A comparison of bending properties for cellular core sandwich panels*. Materials Sciences and Applications, 2013. **4**(08): p. 471.
74. Frazier, W.E., *Metal additive manufacturing: a review*. Journal of Materials Engineering and Performance, 2014. **23**(6): p. 1917-1928.
75. Kelly, S.M., *Thermal and microstructure modeling of metal deposition processes with application to Ti-6Al-4V*. 2004, Virginia Tech.

76. Kelly, S. and S. Kampe, *Microstructural evolution in laser-deposited multilayer Ti-6Al-4V builds: Part II. Thermal modeling*. Metallurgical and Materials Transactions A, 2004. **35**(6): p. 1869-1879.
77. Gu, D., et al., *Laser additive manufacturing of metallic components: materials, processes and mechanisms*. International materials reviews, 2012. **57**(3): p. 133-164.
78. Murr, L.E., et al., *Metal fabrication by additive manufacturing using laser and electron beam melting technologies*. Journal of Materials Science & Technology, 2012. **28**(1): p. 1-14.
79. Yan, M. and P. Yu, *An Overview of Densification, Microstructure and Mechanical Property of Additively Manufactured Ti-6Al-4V—Comparison among Selective Laser Melting, Electron Beam Melting, Laser Metal Deposition and Selective Laser Sintering, and with Conventional Powder*, in *Sintering Techniques of Materials*. 2015, InTech.
80. Vilaro, T., C. Colin, and J.-D. Bartout, *As-fabricated and heat-treated microstructures of the Ti-6Al-4V alloy processed by selective laser melting*. Metallurgical and Materials Transactions A, 2011. **42**(10): p. 3190-3199.
81. Zhang, B., H. Liao, and C. Coddet, *Microstructure evolution and density behavior of CP Ti parts elaborated by Self-developed vacuum selective laser melting system*. Applied Surface Science, 2013. **279**: p. 310-316.
82. Yang, J., et al., *Formation and control of martensite in Ti-6Al-4V alloy produced by selective laser melting*. Materials & Design, 2016. **108**: p. 308-318.

83. Dilip, J., et al., *Influence of processing parameters on the evolution of melt pool, porosity, and microstructures in Ti-6Al-4V alloy parts fabricated by selective laser melting*. Progress in Additive Manufacturing, 2017: p. 1-11.
84. Thijs, L., et al., *A study of the microstructural evolution during selective laser melting of Ti-6Al-4V*. Acta Materialia, 2010. **58**(9): p. 3303-3312.
85. Akman, E., et al., *Laser welding of Ti6Al4V titanium alloys*. Journal of materials processing technology, 2009. **209**(8): p. 3705-3713.
86. Leyens, C. and M. Peters, *Titanium and titanium alloys: fundamentals and applications*. 2003: John Wiley & Sons.
87. Majdic, M. and G. Ziegler, *Effect of the Metastable Beta-Phase on Phase Transformations in the Ti Alloy TiAl 16 V 4*. Z. Metallkunde, 1973. **64**(11): p. 751-758.
88. Ahmed, T. and H. Rack, *Phase transformations during cooling in $\alpha + \beta$ titanium alloys*. Materials Science and Engineering: A, 1998. **243**(1): p. 206-211.
89. Xu, W., et al., *Ti-6Al-4V additively manufactured by selective laser melting with superior mechanical properties*. JOM, 2015. **67**(3): p. 668-673.
90. Vilaro, T., et al., *Microstructural and mechanical approaches of the selective laser melting process applied to a nickel-base superalloy*. Materials Science and Engineering: A, 2012. **534**: p. 446-451.
91. Rafi, H.K., T.L. Starr, and B.E. Stucker, *A comparison of the tensile, fatigue, and fracture behavior of Ti-6Al-4V and 15-5 PH stainless steel parts made by selective laser melting*. The International Journal of Advanced Manufacturing Technology, 2013. **69**(5-8): p. 1299-1309.
92. *ASM Handbook*. ASM International, Materials Park, Ohio. **Vol. 1**: p. 2071.

