

Laser Additive Manufacturing of Soft Magnetic Cores for Rotating Electrical Machinery: Materials Development and Part Design

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Degree: PhD in Mechanical Engineering

Contents

Acknowledgements.....	ix
List of Publications	x
Abstract.....	xii
1. Introduction.....	1
2. Impact of AM on Electrical Machine Development: A Feasibility Study Based on the Literature Review	5
2.1 Introduction.....	5
2.2 Additive Manufacturing Technologies	5
2.2.1 <i>Electron Beam Melting</i>	11
2.2.2 <i>Selective Laser Melting</i>	13
2.3 Elements of Electrical Machine Construction and Design	14
2.3.1 <i>Electrical Machine Construction</i>	14
2.3.2 <i>Electrical Machine Design</i>	18
2.4 Soft Magnetic Cores	20
2.4.1 <i>Iron-silicon alloys (Fe-Si, or silicon steels)</i>	22
2.4.2 <i>Iron-Cobalt alloys (Fe-Co)</i>	23
2.4.3 <i>Nickel-Iron alloys (Fe-Ni)</i>	23
2.4.4 <i>Soft ferrites</i>	23
2.4.6 <i>Amorphous Alloys (or Metallic Glasses)</i>	24
2.4.7 <i>Potential for Machine Core Production Using AM Technologies</i>	25
2.5 Permanent Magnets	31
2.5.1 <i>Potential for Permanent Magnet Production Using AM Technologies</i>	33

2.6	Research Scope, Aim and Novelty	34
2.6.1	<i>Research Scope</i>	34
2.6.2	<i>Research Aim</i>	35
2.6.3	<i>Research Novelty</i>	35
2.6.4	<i>Research Objectives</i>	36
3.	Fe-Si Alloys: Metallurgy, Magnetic Properties and Manufacturing Processes ...	39
3.1	Introduction.....	39
3.2	Metallurgical Effects on the Magnetic Properties of Fe-Si Magnets.....	39
3.3	Effects of Si Addition on the Properties of Fe-Si	46
3.4	Existing Processing Routes.....	50
3.5	Summary.....	54
4.	Physical Metallurgy of SLM	56
4.1	Introduction.....	56
4.2	Melt-Pool Dynamics and Pore Formation	56
4.3	Residual Stress and Crack Formation.....	61
4.4	Microstructural Evolution.....	65
4.5	Effect of SLM Process Parameters on Build Quality, Microstructures and Functional Behaviour.....	73
4.5.1	<i>Laser Power</i>	73
4.5.2	<i>Laser Scan Speed</i>	74
4.5.3	<i>Hatch Spacing</i>	76
4.5.4	<i>Layer Thickness</i>	77
4.5.5	<i>Laser Energy Density</i>	77

4.5.6	<i>Laser Scan Strategy</i>	79
4.6	Summary	81
5.	Topological Design Optimisation Methods in Structural and Electromagnetic Applications	83
5.1	Introduction.....	83
5.2	Introduction to Topology Optimisation	83
5.3	Problem Statement for Stiffness Optimisation	84
5.4	Topology Optimisation Methods	85
5.5	Structural TO Using Material Interpolation Scheme	87
5.6	The BESO Method for Structural TO	90
5.7	Other Methods	94
5.7.1	<i>The Homogenization Method</i>	94
5.7.2	<i>The Bubble Method (Topological Gradient)</i>	97
5.8	Topology Optimisation in Electrical Machinery Design	97
5.9	TO for Additive Manufacturing.....	101
5.10	Summary.....	102
6.	Experimental Methods for the Selective Laser Melting of Silicon Iron and Its Characterisation	103
6.1	Introduction.....	103
6.2	Description of Selective Laser Melting System	103
6.3	High-Si Steel Powder	104
6.4	Sample Types.....	105
6.5	System Settings for Silicon Iron Processing.....	107
6.6	Heat-Treatment of the Manufactured Parts.....	109

6.7	Sample Preparation for Characterisation	110
6.7.1	<i>Sample Preparation for Microstructural Characterisation</i>	110
6.7.2	<i>Sample Preparation for Magnetic Characterisation</i>	111
6.8	Build Quality and Microstructural Characterisation	112
6.8.1	<i>Optical Microscopy Studies</i>	112
6.8.2	<i>Studies Employing Scanning Electron Microscopy</i>	113
6.8.3	<i>Micro-Hardness Measurement</i>	113
6.9	Solid Phase Determination and Micro-Stress Measurement Using X-Ray Diffraction	113
6.9.1	<i>X-Ray Diffraction Principles and System Description</i>	113
6.9.2	<i>Use of X-Ray Diffraction for Solid Phase Determination</i>	115
6.9.3	<i>Use of X-Ray Diffraction for Lattice Parameter Determination</i>	115
6.9.4	<i>Use of X-Ray Diffraction for Micro-Stress Measurement</i>	115
6.10	Crystallographic Texture Characterisation Using Electron Backscatter Diffraction	117
6.10.1	<i>Electron Backscatter Diffraction Principles and System Description</i>	117
6.10.2	<i>Presentation of the Crystallographic Texture Results</i>	119
6.11	Characterisation of the Magnetic Properties	121
6.11.1	<i>Measurement of the DC Magnetic Properties</i>	122
6.11.2	<i>Measurement of the Dynamic Magnetic Properties</i>	125
6.11.3	<i>Post-Processing of the Acquired DC Magnetisation Data</i>	127
6.12	Summary	129
7.	Metallurgy of High-Silicon Steel Parts Produced by Selective Laser Melting..	130
7.1	Introduction	130

7.2	Results of Powder Characterisation.....	130
7.3	Study of Porosity and Crack Formation	131
7.3.1	<i>Pore and Crack Observation Results</i>	131
7.3.2	<i>Origin of Defects</i>	138
7.3.3	<i>Effect of Different Power-Speed Combinations on Build Quality</i>	140
7.4	Study of Solid Phase Formation and Microstructures	141
7.4.1	<i>XRD Studies</i>	141
7.4.2	<i>Metallographic Studies</i>	143
7.4.3	<i>Origin of Solid Phase Formation</i>	148
7.5	Study of Crystallographic Texture.....	149
7.5.1	<i>EBSD Studies</i>	149
7.5.2	<i>Origin of Texture</i>	152
7.6	Effect of Annealing.....	154
7.6.1	<i>Metallographic Studies</i>	154
7.6.2	<i>XRD Studies</i>	155
7.6.3	<i>EBSD Studies</i>	157
7.7	Summary.....	160
8.	Magnetic Properties of High-Silicon Steel Produced by Selective Laser Melting: Effect of Laser Energy Input and Annealing Temperature.....	162
8.1	Introduction.....	162
8.2	Stress-Relief of SLM Cubic Samples	162
8.3	Effect of Laser Energy Input on the Magnetic Properties	166
8.4	Effect of Annealing.....	173

8.5	Summary.....	176
9.	Methodology for Electrical Machine Rotor Core Design for Production Using Laser Additive Manufacturing.....	178
9.1	Introduction.....	178
9.2	Design Workflow.....	179
9.3	Electrical Machine Analysis Using the Finite Element Method.....	181
9.3.1	<i>Machine Specifications and Initial Design</i>	181
9.3.2	<i>Material Properties</i>	182
9.3.3	<i>Finite Element Model Setup: 2D Case</i>	186
9.3.4	<i>Finite Element Model Setup: 3D Case</i>	187
9.4	Rotor Core TO using the BESO Method	189
9.4.1	<i>Structural Problem</i>	189
9.4.2	<i>Magnetostatic Problem</i>	190
9.4.3	<i>Combining Structural and Magnetostatic Sensitivities</i>	195
9.4.4	<i>Adopted Evolutionary Parameters</i>	197
9.5	Summary.....	198
10.	Electric Machine Rotor Core Topology Optimisation and Manufacturing by Selective Laser Melting	199
10.1	Introduction.....	199
10.2	Magnetostatic Topology Optimisation	199
10.2.1	<i>Initial Motor Performance</i>	199
10.2.2	<i>Linear Magnetic Properties</i>	203
10.2.3	<i>Nonlinear Magnetic Properties</i>	205
10.3	Joint Structural-Magnetostatic TO.....	209

10.3.1	<i>Two-Dimensional Studies</i>	209
10.3.2	<i>Three-Dimensional Studies</i>	213
10.4	Rotor Manufacturing Using SLM: Preliminary Results	216
10.4.1	<i>Process Parameter Optimisation</i>	216
10.4.2	<i>Rotor Manufacturing</i>	220
10.5	Summary	223
11.	General Discussion	225
11.1	Main Contributions in Material Processing and Characterisation	226
11.2	Main Contributions in Component Design	228
12.	Conclusions and Further Work	230
12.1	Further Work	231
12.1.1	<i>Material Processing</i>	231
12.1.2	<i>Component Design</i>	232
	References	234

Acknowledgements

First of all, I would like to thank my family, especially my mother Brunella, my father Luciano and my sister Carolina, for their unconditional love and support during all these years. Working on my PhD was a challenging task, which made it sometimes difficult for me to enjoy the few moments I could share with my family, and to show you how much your presence and support means to me. Therefore, this is the occasion to thank you, for accepting me for who I am, for patiently listening to my frequent whining, for never making me feel lonely, and for being the best family one could wish for. I love you, and I hope I could make you proud.

I would also like to thank all of my dear dear friends Ale, Anvy, Bobo, Davidino, Flo, Seb, Sem and Zucca. Despite being spread across different countries (Italy, U.K. and France... and even Canada at some point!), I am happy to say that we always managed to keep in touch and share some unforgettable experiences. I literally don't understand how you can still stand my frequent disappearances. You don't imagine how blessed I feel for having you in my life, you taught me a lot throughout our years of friendship, and you made me the person I am today.

A special thank you also to the people that supported my daily work at Nottingham: my supervisors - Ian, Richard and Chris - the lab technicians – Mark, Mark and Joe – and all of the other amazing professionals I have met during my studies.

Last but not least, I would like to thank my wonderful partner Olympia, an amazingly clever, beautiful and loving woman that I wish I met earlier. You made the last, and most difficult, months of my PhD studies seem more bearable, and on a daily basis you manage to remind me (without even having to say it) how the little things in life are, ultimately, what really make us happy. I love you.

List of Publications

Journals:

- 1) M. Garibaldi, I. Ashcroft, M. Simonelli, R. Hague, *Metallurgy of high-silicon steel parts produced using Selective Laser Melting*, Acta Materialia, 110 (2016), 207-216.
- 2) M. Garibaldi, I. Ashcroft, J. Lemke, M. Simonelli, R. Hague, *Effect of Annealing on the Microstructures and Magnetic Properties of Soft Magnetic Fe-Si Produced via Laser Additive Manufacturing*, Scripta Materialia, 142 (2018), 121-125.
- 3) J. Lemke, M. Simonelli, M. Garibaldi, I. Ashcroft, M. Vedani, R. Wildman, C. Tuck, R. Hague, *Calorimetric Study and Microstructure Analysis of the Order-Disorder Phase Transformation in Silicon Steel Built by SLM*, Journal of Alloys and Compounds, 722 (2017), 293-301.
- 4) M. Garibaldi, I. Ashcroft, N. Hillier, S. Hammond, R. Hague, *Relationship between Process Parameters and Magnetic Properties in Selective Laser Melted Fe-6.9%Si Soft Magnets*, Materials Characterisation, accepted for publication, in press.
- 5) M. Garibaldi, C. Gerada, I. Ashcroft, R. Hague, *Free-Form Design of Electrical Machine Soft Magnetic Cores for Production Using Laser Additive Manufacturing*, submitted for publication.

Conferences:

- 1) M. Garibaldi, I. Ashcroft, C. Gerada, N. Hillier, M. Hall, R. Hague, *Microstructure and Magnetic Properties of High-Silicon Steel Produced by Additive Manufacturing*, 7th International Conference on Magnetism and Metallurgy (WMM2016), Rome, June 2016.

- 2) M. Garibaldi, C. Gerada, I. Ashcroft, R. Hague, H. Morvan, The Impact of Additive Manufacturing on the Development of Electrical Machines for MEA Applications: A Feasibility study, More Electric Aircraft Conference (MEA 2015), Toulouse, February 2015.
- 3) M. Garibaldi, I. Ashcroft, R. Hague, *Challenges and Opportunities in the Additive Manufacture of Soft Magnetic Cores for Use in Rotating Electrical Machinery: Materials, Processes, and Component Design*, 146th Annual Meeting of The Minerals, Metals & Materials Society (TMS 2017), San Diego, March 2017 (Invited presentation, abstract only).

Abstract

This research work addresses the application of Additive Manufacturing (AM) technologies in novel electrical machinery. The unrivalled design freedom offered by AM has the potential to revolutionise the way rotating electrical motors are designed and manufactured. The thesis investigates the possibility offered by AM to advance the design of electrical machines for lightweight and high performance, with potential in several industrial sectors, including automotive and aerospace. In particular, we investigate how the performance of electrical motors can be improved by manufacturing the soft magnetic rotor core using Selective Laser Melting (SLM) and how the consequent design choices affect the component's performance from a torque-to-weight perspective. First, the metallurgical and soft magnetic properties of high-silicon steel (Fe-6.7%wt.Si) parts produced by Selective Laser Melting (SLM) will be characterised and discussed as function of the processing parameters and post-processing heat-treatments. The results of this research show that, by means of SLM, high-silicon steel parts with permeability (above 24000) and hysteresis loss (coercivity below 20 A/m) comparable to the ones of laminated high-silicon iron can be produced. Then, FEM-based modelling and Topology Optimisation (TO) will be employed to design the rotor core of a surface-mount Permanent Magnet (PM) machine in order to achieve maximum torque-to-weight ratio while maintaining structural integrity. Importantly, three-dimensional FEA results show that the weight of an existing PM rotor core can be slashed by 50% without affecting its torque performance when the proposed TO scheme is employed and the actual mechanical and magnetic properties of the SLM material are considered. Finally, we suggest that further research should be aimed at extending the range of applicability of the proposed TO scheme to other machine topologies (i.e., Synchronous Reluctance), as well as at the adjustment of the alloy's chemistry for improved ductility to avoid the risk of in-process cracking.

1. Introduction

Since its invention in the first half of the 19th Century, electrical machinery has established itself as a fundamental component in several industrial sectors. In the form of motors, electrical machines are employed to transform magnetic energy into rotational motion. In the form of generators, electrical machines transform a mechanical torque into an electromotive force. Importantly, the demand for vehicle electrification in both the automotive and aerospace industries is deemed to increase over the next decades owing to ever stricter emission norms over climate change concerns (Howey, 2012) and the need for lower operating and maintenance costs (Howse, 2003). It is not surprising then that the research efforts towards lighter and more efficient electrical machines are growing and span multiple disciplines, including those of materials, production technology and component design (Krings et al.).

In recent years it has been shown that in structural applications component weight reduction and efficiency improvement through geometrical complexity can be achieved by means of Additive Manufacturing (AM) technologies, especially in the form of powder bed fusion systems, such as Selective Laser Sintering (SLS), Selective Laser Melting (SLM) and Electron Beam Melting (EBM) (Hopkinson et al., 2006). Unlike conventional methods that are either formative or subtractive, AM technologies build up components layer by layer. The success of the AM technologies derives from the possibility of creating sound, near-net parts of almost any geometry with little or no added cost. Therefore, AM is giving the possibility to design and manufacture products around the actual target performance. In this respect, the combination of AM technologies and Finite Element Analysis (FEA)-based free-form design

methodologies, especially Topology Optimisation (TO), is proving especially valuable in achieving significant material saving and component weight reduction. Therefore, AM is emerging across a broad range of sectors, including automotive, medical and aerospace, for the creation of functional parts such as the examples reported in Figure 1-1, which were taken from a number of industrial and academic sources. In particular, the AlSi10Mg motorbike frame in Figure 1-1(b) is an example of how AM can be used in combination with TO in order to achieve substantial weight savings.



(a)



(b)



(c)



(d)

Figure 1-1: Examples of metal employment in a number of applications and industrial sectors. Jet engine fuel injector by Morris Technologies (Zelinski, 2009) (a), motorbike frame by Airbus APWorks (Airbus APWorks GmbH, 2016) (b), patient-specific implant by EOS (EOS GmbH, 2015) (c) and aerospace bracket used on the Airbus A350 XWB (Martin, 2016).

Despite the employment of AM technologies for processing a vast range of dielectric and metallic materials, including several Steels, and Titanium, Cobalt and Aluminum alloys (Thijs et al., 2013a, Simonelli et al., 2014, Niendorf et al., 2013), its application to materials potentially relevant to the manufacturing of electrical machines seems limited to a few recent contributions involving the processing of soft and hard ferromagnetic materials (Zhang et al., 2012, Huber et al., 2017). The lack of extensive research on the potential of AM on the development of electrical machines indicates a gap in the literature, as it seems that AM has the potential to open the path to a new generation of high torque-to-weight devices, with a consequent direct impact on the development of more-electric vehicles. Therefore, the present research aims at investigating the advantages of using AM for the development of electrical machines, both from the materials and design perspectives. While the best way to fully exploit the disruptive potential of AM would be through the design of completely new machine topologies, this might prove challenging. Therefore, we choose to try and improve an existing design, and the selection of candidate machine parts for fabrication using AM technologies is performed taking into account the following issues:

- 1) the availability, costs and manufacturability via AM of the materials they are composed of;
- 2) the maturity of the AM technologies available;
- 3) the potential for performance improvement through geometry design optimisation, which is arguably the main advantage AM technologies can bring.

Based on the above criteria and on a thorough literature review, the rotor core of a synchronous machine is selected in this research for re-design for AM, and a soft magnetic high-silicon steel alloy is developed to produce the re-designed component by means of SLM. After performing an investigation of the effect of AM processing on the microstructural and mechanical properties of high-silicon steel, the rotor core is re-designed considering the material properties of the additively manufactured material in order to improve the overall torque density (i.e., the torque-to-weight-ratio) of the electrical machine.

This thesis is structured as follows. In Chapter 2 we carry out a feasibility study based on the literature review, aimed at providing guidelines towards the selection of a meaningful target application for this research. This will lead to the definition of the specific aim and objectives of this research. In Chapter 3, 4 and 5, respectively, we cover some fundamental notions in the fields of silicon steels, SLM and TO, which will provide the background knowledge needed to understand and interpret the results of this research. In Chapter 6, the experimental methods used for the processing, post-processing and characterisation of high-silicon steel are presented. The results of the metallurgical and magnetic characterisation of high-silicon steel are covered in Chapter 7 and 8, respectively. Chapter 9 presents the TO scheme developed to carry out the functional design of the electrical machine's rotor core, while in Chapter 10 the results of the TO-based rotor design are presented and discussed. Finally, a general discussion and conclusions of this work are presented in Chapter 11, where an outlook to the future developments of this work are also proposed.

2. Impact of AM on Electrical Machine Development:

A Feasibility Study Based on the Literature Review

2.1 Introduction

A feasibility study trying to shed light on the potential for electrical machine construction by means of AM technology is proposed in this chapter, based on the relevant literature. This chapter starts with an overview of the main AM technologies available on the market, followed by an insight into the electrical machine technology and main machine types. Then, we describe the active parts of a typical electrical machine, namely, the soft magnetic cores and the permanent magnets, with special regard to the relevant materials. Electrical conductors (windings and rotor cages made out of an electrically conductive material, usually copper or aluminium, and an insulating material coating) are also fundamental components of electrical machines. However, these will be excluded from the study, due to the lack of mature AM technologies able to produce multi-material end-use parts. The feasibility study will allow us to conclude that the soft magnetic core of synchronous machine rotors has the greatest potential for manufacturing using AM technologies, and that Selective Laser Melting (SLM) represents the best option to achieve this goal. This conclusion will allow us, in turn, to define the scope, aim and novelty of this research.

2.2 Additive Manufacturing Technologies

The term “additive manufacturing” (AM) is used to indicate the class of technologies that produce functional components from a digital model file by

adding material layer-upon-layer (ASTM, 2012). AM presents a radically different approach as compared to forming or subtractive methodologies and for this reason it presents several advantages. Importantly, AM increases greatly the design freedom, thus offering exciting possibilities as products with new functional geometries can be created at no added cost. In particular, it has been shown that AM helps to reduce the production costs for components that would be extremely costly if manufactured with conventional processes (Baumers et al., 2016b). Since AM technologies add material only where it is needed, no production line is needed to make specific tools, moulds/dies, etc. This makes AM very valuable for low-volume production, where the technology is well-known to compare favourably against traditional manufacturing technologies, such as injection moulding, as shown in Figure 2-1 (Ruffo et al., 2006). In low-to-medium volume production, AM technologies are beneficial where a part's performance gain through geometrical design is more important than its per-unit cost. In this respect, the aerospace industry has been showing increasing interest in the AM technologies for the production of end-use parts and assemblies (GE Aviation, 2016). Another advantage of AM is represented by the simplification of the supply chain, since entire assemblies can be produced within the same AM system, thus avoiding inventory and transportation costs (Thomas, 2016). As far as environmental sustainability is concerned, the biggest opportunities offered by AM mainly regard the compression of the supply chain and the savings in fuel consumption that efficiently-designed components can produce in automotive and aerospace applications. On the other hand, the efficiency and total energy consumption of the AM processes are mainly dependent on the

specific system and materials considered. This is largely due to the fact that the energy needed to melt material varies greatly among materials. Another issue affecting sustainability with polymer powder-bed systems (laser sintering) is the limited recyclability of the currently employed powders (Hague and Reeves, 2013).

One of the barriers to AM technology adoption is the low productivity of current AM platforms compared to traditional manufacturing, which is regarded as an hindering factor to achieving economies of scales (Baumers et al., 2016a). Other important barriers to AM adoption are the high machine and raw materials costs, the difficulties in achieving AM-friendly designs (e.g., lattice structures) using traditional design software and the limited number of materials that can currently be processed using AM systems (Hague and Reeves, 2013).

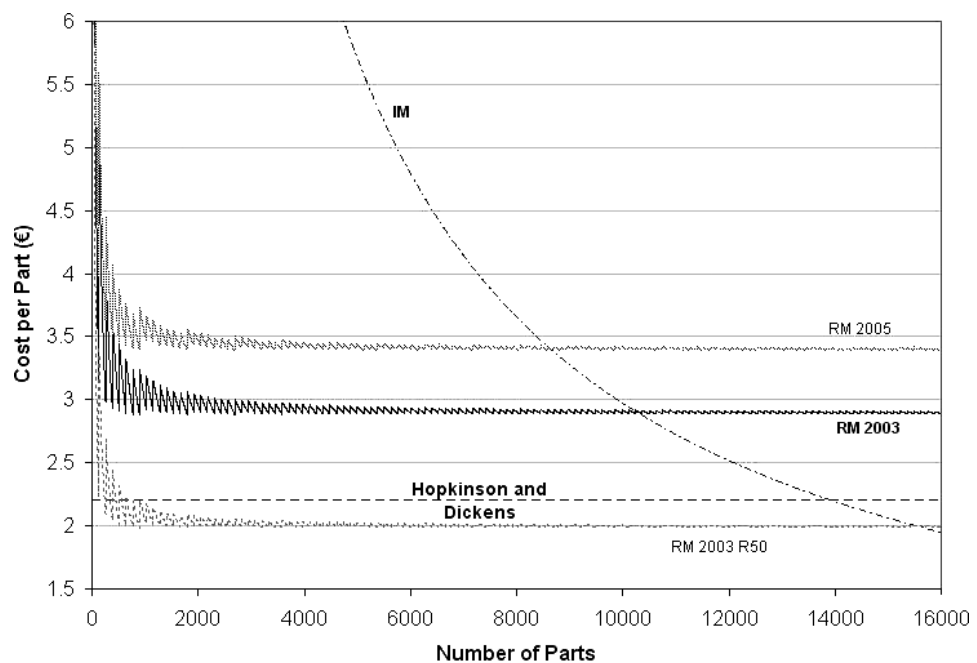


Figure 2-1: Comparison of the cost per part of laser sintering, an AM technology, and traditional injection moulding (IM), according to the models proposed by different sources (Ruffo et al., 2006).

While the AM processes vary greatly in terms of the materials processed and layer deposition strategies, the work-flow of all additive technologies can be broken down into seven basic steps (Gibson et al., 2010):

- 1) **Creation of the digital model** corresponding to the object that is intended to print. The digital model of the product is normally created using any CAD software. Because the model is created layer-by-layer, freeform design strategies, such as topology optimisation schemes, can be implemented as part of the design stage;
- 2) **Conversion of the CAD model into the Standard Tessellation Language (.STL) format.** This operation is normally carried out through any standard CAD software. The conversion involves the model's surface being replaced by a mesh of triangles. Storage of the .STL files takes place in terms of the unit normal and vertices of the generated triangles. Once in the .STL format, the model is oriented in the AM building platform coordinate system. Any overhangs present in the model and typically the base of the STL mesh are then automatically (or manually) supported by auxiliary structures (so-called support structures). The supported STL model is finally "sliced" into a number of layers;
- 3) **Loading of the STL model in the AM machine environment.** Once the STL model has been oriented, supported and sliced into horizontal layers, it can be transferred into the manufacturing machine environment. Different parts can be added onto the same build platform for simultaneous fabrication;

- 4) **Set-up of the process parameters.** The manufacturing process parameters, such as heat-source power, speed, etc. are setup in relation to the desired final properties of the built component/s, such as low surface roughness, low porosity, etc., as well as on a number of other constraints, such as productivity. The optimal parameters are generally unique for each processed material-manufacturing platform-part geometry combination, and should be determined through a stage of systematic parameter optimisation;
- 5) **Building stage.** The process is then launched and the part is created layer-by-layer, by depositing powder or wire material (depending on the AM technology employed), according to the corresponding CAD model;
- 6) **Removal of the end-use component from the building platform.** Once the “build” is completed, it will be detached from the build platform and supporting structures;
- 7) **Part post-processing.** This last stage refers to a number of finishing procedures necessary for a part to become operational. These may include: milling, polishing, thermal and/or mechanical treatments.

Table 2-1 shows the main AM technologies commercially available at the moment for producing end-use parts, classified depending on the layer deposition strategy employed. The standard materials that can be processed with each technology are also reported in the table (Gibson et al., 2010, Arcam AB, 2016, Sciaky Inc., 2017). Note that the list does not include materials processed at fundamental research levels. Among the industrial processes

included in the list, Direct Energy Deposition (DED), despite being the only one capable of processing multiple materials contemporarily among the existing industrial AM technologies, is typically employed for repairing and maintaining structural parts, rather than as a fully-developed AM technology (Gibson et al., 2010), owing to the poor resolution of the end parts, which normally require extensive machining before use, as the turbine blade example reported in Figure 2-2 suggests. Fusion Deposition Modelling (FDM) and Selective Laser Sintering (SLS), despite being especially employed for prototyping purposes, are also being employed for the fabrication of end-use parts. However, they are limited to polymeric materials, as Table 2-1 indicates.

Given the aim of this chapter, which is to establish the potential for electrical machine part manufacture using AM, only the AM processes capable of processing high-density metals/alloys with sufficiently good surface finish will be described in more depth. The AM technologies of interest for this work therefore are: Selective Laser Melting (SLM) and Electron Beam Melting (EBM). The technical aspects of these two processes, as well as a comparative assessment of the two technologies, are described in the following sections.