93. *ASM Handbook*. ASM International, Materials Park, Ohio. **Vol. 2**: p. 1894.
94. Vrancken, B., et al., *Heat treatment of Ti6Al4V produced by Selective Laser Melting: Microstructure and mechanical properties*. Journal of Alloys and Compounds, 2012. **541**: p. 177-185.
95. Gong, H., *Generation and detection of defects in metallic parts fabricated by selective laser melting and electron beam melting and their effects on mechanical properties*. Dissertation, 2013.
96. Gong, H., et al. *Defect Morphology in Ti-6Al-4V Parts Fabricated by Selective Laser Melting and Electron Beam Melting*. in *24rd Annual International Solid Freeform Fabrication Symposium—An Additive Manufacturing Conference, Austin, TX, Aug.* 2013.
97. Beuth, J. and N. Klingbeil, *The role of process variables in laser-based direct metal solid freeform fabrication*. JOM Journal of the Minerals, Metals and Materials Society, 2001. **53**(9): p. 36-39.
98. Gong, H., et al. *Melt pool characterization for selective laser melting of Ti-6Al-4V pre-alloyed powder*. in *Solid freeform fabrication symposium*. 2014.
99. Tang, M., P.C. Pistorius, and J.L. Beuth, *Prediction of lack-of-fusion porosity for powder bed fusion*. Additive Manufacturing, 2017. **14**: p. 39-48.
100. Martukanitz, R., et al., *Toward an integrated computational system for describing the additive manufacturing process for metallic materials*. Additive Manufacturing, 2014. **1**: p. 52-63.
101. Petrick, I.J. and T.W. Simpson, *3D printing disrupts manufacturing: how economies of one create new rules of competition*. Research-Technology Management, 2013. **56**(6): p. 12-16.

102. Baufeld, B., O. Van der Biest, and R. Gault, *Additive manufacturing of Ti-6Al-4V components by shaped metal deposition: microstructure and mechanical properties*. Materials & Design, 2010. **31**: p. S106-S111.
103. Rosenthal, I., A. Stern, and N. Frage, *Microstructure and mechanical properties of AlSi10Mg parts produced by the laser beam additive manufacturing (AM) technology*. Metallography, Microstructure, and Analysis, 2014. **3**(6): p. 448-453.
104. Anam, M.A., et al. *Effect of scan pattern on the microstructural evolution of Inconel 625 during selective laser melting*. in *Proceedings of 25th Annual International Solid Freeform Fabrication Symposium*. 2014.
105. Rafi, H., et al., *Microstructures and mechanical properties of Ti6Al4V parts fabricated by selective laser melting and electron beam melting*. Journal of materials engineering and performance, 2013. **22**(12): p. 3872-3883.
106. Hayashi, T., *Dynamic Buckling of Elastic Bars: 1st Report, The Case of Low Velocity Impact*. Bulletin of JSME, 1972. **15**(88): p. 1167-1175.
107. Kenny, S., N. Pegg, and F. Taheri, *Dynamic elastic buckling of a slender beam with geometric imperfections subject to an axial impulse*. Finite elements in analysis and design, 2000. **35**(3): p. 227-246.
108. Bell, J., *The dynamic buckling of rods at large plastic strain*. Acta mechanica, 1988. **74**(1-4): p. 51-67.
109. Kenny, S., F. Taheri, and N. Pegg, *Experimental investigations on the dynamic plastic buckling of a slender beam subject to axial impact*. International journal of impact engineering, 2002. **27**(1): p. 1-17.

110. Hsu, S. and N. Jones, *Quasi-static and dynamic axial crushing of thin-walled circular stainless steel, mild steel and aluminium alloy tubes*. International Journal of Crashworthiness, 2004. **9**(2): p. 195-217.
111. Xu, X., et al., *Dynamic local and global buckling of cylindrical shells under axial impact*. Engineering Structures, 2009. **31**(5): p. 1132-1140.
112. Krauthammer, T., et al., *Size effect for high-strength concrete cylinders subjected to axial impact*. International Journal of Impact Engineering, 2003. **28**(9): p. 1001-1016.
113. Foo, C.C., G.B. Chai, and L.K. Seah, *Mechanical properties of Nomex material and Nomex honeycomb structure*. Composite structures, 2007. **80**(4): p. 588-594.
114. Meraghni, F., F. Desrumaux, and M. Benzeggagh, *Mechanical behaviour of cellular core for structural sandwich panels*. Composites Part A: Applied Science and Manufacturing, 1999. **30**(6): p. 767-779.
115. Duarte, I., M. Vesenjak, and L. Krstulović-Opara, *Dynamic and quasi-static bending behaviour of thin-walled aluminium tubes filled with aluminium foam*. Composite Structures, 2014. **109**: p. 48-56.
116. Harte, A.-M., N. Fleck, and M. Ashby, *The fatigue strength of sandwich beams with an aluminium alloy foam core*. International Journal of Fatigue, 2001. **23**(6): p. 499-507.
117. Daniel, I.M. and J.L. Abot, *Fabrication, testing and analysis of composite sandwich beams*. Composites Science and Technology, 2000. **60**(12): p. 2455-2463.
118. Junior, S.A.R., J.L. Ferracane, and A. Della Bona, *Flexural strength and Weibull analysis of a microhybrid and a nanofill composite evaluated by 3-and 4-point bending tests*. Dental materials, 2008. **24**(3): p. 426-431.

119. Baker, W., T. Togami, and J. Weydert, *Static and dynamic properties of high-density metal honeycombs*. International Journal of Impact Engineering, 1998. **21**(3): p. 149-163.
120. Sharpe Jr, W.N., *Mechanical properties of MEMS materials*. The MEMS handbook, 2001. **20013566**.
121. Srikar, V. and S. Spearing, *A critical review of microscale mechanical testing methods used in the design of microelectromechanical systems*. Experimental mechanics, 2003. **43**(3): p. 238-247.
122. Jadaan, O.M., et al., *Probabilistic Weibull behavior and mechanical properties of MEMS brittle materials*. Journal of materials science, 2003. **38**(20): p. 4087-4113.
123. Tsuchiya, T., et al., *Specimen size effect on tensile strength of surface-micromachined polycrystalline silicon thin films*. Journal of Microelectromechanical Systems, 1998. **7**(1): p. 106-113.
124. Chen, K.S., A. Ayon, and S.M. Spearing, *Controlling and testing the fracture strength of silicon on the mesoscale*. Journal of the American Ceramic Society, 2000. **83**(6): p. 1476-1484.
125. Kraft, O. and C.A. Volkert, *Mechanical testing of thin films and small structures*. Advanced Engineering Materials, 2001. **3**(3): p. 99-110.
126. Weihs, T., et al., *Mechanical deflection of cantilever microbeams: A new technique for testing the mechanical properties of thin films*. Journal of Materials Research, 1988. **3**(05): p. 931-942.
127. Oliver, W.C. and G.M. Pharr, *An improved technique for determining hardness and elastic modulus using load and displacement sensing indentation experiments*. Journal of materials research, 1992. **7**(06): p. 1564-1583.

128. Pharr, G., *Measurement of mechanical properties by ultra-low load indentation*. Materials Science and Engineering: A, 1998. **253**(1): p. 151-159.
129. Harding, D., W. Oliver, and G. Pharr. *Cracking during nanoindentation and its use in the measurement of fracture toughness*. in *MRS Proceedings*. 1994. Cambridge Univ Press.
130. Gibson, I., D. Rosen, and B. Stucker, *Additive manufacturing technologies: 3D printing, rapid prototyping, and direct digital manufacturing*. 2014: Springer.
131. Teng, C., et al., *Simulating Melt Pool Shape and Lack of Fusion Porosity for Selective Laser Melting of Cobalt Chromium Components*. Journal of Manufacturing Science and Engineering, 2017. **139**(1): p. 011009.
132. Kruth, J.-P., et al., *Selective laser melting of iron-based powder*. Journal of Materials Processing Technology, 2004. **149**(1): p. 616-622.
133. Yan, C., et al., *Microstructure and mechanical properties of aluminium alloy cellular lattice structures manufactured by direct metal laser sintering*. Materials Science and Engineering: A, 2015. **628**: p. 238-246.
134. EOSGmbH. *EOS Titanium Ti64 Material data sheet*. Available from: www.eos.info.
135. Elahinia, M., et al., *Fabrication of NiTi through additive manufacturing: A review*. Progress in Materials Science, 2016. **83**: p. 630-663.
136. Lenel, F.V., *Powder metallurgy: principles and applications*. 1980: Metal Powder Industry.
137. Habijan, T., et al., *The biocompatibility of dense and porous Nickel–Titanium produced by selective laser melting*. Materials Science and Engineering: C, 2013. **33**(1): p. 419-426.