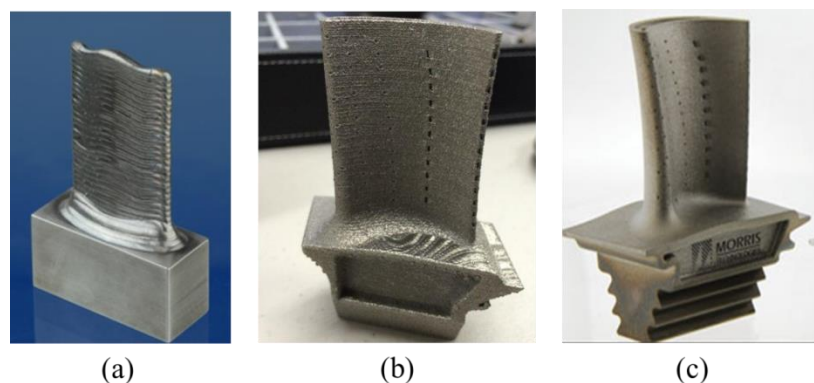


Figure 2-2: comparison of turbine blades manufactured using DED (a), EBM (b) and SLM (c). Images were taken from www.epma.com/additive-manufacturing, (Arcam AB, 2016) and www.morristechpartners.com/, respectively.

Table 2-1: map of main AM technologies for the fabrication of end-use parts (Gibson et al., 2010, Arcam AB, 2016, Sciaky Inc., 2017, Exone, 2017)

Deposition Method	Technologies	Processed Materials
<i>Powder Bed</i>	Selective Laser Sintering (SLS)	Polyamides, polystyrenes, thermoplastic elastomers, some composites (e.g., glass-filled Nylon)
	Selective Laser Melting (SLM)	Maraging steels, stainless steels, Ti, Ti-alloys, Al-alloys, Ni-Ti, Co-Cr, Au, Ag, Inconel, Bronze Alloys, Cu-Alloys
	Electron Beam Melting (EBM)	Ti-6Al-4V, Ti-6Al-4V ELI, Ti Grade 2, Co-Cr
<i>Directed Energy Deposition</i> (Fed-wire or blown powder)	Laser/Electron Engineered Net Shape (LENS, EBAM)	Ti-, Ti-alloys, Inconel, Ta, W, Nb, Cu,, Cu-Ni, Steels, Al, Zircalloy
<i>Binder-Jetting</i>	Binder-Jetting (BJ)	Stainless Steels, Inconel, Fe-Cr-Al, Co-Cr, W-Carbide
<i>Filament Extrusion</i>	Fused Deposition Modelling (FDM)	ABS, PLA

2.2.1 Electron Beam Melting

Electron Beam Melting (EBM) was developed and commercialised in 2001 by Arcam AB (Arcam AB, 2016). Figure 2-3 shows the schematic of the EBM process and a corresponding commercial system, the Arcam Q10. EBM uses a high-power electron beam (typically 4 kW) to scan layers of powder, one at a time. Because gases in the build chamber would interfere with the electron beam, EBM is carried out in vacuum (typical pressure is 10^{-4} - 10^{-5} mbar). By feeding helium to the work area during the melting process the pressure inside the system is increased to approximately 1 Pa in order to avoid electrical charging of the powder particles (Herzog et al., 2016). The build platform is kept at high temperature (typically 500°C to 700°C) to promote in-process stress-relief and avoid residual stress-related issues, such as part warping and cracking, and therefore reduce post-processing time. Furthermore, build

platform pre-heating affects the microstructure of the final part, and large grain sizes can be achieved when temperatures close to 700°C are employed (Herzog et al., 2016). In practice, during EBM each layer of powder, whose typical thickness amounts to 50-200µm, is scanned twice. The first initial scan is carried out using low beam current and high scan rates to heat up the powder bed for partial sintering. Full melting of the powder is then achieved with a second scan, typically conducted by a high-power beam with scan speeds up to 8000 mm/s. Owing to the power of the electron beam, the typical build rate for EBM systems amounts to 55-80 mm³ /s (for Ti-6Al-4V) (Arcam AB, 2016), which makes it one of the most productive systems in the AM field (Baumers et al., 2016a). The relatively large spot size of the electron beam hitting the powder bed (~200µm) causes the surface roughness of EBM parts to be relatively high as compared to SLM, as shown in Figure 2-4. As of 2017, Arcam AB holds intellectual property rights for the EBM systems it produces.

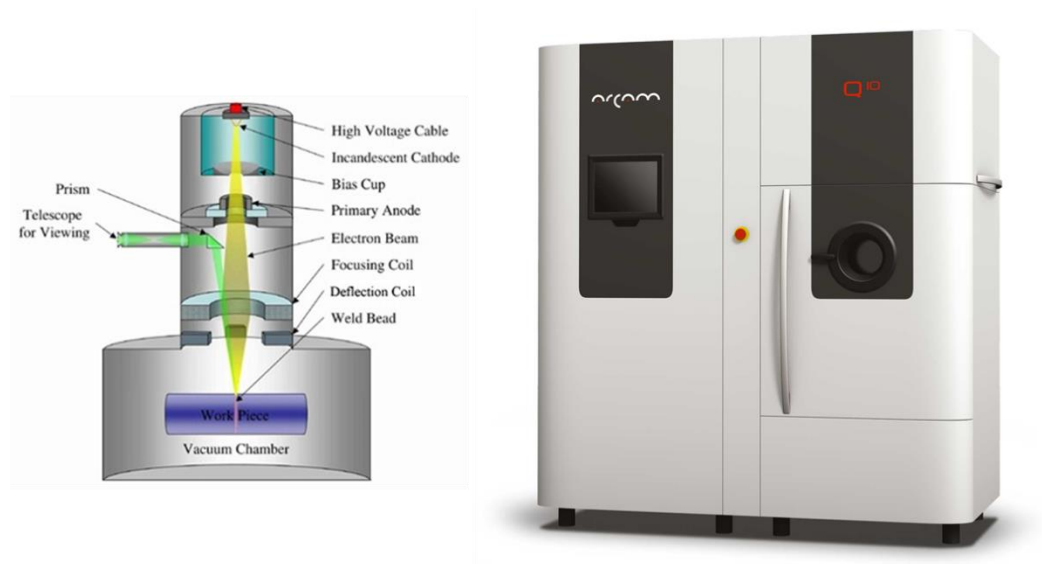


Figure 2-3: Electron Beam Melting (EBM) schematics from www.popular3dprinters.com/ (left) and example of commercial model Q10 by Arcam AB (Arcam AB, 2016).

2.2.2 *Selective Laser Melting*

In Selective Laser Melting (SLM) the energy for layer processing is provided by a high-power laser, typically between 100 to 1000W. In SLM, the chamber vacuum of EBM is replaced by either Ar or N, which in addition to providing oxidation protection by evacuating the oxygen from the chamber provides a heat conduction route for component cooling. During the SLM build process, the build platform is normally heated to 100°C and roughly maintained at this temperature. Therefore, the build platform is considerably cooler than in the EBM build environment, thus making SLM parts cool significantly more rapidly than EBM parts. This in turn, exposes SLM parts to the formation of residual stresses, which make heat-driven stress relief a necessary step in the post-processing stage, in order to improve the materials' functional behaviour. A problem affecting SLM components, however, is the phenomenon of in-process cracking, due to the magnitude of the residual stresses induced by the localised rapid heating/cooling sequences increasing beyond the ultimate strength threshold of the processed material, as shown in Figure 2-5. The laser beam scans at speeds lower than those of the electron beam in EBM, between 800 and 1200 mm/s (Murr et al., 2012). This affects the productivity of the of the SLM process, which is lower than for EBM (Baumers et al., 2016a).

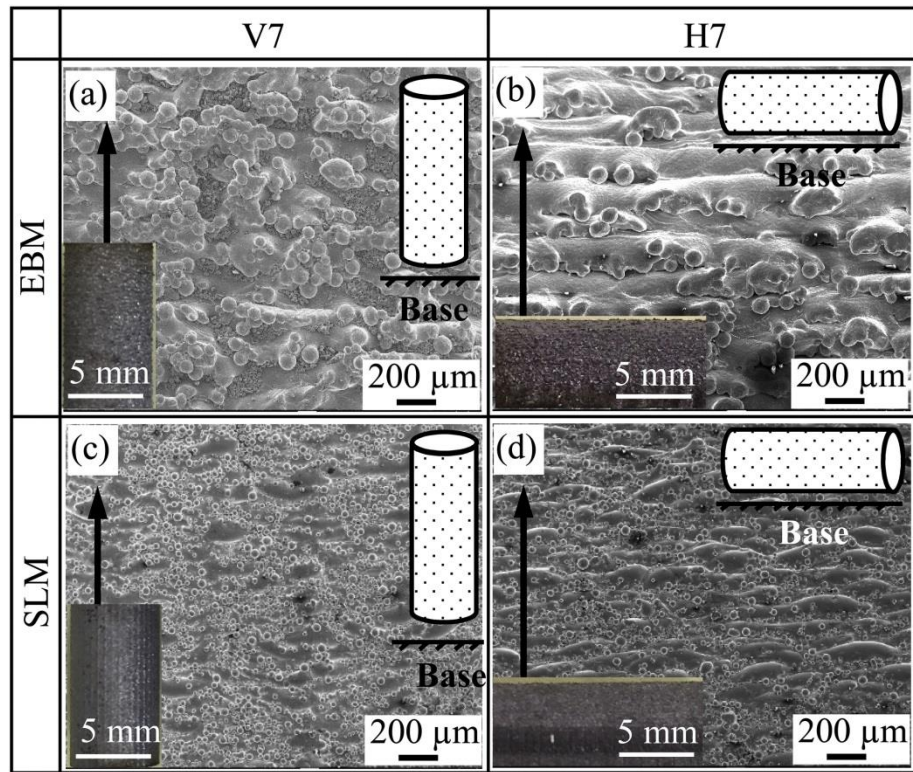


Figure 2-4: SEM micrographs comparing the surface finish of as-built EBM (a and b) and SLM Ti-6Al-4V cylinders (c and d), from (Zhao et al., 2016).

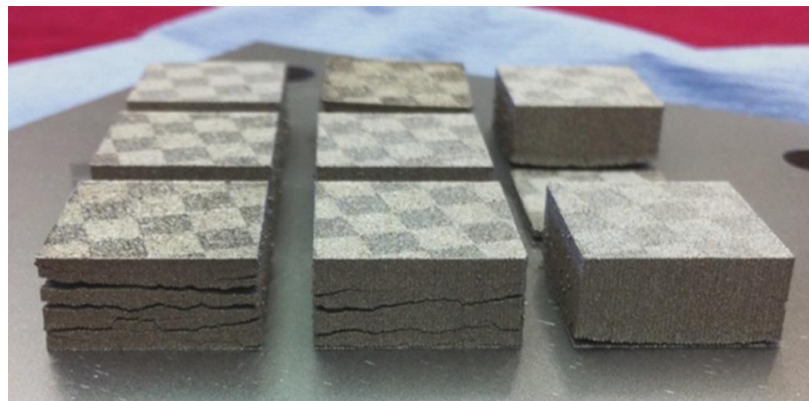


Figure 2-5: Part delamination due to crack formation and propagation in SLM high speed steel, from (Kempen et al., 2014).

2.3 Elements of Electrical Machine Construction and Design

2.3.1 Electrical Machine Construction

For simplicity, this discussion only considers electric motors, since it is well known that the same underlying principles are also valid in the case of

generators. The exploded view of an (induction) electric motor in Figure 2-6 shows the active components and materials of a traditional electrical machine. In the static part, or stator, a rotating magnetic field is generated by means of an array of electromagnets. The magnetic field produced in the stator interacts with the rotating part, namely the rotor, resulting in the production of a mechanical torque. Both the stator and the rotor consist of soft magnetic cores, whose function is to amplify and guide the magnetic field. The stator core presents protuberances, around which copper field coils are wound to produce an array of field-generating electromagnets. The specific electromagnetic nature of the interaction between the stator's magnetic field and the rotor determines the topology of the motor.

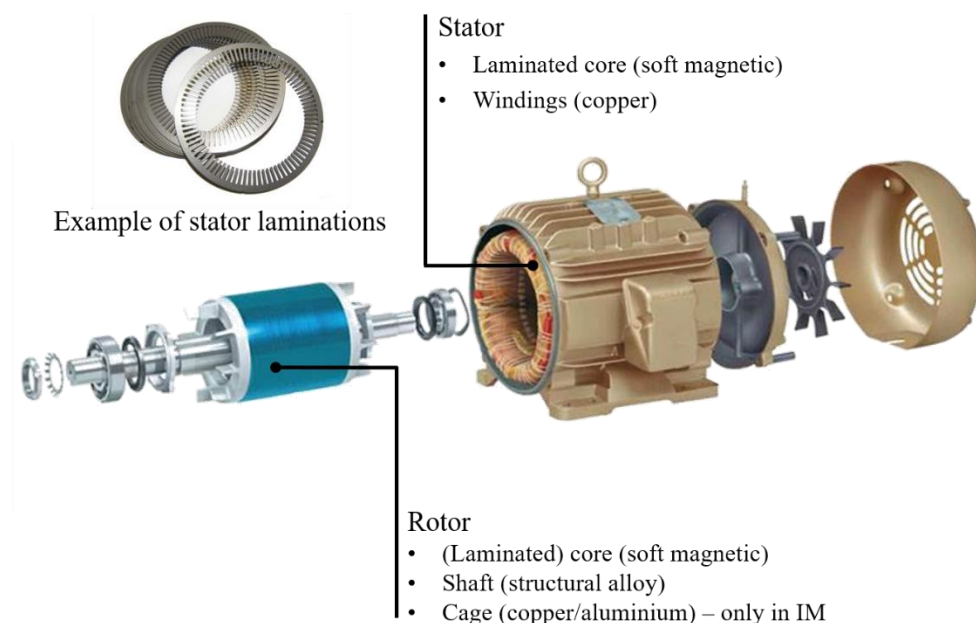


Figure 2-6: Exploded view of standard 3-phase Induction Machine. The parts making out the stator and rotor are also listed. Exploded motor view from www.aproco.nl/.

Since the development of the DC motor (an example DC rotor is shown in Figure 2-7 (a)) in the first half of the 1800s (Doppelbauer, 1822), a broad variety of machine topologies and constructions has been introduced. These

can belong to two broad categories, based on the relative rotation speed of the rotor with respect to that of the stator's magnetic field. When the rotor and the magnetic field rotate at the same speed, the machine is called synchronous, whereas the term asynchronous is used for machines characterised by a rotor and a stator's magnetic field rotating at different speeds. The advantages and disadvantages associated with the different machine types make them suitable in different applicative fields, power ranges and operational/maintenance conditions. In hybrid/all-electric transportation, the main requirements are for high power density, high efficiency, as well as compactness and reliability (Finken et al., 2008). In this respect, there seems to be agreement in the literature that the brushless Permanent Magnet (PM) machine, the Switched/Synchronous Reluctance machine (SR, SyR, respectively) and the Induction Machine (IM) present the highest potential (Finken et al., 2008, Cao et al., 2012, Galea et al., 2014). While the first two machine types are examples of synchronous machines, the IM is the most popular representative of the asynchronous category.

In the IM the interaction relies on the electrical current induced inside a number of electrically conductive aluminium or copper bars/windings placed within the rotor core by the rotating magnetic field. The induced current produces a magnetic field that “chases” the original stator field, thus creating mechanical work. It is important to note that in order for the stator field to induce a current inside the rotor, hence a torque, Faraday-Lenz law requires that the former and the latter rotate at different speeds (i.e., the IM is an asynchronous machine).

In the Permanent Magnet (PM) motor, hard magnetic material (rare-earth materials, aluminium-nickel-cobalt compounds or hard ferrites) is placed within, or on the surface of the rotor core. In this case a torque is generated by the attraction between the opposite poles of the permanent magnets and stator field. An example of a typical surface-mount PM rotor design is shown in Figure 2-7(b).

In the SyR and SR motors (Figure 2-7 (c) and (d), respectively), the rotor is only composed of a soft magnetic core, and therefore represents the simplest and cheapest form of rotor. Therefore, for reluctance machines torque generation solely relies on the tendency of the rotor to adjust its position according to the path of minimum reluctance. It follows that motor performance solely relies on the quality of the soft magnetic materials employed, as well as on the geometrical design of the rotor and stator cores (Chapman, 2005). The SyR and SR differ in that the SR rotor divides its rotation into a number of steps, each consisting in the rotor aligning with a stator pole, thus effectively representing an example of stepper motor.

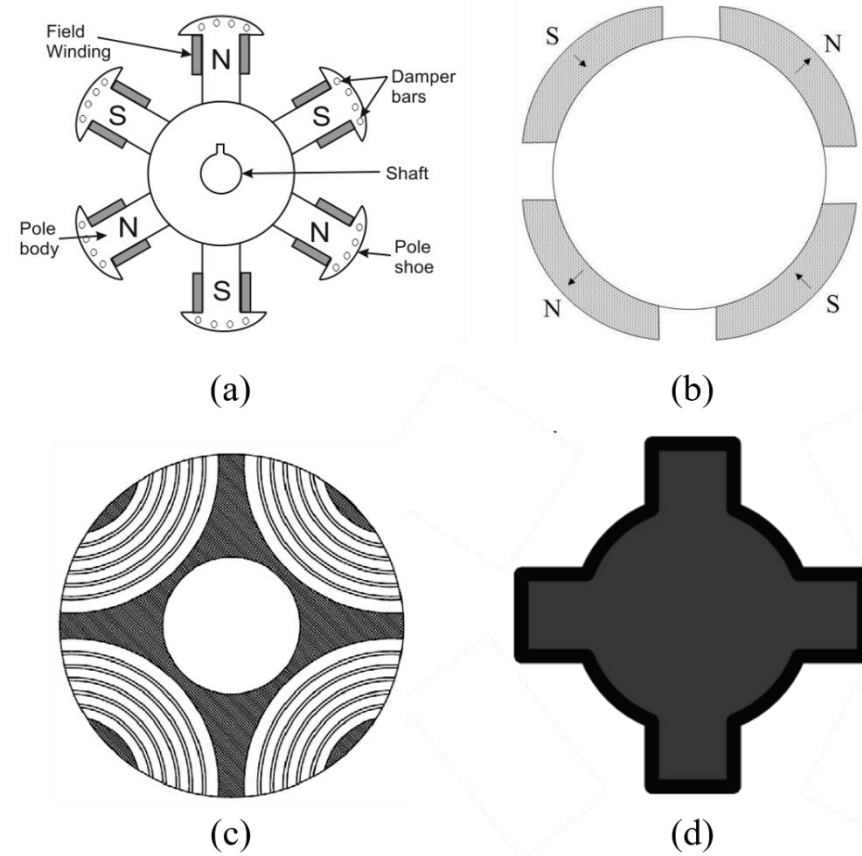


Figure 2-7: The cross-sections of four typical rotor topologies: (a) salient pole DC with armature windings, (b) Surface-Mount Permanent Magnet (PM), (c) Synchronous Reluctance (SyR), (d) Switched Reluctance (SR). Artwork adapted from steemit.com/.

2.3.2 Electrical Machine Design

The geometrical design of the active components in an electrical machine is generally carried out as a size optimisation problem, whereby the choice of a part's dimension is driven by the desired electromagnetic performance (represented by a power or torque target). While analytical methods represent the traditional approach to electrical machine design (Lipo, 1991), more recently mathematical optimisation techniques have started to establish in the literature (Vas, 1999). The use of optimisation mainly regards the application of evolutionary schemes, especially Genetic Algorithms in multi-objective configurations (Bianchi et al., 1998, Hamiti et al., 2011, Cupertino et al.,

2014), for the selection of geometrical dimensions, especially for machine types whose geometry is described by a high number of variables (e.g., the rotor of SyR machines). However, all of the above-mentioned approaches require an initial “template” design, on the basis of which the final component dimensions are defined. In order to overcome the limitations of size optimisation, Topology Optimisation (TO) has started to emerge as a powerful alternative for electrical machine design.

A great deal of research employing TO has been produced in the structural engineering sector since the pioneering work of Bendsoe and Kikuchi in 1988 (Bendsøe and Kikuchi, 1988). A clear advantage of TO over classical size and shape optimization is the possibility of achieving truly optimal designs. Therefore, Since the article of Dyck and colleagues in 1996 (Dyck and Lowther, 1996), TO has started to emerge as a tool for the free-form design of electromagnetic devices. At the end of the 2000’s the first applications of TO for the electromagnetic design of rotating electrical machines began to appear in the literature. Most of the published work dealt with 2D problems (Gangl and Langer, 2012, Lee et al., 2010, Kim and Park, 2010), and among the few examples of 3D TO for electrical machines we report the design of a stator pole piece in a BLDC motor for manufacturing using SMCs (Kim et al., 2009), and the design of a single-phase 6/4 SR rotor (Lee et al., 2010). In both papers the authors aimed to minimise the torque ripple affecting the motor, while satisfying motor weight constraints, but adopted different TO methodologies to tackle the problem. The TO methodologies for the electromagnetic design of electrical machines and other electromechanical devices employed in the literature are described in Section 5.8.

2.4 Soft Magnetic Cores

Soft magnets are ferromagnetic materials that almost completely lose their magnetization when the applied magnetic field is reduced to zero. In other words, they are characterized by a narrow induction-applied field loop (or B-H loop). An example of soft magnetic material magnetization curve is shown in Figure 2-8, where the main properties are also indicated. A comparison among some popular types of soft and permanent magnetic materials is summarized in Table 2-2, which reports the fundamental electromagnetic and mechanical properties, based on the data from various sources.

Good core soft magnets for motor applications are characterized by the following properties:

- high magnetic permeability μ : this allows the core to achieve high inductions (B) at low applied fields (H), hence increasing the efficiency of the machine;
- high saturation magnetization: this allows the material to achieve high B. A higher B produces higher torque, given the same machine design and dimensions;
- low magnetic coercivity: this is necessary in order to reduce the hysteresis-related losses, since the latter equal the area contoured by the B-H loop;
- high electrical resistivity: this is required in order to limit the propagation of the eddy currents induced by a time-varying B inside a core, and therefore reduce joule-heating-related power losses.

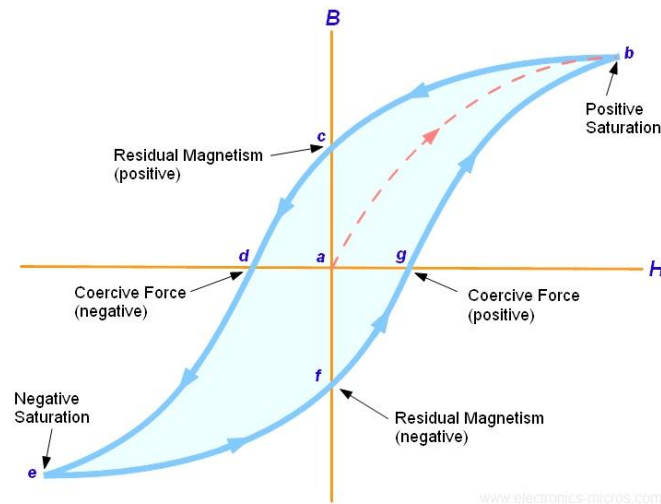


Figure 2-8: typical B-H loop (or hysteresis loop) for a soft magnetic material. The red dashed line represents the initial magnetisation curve, i.e., the magnetisation path experienced by the material when starting from a demagnetised state. The steeper the initial magnetisation curve of a material, the higher its relative permeability μ since $\mu=B/H$. Adapted from <http://nptel.ac.in/>

As an alternative to the use of high-resistivity materials, which often lack of good ferromagnetic and mechanical properties, as shown later on in this chapter for the case of soft ferrites, the propagation of eddy currents is generally limited by constructing the cores as stacks of varnish-insulated laminations, typically 0.1-to-1mm thick, as schematically shown in Figure 2-9. In order to avoid the translational symmetry that the lamination imposes on the designs, other solutions have been envisaged to limit eddy-current propagation, especially in the form of polymer-coated iron particles, also known as soft magnetic composites (SMC), which have been emerging thanks to the advances in powder metallurgy technologies (Shokrollahi and Janghorban, 2007). In synchronous machines the rotor flux variation is minimal, thus making their fabrication as solid structures using high-performance metallic alloys viable. This is particularly the case in high-speed applications, where mechanical integrity is harder to guarantee using laminated structures (Ikaheimo et al., 2014).

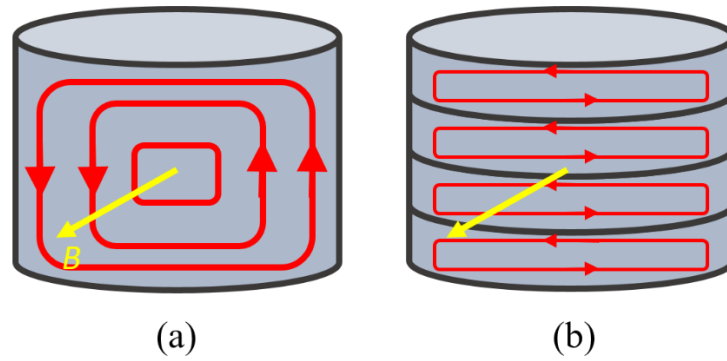


Figure 2-9: Propagation of magnetic field-induced currents in electrically conductive materials in solid form (a) and in laminated form (b).

2.4.1 Iron-silicon alloys (*Fe-Si, or silicon steels*)

Alloying Fe with Si in the concentration range 2-6.5%wt. is the most popular solution for core construction, owing to the desirable electromagnetic properties of low hysteresis losses, high resistivity, high permeability and thermal stability, associated with low costs. Owing to their popularity for machine construction, laminated Fe-Si alloys are often referred to as electrical steels (Beckley, 2002). Importantly, within the family of silicon steels, Fe–6.5%wt. Si has a high potential in magnetic device applications due to its very low hysteresis and eddy-current losses (Bozorth, 1993b). However, its employment has been limited due to its poor workability and to this day the only commercially available Fe–6.5%wt. thin sheet material is represented by JNEX, produced by JFE Steels (Namikawa et al., 2005). The main draw-back of JNEX is cost and pollution, which can be significant (He et al., 2008). High-silicon steels (Si content $\geq 6.5\%$ wt.), often with the addition of aluminium (Sendust) are produced in powder form and employed as compacted cores for applications in the frequency range 50-100 kHz (Hilzinger and Rodewald, 2013). The metallurgical and magnetic properties of Fe-Si, as well as the most common processing routes, are described in more detail in Chapter 3.

2.4.2 Iron-Cobalt alloys (*Fe-Co*)

Especially in the equiatomic composition form, i.e., Fe-50wt.Co, these alloys are characterised by the highest saturation magnetisation and temperature of ferromagnetic-to-paramagnetic transition, thus making them particularly valuable in applications where thermal stability and compactness are paramount, and the high costs associated with Co are secondary (e.g., the aerospace industry) (Hilzinger and Rodewald, 2013).

2.4.3 Nickel-Iron alloys (*Fe-Ni*)

Fe-Ni alloys are particularly valuable in their face-centered cubic state because of their very high permeability. While their employment is normally limited to a few regions of Ni content (especially 80wt. and 50wt.), Fe-80wt.Ni presents an extremely high permeability, hence the name permalloy. The low saturation magnetisation and electrical resistivities are the main draw-back of this class of alloys (Hilzinger and Rodewald, 2013).