138. Cheng, Y.-T. and C.-M. Cheng, *What is indentation hardness?* Surface and Coatings Technology, 2000. **133**: p. 417-424.
139. Kruth, J.-P., et al., *Rapid manufacturing of dental prostheses by means of selective laser sintering/melting*. Proceedings of the AFPR, 2005.
140. Cheng, B., S. Shrestha, and K. Chou, *Stress and deformation evaluations of scanning strategy effect in selective laser melting*. Additive Manufacturing, 2016. **12**: p. 240-251.
141. Carter, L.N., et al., *The influence of the laser scan strategy on grain structure and cracking behaviour in SLM powder-bed fabricated nickel superalloy*. Journal of Alloys and Compounds, 2014. **615**: p. 338-347.
142. Yadroitsev, I., et al., *Single track formation in selective laser melting of metal powders*. Journal of Materials Processing Technology, 2010. **210**(12): p. 1624-1631.
143. Teng, C., et al., *A review of defect modeling in laser material processing*. Additive Manufacturing, 2017.
144. Gu, D. and Y. Shen, *Balling phenomena in direct laser sintering of stainless steel powder: Metallurgical mechanisms and control methods*. Materials & Design, 2009. **30**(8): p. 2903-2910.
145. Kempen, K., et al. *Process optimization and microstructural analysis for selective laser melting of AlSi10Mg*. in *Solid Freeform Fabrication Symposium*. 2011.
146. Yuan, P. and D. Gu, *Molten pool behaviour and its physical mechanism during selective laser melting of TiC/AlSi10Mg nanocomposites: simulation and experiments*. Journal of Physics D: Applied Physics, 2015. **48**(3): p. 035303.

147. Khairallah, S.A., et al., *Laser powder-bed fusion additive manufacturing: Physics of complex melt flow and formation mechanisms of pores, spatter, and denudation zones*. Acta Materialia, 2016. **108**: p. 36-45.
148. Margolin, H. and P. Farrar, *The physical metallurgy of titanium alloys*. Ocean Engineering, 1969. **1**(3): p. 329IN19335-334IN20345.
149. Yang, J., et al., *Role of molten pool mode on formability, microstructure and mechanical properties of selective laser melted Ti-6Al-4V alloy*. Materials & Design, 2016. **110**: p. 558-570.
150. Verhaeghe, F., et al., *A pragmatic model for selective laser melting with evaporation*. Acta Materialia, 2009. **57**(20): p. 6006-6012.
151. Gao, X.-L., et al., *Porosity and microstructure in pulsed Nd: YAG laser welded Ti6Al4V sheet*. Journal of Materials Processing Technology, 2014. **214**(7): p. 1316-1325.
152. Mackwood, A. and R. Crafer, *Thermal modelling of laser welding and related processes: a literature review*. Optics & Laser Technology, 2005. **37**(2): p. 99-115.
153. Ramirez, D., et al., *Open-cellular copper structures fabricated by additive manufacturing using electron beam melting*. Materials Science and Engineering: A, 2011. **528**(16): p. 5379-5386.
154. Gong, H., et al., *Analysis of defect generation in Ti-6Al-4V parts made using powder bed fusion additive manufacturing processes*. Additive Manufacturing, 2014. **1**: p. 87-98.
155. Gu, H., et al. *Influences of energy density on porosity and microstructure of selective laser melted 17-4PH stainless steel*. in *2013 Solid Freeform Fabrication Symposium*. 2013.

156. Collings, E., *The physical metallurgy of titanium alloys*. American Society for Metals, 1984, 1984: p. 261.
157. Brandl, E., et al., *Deposition of Ti-6Al-4V using laser and wire, part I: Microstructural properties of single beads*. Surface and Coatings Technology, 2011. **206**(6): p. 1120-1129.
158. Murr, L., et al., *Microstructures and mechanical properties of electron beam-rapid manufactured Ti-6Al-4V biomedical prototypes compared to wrought Ti-6Al-4V*. Materials characterization, 2009. **60**(2): p. 96-105.
159. Murr, L., et al., *Microstructure and mechanical behavior of Ti-6Al-4V produced by rapid-layer manufacturing, for biomedical applications*. Journal of the mechanical behavior of biomedical materials, 2009. **2**(1): p. 20-32.
160. Murr, L., et al., *Microstructure evolution associated with adiabatic shear bands and shear band failure in ballistic plug formation in Ti-6Al-4V targets*. Materials Science and Engineering: A, 2009. **516**(1): p. 205-216.
161. Facchini, L., et al., *Microstructure and mechanical properties of Ti-6Al-4V produced by electron beam melting of pre-alloyed powders*. Rapid Prototyping Journal, 2009. **15**(3): p. 171-178.
162. Vandenbroucke, B. and J.-P. Kruth, *Selective laser melting of biocompatible metals for rapid manufacturing of medical parts*. Rapid Prototyping Journal, 2007. **13**(4): p. 196-203.
163. *ASTM B265, Standard Specification for Titanium and Titanium Alloy Strip, Sheet, and Plate*. 2015, ASTM International, West Conshohocken, PA.
164. Niinomi, M., *Mechanical properties of biomedical titanium alloys*. Materials Science and Engineering: A, 1998. **243**(1): p. 231-236.

165. Meyers, M.A. and K.K. Chawla, *Mechanical behavior of materials*. Vol. 2. 2009: Cambridge University Press Cambridge.
166. *ASTM E8/E8M-11, Standard Test Method for Tension Testing of Metallic Materials*. ASTM International, West Conshohocken, PA, 2013.
167. Callister, W.D. and D.G. Rethwisch, *Materials science and engineering*. Vol. 5. 2011: John Wiley & Sons NY.
168. Enjo, T., T. Kuroda, and M. NISHIZAWA, *Microstructure and mechanical properties in weld heat affected zone of titanium alloy (materials, metallurgy & weldability)*. Transactions of JWRI, 1988. **17**(2): p. 393-398.
169. Zafari, A., et al., *Achieving fine beta grain structure in a metastable beta titanium alloy through multiple forging-annealing cycles*. Metallurgical and Materials Transactions A, 2016. **47**(7): p. 3633-3648.
170. Ashby, M., C. Gandhi, and D. Taplin, *Overview No. 3 Fracture-mechanism maps and their construction for fcc metals and alloys*. Acta metallurgica, 1979. **27**(5): p. 699-729.
171. Hooke, R., *Micrographia: Or Some Physiological Descriptions of Minute Bodies Made by Magnifying Glasses. With Observations and Inquiries Thereupon*. By R. Hooke, Fellow of the Royal Society. 1665: Jo. Martyn, and Ja. Allestry, printers to the Royal Society, and are to be sold at their shop at the Bell in S. Paul's church-yard.
172. Yang, L., et al., *An investigation of thin feature generation in direct metal laser sintering systems*. Solid freeform fabrication symposium, 2014.
173. Qiu, C., et al., *On the role of melt flow into the surface structure and porosity development during selective laser melting*. Acta Materialia, 2015. **96**: p. 72-79.

174. Sallica-Leva, E., A. Jardini, and J. Fogagnolo, *Microstructure and mechanical behavior of porous Ti-6Al-4V parts obtained by selective laser melting*. Journal of the mechanical behavior of biomedical materials, 2013. **26**: p. 98-108.
175. Fukuda, A., et al., *Osteoinduction of porous Ti implants with a channel structure fabricated by selective laser melting*. Acta Biomaterialia, 2011. **7**(5): p. 2327-2336.
176. Maskery, I., et al., *Mechanical properties of Ti-6Al-4V selectively laser melted parts with body-centred-cubic lattices of varying cell size*. Experimental Mechanics, 2015. **55**(7): p. 1261-1272.
177. Mazur, M., et al., *Deformation and failure behaviour of Ti-6Al-4V lattice structures manufactured by selective laser melting (SLM)*. The International Journal of Advanced Manufacturing Technology, 2016. **84**(5-8): p. 1391-1411.
178. Speirs, M., et al., *The effect of pore geometry on the mechanical properties of selective laser melted Ti-13Nb-13Zr scaffolds*. Procedia CIRP, 2013. **5**: p. 79-82.
179. Yang, L., et al., *Design for Additively Manufactured Lightweight Structure: A Perspective*.
180. Shen, Y., et al., *The mechanical properties of sandwich structures based on metal lattice architectures*. Journal of Sandwich Structures & Materials, 2010. **12**(2): p. 159-180.
181. Hernández-Nava, E., et al., *The effect of defects on the mechanical response of Ti-6Al-4V cubic lattice structures fabricated by electron beam melting*. Acta Materialia, 2016. **108**: p. 279-292.
182. Malek, N.A., et al. *Critical evaluation on structural stiffness of porous cellular structure of cobalt chromium alloy*. in *IOP Conference Series: Materials Science and Engineering*. 2015. IOP Publishing.