2.4.4 Soft ferrites

Iron oxide Fe_3O_4 is a ferromagnetic ceramic and presents an electrical resistivity that is several orders of magnitude higher compared to metallic alloys. Therefore, soft ferrites are particularly suitable for very high frequency applications (up to 100 MHz for the NiZn grades). While soft ferrites are generally characterised by low saturation magnetisations, they can be doped with other oxides, especially ZnO and MnO to obtain relatively high permeability, which is comparable with that of Fe-Si (Hilzinger and Rodewald, 2013).

2.4.5 Soft magnetic composites (SMCs)

Iron particles can be coated with a layer of polymeric insulator to be compacted using powder metallurgy technology. The main advantages of SMCs are the very low eddy current losses and the possibility of extending the design of electrical machines to three-dimensions, which has prompted the emergence of novel, three-dimensional machine core topologies, such as the claw-pole rotor (Figure 2-10) and research on freeform design strategies, such as topology optimization (Kim et al., 2009). However, the limited magnetization properties of SMCs (characterized by low permeability and low saturation magnetization) have limited the use of this new class of materials to high-frequency applications.



Figure 2-10: example of claw-pole rotor produced using SMCs. Adapted from www.shw.sh.cn

2.4.6 Amorphous Alloys (or Metallic Glasses)

Amorphous alloys for soft magnetic applications are normally obtained using compositions of transition metals (Fe, Co and Ni) and metalloids (normally B and Si). Owing to the lack of long range atomic order and, therefore, of grain boundaries and any other lattice-related defects, amorphous alloys display excellent soft magnetic properties, which include high permeability and near-

zero hysteresis losses (especially for Co-based compositions), and higher electrical resistivity than typical soft magnetic crystalline alloys. While the saturation magnetisation depends on the metalloid content, this is in general relatively high for Fe-based amorphous alloys, such as Fe-Si-B. While the production of amorphous alloys is limited to ribbons, owing to the intrinsic need to achieve extremely high solidification rates, methods for the preparation of continuous lengths have been proposed, and this has prompted ongoing commercial interest (Hilzinger and Rodewald, 2013). Production costs also represent another obstacle hindering the adoption of amorphous alloys, although efforts towards cheaper alloys are being undertaken, for example through the replacement of the most expensive elements, especially B, with more viable substituents (Kurniawan et al., 2015).

2.4.7 Potential for Machine Core Production Using AM Technologies

Design and Processing considerations: Since the performance of soft magnetic cores is highly dependent on their geometrical design, and as they represent a high portion of the total machine weight, the use of AM for the production of innovative core designs that could significantly improve the power-to-weight ratio characteristics of the motor by optimising the distribution of the active material is arguably seen as a potential game-changer. The potential advantages of three-dimensional core design have been discussed in the literature, and a number of unconventional solutions in this regard have been recently proposed (Ikaheimo et al., 2014, Ertugrul et al., 2015). However, it is common that electrical machine cores created using classical manufacturing processes (e.g., laminate punching or alloy powder pressing) are

designed to facilitate the productivity of the fabrication process; the material used for the parts, the desired shape and features of the part and even the material properties of the processing machine must all be taken into account. Therefore, the typical design approach is to start from template geometries (e.g., a C-shaped flux barrier in a SyR machine) and modify the associated size, depending on the application at hand and the required performance. The resulting component is thus far from optimal. Given the capability of AM of creating complex structures without the limits of conventional manufacturing techniques, a different approach to design should be adopted, which is not limited to the optimisation of predefined sizes. In this respect, Topology Optimisation (TO) stands out as an appropriate tool for the creation of 3D designs that are as close as possible to unconstrained optimality and thus fully exploit the capabilities of AM. TO is a branch of structural design that optimises the distribution of material within a structure (in the form of a FEA model) without the need to start from a template design (Bendsoe and Sigmund, 2013). Deterministic and stochastic optimisation algorithms have been developed in the literature for the topological design of soft magnetic cores in both two- and three-dimensions (Kim et al., 2009, Labbe and Dehez, 2010, Lee and Kikuchi, 2010, Okamoto et al., 2006), including the geometrical optimisation of a Switched Reluctance rotor core (Lee et al., 2010). However, all of the mentioned attempts of free-form do not go beyond the proof-of-concept stage, and thus exclude any real world-related considerations, including manufacturing-related issues. The issue of TO implementation for the design of real life objects to be fabricated using AM has only been discussed for structural engineering problems in (Brackett et al., 2011), but the

guidelines provided are potentially applicable to electromagnetic devices as well. A thorough review of the most popular TO methods is proposed in Chapter 5.

Layer-by-layer manufacturing is not suitable for thin sheet fabrication, due to the well-known thermal gradient-induced warping and high-surface roughness affecting additively manufactured structures. Therefore, materials with high electrical resistivity could be employed for the additive manufacture of stators and asynchronous rotors, which are exposed to AC flux during operation. In this respect, two possibilities can be considered:

- Processing ferrite powder in association with a high radiation power powder bed system (SLM or EBM). Successful production of ceramic material using SLM has only been achieved by researchers at the Fraunhofer Institute (Wilkes et al., 2013). While highly-dense, crack-free ceramic specimens have been produced in the literature using an ad-hoc, in-house SLM system that allows pre-heating the build platform at very high temperatures ($>1500^{\circ}\text{C}$) (Wilkes et al., 2013), the technology is not available on the market. Furthermore, the poor surface finish generally characterising the manufactured components, as shown in the example of Figure 2-11, requires extensive post-processing, which would be difficult for parts characterised by re-entrant geometries, such as a rotating machine's stator;
- Processing insulated iron particles (such as SMCs) could be performed using powder bed systems, but these should be equipped with a low-power heat-source, in order to produce particle bonding without prompting the evaporation of the coating layer.

The above-mentioned solutions, besides the processing problematics outlined, would hardly compete with existing laminated counterparts, owing to the previously described poor material properties of the materials involved, thus reducing the potential interest for AM adoption.



Figure 2-11: As-built Zirconia-Alumina component made by SLM (Wilkes et al., 2013).

As we previously mentioned in this chapter, in synchronous machines the flux variation is minimal in the rotor, thus making the fabrication of the latter as a solid structure using high-performance metallic alloys viable. Therefore, the rotors of SyR and PM machines appear to be particularly good candidates for production using AM technologies. A traditional SyR rotor is characterised by a very simple structure, composed entirely of soft magnetic material. Neither conducting material nor permanent magnets are involved, thus making it suitable for processing using single-material processes, especially powder bed laser and electron beam systems (SLM and EBM). Since the performance of the SyR machine mainly relies on the geometry of the flux barriers and on the magnetic properties of the soft magnetic material employed, designing its rotor for AM using free-form methodologies, such as TO, could clearly lead to the improvement of the machine's electromechanical performance. Similar

considerations stand for the PM rotor, although in the case of buried-magnet design (refer to Figure 2-6 (d)) the freeform design strategy should include constraints to aid permanent magnet insertion. Finally, in both the SyR and PM cases, well-known solutions for the improvement of the torque profile, such as rotor axial skewing (Bomela and Kamper, 2002) would be straight-forward using AM technologies for fabrication.

Materials Considerations: Regarding the choice of the most suitable soft magnetic alloys for production using AM technology, it is first important to note that despite the synchronous nature of the SyR and PM rotors, the flux they “see” during operation is not perfectly constant, owing to the shape of the stator, characterised by protruding poles. Therefore, the selection of soft magnetic materials characterised by relatively high electrical resistivity should be preferred, in order to limit the power losses. In this respect highly resistive alloys, such as high-silicon steels (e.g., Fe-6.5%wt. Si) and amorphous alloys represent interesting candidates.

A second consideration on material choice is that, as the main advantage that AM can bring to machine design is the removal of inefficient material, materials characterised by a relatively high saturation magnetisation should be preferred. By using rapidly-saturating soft magnets, such as Fe-Ni alloys, larger cores are necessary to compensate for the low induction and generate the required magnetic flux, thus leaving little room for improvement through geometry optimisation. In this respect, Fe-Si and Co-Fe alloys should be preferred over Fe-Ni alloys, SMCs or soft ferrites, as the magnetisation at saturation data reported in Table 2-2 suggest.

A final consideration regards the potential processability of the alloys using AM. In this respect, Fe-Ni alloys have already been processed using SLM technology in the literature, starting from elemental Fe and Ni mixed powders (Zhang et al., 2012, Zhang et al., 2013). Cobalt-based alloys with structural functions, especially Co-Cr, are being extensively processed using powder bed AM technologies in the dental implant sector, thus hinting at the suitability of Co-based soft magnetic cores for SLM/EBM processing. Fe-Si alloys represent excellent candidates for processing using SLM also from a processability perspective, because of the well-known suitability of iron-based powders for processing using powder bed fusion technologies (Rombouts et al., 2006). In addition, AM might viably process 6.5%Si steels, which are known for their good magnetic properties, such as high permeability, low hysteresis loss, high electrical resistivity, and low magnetostriction. The commercial application of these alloys has been limited due to their poor ductility, which makes their processing difficult; however, the tool-free, layer manufacturing approach of AM could provide a solution to the processing issue.

Table 2-2: Some magnetic and mechanical properties of some popular soft and hard magnetic materials (Hilzinger and Rodewald, 2013, Kollar et al., 2010, Gavrilă and Ionita, 2002): relative permeability (μ_i), coercivity (H_c), remanence (B_r), saturation polarisation (J_s), electrical resistivity (ρ) and yield strength (σ_Y).

		μ_i	H_c (A/m)	B_r (T)	J_s (T)	ρ ($\mu\Omega \cdot \text{cm}$)	σ_Y (MPa)
Soft	Fe-3%wt.Si	1000	30-80	-	2.03	47	37
	Fe-6.5%wt.Si	1500	15-18	-	1.81	85	80
	Fe50%wt.Ni	$10 \cdot 10^3$	4	-	1.6	48	152
	Fe50%wt.Co	$15 \cdot 10^3$	40	-	2.35	65	500
	SMCs (Somaloy 700)	350	70	-	2.02	500	65
	Mn-Zn Ferrites	$15 \cdot 10^3$	5-20	-	0.35- 0.55	1500	48
	Amorphous Fe-Si-B	$10 \cdot 10^3$	2-10	-	1.5-1.6	130	1666
Hard	NdFeB (sintered)	-	$(1-3) \cdot 10^6$	1.1-1.5	1.61	-	8
	BaSr Ferrites	-	$(2-3) \cdot 10^5$	0.4	0.48	-	-

2.5 Permanent Magnets

The introduction of strong rare-earth-based permanent magnets has played a pivotal role in compacting machine design, and as of today Synchronous PM machines represent the preferred choice when high torque density and compactness are desired (Finken et al., 2008). As shown in the schematics of Figure 2-12, a permanent magnet keeps its polarisation after the magnetising field is removed. In order for this to be true, a hard magnet should exhibit a high remanent polarisation and a strong coercivity. In order to achieve these properties, an almost perfect crystallographic texture is required, alongside a high saturation polarisation of the alloy and high magneto-crystalline anisotropy. A comparison among some popular types of soft and permanent

magnetic materials is summarized in Table 2-2, which reports the fundamental electromagnetic and mechanical properties, based on the data from various sources. The main classes of hard magnets are Al-Ni-Co, Fe-Cr-Co, ferrites and rare earth-based magnets, which today represent the solution characterised by the highest energy product and high resistance to demagnetisation. Among the most popular rare-earth based alloys are Nd-Fe-B and Sm-Co. To this day rare-earth permanent magnets represent the preferred solution unless product cost is a major issue over magnetic performance (Hilzinger and Rodewald, 2013).

Typical processing routes for permanent magnets are powder compaction (sintering) and injection (or compression) bonding. Bonded magnets are composed of granular magnet material and non-magnetic polymeric material and are therefore less suitable for applications where a high energy product is important. For this reason, the permanent magnets for use in electrical machines are produced by sintering the granular magnet material. This technique, in turn, presents major challenges due to the highly abrasive nature of the permanent magnet materials. The strong crystallographic texture, which is important to guarantee high remanence, is obtained by a number of metallurgical processes, for example by anisotropic precipitation after severe cold forming, or by anisotropic grain growth during hot deformation (Hilzinger and Rodewald, 2013).

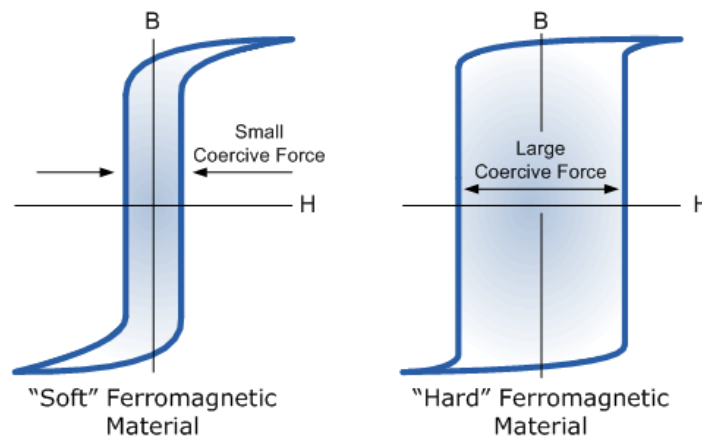


Figure 2-12: Hysteresis loop comparison between soft and hard-magnetic materials.

2.5.1 Potential for Permanent Magnet Production Using AM Technologies

Powder bed fusion processes such as SLM could provide a means to produce strong crystallographic texture in hard magnetic alloys, as this has been shown to be the case for steel (Niendorf et al., 2013). However, apart from Fe-Cr-Co, which exhibits ductile behaviour, most of the relevant families of hard magnets are brittle, thus making them prone to cracking when processing using a localised heat source, such as a laser or an electron beam. Furthermore, one of the most interesting hard magnetic alloys, Nd-Fe-B powder is very fine and when dry can ignite spontaneously, thus making its employment in atomised form particularly dangerous. Therefore, it seems that, despite the potential interest of hard magnet processing using AM technology, the technical challenges appear strong given current AM machine constructions.

2.6 Research Scope, Aim and Novelty

2.6.1 Research Scope

On the basis of the feasibility study carried out in this chapter, the scope of the present research is outlined. The soft magnetic core is preferred over permanent magnets as the best candidate for development using powder bed fusion AM technology, owing to the former's lower material costs and availability, especially when compared with rare earth-based permanent magnets. Another factor is the unsuitability of the AM technology for processing the permanent magnets, since both rare earth-based and ferrite-based permanent magnets are characterised by very low tensile strength (see Table 2-2), which makes them unsuitable for processing using powder bed fusion, especially if laser-based, owing to the high thermal stresses induced in the part during processing. High-silicon steel, i.e., Fe-Si alloy with Si content around 6.5%, is selected as the raw material of choice, owing to its high electrical resistivity, which helps counteracting the eddy currents, good magnetic properties and good raw material availability. Moreover, the poor workability of the alloy makes it suitable for showing the advantages of tool-free manufacturing over traditional manufacturing methods. As regards the AM technology used for processing the raw material, SLM is preferred over EBM, for both technical and economic reasons. SLM has the ability to produce geometrical details with higher resolution and lower surface roughness than EBM, owing to the lower beam spot size and layer thickness. This advantage of SLM over EBM is particularly important because it reduces the need for part post-processing, which can be particularly time-consuming and costly if the

access to the surfaces is not straight-forward due to the geometrical intricacy of the part.

Finally, we recognise that in order to demonstrate the full potential of AM in electrical machine construction, material development and part design need to be carried out in parallel. Therefore, the present work also covers aspects of rotor core geometry optimisation. Part design for AM is best tackled using TO, which arguably represents the most powerful design tool to effectively exploit the design freedom offered by AM. As a case study, the rotor core of a synchronous machine is selected as the most suitable application because of the possibility to produce it as a solid component, rather than as a lamination stack, owing to the near-static nature of the magnetic flux inside it. We argue that the high-resistivity of the high-silicon steel selected for this research also helps counteracting the eddy currents propagating across the solid component.

2.6.2 Research Aim

The present research work aims to improve the torque/power density of a synchronous electrical machine (brushless PM or SyR) by optimising its rotor's core three-dimensional geometry, and use SLM technology to produce the optimised geometry using functional soft magnetic high-silicon steel, which is also developed as part of this research work. Achieving this goal will in turn allow demonstrating the applicability and benefits of AM powder-bed technology in the development of electrical machinery.

2.6.3 Research Novelty

To the author's knowledge, the novelty of the proposed research is twofold. First, the production of soft magnetic material for electromechanical

applications by means of AM technology has never been attempted in the literature. Second, the use of TO for the design of real-world rotor cores has also never been addressed. While TO approaches have been employed for the magnetostatic design of magnetic cores, multi-physical TO for the design of real-world components is still unexplored to the authors' knowledge. The above mentioned innovations are achieved through the fulfilment of intermediate research objectives, which measure the progress towards meeting the aim.

2.6.4 Research Objectives

The objectives of this research are outlined in the points below and summarised in the flowchart of Figure 2-13, which also helps to visualise their interconnection:

1. Investigate the possibility of processing high-silicon steel for the fabrication of electrical motor cores via SLM:

- 1.1 Select the processing and post-processing parameters that achieve the best materials properties with respect to the target performance;
- 1.2 Provide a comparison between the SLM-processed material and the same material processed through other methods (e.g., chemical vapour deposition), with respect to the electromagnetic properties.

2. Investigate the possibility of improving the performance of a synchronous rotor core through the development of 3D designs that are not constrained by manufacturing:

- 2.1 Implement a TO-based strategy aimed at improving the geometry of an electromechanical component for optimal torque-to-weight ratio, while maintaining structural integrity;
 - 2.2 Test the proposed methodology in two and three-dimensions on the selected case study. The latter should be an existing synchronous rotor core design;
 - 2.3 The magnetic properties of the 3D-printed material are integrated into the design process, and assess how both the design strategy and material properties affect the performance of the device as compared to the initial design.
3. **Manufacturing of the final rotor design:** Once the above list of objectives is fulfilled, the ultimate task is the fabrication by SLM of the designed rotor. The prototype shall be employed in future research for validating experimentally the electromechanical performance of the proposed solution and provide a direct comparison with the classically manufactured counterparts.

AIM

Improve Motor Torque Density through Additive Manufacturing of Rotor Core

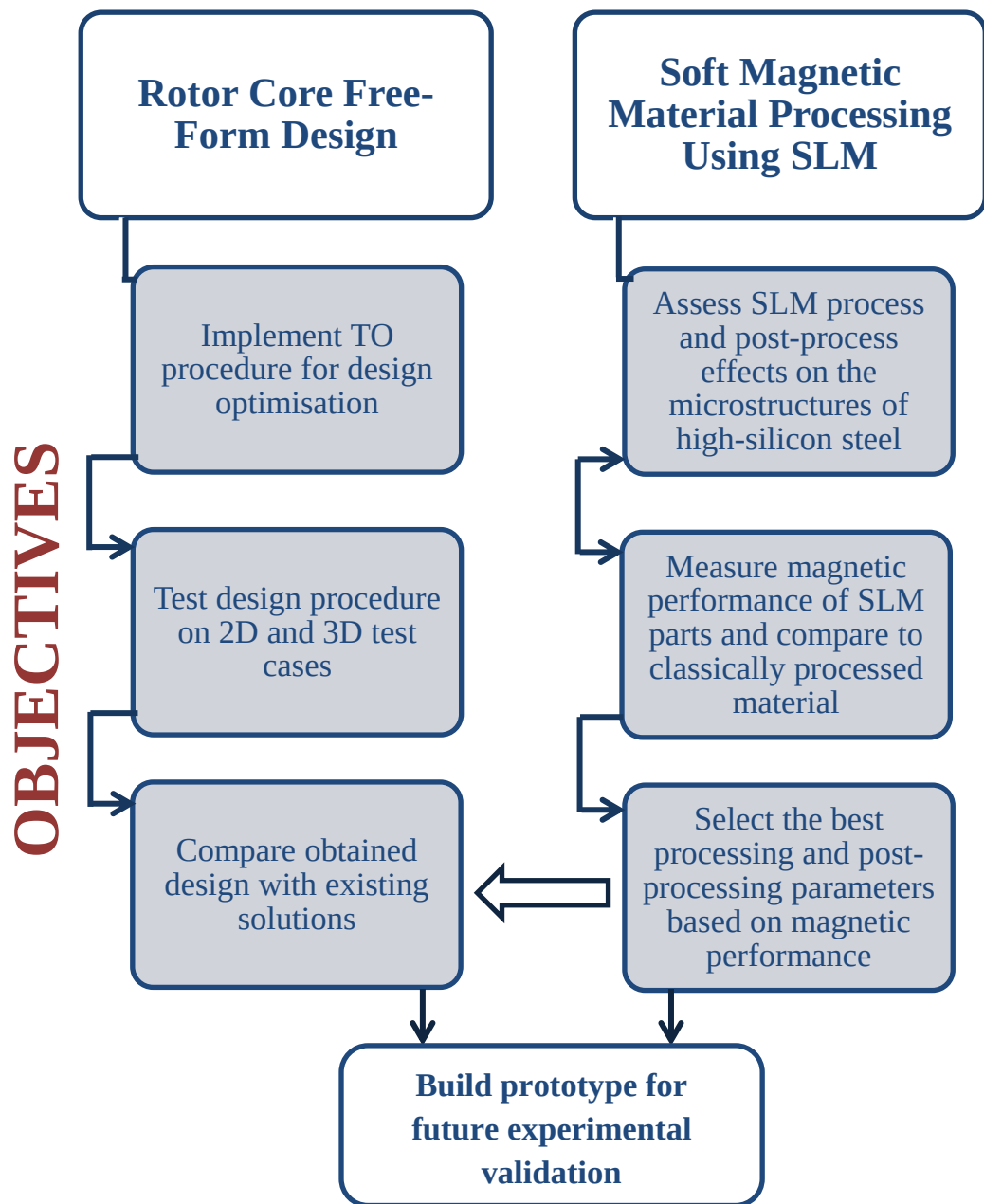


Figure 2-13: Flowchart resuming the overall aim of the present research and the intermediate objectives pursued to achieve it. Arrows specify the input/output relationship between objectives.

3. Fe-Si Alloys: Metallurgy, Magnetic Properties and Manufacturing Processes

3.1 Introduction

The Fe-Si binary system is arguably most popular materials in the electromechanical industry. Therefore, a rich literature exists that covers all of the physical, technological and industrial aspects of the alloy. Some of the fundamental notions on the relationship between metallurgy, production processes and functional properties of the Fe-Si system are summarised in this chapter. This will serve as the basis to understand and interpret the metallurgical and electromagnetic properties observed for the Fe-Si alloy processed via SLM as part of the present research work.

3.2 Metallurgical Effects on the Magnetic Properties of Fe-Si Magnets

At Si content $>2\%$, the Fe-Si system (phase diagram shown in Figure 3-1) is body centered cubic (bcc) and below the Curie temperature it presents ferromagnetic properties. However, like all soft magnets, Fe-Si is characterised by no net magnetisation when no external field is applied. This is because the elementary magnets are not aligned parallel to each other throughout the material, but rather subdivided into a number of so-called “magnetic domains”, i.e., small areas, generally ranging from 0.1 to 1mm in size, within which the individual magnetic moments are aligned by exchange coupling energy. Each magnetic domain represents a magnetic dipole, whose spatial alignment with respect to the adjacent magnetic domains produces a net zero magnetisation at a macroscopic level. The subdivision into magnetic domains allows the

material to minimise its magnetic stray energy content, which normally represents the highest energy term (Hilzinger and Rodewald, 2013). The boundaries between the magnetic domains are called domain walls, transition layers within which the magnetisation direction gradually rotates from the direction of one domain to that of the adjacent domain.

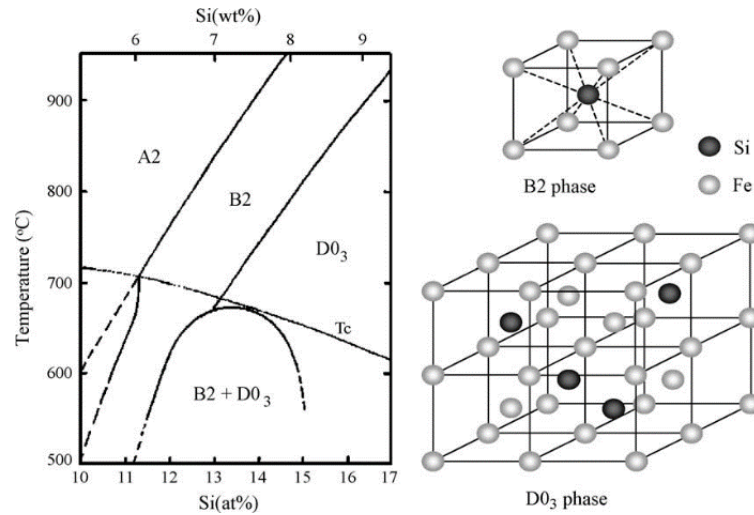


Figure 3-1: Binary Fe-Si phase diagram (left) and atomic structures of the ordered phases B2 and D0₃ (right), from (Swann et al., 1975)

In crystalline Fe-Si, different crystallographic directions are normally associated with different magnetic properties. Figure 3-2(a) shows the magnetisation curves of a single crystal Fe-Si specimen, when the applied field is oriented along the $\langle 100 \rangle$, $\langle 110 \rangle$ and $\langle 111 \rangle$ crystallographic axis families. It can be seen that the former crystallographic direction presents the best magnetisation properties, thus often being referred to as the “easy axis” of magnetisation. The $\langle 101 \rangle$ is the medium axis, while the $\langle 111 \rangle$ is the hard axis of magnetisation. Since the magnetic moments within each magnetic domain align according to the easy axis of magnetisation and owing to the symmetry of the bcc systems, adjacent domain walls are oriented either perpendicularly or at 180° to each other in Fe-Si alloys.

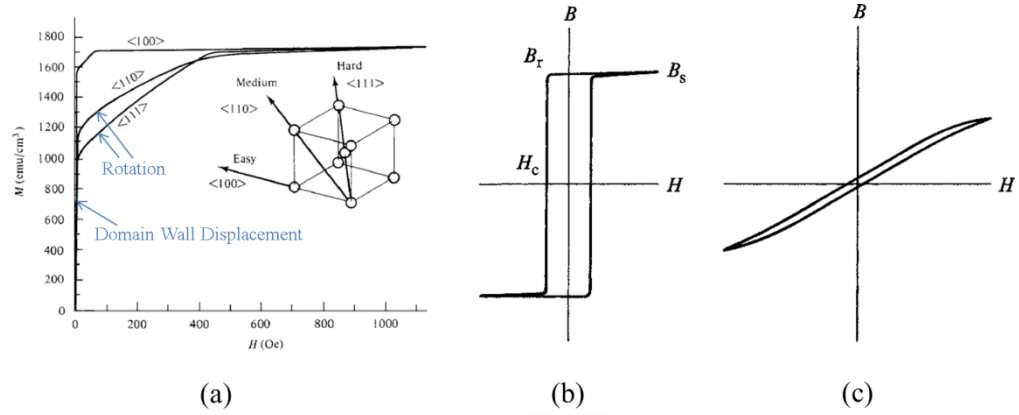


Figure 3-2: Dependence of magnetic properties of Fe-Si along different crystallographic directions and depending on the magnetisation mechanism involved. The magnetisation curves of a Fe-Si single crystal along different crystallographic axes is in (a). The hysteresis loops of ideal samples magnetised via domain wall displacement and magnetic moment rotation are in (b) and (c), respectively.