183. Heinl, P., et al., *Cellular Ti–6Al–4V structures with interconnected macro porosity for bone implants fabricated by selective electron beam melting*. Acta biomaterialia, 2008. **4**(5): p. 1536-1544.
184. Geetha, M., et al., *Ti based biomaterials, the ultimate choice for orthopaedic implants—a review*. Progress in materials science, 2009. **54**(3): p. 397-425.
185. Attar, H., et al., *Mechanical behavior of porous commercially pure Ti and Ti–TiB composite materials manufactured by selective laser melting*. Materials Science and Engineering: A, 2015. **625**: p. 350-356.
186. Weißmann, V., et al., *Comparison of Single Ti6Al4V Struts Made Using Selective Laser Melting and Electron Beam Melting Subject to Part Orientation*. Metals, 2017. **7**(3): p. 91.
187. Abele, E., et al., *Selective laser melting for manufacturing of thin-walled porous elements*. Journal of Materials Processing Technology, 2015. **215**: p. 114-122.
188. Song, B., et al., *Effects of processing parameters on microstructure and mechanical property of selective laser melted Ti6Al4V*. Materials & Design, 2012. **35**: p. 120-125.
189. Hollander, D.A., et al., *Structural, mechanical and in vitro characterization of individually structured Ti–6Al–4V produced by direct laser forming*. Biomaterials, 2006. **27**(7): p. 955-963.
190. Van Bael, S., et al., *Micro-CT-based improvement of geometrical and mechanical controllability of selective laser melted Ti6Al4V porous structures*. Materials Science and Engineering: A, 2011. **528**(24): p. 7423-7431.

191. Li, R., et al., *Balling behavior of stainless steel and nickel powder during selective laser melting process*. The International Journal of Advanced Manufacturing Technology, 2012. **59**(9-12): p. 1025-1035.
192. Zhang, S., et al. *Property evaluation of metal cellular strut structures via powder bed fusion AM*. in *Proceedings of Solid Freeform Fabrication (SFF) Symposium, Austin, TX, USA*. 2015.
193. Facchini, L., et al., *Ductility of a Ti-6Al-4V alloy produced by selective laser melting of prealloyed powders*. Rapid Prototyping Journal, 2010. **16**(6): p. 450-459.
194. *ASTM E112-12, Standard Test Methods for Determining Average Grain Size*. ASTM International, West Conshohocken, PA, 2013.
195. Filip, R., et al., *The effect of microstructure on the mechanical properties of two-phase titanium alloys*. Journal of Materials Processing Technology, 2003. **133**(1): p. 84-89.
196. Brandl, E., et al., *Mechanical properties of additive manufactured titanium (Ti-6Al-4V) blocks deposited by a solid-state laser and wire*. Materials & Design, 2011. **32**(10): p. 4665-4675.
197. *ASTM F1108 - 97a, Standard Specification for Ti6Al4V Alloy Castings for Surgical Implants (UNS R56406)*. ASTM International, West Conshohocken, PA, 1997.
198. *ASTM F 1472-02a, Standard Specification for Wrought Titanium-6Aluminum-4Vanadium Alloy for Surgical Implant Applications (UNS R56400)*. ASTM International, West Conshohocken, PA, 2003.
199. Kobryn, P.A. and S. Semiatin, *Microstructure and texture evolution during solidification processing of Ti-6Al-4V*. Journal of Materials Processing Technology, 2003. **135**(2): p. 330-339.

200. Wulpi, D., *Understanding How Components Fail*, American Society for Metals. Metals Park (OH, USA), 1985: p. 163.
201. *ASTM E384-17, Standard Test Method for Microindentation Hardness of Materials*. ASTM International, West Conshohocken, PA, 2012.
202. Kodiyalam, S., *Geometry and optimization techniques for structural design*. 1993: Computatio.
203. Maheshwaraa Namasivayam, U. and C. Conner Seepersad, *Topology design and freeform fabrication of deployable structures with lattice skins*. Rapid Prototyping Journal, 2011. **17**(1): p. 5-16.
204. Chahine, G., P. Smith, and R. Kovacevic. *Application of topology optimization in modern additive manufacturing*. in *Solid Freeform Fabrication Symposium*. 2010.
205. Fey, N.P., et al. *Topology optimization and freeform fabrication framework for developing prosthetic feet*. in *Solid Freeform Fabrication Symposium*. 2009.
206. Brackett, D., I. Ashcroft, and R. Hague. *Topology optimization for additive manufacturing*. in *Proceedings of the solid freeform fabrication symposium, Austin, TX*. 2011. S.
207. Gu, W. *On challenges and solutions of topology optimization for aerospace structural design*. in *10th World Congress on Structural and Multidisciplinary Optimization, May*. 2013.
208. Johnston, S.R., et al., *Analysis of mesostructure unit cells comprised of octet-truss structures*. 2006.
209. Mullen, L., et al., *Selective laser melting: A unit cell approach for the manufacture of porous, titanium, bone in - growth constructs, suitable for orthopedic*