When an external field H is applied to a magnetic material the magnetisation process is a combination of two elementary mechanisms: the mechanism of magnetisation via rotation of magnetic moments, and the mechanism of domain wall displacement (Hilzinger and Rodewald, 2013). When the applied field is parallel to the easy axes, the domains with favourable orientation (i.e., those whose easy axis is at 0° to the field) will tend to grow via domain wall displacement, at the expense of domains oriented unfavourably, i.e., 90° or 180° apart. Material saturation is reached when all domains have been swept by the domain walls. In real polycrystalline materials, the ideal motion of domain walls described above is hindered by defects in the microstructure, in particular:

- grain boundaries;
- stress centres;
- point and line defects;
- non-magnetic inclusions: precipitates, porosities, etc.

The above inhomogeneities exert a pinning force on the domain walls, since they allow the domain walls to reduce the energy associated with their volume (Hilzinger and Rodewald, 2013). The strength of the pinning force has been shown to be highest when the size of the inhomogeneity is similar to the size of the domain wall. On applying a field H , the domain walls start moving reversibly until the magnetomotive force is strong enough to overcome the weakest pinning force exerted by some of the defects, thus causing an irreversible domain wall “jump” (known as Barkausen jump). As the field increases further, it will be able to break the walls away from stronger and stronger pinning sites until all of the obstacles have been overcome. Therefore, the mechanism of domain wall motion in real materials is irreversible, since energy is dissipated at each Barkausen jump, thus producing a hysteresis loop upon one complete cycle of magnetisation/demagnetisation. In particular, the coercivity can be expressed as the sum of three components (Hilzinger and Rodewald, 2013):

$$H_c = H_{co} + H_{cg} + H_{ci} \quad (3.1)$$

where H_{co} is the contribution of residual stresses, H_{cg} the contribution of grain boundaries and H_{ci} the contribution of non-magnetic inclusions, such as precipitates and porosities. Importantly, in standard polycrystalline materials, i.e., materials characterised by grain sizes $>1\mu\text{m}$, the grain boundary contribution to coercivity H_{cg} is inversely proportional to the grain size D , according to (Hilzinger and Rodewald, 2013):

$$H_{cg} = \text{constant}/D \quad (3.2)$$

Viceversa, the initial permeability μ_i obtained when magnetisation occurs via domain wall motion is proportional to the grain size, according to (Hilzinger and Rodewald, 2013):

$$\mu_i = \text{constant} \cdot D \quad (3.3)$$

As we have mentioned, pores and cracks also represent pinning sites impeding the movement of the magnetic domain walls, thus decreasing the permeability and increasing the coercivity and remanence, as well as the total power losses. If large enough, they represent airgaps from which demagnetisation fields emanate, thus resulting in sheared B-H curves. With an increased degree of shearing the measured permeability and remanence are diminished and do not correspond to the material's intrinsic ones. On the other hand, shearing does not affect the coercivity (Hilzinger and Rodewald, 2013).

When magnetic field direction and easy axis of magnetisation are misaligned, ideally rotation of the magnetic moments is the mechanism guiding the magnetisation process. The extent of rotation of the magnetic domains out of the easy axis given a fixed H is inversely proportional to the specific anisotropy constant of the material. Therefore, magnetisation via rotation is characterised by a constant permeability (i.e., by a linear magnetisation curve), which for Fe-Si alloys is in general significantly lower than the maximum permeability achieved via domain wall displacement. Ideally, samples magnetising solely via domain wall displacement are described by a square hysteresis loop similar to the one in Figure 3-2(b), characterised by high remanence. Figure 3-2(c) shows the ideal loop created when the magnetisation occurs solely via rotation. The loop associated with rotation is S-shaped (i.e., characterised by a low remanence) and narrower (i.e., low coercivity) than the

B-H loop in Figure 3-2(b), owing to the reversibility of the rotation process. In a real material both rotation and domain wall motion phenomena always coexist. In particular, if the field is applied along the medium or hard axes of a crystal, domain wall motion continues until the only domains left are characterised by equal potential energy (Cullity and Graham, 2011). As the applied field is increased further, enough energy is provided to counteract the anisotropic energy and thus rotate the magnetic moments of the remaining misaligned domains out of the easy axis direction. Therefore, in Figure 3-2(a) the change in steepness at high applied fields in the magnetisation curves along $\langle 110 \rangle$ and $\langle 111 \rangle$ is attributed to a change from domain wall motion to magnetic moment rotation.

The above discussion suggests that the crystallographic texture of a material affects the preferred mode of magnetisation/demagnetisation and eventually the shape of the quasi-static B-H loop (Hilzinger and Rodewald, 2013). In particular, in a polycrystalline sample whose grains are characterised by a preferential crystallographic orientation (strong crystallographic texture), if the external field is applied in the direction of the grains' easy axis of magnetisation, fast magnetisation through domain wall motion will result in a B-H loop characterised by high maximum permeability, high remanence and, importantly, low hysteresis loss. On the other hand, if the applied field and the direction of preferential grain orientation are misaligned, the material has low remanence, and is characterised by a low, constant permeability (Hilzinger and Rodewald, 2013) and, importantly, by a high hysteresis loss (Bozorth, 1993a). Since it has been found that the difference in loss between the direction of preferential orientation and the transverse direction is mainly generated in the

high induction region, it has been argued that some magnetisation mechanisms other than rotation have to be involved in that region, such as domain nucleation and annihilation, since rotation is a lossless process (Landgraf et al., 1999). Figure 3-3 shows how the total loss in a strong-textured Fe-Si lamination varies with the angle between the direction of the applied field and the direction of preferred <001> orientation (Mori et al., 2011).

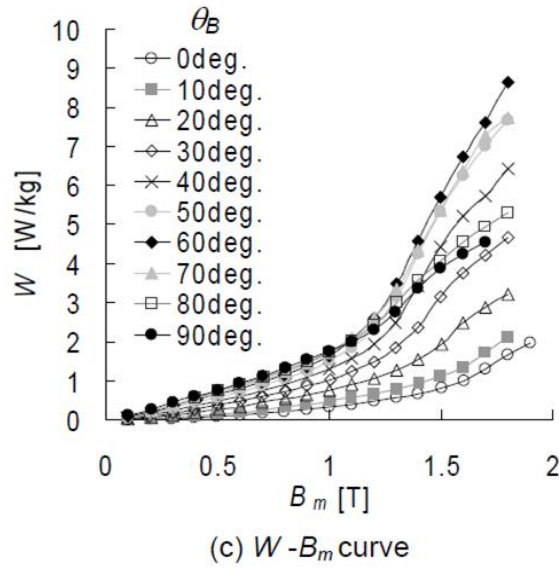


Figure 3-3: Variation of power loss (W) with maximum induction (B_m) in 3.2%Si electrical steel with strong texture along the rolling direction. The losses are measured for various orientations (θ_B) between the applied field and the direction of preferred <001> orientation. From (Mori et al., 2011).

Figure 3-4 shows how the hysteresis, eddy current and total losses change with increasing grain size. While it is true that small grains produce high hysteresis losses, beyond a certain size grains have domain walls at a wider spacing such that the extra losses associated with their required rapid movement can jeopardise the efficiency of the material. Therefore, it has been suggested that at 50 Hz the grain size for transformer electrical steels should not exceed 5–7mm, whereas for frequencies >400 Hz grains of size below 100 μ m are

advantageous. For motor steels of a non-oriented variety it has always been appropriate to grow grains as large as practicable as the metallurgy of production could not give grains large enough to have inconveniently large domain wall distances at the typical frequencies employed. Therefore, for motor applications the main challenge consists in obtaining an appropriate crystallographic texture so that the properties of the material are substantially the same in all directions in the plane of the sheet while avoiding the appearance of many $\langle 111 \rangle$ fibres in the plane of the applied magnetic field (Beckley, 2002).

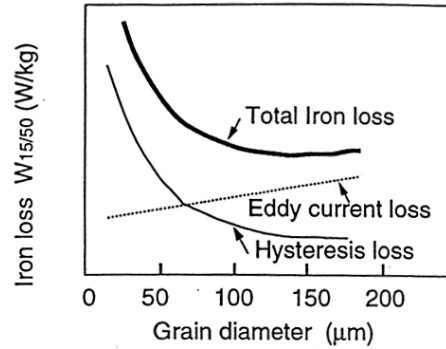


Figure 3-4: Loss components variation with grain size, at flux density of 1.5T and excitation frequency of 50Hz in a non-oriented electrical steel. Taken from (Hilinski and Johnston, 2014).

3.3 Effects of Si Addition on the Properties of Fe-Si

Figure 3-5 shows the effect of Si on various magnetic properties of Fe, namely the magnetisation at saturation (Figure 3-5(a)), total losses at excitation frequency of 60Hz (Figure 3-5(b)), as well as maximum permeability, coercivity and hysteresis loss (Figure 3-5(c)). The curves show that while the magnetisation at saturation decreases with increasing Si content, all of the other magnetic properties (permeability, coercivity and total losses) are positively affected by the addition of Si to the Fe matrix. In particular, Fe-Si alloy

containing about 6.5%-Si are shown to be characterised by the lowest total losses (Bozorth, 1993a).

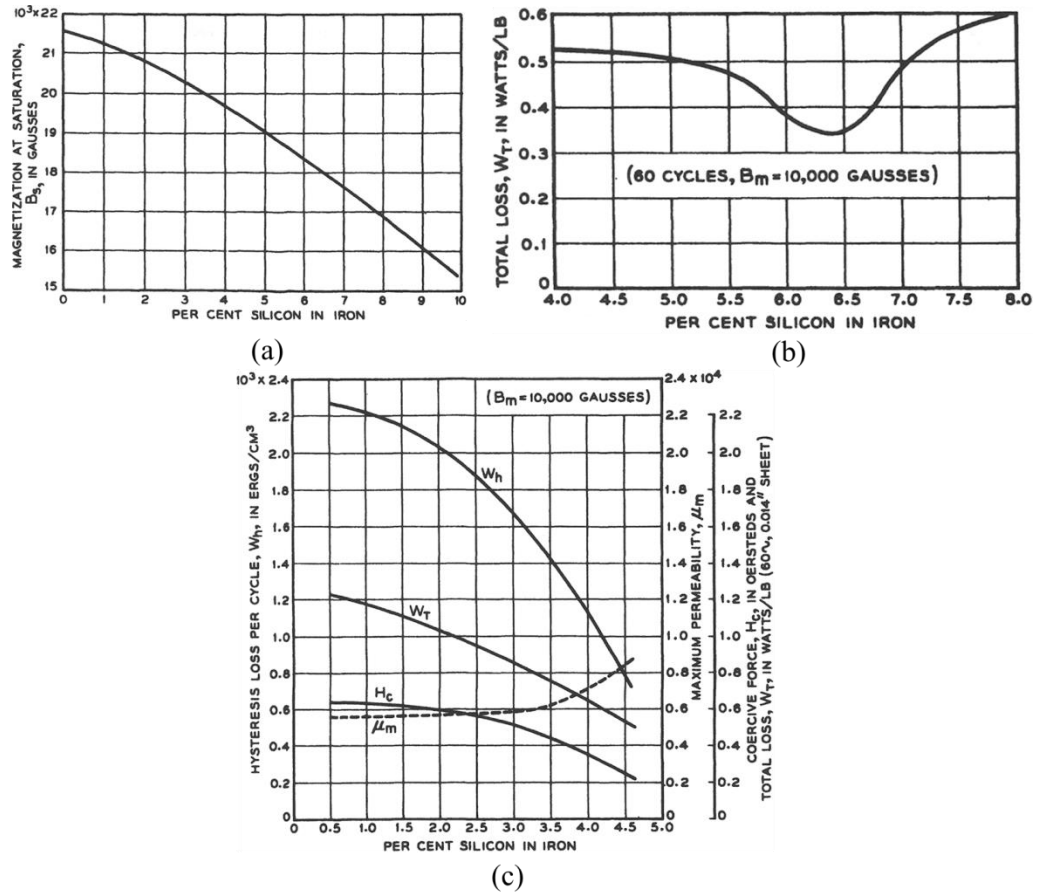


Figure 3-5: Effect of Si content on some magnetic properties of Fe. Magnetisation saturation (a), total loss at 60Hz (b), and permeability (μ_m), coercivity (H_c) and hysteresis loss (W_H) (c). Images adapted from (Bozorth, 1993a).

The markedly beneficial effect of Si on the total losses of Fe is due to the increase in electrical resistivity it causes, as Figure 3-6(a) and (b) show. Figure 3-6(b) also shows that compared to all of the other alloying elements of interest for magnetic applications, Si is the element producing the highest increase in electrical resistivity (Heck, 2013). Adding Si to Fe also causes a monotonic decrease in magnetocrystalline anisotropy, while the magnetostriction constant shows a non-monotonic behaviour and passes zero in correspondence of 6.5%Si (Hilzinger and Rodewald, 2013). Therefore, owing to its low

electromagnetic losses and near-zero magnetostriction, Fe-Si with around 6.5%Si is often considered to be ideal.

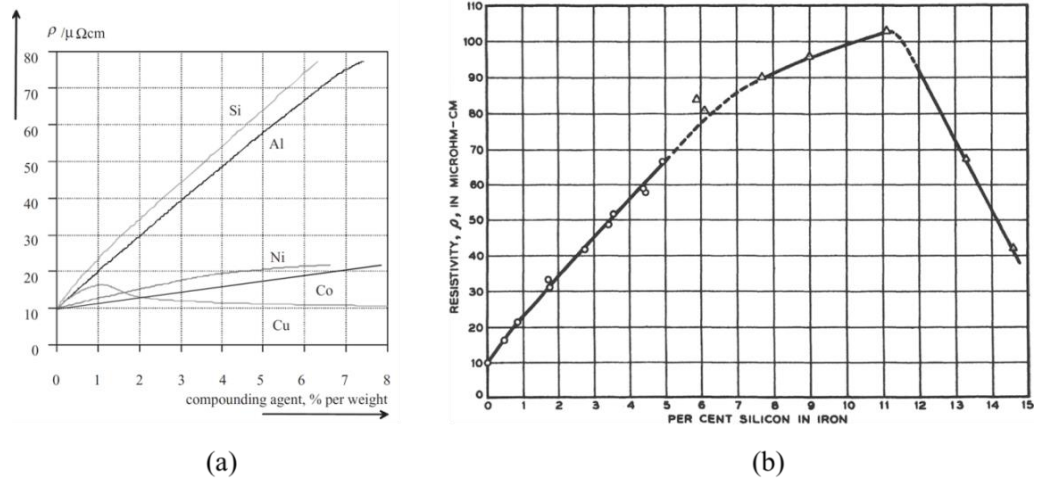


Figure 3-6: In (a) the effect of different elements on the resistivity of Fe for compounding content ranging from 0-8%wt., adapted from (Heck, 2013). In (b), the resistivity of Fe for Si contents ranging from 0-15%wt. (b), adapted from (Bozorth, 1993a).

The dependence of the mechanical properties of Fe-Si on Si content is illustrated in Figure 3-7. Although yield and tensile strengths increase until 4.5%, they suddenly start deteriorating at higher Si contents. Importantly, ductility quickly goes to 0 in the vicinity of 5% Si. Because of the brittle nature of high-Si steels, processing this material (e.g., by rolling and punching) becomes difficult. Therefore, practically all commercial Fe-Si cores contain less than 3.5%Si.

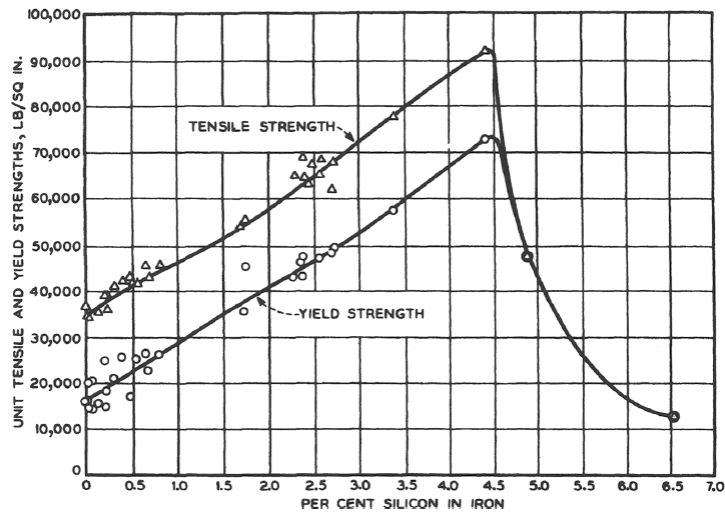


Figure 3-7: Effect of composition on the mechanical properties of Fe-Si. Both tensile and yield strength improve as Si content increase until 4.5% Si. As Si increases further, mechanical properties experience a sudden deterioration. At 5% the ductility of the material is practically zero. Image from (Bozorth, 1993a).

We have previously mentioned that assuming negligible carbon content, above 2%wt. Si, Fe-Si is bcc-structured over the whole solid temperature range. However, for Si content >5%wt., while cooling from the melt the material undergoes two ordering phase transitions, i.e. transformations that involve the re-arrangement of Si in the Fe lattice. Figure 3-1 (Swann et al., 1975) shows the Fe-Si phase diagram in the 5-9%wt.Si content range. At around 6.5%wt. Si content, disordered ferritic solid solution, commonly referred to as A2, transforms into a short range ordered B2 phase at around 820°C, with the previously randomly distributed Si atoms now occupying the centre position of the bcc cells, and the Fe atoms located at the cell corners. Therefore, the B2 phase has a stoichiometry of 1:1 Fe to Si atoms, as the schematics in Figure 1 also shows. Cooling further, below the magnetic transition (at around 680°C), another ordering transformation takes place at a temperature below 650°C. The B2 phase and some of the residual A2 transform into B2+D0₃. The

stoichiometry of $D0_3$ is 1:2, with the centre position of every second bcc cell occupied by a Si atom (refer to schematic in Figure 1). Owing to the different stoichiometry, longer-range ordering can be achieved for $D0_3$ in comparison to the B2 within the considered Si-content range (Swann et al., 1975). The ordering arrangement of the solute affects the physical properties of the high silicon steels and has especially negative implications on the mechanical properties, for it increases the hardness and hence the brittleness of the material, reducing drastically its workability. Therefore, the phenomenon of ordering explains the drop in ductility at Si concentrations above 5%wt. On the other hand, the effect of ordering on the electromagnetic properties of Fe-Si has not yet been fully elucidated. Some authors have reported that ordering can have a beneficial effect on the magnetisation properties of the alloy, although the effect of ordering on magnetic performance is in general negligible compared to other metallurgical effects, especially residual stress relief and grain growth (Faudot et al., 1989, Bolfarini et al., 2008). Other authors however have shown how different morphologies and volume fractions of B2/ $D0_3$ can either positively or negatively affect the magnetisation and loss properties (Bi et al., 1996).

3.4 Existing Processing Routes

Silicon steels are usually rolled in the form of laminates, which are subsequently punched to obtain the desired cross-sectional shape. Laminates are finally stacked together using insulating varnish to create the final component. Fe-Si laminates are well-known in the industry as electrical steels. As previously mentioned in Section 1.3, lamination substantially reduces the eddy current losses, especially in components where magnetic flux changes

with time (e.g., in the stator core of rotating electrical machines). Laminates can be obtained by both hot and cold rolling, although the latter is most popular for the improved sheet flatness it produces. Furthermore, a sequence of controlled hot and cold rolling steps can be employed to produce strong $\langle 001 \rangle$ texture along the rolling direction and $\langle 101 \rangle$ texture along the transverse and normal directions (cube-on-edge or Goss texture). As explained in Section 3.2, since the $\langle 001 \rangle$ crystallographic axis family in b.c.c. systems represents the easy axis of magnetisation, grain orientated electrical steels are characterised by high magnetic permeability and low core losses along the rolling direction (Beckley, 2002), as shown in Figure 3-8. Grain oriented steels are widely employed in the production of transformer cores, as they are suited to the simple magnetic path flowing through this class of devices. The more complex magnetic field paths that characterise rotating electrical machines require the properties of the material to be substantially the same in all directions in the plane of the sheet, hence, grain oriented electrical steels are normally not useful for the construction of machine cores, with the exception of specific designs and unconventional machine constructions, as outlined in (Kowal et al., 2010).

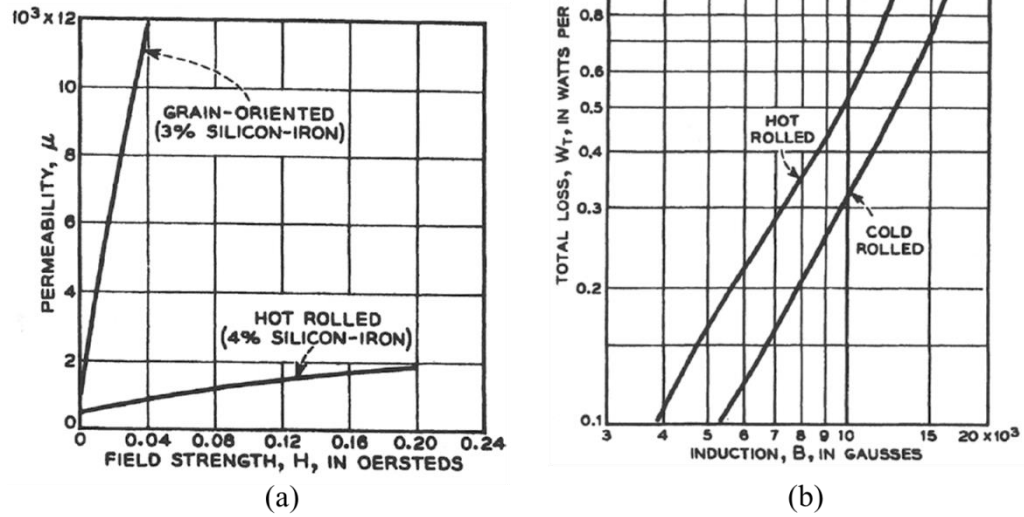


Figure 3-8: Comparison of magnetic properties in non-oriented and grain-oriented electrical steel. In (a) the permeability at low fields is compared, while in (b) the term of comparison is the total loss at 60Hz. From (Bozorth, 1993a).

Owing to their poor workability, Fe-Si alloys containing more than 3.5% Si are difficult to roll. A recent technique developed by JFE Steel employs chemical vapour deposition to add silicon by diffusion after the lamination has been created (Namikawa et al., 2005). The two 6.5%Si lamination types obtained through this technique, namely JNEX and JNHF, represent the only examples of commercialised high-Si electrical steel. However, (He et al., 2008) have pointed out the significant polluting nature of the CVD process. The general process for high-Si electrical steel production via CVD is schematically represented in (Namikawa et al., 2005). Other attempts at high-silicon steel processing have been explored in the literature, in the form of powder cores (Jang et al., 2008, Lee et al., 2008), melt-spun ribbons (Faudot et al., 1989), cast strips (Liu et al., 2012) and electron-beam physical vapour deposited

sheets (He et al., 2008). The first method consists of coating the Fe-Si powder using insulating varnish, before compressing it into the desired shape. Since the powders are insulated, it has been shown that adding Si does not provide advantages in terms of the electromagnetic performance (Lee et al., 2008). As regards the cast strips, melt-spinning and electron-beam physical vapour deposition, they rely on high cooling rates to suppress the DO₃/B₂-type ordering reaction and produce thin ribbons with a certain degree of ductility. However, to date none of these techniques have been able to achieve industrialisation.

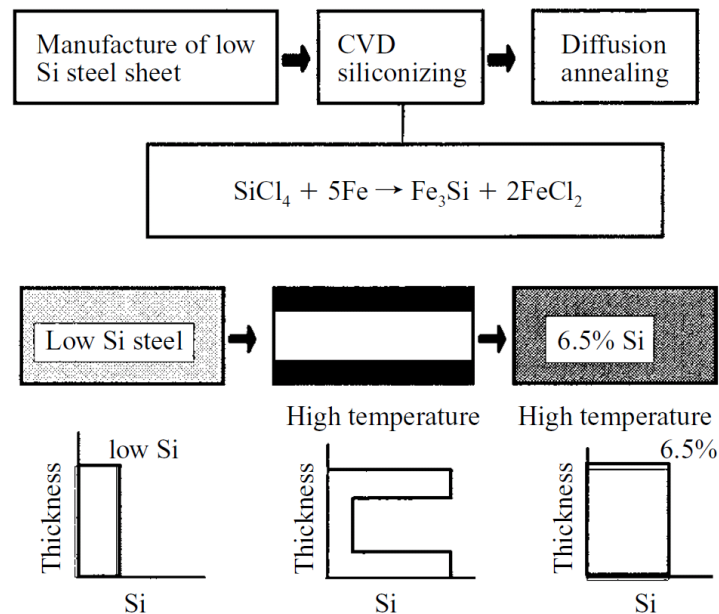


Figure 3-9: The JFE's manufacturing process for the production of laminated Fe-6.5%Si via CVD (Namikawa et al., 2005).

After manufacturing, Fe-Si parts typically require heat-treatment in order to:

- 1) Remove the residual stresses introduced during the manufacturing process;
- 2) Promote grain growth;
- 3) Control the crystallographic texture;

4) Control phase transformations (i.e., phase ordering for high-Si steels).

In order to remove the residual stresses, rolled Fe-Si samples are typically stress-relief annealed at temperatures ranging between 700 and 800°C (Landgraf et al., 2003, Beckley, 2002). Grain growth is achieved through recrystallisation of the strained grains through annealing above a certain critical temperature, which is dependent on the extent of strain energy stored during manufacturing (Beckley, 2002). Furthermore, some authors have found that increasing the Si-content from 3 to 3.7% in commercial electrical steels can retard recrystallisation and grain growth (Nakashima et al., 1997). The optical micrographs after hot-rolling and after annealing for the two Si contents considered in (Nakashima et al., 1997) are shown in Figure 3-10. In high-Si steels, grain-growth retardation has also been observed and attributed to the onset of ordering (Faudot et al., 1989). Given the particular interest of this research in rapidly-quenched high-Si steel, in Figure 3-11 we report the results presented in (Faudot et al., 1989) on the dependence of grain size with annealing temperature in melt-spun Fe-6.5wt.Si ribbons. These results will be used in the present research to guide our heat-treating strategy, which shall be outlined in Section 7.6.

3.5 Summary

This chapter has reviewed the main notions in the field of Fe-Si alloys, covering the relationship between metallurgy, magnetic properties and processing/post-processing. Importantly, the addition of Si up to 6.5 wt.% improves the magnetic properties and electrical resistivity, but embrittles the solid solution, thus making its processing using traditional manufacturing methods (rolling and punching) difficult for Si contents above 3 wt.%.

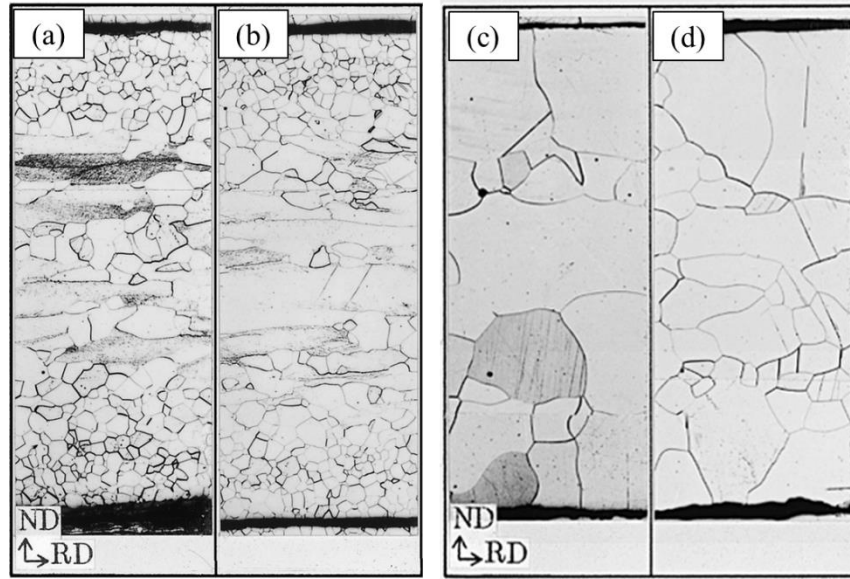


Figure 3-10: Optical micrographs showing the effect of annealing and Si content on the grain-size of grain-oriented electrical steel. Grain structures after hot rolling for Si-content of 3% and 3.7% are shown in (a) and (b), respectively. In (c) and (d), the grain structures after annealing at 1200°C are shown for Si-content of 3% and 3.7%, respectively. The plane analysed is normal to the sheet surface and parallel to the rolling direction. Images adapted from (Nakashima et al., 1997).

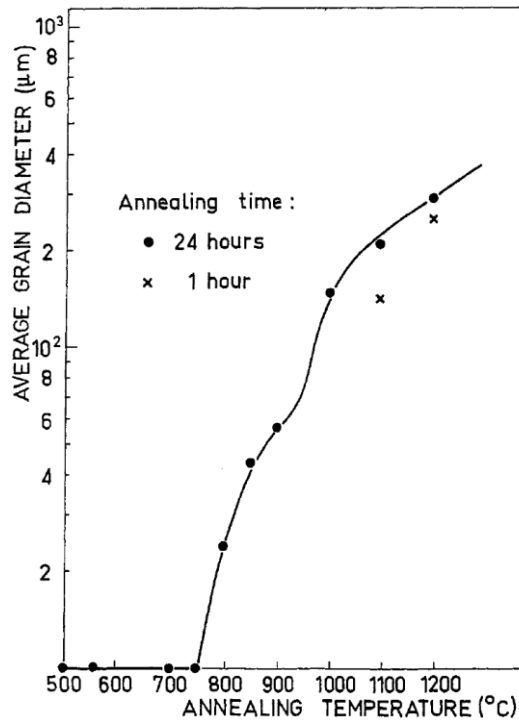


Figure 3-11: Effect of annealing temperature on the grain size of melt-spun Fe-6.5%wt.Si ribbon, from (Faudot et al., 1989).

4. Physical Metallurgy of SLM

4.1 Introduction

Selective Laser Melting (SLM) is a highly complex process, involving multiple metallurgical mechanisms and transport phenomena, which determine the ultimate density, internal stress state and crystallography of the built part. These features, in turn, are paramount in determining the functional behaviour of the part. The metallurgical mechanisms behind defect formation, internal stress and microstructure development will be described in the next section, followed by a presentation of the literature findings on the effects of the main SLM processing parameters on part quality and microstructure. The notions outlined in this chapter will serve as a basis to interpret the material properties of the high-silicon steel alloy processed via SLM as part of this research work.

4.2 Melt-Pool Dynamics and Pore Formation

The melting process in SLM is characterised by the local interaction between the powder bed and the fast-moving heat source (the laser), which results in the rapid heating and melting of the metal powder and leads to the formation of a melt-pool and a heat-affected zone, as described in (Kruth et al., 2010) and represented here in Figure 4-1 using a schematics from the same authors.

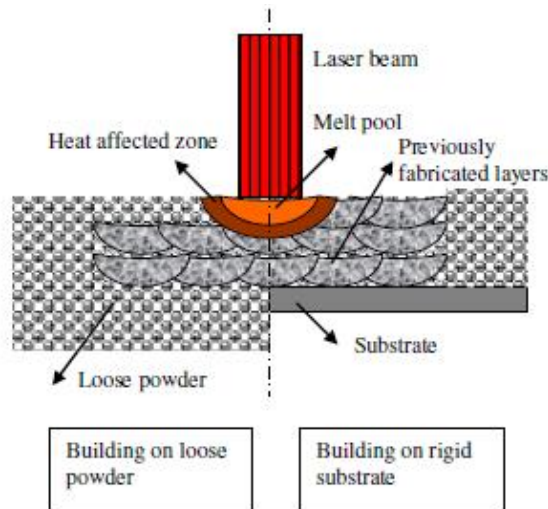


Figure 4-1: Melt-pool and heat-affected zone formation during the SLM process. Schematics from (Kruth et al., 2010).

The formation of pores upon melt-pool solidification can hinder the densification of laser- (and electron-) processed parts. At low laser energy densities, defective bonding between melt-pools can occur owing to either incomplete melting of the base powder or to poor wettability/high surface tension of the melt-pool, and leads to the formation of large, irregularly-shaped voids (Vilaro et al., 2011, Aboulkhair et al., 2014, Kruth et al., 2004, Al-Bermani et al., 2010). A surface tension-based mechanism leading to pore formation is known as “balling”, which concerns the spherodization of the molten metal. Balling has been extensively studied in the SLM literature (Kruth et al., 2004, Yadroitsev et al., 2007, Khairallah et al., 2016, Li et al., 2010). Balling results in a rough, irregular surface that obstructs smooth layer deposition and thus increases the chance of pore formation in additional layers. Surface tension can cause balling by preventing the liquid from wetting the underlying substrate, and/or by causing the onset of Plateau-Rayleigh instabilities leading to the break-up of the molten track into spheroids when the latter’s length-to-width ratio l/d is too high relative to the viscosity of the

liquid, as shown in Figure 4-2 (Kruth et al., 2004). Another surface tension-related mechanism potentially responsible for pore formation is linked to the presence of steep thermal and, consequently, surface tension gradients, which lead to the formation of thermal and capillary-driven convective flows inside the melt-pool (Rombouts et al., 2006, Khairallah et al., 2016, Matthews et al., 2016). Depending on the variation of liquid's surface tension with temperature, which is normally negative, but can become positive depending on the specific composition of the alloy (David et al., 2003), either inward or outward radial convective flows arise in a stationary melt-pool. It is well-known in both the SLM (Rombouts et al., 2006) and welding community (David et al., 2003), that an outwards flow results in a broad, shallow melt-pool, while an inwards flow results in a deeper but narrower melt-pool, as shown in Figure 4-3. A narrower melt-pool, in turn, is more likely to result in Plateau-Rayleigh instabilities and, hence, balling.

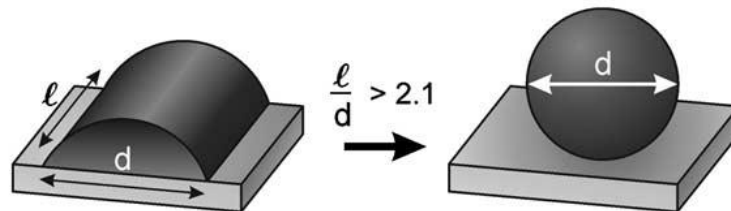


Figure 4-2: sketch adapted from (Kruth et al., 2004), showing the break-up of a half-cylindrical track into a number of spheroids when its length-to-width ratio is too high, according to Plateau-Rayleigh theory.

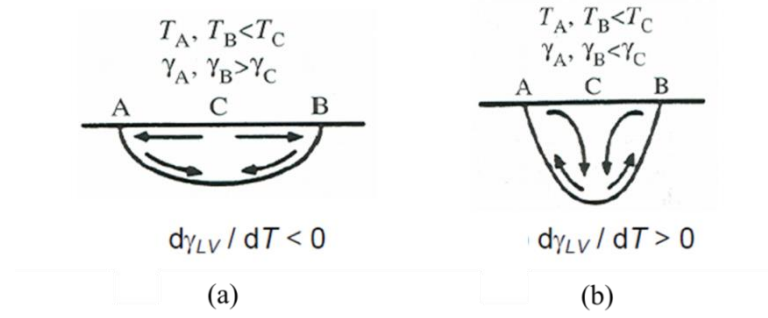


Figure 4-3: Surface tension gradient-induced capillary flows in the melt-pool can be directed radially outward (a) or inward (b), depending on whether the variation of the surface tension with temperature is negative or positive, respectively. Adapted from (Kruth et al., 2010).

If the laser energy density is high enough to cause the material to evaporate, the evaporated powder particles expand and generate a recoil pressure on the melt-pool (Kruth et al., 2004). Recoil pressure is known from the laser welding literature to cause a qualitative change in the melt-pool shape, which turns from flat (conduction-mode welding) to protuberant (keyhole welding) (Dahotre and Harimkar, 2008). At even higher incident intensities the vapour turns into plasma through interaction with the incoming laser radiation. While low recoil pressures facilitate the flattening of the melt in SLM, thus favouring smoother layer deposition, high pressures cause material removal by melt expulsion, with the consequent formation of porosities (Khairallah et al., 2016) and potential modifications in the chemical composition of the final build (Simonelli et al., 2015). Furthermore, in both the SLM and laser welding literature unstable collapse of the keyhole during cooling has been associated with the formation of pores (Pastor et al., 1999, Aboulkhair et al., 2014, Gong et al., 2014). Gas bubble entrapment at the tip of the keyhole is the origin of keyhole porosities, which are usually small and spherical in shape, as opposed to the larger and more irregularly-shaped bonding defects that form at low laser energy densities. The micrographs in Figure 4-4(a) and (b) show examples of

irregular and spherical porosities in SLM materials, from Aboulkhair et al. (2014) and Gong et al. (2014), respectively. An example of spherical porosity in laser-weld Al is also included for comparison in Figure 4-4(c) from Pastor et al. (1999).

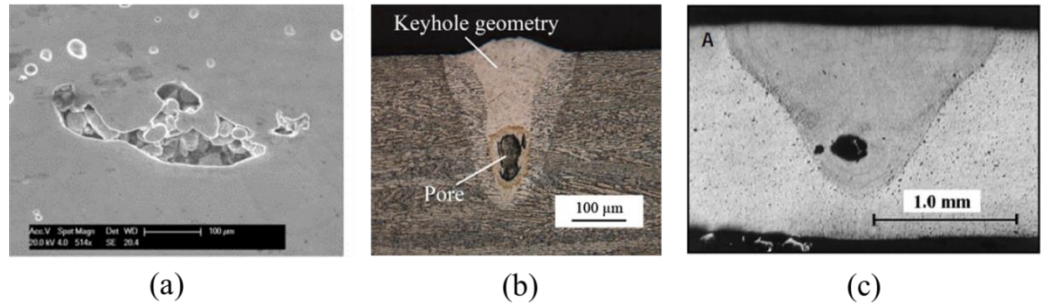


Figure 4-4: Bonding defect (a) and spherical porosity (b) in SLM Al-10Si-Mg and Ti-6Al-4V, respectively. In (c), an example of spherical porosity in autogenous laser-welding of Al is proposed. Images from Aboulkhair et al. (2014), Gong et al. (2014) and Pastor et al. (1999).

A final phenomenon that can lead to the formation of porosities is that of powder denudation, which entails the clearing of powder around a laser-processed track, as shown in the optical image of Figure 4-5. Powder denudation is among the most complex phenomena in SLM, as it involves a number of competing mechanisms, including surface tension-driven and recoil pressure-driven effects (Matthews et al., 2016). The authors suggest that, while denudation effects can occur at both low and high recoil pressure, at low pressures the metal vapour flux emitted from the melt-pool may sufficiently expand into a wide plume, thus promoting outwards particle motion, which results in a wider denudation area, as shown in the optical image and schematics of Figure 4-5.

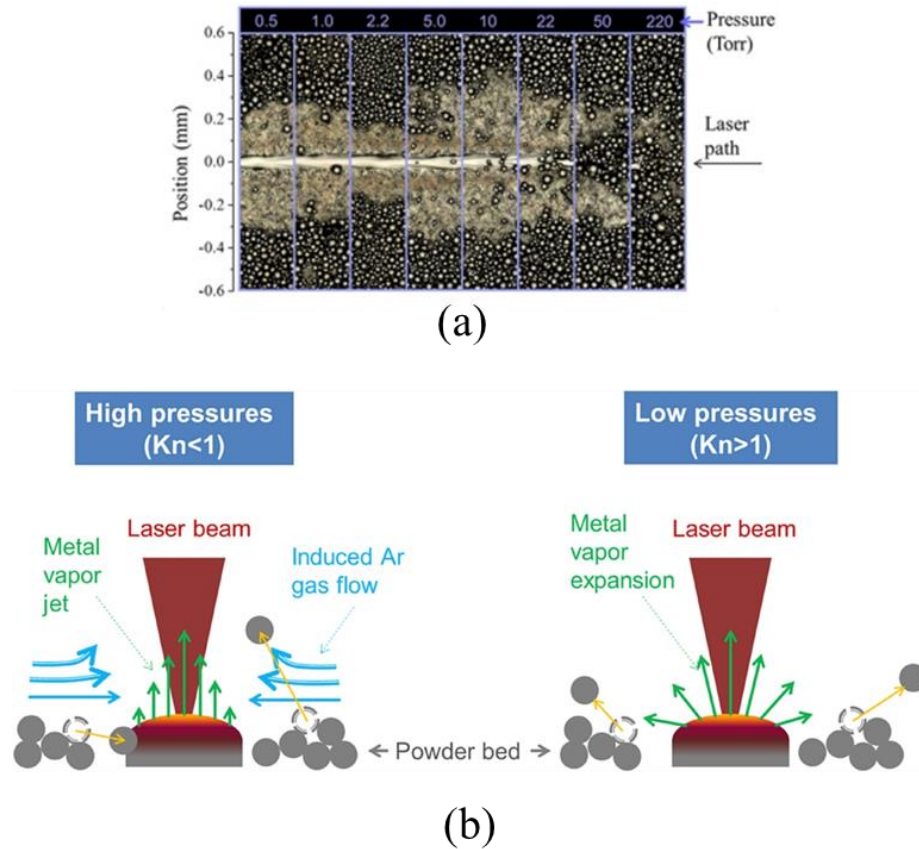


Figure 4-5: Powder denudation phenomenon in SLM. An optical top view of a single track test case, demonstrating how denudation extent is affected by local gas pressure is shown in (a). The schematics in (b) show the effects of high (left) and low (right) vapour pressure on particle motion around the molten track. Adapted from (Matthews et al., 2016).

4.3 Residual Stress and Crack Formation

Alongside porosity, residual stresses are another important determinant of part performance, and the primary drawback of manufacturing a part using SLM, since they jeopardise the structural integrity of the final part by producing part distortion, warping and possibly failure by crack formation *during* the manufacturing process. As described, among others, in (Merzelis and Kruth, 2006), SLM introduces tensile residual stresses in the processed part owing to the localised application of high laser energy over a very short time interval (in the order of the μ s) which, alongside the generally poor heat conductivity of

the powder bed, develops a high thermal gradient around the laser spot. During the heating stage, non-uniform expansion takes place and eventually the heat affected zone (HAZ) – whose tendency to expand is inhibited by the surrounding material – is plastically compressed. During the cooling stage, the upper layer's tendency to contract is counteracted by the underlying layers, which are anchored to the base plate, thus creating tensile residual stresses. The above-described thermo-mechanical mechanism is better known as the Thermal Gradient Mechanism (TGM) and is exploited in laser forming for the controlled bending of metal plates (Shi et al., 2007). Here, the mechanism is represented in Figure 4-6 using a graphics from (Kruth et al., 2004).

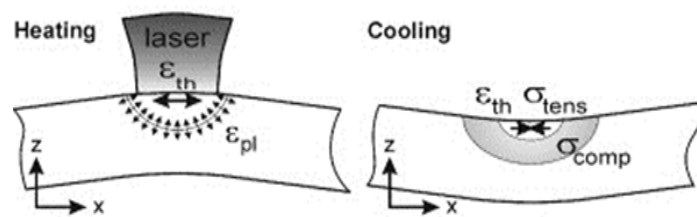


Figure 4-6: The two phases in the Thermal Gradient Mechanism (TGM), from (Kruth et al., 2004).

A number of strategies have been employed in the literature to quantitatively evaluate stress distribution within a manufactured part with respect to sample and base-plate geometry and process parameter settings. Direct measurements of the residual stresses (generally of Type I) have been performed using X-Ray Diffraction (XRD) (Mercelis and Kruth, 2006), sample deflection measurement (Kruth et al., 2012, Pohl et al., 2001) and destructive methods, such as hole-drilling (Mercelis and Kruth, 2006). Numerical analysis, especially using the Finite Element Method (FEM), is also widely employed as it allows for residual stress formation prediction, thus providing a tool for understanding the

dynamics of stress induction and possibly for their minimisation in the design stage (Nickel et al., 2001, Parry et al., 2016).

Owing to the layer-wise nature of the additive process, TGM has been shown to develop a tensile stress state that is predominantly planar, with the two principal directions of the stress tensor lying in the scanning plane. As known from experience in laser forming (Shi et al., 2007), the stress develops mainly along the scan vector direction, owing to the distribution of the thermal gradient magnitude. Furthermore, stress has been shown to increase with sample height owing to stress accumulation as new layers are added (Mercelis and Kruth, 2006, Pohl et al., 2001, Kruth et al., 2012). As the as-built part is separated from the base plate, stress relaxation takes place, which leads to a significant reduction in the internal stress, as shown in Figure 4-7, where the x and y components of the normal stress (σ_{xx} and σ_{yy} , respectively) on the top surface of an SLM sample processed using an alternating x-y scanning pattern (refer to Section 4.5 for a list and description of the most popular scanning patterns) were measured by XRD (Mercelis and Kruth, 2006). The authors' findings suggest that base-plate removal can relax more than half of the initial residual stress in the part. With regards to the distribution of the normal planar stress components σ_{xx} and σ_{yy} , while initial analytical studies had assumed that the variation of the normal stress across the part's width can be neglected (Mercelis and Kruth, 2006), more recent, FEM-based studies (Parry et al., 2016) have shown that the stress field is more complicated, and characterised by a non-uniform distribution, as shown in the simulation results in Figure 4-8 for a single layer 3mm x 3mm test case produced using a uni-directional scan strategy with scan vector oriented along the y-axis. The figure shows that the

σ_{yy} , i.e., the longitudinal stress component (Figure 4-8(a)) is substantially higher than the transversal component (σ_{xx} , in Figure 4-8(b)) owing to the chosen scan strategy. However, regions of lower transverse and longitudinal stresses are situated towards the end of the scan vectors.

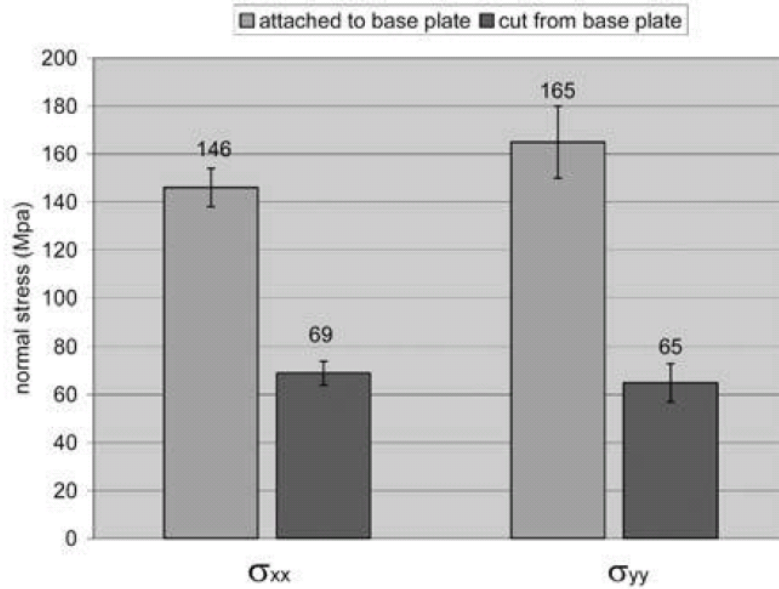


Figure 4-7: Effect of base-plate removal on top-surface tensile stress magnitude, as measured using XRD (Mercelis and Kruth, 2006).

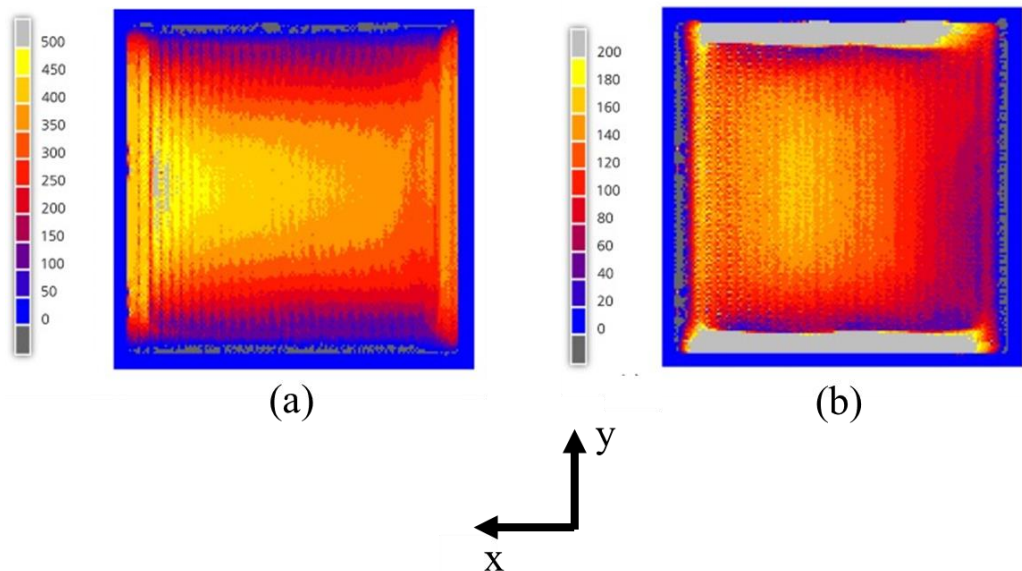


Figure 4-8: FEM simulation results showing the top view of normal stress components σ_{yy} (a) and σ_{xx} (b) for a single-layer 3mmx3mm test-case obtained using unidirectional scan strategy. Adapted from (Parry et al., 2016).

4.4 Microstructural Evolution

It is well known from autogenous-weld-metal solidification theory (David and Vitek, 1989, David et al., 2003) that as the laser-induced melt-pool solidifies, grain-structure development is determined by thermodynamic, kinetic and geometrical factors. The regions whose microstructures are affected by the heat-source are the melt-pool (fusion-zone) and the heat-affected zone (HAZ), where the temperature are sufficiently high to trigger diffusion-based transformations (David et al., 2003). The fusion zone and the HAZ are represented in the schematics of Figure 4-9.

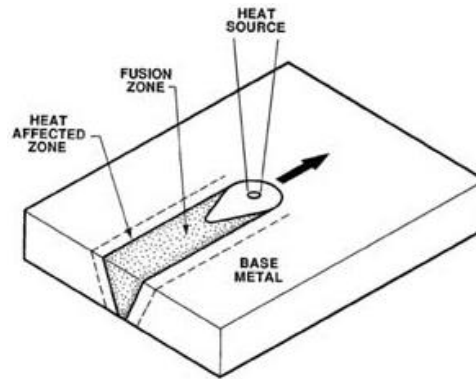


Figure 4-9: fusion zone (melt-pool) and heat-affected zone (HAZ) in a weld metal, from (David et al., 2003).

Grain morphological and crystallographic textures are mainly determined by the melt-pool shape and the structure of the substrate metal. In autogenous welding the melt-pool is formed by melting the upper region of the underlying metal, which is of a composition similar to that of the liquid. The partial grain melting in the base metal provides a substrate for solidification, and this can potentially allow the grain to grow epitaxially. However, the melt-pool shape, whose origins were described in Section 4.2, determines the direction of the thermal gradient (which lies perpendicular the melt-pool boundary) and,

consequently, that of the solidification front. Since crystals have preferred “easy growth” directions (e.g., the $\langle 001 \rangle$ family in cubic crystals), the optimum orientation for the grains in the base metal to grow epitaxially is one in which the easy growth direction is perpendicular to the solidification front. It follows that the melt-pool shape has a critical influence in determining whether a partially molten grain in the substrate will grow epitaxially, or new nuclei will form according to a grain-growth selection process (David and Vitek, 1989). Furthermore, depending on the shape of the melt-pool, the local direction of the thermal gradient shifts as this moves towards the melt-pool center-line, and such gradual change in orientation can cause a number of different scenarios, including the growth of curved columnar grains, or the transition from columnar to equi-axed grains as the solidification front reaches the centreline, mainly depending on the character of the solidification structures and the propensity to nucleation (David and Vitek, 1989). Two micrographs showing typical microstructures of a weld metal in the vertical and horizontal cross-sections of the weld are shown in Figure 4-10 from (David and Vitek, 1989). The grain-growth selection process is visible in the frontal plane (Figure 4-10(a)), while the formation of curved columnar grains in the top plane of the sample, which indicates their adaptation to the change in direction of the thermal gradient towards the melt-pool centreline, is shown in Figure 4-10(b).

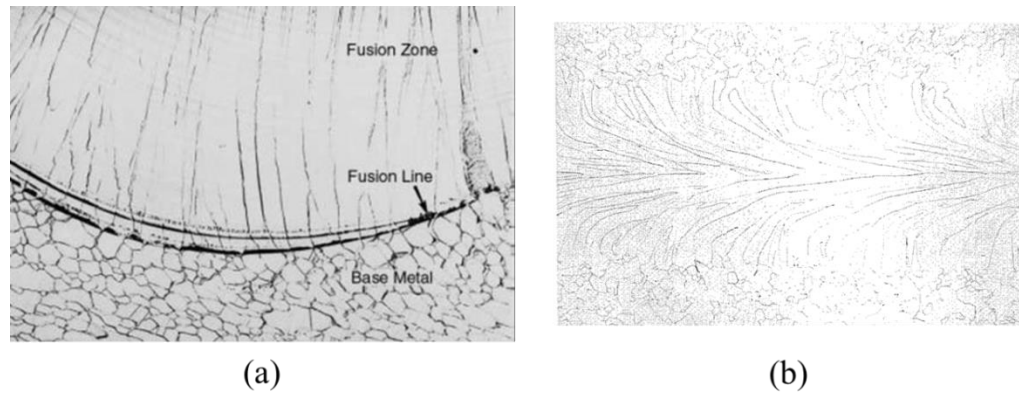


Figure 4-10: Optical micrographs of an iridium alloy weld showing the microstructure in the frontal plane of the sample (a) and in the top plane (b), from (David et al., 2003)

The cooling rate and alloy composition mainly determine the solidification structures, which can be of planar, cellular or dendritic nature (David and Vitek, 1989). These structures, shown schematically in Figure 4-11 will decide the compositional distribution in the solidified alloy and therefore contribute to its functional behaviour. At equilibrium (i.e., at relatively slow solidification speeds), the thermal stability of the solid-liquid interface is paramount in determining which one of the three growth modalities will occur: in particular, a stable interface is characterised by a positive temperature gradient ahead of the interface and will result in a planar structure Figure 4-11(a), in which the compositional distribution is homogeneous, i.e., all solutes are in the solid solution. While in pure metals thermal undercooling at fast cooling rates is the sole determinant of interface stability, in alloys the interface can become unstable owing to constitutional super-cooling of the liquid front, deriving from the segregation of the solute in the liquid phase, whose entity depends on the alloy's phase diagram. Depending on the magnitude of the super-cooling, the planar front can develop single-armed protrusions thus resulting in a

cellular structure (Figure 4-11(b)), or branched protrusions, thus resulting in dendritic morphologies (Figure 4-11(c)) (Campbell, 2012).