- applications. II. Randomized structures.* Journal of Biomedical Materials Research Part B: Applied Biomaterials, 2010. **92**(1): p. 178-188.
210. Qiu, C., et al., *Influence of processing conditions on strut structure and compressive properties of cellular lattice structures fabricated by selective laser melting.* Materials Science and Engineering: A, 2015. **628**: p. 188-197.
 211. Yang, L., *Experimental-assisted design development for an octahedral cellular structure using additive manufacturing.* Rapid Prototyping Journal, 2015. **21**(2): p. 168-176.
 212. Roux, S. and E. Guyon, *Mechanical percolation: a small beam lattice study.* Journal de Physique Lettres, 1985. **46**(21): p. 999-1004.
 213. Hassani, B. and E. Hinton, *A review of homogenization and topology optimization II—analytical and numerical solution of homogenization equations.* Computers & structures, 1998. **69**(6): p. 719-738.
 214. Oleinik, O., *On homogenization problems,* in *Trends and applications of pure mathematics to mechanics.* 1984, Springer. p. 248-272.
 215. Cheng, K.-T. and N. Olhoff, *An investigation concerning optimal design of solid elastic plates.* International Journal of Solids and Structures, 1981. **17**(3): p. 305-323.
 216. Rozvany, G., et al., *Least-weight design of perforated elastic plates—I.* International journal of solids and structures, 1987. **23**(4): p. 521-536.
 217. Fish, J., et al., *Computational plasticity for composite structures based on mathematical homogenization: Theory and practice.* Computer Methods in Applied Mechanics and Engineering, 1997. **148**(1-2): p. 53-73.

- 218. Sigmund, O., *Materials with prescribed constitutive parameters: an inverse homogenization problem*. International Journal of Solids and Structures, 1994. **31**(17): p. 2313-2329.
- 219. Tantikom, K., T. Aizawa, and T. Mukai, *Symmetric and asymmetric deformation transition in the regularly cell-structured materials. Part I: experimental study*. International Journal of Solids and Structures, 2005. **42**(8): p. 2199-2210.
- 220. Dai, G. and W. Zhang, *Size effects of basic cell in static analysis of sandwich beams*. International Journal of Solids and Structures, 2008. **45**(9): p. 2512-2533.
- 221. Yang, L., et al., *Additive Manufacturing of Metals: The Technology, Materials, Design and Production*. 2017, Springer.

APPENDICES

Appendix A: Formation success rate summary

A.1 15° Struts

a. Contour & BO=0

b. Contour + Hatch & BO=0

c. Contour & BO = 40 μm

d. Contour + Hatch & BO = 40 μm

A.2 45° struts

a. Contour & BO=0

b. Contour + Hatch & BO=0

c. Contour & BO = 40 μm

d. Contour + Hatch & BO = 40 μm

A.3 75° struts

a. Contour & BO=0

b. Contour + Hatch & BO=0

c. Contour & BO = 40 μm

d. Contour + Hatch & BO = 40 μm

Appendix B: Analysis on the formation quality of thin struts

B.1 Analysis on 15° struts

B.2 Analysis on 45° struts

B.3 Analysis on 75° struts

Appendix C: Tensile stress-time curves for thin strut specimens

60 W, 300 mm/s:

60 W, 400 mm/s:

15°

45°

75°

0.6 mm

0.5 mm

0.4 mm

Curve raw data damage

0.3 mm

Curve raw data damage

0.2 mm

60 W, 500 mm/s:

80 W, 300 mm/s:

80 W, 400 mm/s:

80 W, 500 mm/s:

100 W, 300 mm/s:

100 W, 400 mm/s:

100 W, 500 mm/s:

CURRICULUM VITAE

Shanshan Zhang

Department of Industrial Engineering
J.B. Speed School of Engineering
University of Louisville
Louisville, KY, 40292

EDUCATION

- Ph.D. Industrial Engineering, University of Louisville, Louisville, KY (Expected: Fall 2017)
Dissertation Title: Design, Analysis, and Application of a Cellular Material/Structure Model for Metal Based Additive Manufacturing Process
 - M.Sc. Material Science, Northeastern University, Shenyang, China (2011)
Thesis Title: Methods of Homogenization on Semi-Continuous Cast Ingots of Al-Mg-Si Alloys
 - B.Sc. Material Science and Engineering, Northeastern University, Shenyang, China (2009)
Thesis Title: Age Forming of Al-Cu-Mg Alloys
-