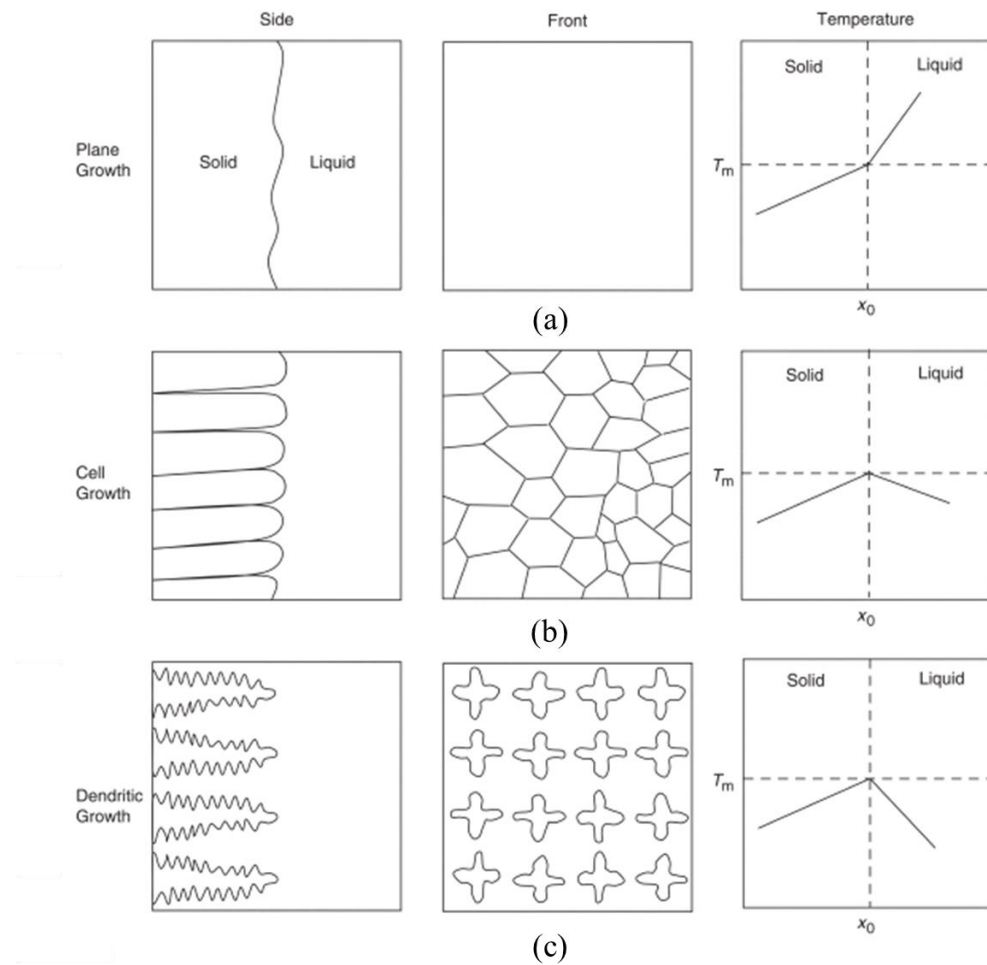


Figure 4-11: Morphologies (left and centre) and thermal conditions (right) for the solidification structures during (epitaxial) grain growth: planar (a), cellular (b) and dendritic (c).

The conditions and mechanisms described in the weld metal literature have been used to interpret the microstructures found in SLM materials (Thijs et al., 2010, Niendorf et al., 2013, Thijs et al., 2013a, Cloots et al., 2016, Vrancken et al., 2014). The typical microstructures encountered in many laser- (and electron-) beam processed alloys are shown schematically in Figure 4-12, which was taken from a recently published review article (Herzog et al., 2016). Owing to the overlap between adjacent scan tracks and layers, the grains near

the surface of the underlying/adjacent solid layer are partially remolten at each new track/layer scan. In particular, at the centreline of the melt-pool, the solidification direction is approximately along the build direction for every added layer. Therefore, epitaxial growth is promoted along directions close to the build direction, and this often leads to the formation of long columnar grains, which can span several hundreds of microns (Thijs et al., 2010, Vrancken et al., 2014, Niendorf et al., 2013, Simonelli et al., 2014). Finer, curved columnar grains are often observed close to the lateral sample's boundaries, owing to the different thermal gradient path within these regions, as also shown in Figure 4-12 (Herzog et al., 2016, Thijs et al., 2013b). Since epitaxial growth is crystallography-driven, the morphological texture is often accompanied by a crystallographic texture, characterised by the $\langle 001 \rangle$ family aligning along the direction of growth for cubic crystals (Niendorf et al., 2013).

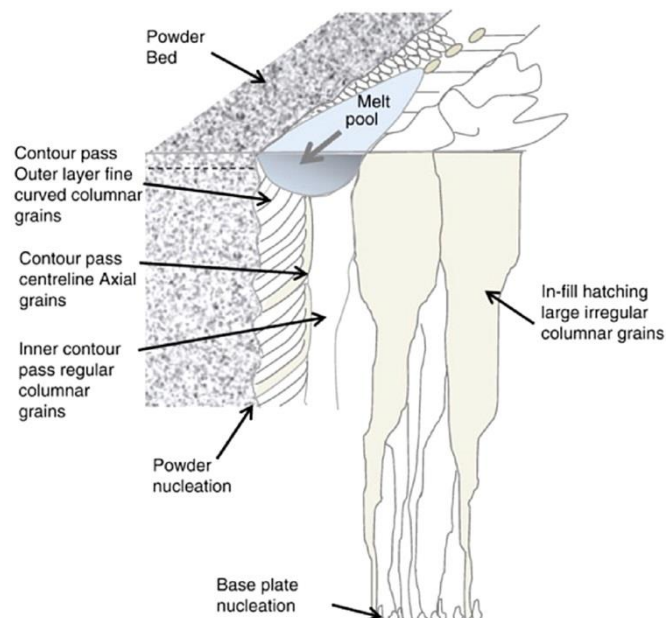
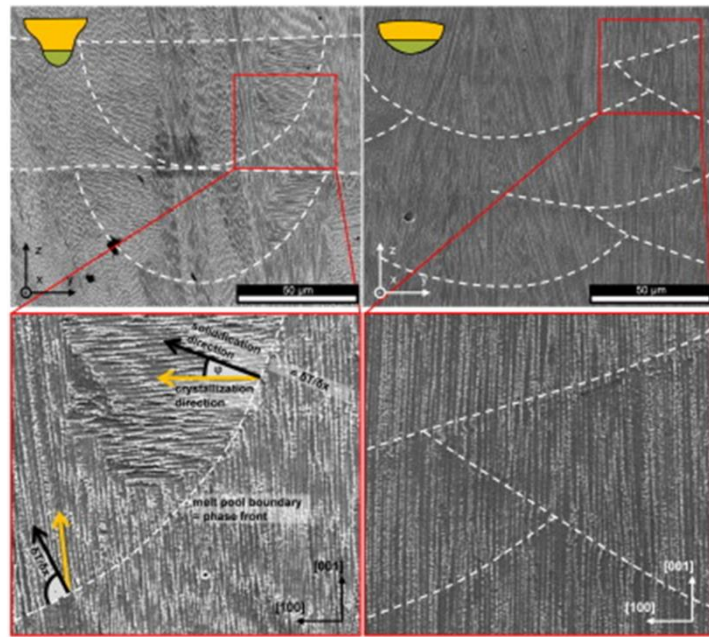
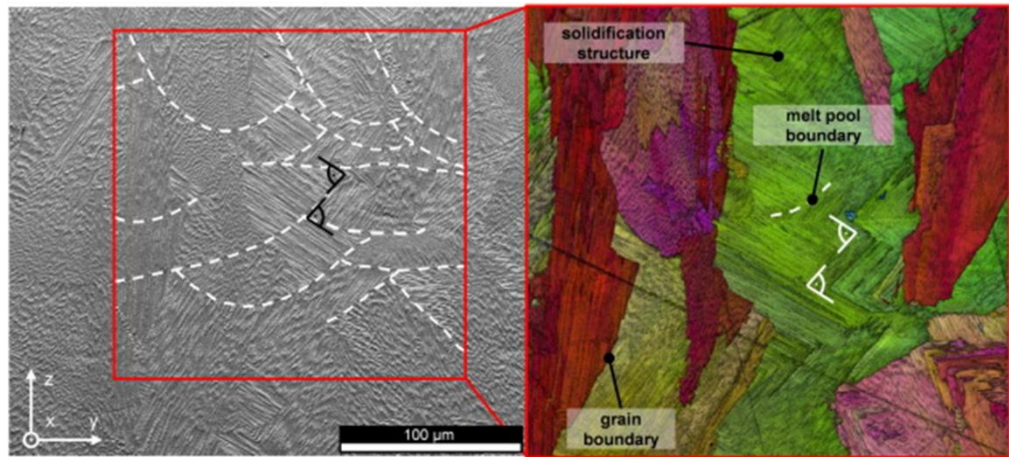


Figure 4-12: Typical microstructures encountered in SLM samples, and their dependence on the spatial location within the sample. Drawing from (Herzog et al., 2016).

In general, the specific geometrical and crystallographic conditions determine the extent and directionality of the epitaxial growth. In (Cloots et al., 2016), the authors show the effects of melt-pool shape on the microstructure of Inconel IN738LC alloy, refer to Figure 4-13(a). In case of shallow melt-pools (right micrographs in Figure 4-13(a)), the cellular solidification structures' direction remains aligned with the build direction along the entire melt-pool boundaries, since higher-angle cells have been remolten when the new layer was deposited. On the other hand (left micrographs in Figure 4-13(a)), when protuberant melt-pools are used, the direction of the cellular structures shifts away from the centreline. The shift however does not occur following the solidification direction, but the symmetry of the crystal's unit cell. Therefore, the crystallographic orientation of the remolten grain is not disturbed and epitaxial elongation across the melt-pool boundaries can occur, as shown in the micrograph and orientation map in Figure 4-13(b).



(a)



(b)

Figure 4-13: The orientation of the solidification structures in SLM IN738LC is affected by the melt-pool structures. In (a) SEM micrographs show that, while the solidification cells remain aligned with build direction along the whole melt-pool boundary in case of shallow melt-pool (right), a shift in direction is observed in case of protuberant melt-pool (left). The micrograph and orientation map in (b) show that the shift in orientation does not hinder grain growth when it occurs along crystallographic directions of easy growth.

As discussed for weld metal, the solidification modalities found in SLM materials are strongly dependent on the thermodynamics of the alloy, since this determines solute micro- and macro-segregation dynamics during solidification

(Vrancken et al., 2014, Cloots et al., 2016). Cellular solidification with Si micro-segregation at the cells' boundaries has been reported for SLM Al-10Si-Mg (Thijs et al., 2013a, Aboulkhair et al., 2014, Brandl et al., 2012). In (Cloots et al., 2016), the micro-segregation of Zr at the cells boundaries following cellular solidification in IN738LC has been found responsible for the observed hot-cracking phenomena. Authors have also reported the presence of multiple solidification modalities within the same melt-pool. In (Vrancken et al., 2014), planar solidification was observed within 5-to-10 μ m from the melt-pool boundary of SLM β -Ti, beyond which cellular solidification takes over.

For a general alloy, the solidification modality, as well as the morphology of the final grain structure, is dependent on two fundamental solidification parameters, which are well-known in the welding literature (David et al., 2003): the local thermal gradient (G) and the solidification front speed (R). The ratio of these two quantities (G/R) determines the morphology of the solidification structures. At the low extreme of the G/R spectrum we find planar solidification, whereas at the high extreme the solidification occurs via the formation of equiaxed dendrites. The product $G \cdot R$ represents the solidification rate and determines the size of the solidification and grain structures. Solidification maps (also known as $G-R$ maps) can be used to visualise the transitions between different solidification modalities (Figure 4-14a) and grain morphologies (Figure 4-14b).

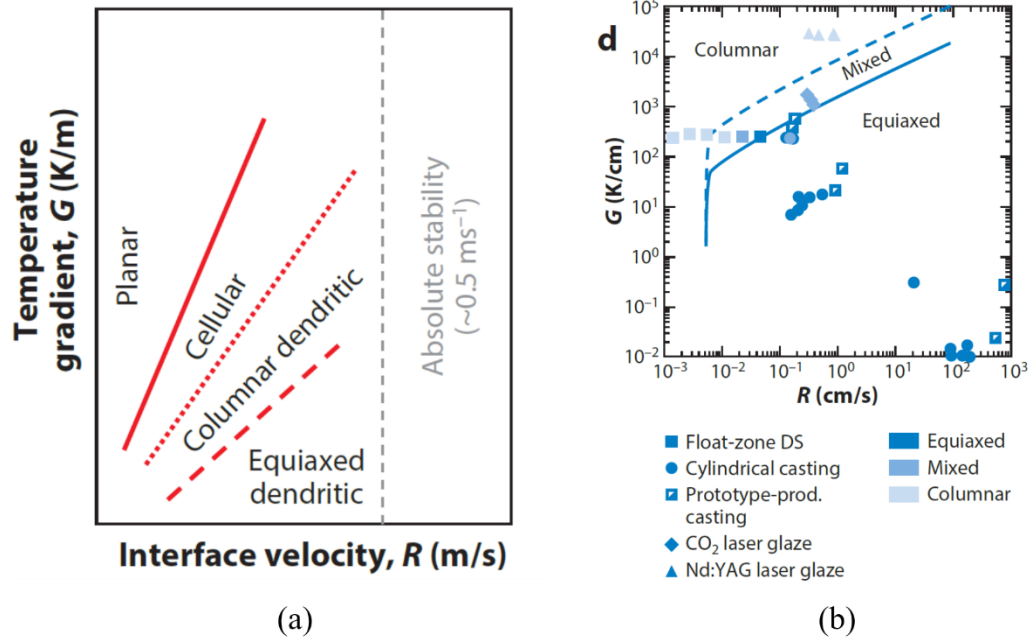


Figure 4-14: Solidification maps for solidification mode (a) and grain morphology identification (b). Images from (Herzog et al., 2016).

4.5 Effect of SLM Process Parameters on Build Quality, Microstructures and Functional Behaviour

Some of the most important SLM processing parameters, namely, laser power, scan speed, hatch spacing, layer thickness, laser energy density and scan pattern strategies are discussed below, particularly in the context of their role in determining a part's quality, microstructure and behaviour. Interactions between the above-mentioned processing parameters are also discussed.

4.5.1 Laser Power

A high laser power promotes both liquid spreading (wetting ability) and melting depth, which in turn improves bonding between molten tracks and layers, ultimately resulting in denser final part. This is demonstrated in (Li et al., 2010) for stainless steel 316L. In (Kruth et al., 2004), the authors argue that if laser power is high enough to induce vaporization, then balling effects are

reduced and the final specimen has higher density. With regard to the effect of laser power on microstructure, Niendorf and colleagues (Niendorf et al., 2013) discuss the possibility of obtaining highly anisotropic austenitic stainless steel structures by replacing a 400W laser beam source with a 1000W one, thanks to the enhancement of the morphological and crystallographic textures along the build direction (Figure 4-15). The relationship is explained by the increased amount of partial grain-remelting occurring when the energy input of the scanning laser is increased, with stronger thermal gradients oriented along the build direction fostering epitaxial growth.

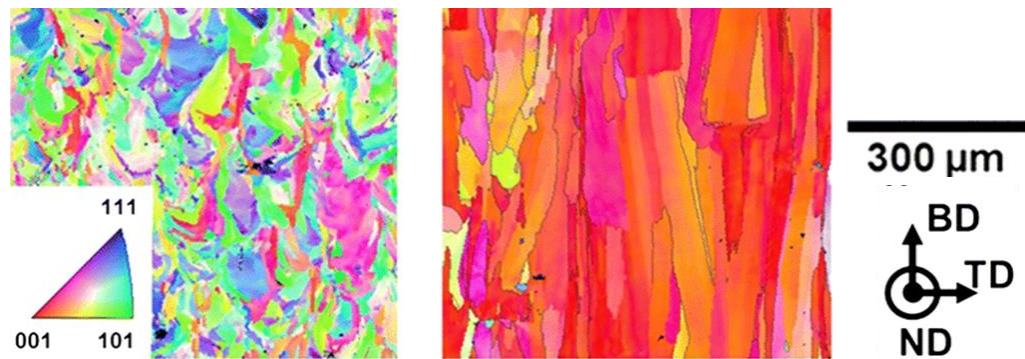


Figure 4-15: Orientation maps showing the microstructure and $\langle 001 \rangle$ crystallographic texture along the build direction and for the sample's front view. On the left, the powder was processed using a 400W, on the right with a 1000W. Material is austenitic stainless steel.

4.5.2 Laser Scan Speed

In general, given a fixed radiation power, slower laser radiation speeds improve the spreading ability of the liquid, which in turns enlarges the width and depth of the scanning track (Yadroitsev et al., 2007, Li et al., 2012). An interesting quantity is the ratio P/v between radiation power P and speed v , which determines the amount of delivered energy per unit length, and is known as effective power, or linear energy input. Many contributions in the SLM literature have focused on the study of the effect of the linear energy input on

part densification (Yadroitsev et al., 2007, Gu et al., 2013, Tsopanos et al., 2010). The results for Inox 904L single tracks (Yadroitsev et al., 2007) and 17-4PH stainless steel cubic samples (Gu et al., 2013) are shown in Figure 4-17(a) and (b), respectively. Many examples of research efforts have systematically discussed the relative contributions towards part densification of the processing laser power and scan speed at fixed energy density. This research indicates that given the same processing linear energy density, different laser power-scan speed combinations can lead to significantly different part porosities. In particular, all authors have shown that a higher laser power is more effective in promoting part densification (refer for example to Figure 4-17(b)). Furthermore, it was shown in (Yadroitsev et al., 2007) that at high linear energy densities, a lower power causes an enlargement of the heat-affected zone, which in turns causes the surrounding powder from outside the scanning track to be dragged into the melt-pool, which in turn enlarges and acquires an irregular shape. For this reason, in the single tracks of Figure 4-17(a), two instability zones, at low and at high energy density, can be observed for the low power case.

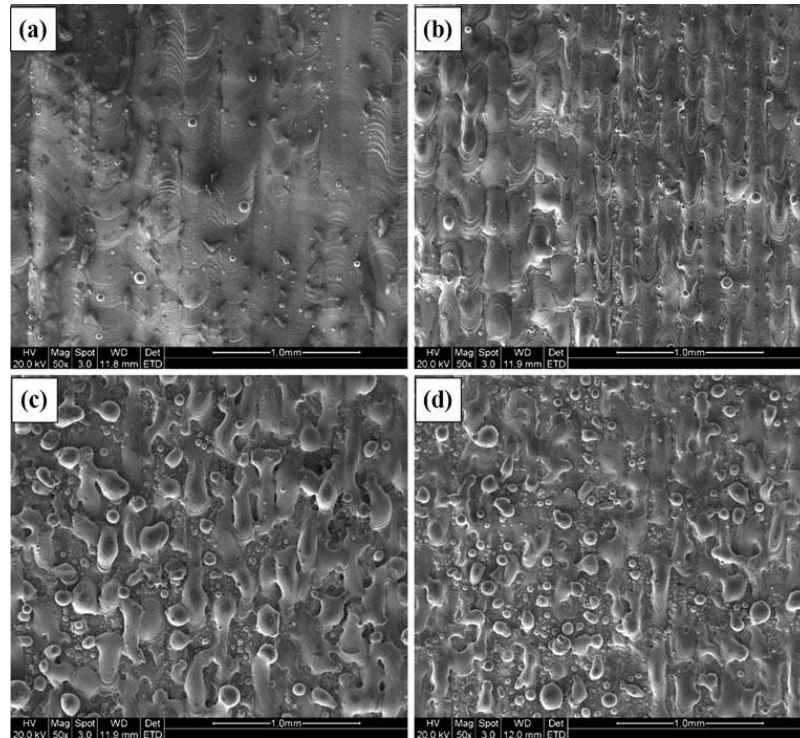


Figure 4-16: Effect of different scan speeds on ball formation in multi-layer SLM 316-L stainless steel. Speeds are 50mm/s (a), 400mm/s (b), 600mm/s (c), 800mm/s (d).

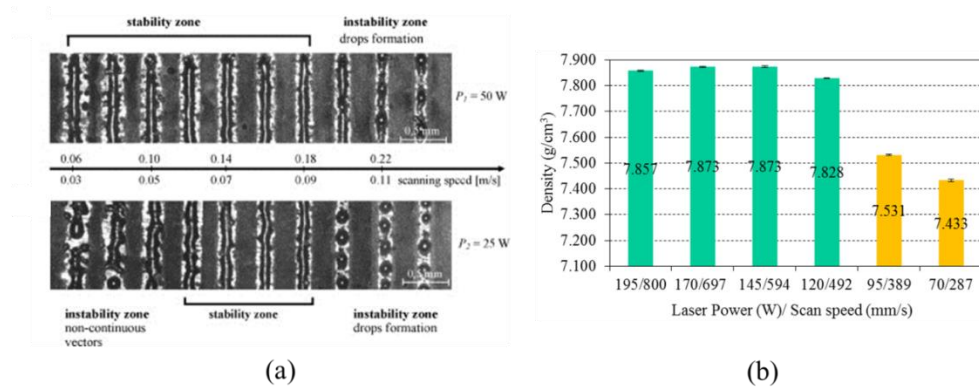


Figure 4-17: Effect of different laser power (P)-scan speed (v) combinations given the same linear energy density (P/v). The photographs in (a) show that on single tracks of Inco 904L, using a higher P allows to achieve a larger stability zone (Yadroitsev et al., 2007). The histograms in (b) show that using higher power when processing 17-4PH steel cubic samples allow to obtain higher parts densities (Gu et al., 2013).

4.5.3 Hatch Spacing

Hatch spacing corresponds to the distance between the centres of two parallel scanning tracks. Its role on densification has been studied extensively for

stainless steel in (Li et al., 2010) and (Yadroitsev et al., 2007). Both articles show that small hatch spacing promotes track overlapping and therefore decreases the amount of pores. Furthermore, the authors of (Yadroitsev et al., 2007) argue that a careful choice of the hatch spacing is important in order to avoid linear void structures associated with the powder denudation effects (described in Section 4.2). The effects of hatch spacing on the induced residual stresses is discussed in (Pohl et al., 2001), where the authors suggest that a smaller hatch spacing produces higher thermal gradients, which ultimately bring about greater residual stresses.

4.5.4 Layer Thickness

A larger layer-thickness value improves productivity, because a lower number of layers are spread on the build platform, but tends to decrease part density if the powder layer is too thick, since no sufficiently low scan speed or high laser power can be found to achieve high density. Furthermore, surface roughness is highly affected by the thickness of the layer, as more powder is available to create spatter. On the other hand, it has been shown that given the same final part density, a larger layer thickness can help reduce residual stress development in Ti-6Al-4V (Kruth et al., 2012). A reasonable compromise often chosen for the layer thickness value for many SLM materials is 30 μ m (Thijs et al., 2010, Kruth et al., 2012).

4.5.5 Laser Energy Density

A useful way of combining all of the above-mentioned parameters is the volumetric energy density, which describes the average laser energy applied to the powder bed per volume of material. For single tracks, the volumetric energy density E_s is:

$$E_s = \frac{P}{v \cdot t \cdot d} \quad (4.1)$$

where P is laser power, v is the scan speed, t is the layer thickness and d is the laser spot diameter.

For multiple-track, multiple-layer builds, a popular expression for the volumetric energy density (E_m) has been proposed in (Thijs et al., 2010) as:

$$E_m = \frac{P}{v \cdot h \cdot t} \quad (4.2)$$

where h is the hatch spacing. Other authors have suggested more advanced expressions for the volumetric energy density in multi-track and multi-layer scenarios, which aim at more precisely taking into account the heat accumulation occurring due to overlap (Di et al., 2012).

A recent contribution (Thomas et al., 2016) proposes a normalised process mapping methodology, where an optimal design window is identified within the space defined by two groups of dimensionless process variables, namely the hatch spacing and the single-track volumetric laser energy density. The authors have shown that by making dimensionless both the energy density using the alloys' density and thermal properties (thermal conductivity, heat capacity and melting temperature), and the hatch spacing using the beam size, a normalised equivalent energy density E_0^* can be derived by dividing the dimensionless quantities. This quantity has been shown to be proportional to the expression in (4.2). Importantly, a constant value of E_0^* will provide an equally energy efficient treatment during powder bed AM, irrespective of the individual processing parameters used, and thus represents a useful tool for process optimisation. Furthermore, the authors show that different E_0^* values lead to different melt-pool geometries and therefore to different

microstructures, which can ultimately produce very different functional behaviours (e.g., different tensile properties). The normalised process map proposed in (Thomas et al., 2016) is shown in Figure 4-18, where the points in the processing space were obtained from the data found in the literature for different alloys and AM powder-based processes. E_0^* isopleths and processability ranges are indicated in the graphics.

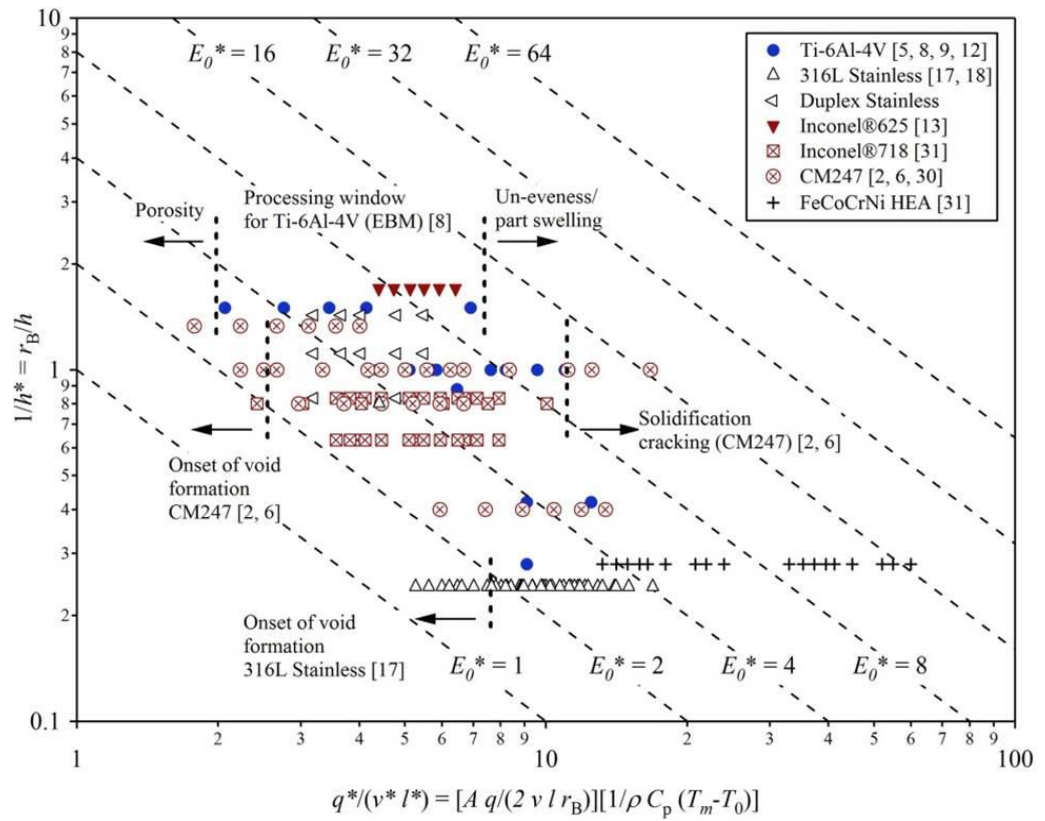
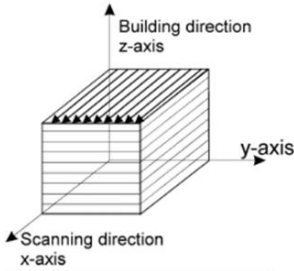
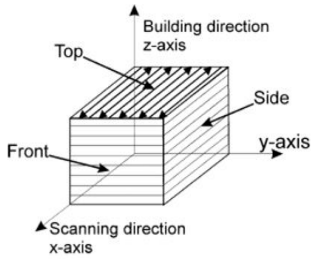
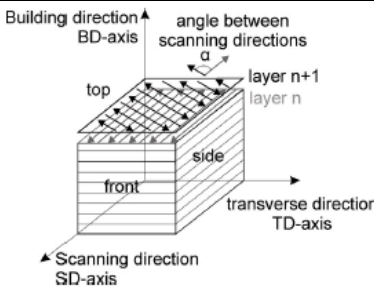
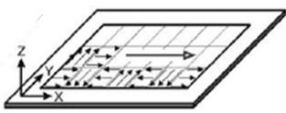
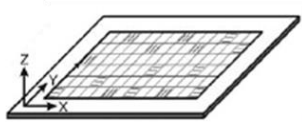
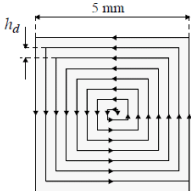
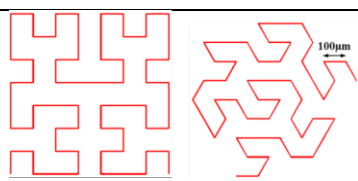


Figure 4-18: Normalised processing diagram for a range of alloys and AM systems, from (Thomas et al., 2016).

4.5.6 Laser Scan Strategy

The strategy employed for the creation of track patterns on the powder surface plays a paramount role in the formation of deformations induced by residual stresses. The most popular scanning patterns are shown in Table 4-1.

Table 4-1: list of some of the most popular scanning strategy categories employed in the literature

Scan Strategy		References
Uni-directional (in x- or y-direction)		(Nickel et al., 2001), (Kruth et al., 2010), (Kruth et al., 2004), (Parry et al., 2016), (Pohl et al., 2001), (Thijs et al., 2010)
Bi-directional (in x- or y-direction)		(Parry et al., 2016), (Kruth et al., 2010), (Thijs et al., 2010), (Jhabvala et al., 2010)
Changing-direction between layers;		(Kruth et al., 2010), (Thijs et al., 2013b), (Thijs et al., 2013a), (Pohl et al., 2001), (Thijs et al., 2010)
Sector-based, successive		(Kruth et al., 2004), (Pohl et al., 2001)
Sector-based, Least Heat Influence		(Kruth et al., 2010), (Thijs et al., 2013a), (Jhabvala et al., 2010)
Spiral		(Jhabvala et al., 2010), (Nickel et al., 2001), (Pohl et al., 2001)
Fractal		(Catchpole-Smith et al.)

In (Kruth et al., 2004), where a large variety among the most popular pattern strategies mentioned were investigated, the authors suggest that sector-based strategies should be preferred to decrease thermal stress-induced deformations. In particular, the successive sector strategy has the advantage of limiting the thermal gradients. More recently, numerical studies using a Ti-6Al-4V thermomechanical model concluded that short scan vectors reduce the residual stress introduced during the SLM process (Parry et al., 2016). More exotic strategies, such as spiral (Jhabvala et al., 2010, Nickel et al., 2001) and fractal patterns (Catchpole-Smith et al.) can potentially provide advantages in stabilising the thermal gradients in order to reduce the residual stresses, these can be hard to obtain for certain part geometries. In (Thijs et al., 2013a), it was shown that changing the scan strategy affects the crystallographic texture of AlSi10Mg. The authors showed that a strong fibre texture obtained by scanning every layer using unidirectional vectors could be changed into a weak cube texture when scan vectors rotated by 90° within and between layers were employed. The effect of the scan strategy on the morphological and crystallographic textures was also discussed in (Thijs et al., 2010) for Ti-6Al-4V, in (Thijs et al., 2013b) for tantalum and in (Vrancken et al., 2014) for β -Ti matrix composites. All authors report that the scan strategy affects both the direction of growth of the grains and their crystallographic texture.

4.6 Summary

In this chapter, the complex physical phenomena taking place in SLM have been presented, and their relationship with build quality and functional behaviour of the final part have been outlined. It has been shown that the stability of the surface-tension driven convective flows and metal evaporation

occurring inside the melt-pool determine the quality of the deposited bead and therefore affect defect formation (porosity and surface roughness). The local application of an intense heat-source develops steep thermal gradients, which result in the formation of residual stress in the solidified material. Residual stress is at the origin of part deformation and, in the most extreme cases crack formation. Weld solidification theory can be used to explain the solidification microstructures observed in SLM materials. As new tracks/layers are deposited, the previously solidified material is partially-remolten. This offers grains the opportunity to grow epitaxially hence forming columnar structures extending for up to several layer thicknesses along the build direction. This is accompanied by a crystallographic texture, since metallic unit cells are characterised by an “easy” axis of solidification. Finally, the effect of the main SLM process parameters on the quality of the build, residual stress development and microstructural evolution has been presented. It has been shown that within the optimal window of laser energy density, some process parameters can be adjusted for productivity without undermining the properties of the build.

5. Topological Design Optimisation Methods in Structural and Electromagnetic Applications

5.1 Introduction

In order to be able to exploit the design freedom of the AM technology, understanding Topology Optimisation (TO) is necessary. Therefore, a presentation of the TO approach to freeform design, as well as an overview of the most popular TO methodologies is provided first in this chapter. Then, we present the attempts of electrical machine design using TO found in the literature. In the last section of the chapter a brief list of guidelines regarding the specific application of TO to AM components is provided.

5.2 Introduction to Topology Optimisation

The flexibility of Finite Element Analysis (FEA) for simulating the multi-physical behaviour of complex structures has prompted the success of this methodology not only for the analysis of the performance of pre-existing components, but also as the fundamental tool for the creation of innovative designs. From an FEA solution, the derived quantities of interests can be fed into an optimisation algorithm. Therefore, the design process is formulated as a mathematical programming problem whereby a function – the objective – measuring the performance of the device is maximised/minimised while satisfying other requirements, in the form of constraint equations. The design formulation consists of an iterative process where the design variables are updated at each step until the system satisfies the desired performance. The design formulation aimed at optimising the distribution of material without any a-priori template describing the geometry of the system is referred to as

Topology Optimisation (TO). This is in contrast to size and shape optimization, where a fixed topological design template is chosen while only the dimensions and/or shapes of the boundaries are modified. This means that TO offers more design flexibility as the creation of an optimised structure can start from a region of empty space or from a block of solid material, since the creation of new boundaries is possible. TO is based on the subdivision of the structure into cells (generally represented by the mesh elements in the FEA model), and one or more design variables are then assigned to each of them. Therefore, with several orders of magnitude more variables than shape optimisation, computational limits will usually lead to preferring a relatively coarse grid. However, once a new design has been found by the TO method, this can be refined using shape optimisation (Bendsoe and Sigmund, 2013).

5.3 Problem Statement for Stiffness Optimisation

Typically the numerical TO problem is solved starting from a meshed model. Therefore, this discussion will consider the discrete formulation of TO, which can be stated as follows:

$$\min_{\mathbf{x}} f(\mathbf{x}, \mathbf{u}(\mathbf{x})) . \quad (5.1)$$

Subject to:

$$- \quad x_e = 0 \text{ or } 1 \quad \text{for } (e = 1, \dots, N), \quad (5.2)$$

$$- \quad V_0 - \sum_{e=1}^N \rho_e v_e = 0 \quad (5.3)$$

$$- \quad \mathbf{f} = \mathbf{K}\mathbf{u} \quad (5.4)$$

$$- \quad \text{Other design constraints (e.g., maximum displacement).}$$

In (5.1) f is the objective function, $\mathbf{x} = [x_1, \dots, x_e, \dots, x_N]$ is the design variable vector, with N the total number of elements inside the design domain. Each elemental design variable x_e represents an element's relative density and can therefore assume the value of 0 for air or 1 for the solid material (i.e., iron). The variable $\mathbf{u}(\mathbf{x})$ is the solution of the equilibrium equation (the nodal displacement matrix in structural problems), which is an implicit function of $\mathbf{x}$. The constraints regard the material's relative density (equation (5.2)), the volume of the solid elements, which should meet a prescribed volume V_0 (equation (5.3)), as well as the equilibrium equation (5.4), where $\mathbf{f}$ is the applied load vector and $\mathbf{K}$ the stiffness matrix. Other application-specific performance requirements can be included in the list of constraints, such as the maximum Von Mises stress, maximum displacement, etc.

In structural TO problems, the goal normally is the maximisation of the structural stiffness (Huang and Xie, 2010), which corresponds to minimising the strain energy. Therefore, the objective f is often represented by the structure's strain energy C , which in finite element form reads:

$$C = \frac{1}{2} \mathbf{f}^T \mathbf{u} = \sum_{e=1}^N \frac{1}{2} \mathbf{u}_e^T \mathbf{K}_e \mathbf{u}_e \quad (5.5)$$

where $\mathbf{K}_e$ is the elemental stiffness matrix.

5.4 Topology Optimisation Methods

The main issue encountered when considering the TO problem as stated in (5.1)-(5.4) is the non-closedness of the solution space, which results in the non-existence of a global solution. This is due to the fact that, given the same prescribed volume, smaller and smaller holes can be introduced in the design to lower the objective function. In FEA-based TO implementation, the closed-

ness issue results in the non-uniqueness of the solution, since a different solution can be found for different finite element mesh sizes due to the tendency of the physically-optimal solution to converge at the microstructural scale (Sigmund and Petersson, 1998). The “visible” consequences of the non-uniqueness of the TO solution are the mesh-dependence and checkerboard issues. Furthermore, the TO problem is in general non-convex, thus different optimisation algorithms normally lead to different local optima. In order to overcome the mentioned issues, a number of methods and algorithms have been proposed in the literature.

The most representative methods proposed over the past few decades for solving the compliance problem include the homogenisation method (Bendsøe and Kikuchi, 1988), the density methods, especially the Solid Isotropic Material with Penalisation (SIMP) method (Rozvany et al., 1992), the Bi-directional Evolutionary Structural Optimisation (BESO) method (Huang and Xie, 2010), the level set method (Wang et al., 2003), and the topological gradient or bubble method (Céa et al., 2000). Stochastic approaches, such as Genetic Algorithms, have also been proposed for structural (and electromagnetic) TO, but they tend to be considered inferior to sensitivity-based methods because of the very high number of function evaluations they require, which make these methods unsuitable for dealing with real-world problems that are generally described by a large number of design variables (Sigmund, 2011).

Some of the most popular sensitivity-based methods are discussed in the following sub-sections, with special regard to the case of compliance minimisation. While the homogenization and topological gradient methods are

only briefly discussed, the SIMP and BESO methods are more thoroughly presented, because of their success in a range of structural and multi-functional applications. An extended version of the BESO methodology will be employed in this research for tackling the multi-objective electromechanical design problem.

5.5 Structural TO Using Material Interpolation Scheme

In order to be able to use gradient information, continuous design variables (i.e., $0 \leq x_e \leq 1$) must be employed by means of material interpolation schemes. In the original Solid Isotropic Material with Penalisation (SIMP) method (Rozvany et al., 1992) the Young's modulus E of the material is interpolated as a function of the element density using a penalisation exponent as follows:

$$E(\rho_e) = E_0 \cdot x_e^p \quad (5.6)$$

where E_0 is the Young's modulus of the solid material (e.g., iron) and p the penalty exponent. The relative elemental material density x_e is allowed to change continuously from 0 to 1. The Poisson's ratio is assumed to be independent of the material's density, while the elemental stiffness matrix $\mathbf{K}_e$ depends on the design variable as:

$$\mathbf{K}_e(\rho_e) = \mathbf{K}_e^0 \cdot x_e^p \quad (5.7)$$

where $\mathbf{K}_e^0$ is the elemental stiffness of the solid material.

If we set the compliance (equation (5.5)) to be the objective function, its derivative with respect to the design variable is:

$$\frac{df}{dx_e} = \frac{1}{2} \frac{d(\mathbf{f}^T \mathbf{u})}{dx_e} = \frac{1}{2} \left(\frac{d\mathbf{f}^T}{dx_e} \mathbf{u} + \mathbf{f}^T \frac{d\mathbf{u}}{dx_e} \right) \quad (5.8)$$

where $\frac{d\mathbf{f}^T}{dx_e}\mathbf{u} = \mathbf{u}^T \frac{d\mathbf{K}}{dx_e}\mathbf{u}$, while the term $\mathbf{f}^T \frac{d\mathbf{u}}{dx_e}$ cannot be solved analytically, owing to the implicit dependence of $\mathbf{u}$ on the design variable. Therefore, the equilibrium equality constraint in (5.4) is added to the objective function in the form of a Lagrange term by introducing the Lagrange multiplier $\boldsymbol{\lambda}$. The resulting Lagrange function reads:

$$f = \mathbf{f}^T \mathbf{u} + \boldsymbol{\lambda}^T (\mathbf{f} - \mathbf{K}\mathbf{u}) \quad (5.9)$$

Since the second term in the above equation always equals zero for equilibrium to be satisfied, $\boldsymbol{\lambda}$ can be chosen freely. Derivation of the above function with respect to the design variable leads to the sensitivity expression:

$$\frac{df}{dx_e} = \frac{1}{2} \frac{d\mathbf{f}^T}{dx_e} \mathbf{u} + \frac{1}{2} \mathbf{f}^T \frac{d\mathbf{u}}{dx_e} + \boldsymbol{\lambda}^T \left[\frac{d\mathbf{f}}{dx_e} - \frac{d\mathbf{K}}{dx_e} \mathbf{u} - \mathbf{K} \frac{d\mathbf{u}}{dx_e} \right] = \quad (5.10)$$

$$= \frac{1}{2} \frac{d\mathbf{f}^T}{dx_e} \mathbf{u} + \left(\frac{1}{2} \mathbf{f}^T - \boldsymbol{\lambda}^T \mathbf{K} \right) \frac{d\mathbf{u}}{dx_e} + \boldsymbol{\lambda}^T \left[\frac{d\mathbf{f}}{dx_e} - \frac{d\mathbf{K}}{dx_e} \mathbf{u} \right] \quad (5.11)$$

By setting $\left(\frac{1}{2} \mathbf{f}^T - \boldsymbol{\lambda}^T \mathbf{K} \right) = 0$ we find that the value of the Lagrange multiplier that makes the unknown $\frac{d\mathbf{u}}{dx_e}$ go to zero is $\boldsymbol{\lambda} = \frac{1}{2} \mathbf{u}$. Therefore, the general sensitivity of the strain energy with respect to the design variable reads:

$$\frac{df}{dx_e} = \frac{d\mathbf{f}}{dx_e} - \frac{1}{2} \mathbf{u}^T \frac{d\mathbf{K}}{dx_e} \mathbf{u} \quad (5.12)$$

In problems where the load $\mathbf{f}$ is design-independent we have $\frac{d\mathbf{f}}{dx_e} = 0$. This leads to the simplified sensitivity expression:

$$\frac{df}{dx_e} = -\frac{1}{2} \mathbf{u}^T \frac{d\mathbf{K}}{dx_e} \mathbf{u} = -\frac{1}{2} \mathbf{u}_e^T \frac{d\mathbf{K}_e}{dx_e} \mathbf{u}_e \quad (5.13)$$

By applying the SIMP interpolation scheme in (5.7), we obtain the explicit formula for the sensitivity:

$$\frac{df}{dx_e} = -\frac{1}{2} p x_e^{p-1} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e \quad (5.14)$$

Therefore, the SIMP approach leads to a smooth, differentiable problem that can be solved efficiently by well-established, gradient-based updating methods such as any of the Optimality Criteria methods, the Method of Moving Asymptotes (MMA) (Sigmund and Maute, 2013) or by other general-purpose optimization algorithms, such as Sequential Linear Programming (SLP) (Lee et al., 2010, Arora, 2004). The latter immediately allows for the inclusion of additional global constraints as required by the specific problem formulation (Lee et al., 2010).

In the general case of design-dependent load, the compliance is non-monotonous. More importantly, it has been demonstrated that using the exponential interpolation scheme in the case of body loads makes it impossible to obtain near solid-void designs (Huang and Xie, 2010). Therefore, a more useful interpolation scheme has been proposed by Stolpe and Svanberg in 2001 (Huang and Xie, 2010):

$$\mathbf{K}_e(x_e) = \mathbf{K}_e^0 \cdot \frac{x_e}{1 + p(1 - x_e)} \quad (5.15)$$

We then consider that a generic body load is applied to each element of the design space, i.e.:

$$\mathbf{f}_e = V_e x_e \rho_0 \bar{\mathbf{f}}_e \quad (5.16)$$

where ρ_0 is the density of the full material and $\bar{\mathbf{f}}_e$ is a vector whose elemental values depend on the specific nature of the body load (usually, either gravitational or centripetal). The final strain energy sensitivity for the body-load case reads:

$$\frac{df}{dx_e} = V_e \rho_0 \bar{\mathbf{f}}_e - \frac{1}{2} \frac{1+p}{[1+p(1-x_e)]^2} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e \quad (5.17)$$

which can be shown to provide near-binary designs.

5.6 The BESO Method for Structural TO

The Bi-directional Evolutionary Structural Optimisation method was introduced in (Yang et al., 1999) for stiffness optimisation using discrete design variables, i.e., the relative density x_e can only assume two bound values, corresponding to two bound materials, normally represented by solid material and air. The following review is mainly based on the book by Huang and Xie (Huang and Xie, 2010), which represents a comprehensive reference for BESO-based TO.

The basic BESO concept is that at each iteration step the elements are sorted based on the values of an elemental “sensitivity number”. The low-ranking solid elements are then switched to air, while the high-ranking air elements are switched to solid. At each iteration i the volume V_i can increase, decrease, or remain constant, depending on whether it is smaller, larger or equal to the prescribed volume V_0 , respectively. The procedure is repeated iteratively until convergence. The sensitivity number used to guide element addition/removal can be derived from the strain energy sensitivity formula in the case of design-independent load (equation (5.14)) as:

$$\alpha_e = -\frac{1}{p} \frac{df}{dx_e} = \begin{cases} \frac{1}{2} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e, & x_e = 1 \\ \frac{1}{2} x_{min}^{p-1} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e, & x_e = x_{min} \end{cases} \quad (5.18)$$

where x_{min} is the relative density assigned to soft elements (i.e., air), normally set to 0.001. Therefore, the sensitivity of the solid elements corresponds to the

elemental strain energy, while the sensitivity of air depends on the chosen penalty exponent. It is important to note that the elemental strain energy can be directly obtained from FEA. This represents an important advantage of the BESO method.

By setting the penalty exponent to infinity, the above sensitivity number becomes:

$$\alpha_e = \begin{cases} \frac{1}{2} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e, & x_e = 1 \\ 0, & x_e = x_{min} \end{cases} \quad (5.19)$$

The above expression corresponds to the “hard-kill” version of BESO, whereby the elements made of air are effectively removed from the design space (i.e., their relative density is set to 0). On the other hand, the sensitivity number definition in (5.18) corresponds to the “soft-kill” BESO method.

In the case of body loads, the soft-kill BESO sensitivity number can be derived from the sensitivity formula in (5.17):

$$\alpha_e = \begin{cases} -\frac{v_e \rho_0 \bar{\mathbf{f}}_e}{p+1} + \frac{1}{2} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e, & x_e = 1 \\ -\frac{v_e \rho_0 \bar{\mathbf{f}}_e}{p+1} + \frac{1}{2[1+p(1-x_{min})]^2} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e, & x_e = x_{min} \end{cases} \quad (5.20)$$

When the penalty exponent tends to infinity, the hard-kill sensitivity number is obtained:

$$\alpha_e = \begin{cases} \frac{1}{2} \mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e, & x_e = 1 \\ 0, & x_e = x_{min} \end{cases} \quad (5.21)$$

Importantly, the comparison between the above equation and equation (5.19) shows that the hard-kill BESO method for design-dependent and design-independent loads is exactly the same. However, it should also be noted that while the hard-kill and soft-kill methods lead to the same qualitative results in

the case of design-independent loads, this is not the case for design-dependent loads, since the value of the penalty factor p affects the monotonicity of the sensitivity number of the solid elements with respect to the design variable, and therefore their relative ranking.

The raw sensitivity numbers need to be processed in order to avoid the issues of mesh-dependence and checkerboard, as previously described for the SIMP method. Therefore, filter schemes have been proposed for the BESO method. The filter presented here is that proposed by Zuo and Xie in (Zuo and Xie, 2015):

$$\alpha_e = \frac{1}{\sum_{l=1}^N (r_{min} - r_{el})} \sum_{k=1}^N (r_{min} - r_{el}) \alpha_k \quad (5.22)$$

where r_{min} is the filter radius and $r_{el} < r_{min}$ is the distance between the centres of element e and l . In order to facilitate convergence, the filtered sensitivity number is then averaged with its historical information in order to stabilise the BESO process. A simple stabilisation scheme can be written as:

$$\alpha_e = w_{i-1} \alpha_e^{i-1} + w_i \alpha_e^i \quad (5.23)$$

where w_{i-1} is the weight assigned to the historical information α_e^{i-1} , and w_i is the weight assigned to the filtered sensitivity number α_e^i at the present iteration step.

At each iteration step i , the sensitivity number threshold α_{th} for element addition/removal is determined based on the target volume at the next iteration step $V_{i+1} = V_i(1 \pm ER)$, with ER being the evolutionary ratio, which is normally comprised between 1% and 2%. In order to ensure that not too many elements are added in a single iteration, if the ratio between the number of

added element and the total element number is greater than a prescribed maximum addition ratio AR_{max} (normally comprised between 1% and 5%), α_{th} is then set as the sensitivity number of the element ranked just below position $AR_{max} \cdot N$.

The BESO method presents a few important advantages over other methods when the TO problem is limited to the minimisation of the strain energy. Importantly, BESO is extremely straight-forward to implement, since it does not require any mathematical programming tools. Furthermore, all of the information required for carrying out the optimisation, i.e., the elemental strain energies, is obtained directly from the FEA solution. Finally, the extension of the method to the case of nonlinear material properties and multiple loads is straight-forward. In the nonlinear structural case, the sensitivity number is normally derived from the complementary work C_w in place of the total strain energy, in order to avoid instabilities at loads smaller than the prescribed one. The objective function to be minimised therefore reads:

$$f = C_w = \lim_{n \rightarrow \infty} \frac{1}{2} \sum_{m=1}^M (\mathbf{f}_m^T - \mathbf{f}_{m-1}^T) (\mathbf{u}_m + \mathbf{u}_{m-1}) \quad (5.24)$$

where $\mathbf{u}_m$ is the displacement vector after the m^{th} quasi-steady state load increment $\mathbf{f}_m^T = \mathbf{f}_{m-1}^T + \frac{\mathbf{f}}{M}$, and M is the total number of load increments. The sensitivity of the above objective function can be shown to be:

$$\frac{df}{dx_e} = C_w = \lim_{n \rightarrow \infty} \frac{1}{2} \sum_{m=1}^M (\mathbf{u}_m^T - \mathbf{u}_{m-1}^T) \left(\frac{d\mathbf{i}_m}{dx_e} + \frac{d\mathbf{i}_{m-1}}{dx_e} \right) \quad (5.25)$$

where $\mathbf{i}_m$ is the internal force vector at the m^{th} load increment. By using the exponential interpolation scheme in (5.6) for expressing $\mathbf{i}_m$, the sensitivity number can be derived as:

$$\begin{aligned}\alpha_e &= -\frac{1}{p} \frac{df}{dx_e} = x_e^{p-1} \lim_{n \rightarrow \infty} \frac{1}{2} \sum_{m=1}^M (\mathbf{u}_m^T - \mathbf{u}_{m-1}^T) (\mathbf{i}_{0,m} + \mathbf{i}_{0,m-1}) \\ &= x_e^{p-1} \lim_{n \rightarrow \infty} \sum_{m=1}^M (\mathbf{E}_m^e - \mathbf{E}_{m-1}^e) = x_e^{p-1} E_I^e\end{aligned}\quad (5.26)$$

where $\mathbf{i}_{0,m}$ is the internal force vector for solid element at m^{th} increment and E^e is the elemental total strain energy.

In the multiple loads scenario, the soft-kill BESO elemental sensitivity number associated with the K loads can be expressed as a weighted sum of the elemental strain energies due to the single loads, i.e.:

$$\alpha_e = -\frac{1}{p} \frac{df}{dx_e} = \begin{cases} \frac{1}{2} \sum_{k=1}^K \omega_k (\mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e)_k, & x_e = 1 \\ \frac{1}{2} x_{min}^{p-1} \sum_{k=1}^K \omega_k (\mathbf{u}_e^T \mathbf{K}_e^0 \mathbf{u}_e)_k, & x_e = x_{min} \end{cases} \quad (5.27)$$

where ω_k is the weight associated with the k^{th} load.

5.7 Other Methods

5.7.1 The Homogenization Method

The homogenisation method is based on the introduction of micro-holes of various shapes (often referred to as “microstructures”), which repeat periodically throughout an object. The seminal work by Bendsoe and Kikuchi (Bendsøe and Kikuchi, 1988) employed rectangular cells with rectangular holes (refer to Figure 5-1(a)), which are described by their size (length a and

width b) and orientation parameters. The advantage of using rectangular unit cells with rectangular holes is the possibility to represent fully solid ($a=b=0$) and fully void regions ($a=b=1$) (Suzuki and Kikuchi, 1991). The homogenization method optimises the geometrical parameters describing the microstructure based on the loading conditions of the problem at hand, and then computes, for each finite element in the model, an equivalent, “homogenized” elasticity tensor corresponding to the parameterised microstructure. The homogenised elasticity tensors are updated according to the optimality criteria for the objective function (Bendsøe and Kikuchi, 1988).

The above description shows that the homogenization method solves the TO problem by “relaxing” it to the form of a sizing problem. While the main merit of this approach is that the solution provides mathematical bounds to the theoretical performance of a structure, a major drawback of the method is that it leads to the formation of porous topologies, i.e., the elements are characterised by an effective relative density $0 < x_e < 1$. This, despite making physical sense since the topologies can be interpreted as made of composite materials, raises the issue of how to manufacture the topologies. In this respect, a number of solutions have been proposed, such as lumping (Bendsøe and Kikuchi, 1988). A second issue of the homogenisation method regards the high number of parameters describing the microstructures, i.e., the hole sizes and orientation, especially in the case of 2D models characterised by fine meshes and 3D models. In this regard, the use of multi-ranked layered microstructures has been proposed, a schematic representation of which is shown in Figure 5-1(b) for a general 2D problem. While the constitutive tensor for the perforated microstructure must be determined numerically, thus increasing the

computing time, the tensor can be established analytically for the layered microstructures (Eschenauer and Olhoff, 2001). Owing to the above mentioned limitations, the homogenization method is of importance because its solution provides the mathematical bounds of a structure's performance (Huang and Xie, 2010).