EMPLOYMENT

- Ph.D. researcher (2012-Current), University of Louisville
 - Selected as the Research assistant of Rapid Prototyping Center in the Department of Industrial Engineering
 - Led the Ph.D. Rapid Prototyping Laboratory research group under the supervision of Dr. Yang
 - Trained fellow students on the knowledge and operation of Additive Manufacturing (AM) machines in Rapid Prototyping Center (RPC) in University of Louisville
 - Mastered the knowledge of Laser Powder Bed Fusion (LPBF) process
 - Modeled the mechanism of the cellular structures using Solidworks
 - Research Assistant (2014-Current), University of Louisville (mentored by Dr. Brent Stucker and Dr. Li Yang)
-

COMPUTER SKILLS

- SOLIDWORKS ➤ MAGICS ➤ AUTOCAD ➤ ORIGIN
 - MATLAB ➤ ANSYS WORKBENCH
 - Microsoft Office Suite (Access, Excel, PowerPoint, Visio, Word)
-

PROFESSIONAL ACTIVITIES

- Reviewer of "International Journal of Mechanics and Materials Design" 2015- present
-

HONORS AND AWARDS

- Acoustic resonance technics Scholarship – University of Louisville (2014 – 2017)
 - Ph.D. fellowship– University of Louisville - (2012 – 2014)
 - Northeastern University Graduate Scholarship (2009-2011)
 - Northeastern University Undergraduate Scholarship (2005-2009)
 - Scholarship from Chinese Academy of Sciences Institute (2008)
 - Selected and Outstanding Student at Northeastern University (2005-2011)
-

ADDITIONAL RESEARCH PROJECTS

- Mechanical behavior of Ti-6Al-4V and CP Ti Dental Implants Fabricated by Selective Laser Melting (SLM or DMLM) and Electron Beam Melting (EBM)
- SLM Process Mapping & Modeling to Correlate Process Parameters on 400W SLM 280 System to Porosity & Build Quality, using 50-micron layers of CoCr Powder (GE Power & Water Research Program)
- Development of a 3D Printing Method for Production of Dental Applications
- Design and Characterization of thin feature cellular structures (NASA)
- Development of a 3D Printing Method for Production of Dental Applications
- Research on Porous Ti Alloy Bio-Materials

PUBLICATIONS

- Shanshan Zhang, Li Yang, “Comparison of characteristics of thin structure and bulky structure in LPBF” (Under review)
- Shanshan Zhang, Li Yang, “Evolution of microstructure of thin structures in LPBF process” (Under review)
- Teng Chong, Haijun Gong, Attila Szabo, J.J.S. Dilip, Katy Ashby, Shanshan Zhang, Nachiket Patil, Deepankar Pal, and Brent Stucker. "Simulating melt pool shape and lack of fusion porosity for selective laser melting of cobalt chromium components." *Journal of Manufacturing Science and Engineering* 139, no. 1 (2017): 011009.
- Hadi Miyanaji, Shanshan Zhang, Austin Lassell, Amirali Zandinejad, Li Yang, “Process development of porcelain ceramic material with binder jetting process for dental applications,” *JOM*, 2016, 68(3), 831-841
- Shanshan Zhang, Li Yang, Brent Stucker, “Property evaluation of metal cellular strut structures via powder bed fusion AM”, *Proceedings of the Solid Freeform Fabrication (SFF) Symposium*, 2015, Austin, TX
- Hadi Miyanaji, Shanshan Zhang, Amir Zandinejad, Li Yang, “Optimal process parameters for 3D printing of dental porcelain structures”, *Proceedings of the SFF Symposium*, 2015, Austin, TX
- Shanshan Zhang, Hadi Miyanaji, Li Yang, Amir Ali Zandinejad, J.J.S. Dilip and Brent Stucker, “An Experimental Study of Ceramic Dental Porcelain Materials Using A 3D Print (3DP) Process”, *Proceedings of the SFF Symposium*, 2014, Austin, TX
- H. Miyanaji, S. Zhang, L. Yang, A. Zandinejad, “A preliminary study of the graded dental porcelain ceramic structures fabricated via binder jetting 3D printing”, *Proceedings of the SFF Symposium*, 2014, Austin, TX
- Li Yang, Shanshan Zhang, Gustavo Oliveira, Brent Stucker, “Development of a 3D Printing Method for Production of Dental Application”, *Proceedings of the SFF Symposium*, 2013, Austin, TX