In the mid 90's Sigmund proposed the inverse homogenization method (Sigmund, 1994), whereby rather than calculating global properties based on the layout of a parameterised unit cell, appropriate unit cell architectures were derived from scratch using TO. This is achieved by minimising the differences between the homogenized global material properties and their target values by means of standard TO techniques, e.g., SIMP.

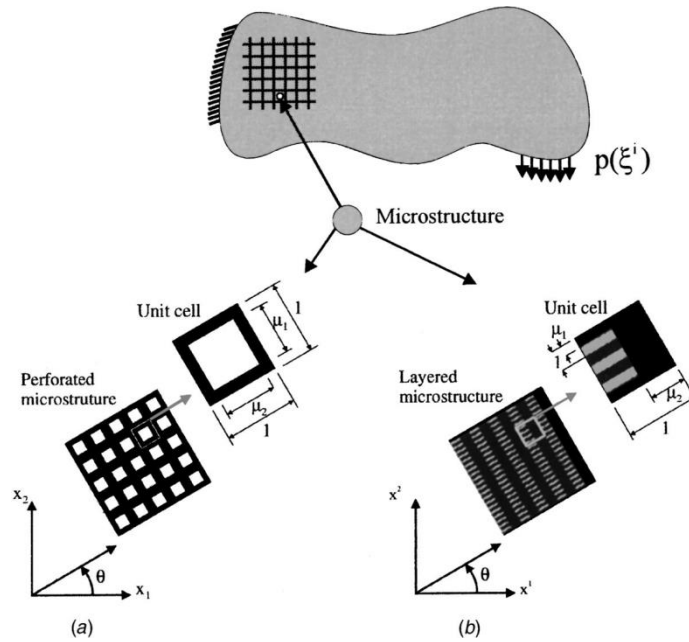


Figure 5-1: Representation of a design domain represented as a homogenized continuum with microstructures. The latter can be characterised by repeating perforated unit cells (a) or layered composite materials (b). From (Eschenauer and Olhoff, 2001).

5.7.2 *The Bubble Method (Topological Gradient)*

The topological derivative for topology and shape design was first introduced in (Eschenauer et al., 1994) and predicts the effect (sensitivity) of introducing an infinitesimal hole in the design domain on the objective function value. While topological derivatives are derived for infinitesimal holes, in numerical implementations the holes have the size of the single finite elements (Kim et al., 2009). While the mathematics involved in the derivation of the topological derivatives is quite complex, it has been argued that for compliance minimisation problems the value of the topological derivatives correspond to that of the standard derivatives as used in the SIMP method (Sigmund and Maute, 2013). However, according to (Céa et al., 2000) the topological gradient represents a better tool for solving structural optimisation problems that do not involve the minimisation of the compliance, but rather, for example, the minimisation of the maximum Von Mises stress. Furthermore, since the level-set of the topological gradient field can be used to guide the enlargement of the created holes, this method can be effectively used for carrying out shape and topological optimisation in the same process (Céa et al., 2000). Examples of the use of the topological gradient can be found in electromagnetics, as will be shown in Section 5.8.

5.8 Topology Optimisation in Electrical Machinery Design

The methods proposed to tackle the TO electromagnetic problem for electromechanical device design can be divided into two main classes: the ones using discrete design variables (Kim et al., 2009, Okamoto et al., 2006), which consider each point in space (i.e., each finite element) as either void or filled with ferromagnetic material, and the density methods (Dyck and Lowther,

1996, Lee and Kikuchi, 2010). The homogenisation method has been investigated in one representative contribution (Yoo et al., 2000). The level-set method has also been employed in only one contribution, for the 2D torque density optimisation of a SyR rotor (Kim and Park, 2010).

Among the methods employing discrete variable approaches, the ON/OFF method first proposed by (Okamoto et al., 2006) has proven attractive for the solution of relatively large, real-world TO problems thanks to its computational efficiency. Similarly to BESO, the ON/OFF method ranks the elements in the design space according to their sensitivity value and their permeability is set to that of air (low ranking elements) or that of iron (high ranking elements). The heuristics have proven very computationally-efficient for optimizing the field flux density in two regions of a large 3D model (around 200,000 brick elements) of a magnetic recording system's writing head with nonlinear magnetic properties (Okamoto et al., 2006). Convergence was achieved by the authors after 288 FEA calculations, corresponding to 81.6 hours. The initial and final designs of the device are shown in Figure 4-8. The main difference between the ON/OFF method by (Okamoto et al., 2006) and the BESO method lies in the implementation of the smoothing scheme, which is referred to as "topology smoother" in the ON/OFF method. According to the topology smoother the density of an element is changed only if the majority of the neighbouring elements (four out of six in the case of brick elements) have a different density. In this respect, the topology smoother appears less robust than the classical BESO/SIMP filtering scheme proposed in (Sigmund and Petersson, 1998).

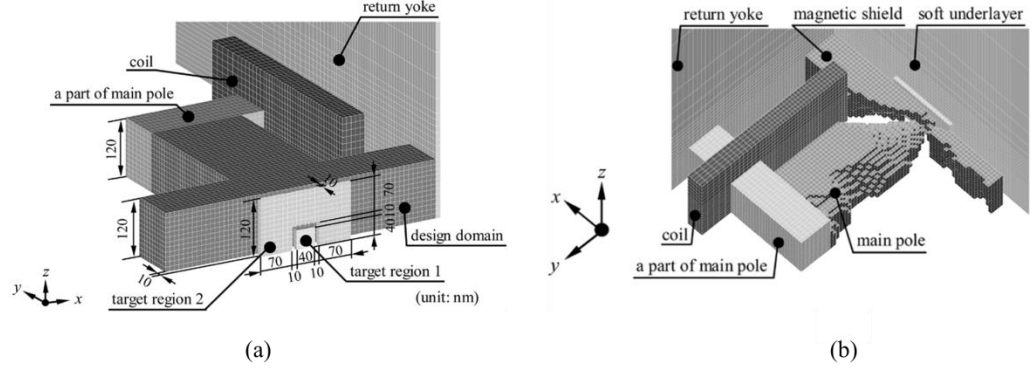


Figure 5-2: Recording system's writing head electromagnetic TO using the ON/OFF method: original design (a) and optimised design (b). Adapted from (Okamoto et al., 2006).

Because of the computational efficiency of the ON/OFF method, Kim and colleagues employed it (Kim et al., 2009) to minimise the torque ripple of a BLDC machine through the TO of the stator teeth. As shown in Figure 5-3(a), the authors have limited the design space to the stator teeth. After the TO step the resulting tooth topology, shown in Figure 5-3(b) was post-processed in CAD to smooth the boundaries, thus achieving the final design (Figure 5-3(c)). The objective function chosen by the authors consisted of a linear combination of the co-energy values at different stator/rotor orientations, according to the expression:

$$f = \sum_k W_{co}^k - W_{co}^o, \quad (5.28)$$

where W_{co}^k is the magnetostatic co-energy at rotor position k , and W_{co}^o is a fixed co-energy value, chosen by the authors to be equal to the target average co-energy over all rotor positions. The sensitivity information used by the authors to guide the element ranking procedure was derived using the topological gradient, which requires the solution of two FEA per iteration (one for the primal and one for the adjoint system). Despite not being able to improve the motor performance, Kim's work represents the only example to employ

electromagnetic TO with the aim of exploiting the design freedom offered by a novel manufacturing process, namely powder metallurgy.

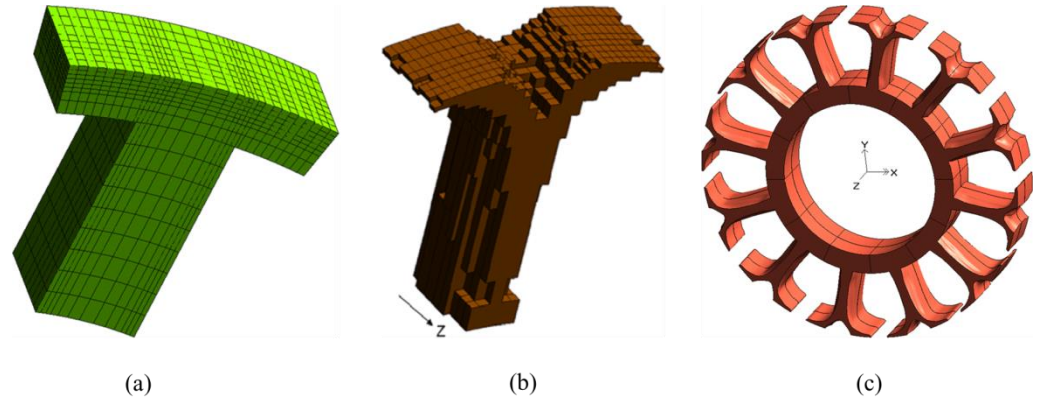


Figure 5-3: Brushless DC stator teeth TO-based design proposed in (Kim et al., 2009). Original tooth design and mesh (a), final topology after TO procedure (b), and full optimised stator after boundary smoothing (c).

A significant contribution employing the SIMP method is that proposed in (Lee et al., 2010), where the rotor topology of an SRM motor with 4-pole rotor and 6-pole stator was optimised both in 2D (Figure 5-4) and 3D (Figure 6-3) with the aim of achieving maximum average torque with minimum torque ripple, while controlling copper losses and motor weight. The interpolation of reluctivity used in (Lee et al., 2010) is based on that of permeability (i.e., the reciprocal of reluctivity) first introduced by (Byun et al., 1999), which reads:

$$\mu(x_e) = \mu_0 \cdot (\mu_s \cdot x_e^p + 1 - x_e^p) \quad (5.29)$$

The general purpose optimisation algorithm SLP was used by the authors in order to include the additional global constraints. For the 2D optimisation, both linear and non-linear relationships between magnetic flux density and magnetic field intensity were considered.

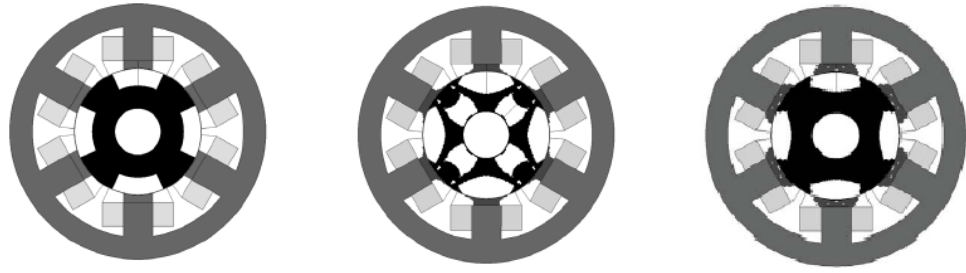


Figure 5-4: Cross-sectional view of a 6/4 SR motor optimised using SIMP. The starting design is on the left, while the optimised designs in the case of linear and non-linear material properties are centre and right, respectively. Images adapted from (Lee et al., 2010).

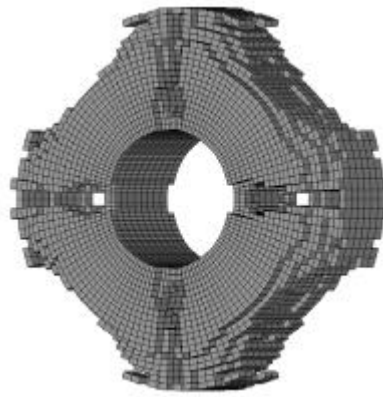


Figure 5-5: SR rotor 3D design after TO using a density method. Adapted from (Lee et al., 2010).

5.9 TO for Additive Manufacturing

The flexibility of AM to fabricate complex 3D shapes is suitable for the unconventional designs issued by TO and therefore has the potential to fully exploit them. The adaptation of the TO process for part production using AM is a topic of ongoing research (A. Aremu et al., 2013, D. Brackett et al., 2011). In particular, the issues investigated regard the practical difficulties/considerations encountered when the TO process is implemented for AM, including:

- Making sure that the mesh size allows for the representation of the optimal topology without becoming computationally prohibitive. In this respect the use of smart mesh refinement is recommended for large models. Among the proposed mesh refinement strategies, octree

decomposition has been shown to prove computationally efficient in 3D problems (Panesar et al., 2017);

- The promotion of self-supporting designs in order to limit the number of supporting structures, whose removal from the part often require lengthy manual post-processing and waste of material. This involves taking actions such as constraining the orientation of the overhangs, usually to angles $>45^\circ$ with respect to the horizontal plane;
- Part preparation for post-processing requirements. The final design should ensure the accessibility of regions in need of post-processing.

Although the issues of TO for AM have not been explored specifically for electromagnetic devices as of yet, it appears that most of the considerations reported above are applicable to this sector as well.

5.10 Summary

In this chapter some of the TO methods most commonly used for strain energy minimisation in structural design have been presented and respective advantages and drawbacks outlined. These methods are: Homogenization, Topological Gradient, SIMP and BESO. For the SIMP and BESO methods, which are the main representatives of continuous and discrete TO approaches, respectively, the fundamental expressions for the sensitivity analysis have been derived. A review of the literature in field of TO for electrical machinery design has also been presented. Importantly, it has been shown that both SIMP and an alternative version of BESO, namely the ON/OFF method, were successfully applied to electromagnetic problems in the literature. The SIMP method is particularly attractive due to its flexibility, while BESO stands out for its simplicity and efficiency.

6. Experimental Methods for the Selective Laser Melting of Silicon Iron and Its Characterisation

6.1 Introduction

This chapter describes the experimental methodology adopted in this research for the manufacture and characterisation of high-silicon steel samples produced by Selective Laser Melting (SLM). A description of the SLM system used, raw material powder and produced sample types is provided first. Then, the strategies employed for material processing (SLM parameter settings) and post-processing (heat-treatment temperatures and times) are outlined. Finally, the steps performed to prepare the samples for characterisation, as well as the experimental techniques used for the characterisation of the metallurgical, mechanical and magnetic properties of the samples are described.

6.2 Description of Selective Laser Melting System

Manufacture of the Fe-6.9%Si samples was carried out using a Realizer SLM50 desktop machine. The technical specifications of this system are listed in Table 6-1. The SLM50 is equipped with a continuous Yb-fibre laser characterised by a maximum nominal power of 100W, wavelength of 1070 nm and minimum laser beam spot size of 30 μm . All of the builds took place on a mild steel platform and under a protective argon atmosphere with an oxygen level below 0.4%. The powder levelling system consists of a double bladed level bar. The level bar wipes the powder bed twice in order to improve the levelling of the layer of powder. The building platform temperature can be set to 200°C to maintain the part at an elevated temperature so as to reduce thermal expansion gradients.

Table 6-1: Some technical details of the Realizer SLM50

SLM50 Technical Specifications	
Build platform Diameter (mm)	70
Maximum Build Height (mm)	40
Layer Thickness (um)	20-100
Power Supply	230 V, 16 A
Weight (Kg)	80

The Realizer SLM50 comes with proprietary control software, which can be used by the user to define, for each layer, the process parameters, namely: the laser power, the hatch spacing, the laser scan speed, the layer thickness and the laser focus position with respect to the level of the powder bed surface. The last parameter determines the size of the beam on the powder bed.

6.3 High-Si Steel Powder

Gas-atomised powder of pre-alloyed high-silicon steel with approximately 6.9% wt. Si (refer to Table 6-2 for a list of the alloying elements) was provided by LPW Technology Ltd., Runcorn, UK. The selected powder presents a Si content that is slightly higher than the ideal 6.5% recommended for electromechanical applications. While the additional Si content is expected to cause a decrease in both magnetisation at saturation and ductility, as previously discussed in Section 3.3, it determines an increase in electrical resistivity, which helps to limit the dynamic losses. This is deemed necessary in light of the final goal of the present research, which is to build an electrical machine rotor core as a solid component rather than in laminated form. Furthermore,

the specific powder was found readily available on the market, which represents an important advantage in terms of cost and lead time. According to the supplier the particle size range is 15 μm to 45 μm . Powder particles size distribution was also assessed using a particle size analyser. The shape of the powder particles was determined using a Scanning Electron Microscope (SEM). Powder particles size distribution was also assessed using a particle size analyser by Malvern Instruments Ltd. (Malvern, UK). The powder was dried at 60 °C for 120 min to remove humidity and improve flowability before processing using Selective Laser Melting.

Table 6-2: Alloying elements concentration range of the employed high-silicon steel powders according to supplier LPW Technologies Ltd.

	Element				
	<i>Si</i>	<i>Mn</i>	<i>C</i>	<i>P</i>	<i>O</i>
Concentration (%wt.)	6.9	0.05	0.01	0.009	0.0065

6.4 Sample Types

Two sets of test samples were produced:

- Standard cubic samples of dimensions 5x5x5 mm were created to analyse porosity, the development of cracks and microstructure;
- Ring samples of height $h=4.5$ mm and 2.5 mm, internal diameter $d=20$ mm and external diameter $D=28$ mm were manufactured for the magnetic measurements. These ring sample dimensions comply with the standards in (IEC, 2000) and (IEC, 2003).

For both sets of test samples, support structures were built to anchor the part to the build platform. This strategy allowed for an easier removal of the part from

the build platform. The orientation of the parts with respect to the horizontal and vertical coordinates is defined in Figure 6-1. In particular, the “lab” coordinates (“x”, “y” and “z” spatial axes) were aligned with the build (BD) and scan directions (SD_x and SD_y). The latter are introduced in the following subsection, where we discuss the laser scanning strategy and process parameters employed.

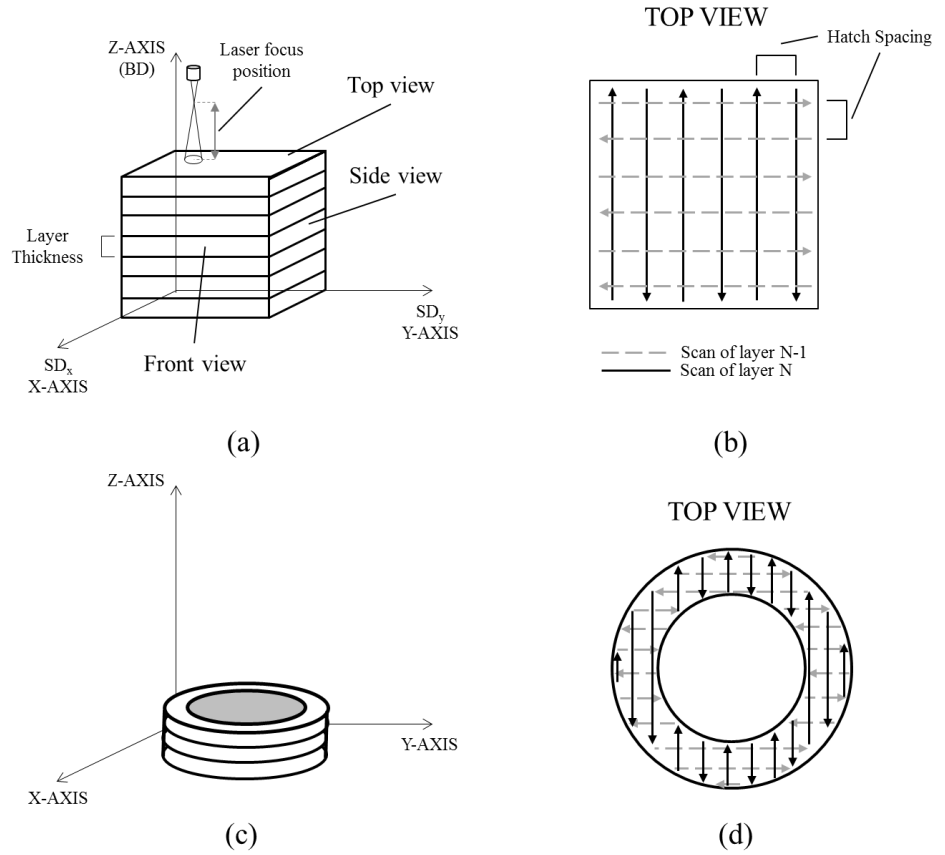


Figure 6-1: Overview of the SLM processing parameters and laser scan strategy. In (a) the different views of the cubic sample are indicated and the definitions of layer thickness and layer focus position are represented. Top view of the cubic sample with the bidirectional scan vectors is provided in (b). Three-dimensional representation and top view with scan vector representation for the ring sample are shown in (c) and (d), respectively.

As shown in Figure 6-1, the ring samples were built with their base parallel to the build platform. The advantages of choosing this orientation of the ring samples are twofold: the first advantage is that it allows us to minimise the amount of auxiliary structures and therefore the need for post-processing. The second and most important advantage arises from the crystallographic texture of the sample and its consequences on the measured magnetic properties of the samples. In the results presented in Section 7.5 it will be shown that SLM prompts the $\langle 001 \rangle$ crystallographic direction, which is also the “easy” axis of magnetisation, to orient along the build direction to produce a fibrous texture. Therefore, the magnetic properties in the horizontal plane of an SLM sample can be expected to be isotropic and close to the optimal ones, since the $\langle 111 \rangle$ directions tend to lie out of the horizontal plane. By building the ring samples with their base parallel to the horizontal plane, we can therefore measure the magnetic properties of the material in its plane of isotropy.

6.5 System Settings for Silicon Iron Processing

For the studies carried out using the Realizer SLM50, laser beam focus position, hatch-spacing, scan distance and layer thickness were kept constant for both cubic and ring samples. The values for the fixed parameters were chosen since they are suggested by the SLM system manufacturer for the processing of stainless steels, which are routinely processed using SLM and are closest to silicon steel in terms of iron content and, therefore, expected thermal and laser energy absorption properties. For manufacturing the cubic samples, laser power (P) and scan speed (v) were varied in order to study the effects of the linear energy input $E=P/v$ on the material properties. As regards the scan strategy, each layer was scanned once using vectors oriented along either the

X-axis (SD_x) or the Y-axis (SD_y), alternatively. This strategy is referred to as bidirectional X-Y. Table 6-3 summarises the fixed process parameter values used, whereas the considered levels of P , v and corresponding E are listed in Table 6-4. The ring samples for magnetic measurement were manufactured at a laser power of 70 W and at three different scan speeds. The choice of the laser power and scan speed values was based on analysis of the results obtained from the study of the cubic samples, as presented in Section 7.3.

Table 6-3: The values of the fixed SLM process parameters employed for manufacturing the cubic samples

SLM Parameter	Values (units)
Laser Beam Focus Position	13.1 (mm)
Hatch Spacing	60 (μm)
Layer thickness	25 (μm)
Scan Strategy	Bidirectional X-Y

Table 6-4: The values of laser power (P), scan speed (v) and corresponding energy input (E) explored for manufacturing the cubic samples.

P (W)					
50		70		90	
v (m/s)	E (J/m)	v	E	v	E
0.083	600	0.125	560	0.125	720
0.125	400	0.167	420	0.167	540
0.250	200	0.250	280	0.250	360
0.5	100	0.5	140	0.5	180
1	50	1	70	1	90

6.6 Heat-Treatment of the Manufactured Parts

In order to study the effect of the processing parameters on the magnetic properties of the near-fully dense parts, the ring specimens with $h=4.5$ mm were heat-treated in argon atmosphere in order to remove the residual stresses introduced during processing, while keeping the grain size unchanged. The stress-relief temperature was chosen based on the results presented in (Faudot et al., 1989), where the authors showed that the grain size of microcrystalline high-silicon steel ribbons was unaffected after a heat-treatment carried out at 700 °C for 24 hours followed by furnace cooling. In order to determine a sufficient dwelling time in order to remove most of the residual stress, the cubic samples were first heat-treated at 700 °C for 1 and 5 hours. The heating rate was set to 30°C/min, and the dwell time was followed by furnace cooling. The micro-residual stress in the heat-treated samples was then determined using X-ray Diffraction (XRD) as described in Section 6.9.4 and the microstructures were observed using SEM as described in Section 6.8.1.

Furthermore, the effect of the annealing temperature on the microstructural evolution and magnetic properties of SLM high-silicon steel was investigated. Therefore, cubic and ring specimens were annealed in Ar atmosphere for 1 hour at four alternative temperatures, namely, 400 °C, 700 °C, 900 °C and 1150 °C, followed by furnace cooling. The studies at variable annealing temperature were performed on samples produced at fixed processing parameters, which were chosen to provide a good compromise between low crack formation and low part porosity.

6.7 Sample Preparation for Characterisation

6.7.1 Sample Preparation for Microstructural Characterisation

In order to investigate porosity, crack formation and microstructural features, the manufactured 5x5x5 mm cubic samples were cut along two perpendicular planes, parallel to the cube faces. The test specimens were cross-sectioned, mounted, and polished. Grinding and polishing of the specimens was performed in reference to the protocol suggested by *Struers* (Struers AS, Ballerup, Denmark) for Fe alloys. Details on the grinding and polishing steps can be found in Table 6-5. For the studies requiring the visualisation of the melt-pool boundaries and solidification structures, after the last polishing step the sample surfaces were chemically etched by exposing them to a solution of 2% Nital (2% nitric acid and 98% ethanol by volume) for 30s-60s.

Table 6-5: Grinding and polishing procedure followed in the present research work

	Wheel/Cloth	Lubricant	Force (N)	Wheel Speed (rpm)	Time (min)
Plane Grinding	220 mesh	Methylated Spirit	30	300	3
	800 mesh	Methylated Spirit	30	300	3
Fine Grinding	1200 mesh	Methylated Spirit	30	300	3
Polishing	MD-Largo	9 μm diamond suspension	10	150	3
	MD-Dac	3 μm diamond suspension	10	150	10
	MD-Floc	1 μm diamond suspension	10	150	15
	MD-Chem	0.04 μm silica suspension	10	150	30-40

6.7.2 Sample Preparation for Magnetic Characterisation

After manufacturing, the top and bottom surfaces of the ring specimens were ground for surface finish in order to reduce measurement uncertainty. Ring samples with initial height of 4.5 mm were ground to a final height of 4 ± 0.01 mm. For ring samples with initial height of 2.5 mm, final height after grinding both surfaces was 2 ± 0.01 mm.

The secondary winding with $N_2 = 15$ turns of insulated copper wire was applied first, and care was taken to wind the cable as uniformly as possible and to minimise the airgap enclosed between the wire and the specimen core. All of the N_2 turns of wire were wound in one single layer on the core. For the magnetising winding, $N_1 = 15$ turns of insulated copper wire were also uniformly wound in one layer.

Before each measurement, each sample was carefully neutralised (demagnetised) using cyclical magnetisation, in order to remove any remanence from previous measurements and/or from exposure to the environmental field. The procedure followed for the neutralisation is shown in Figure 6-2. The ring was placed at the centre of a straight solenoid and exposed to an AC magnetic field at 50 Hz, whose amplitude was steadily reduced to zero. As represented in Figure 6-2, the straight magnetic field inside the solenoid is ineffective in uniformly magnetise/demagnetise the toroidal sample. Therefore, for each sample the neutralisation procedure was repeated twice, at 90° angles. It is noteworthy that given the solid construction of the ring samples, a frequency as high as 50Hz may produce eddy currents strong enough to prevent the magnetic field from penetrating all the way into the sample, thus reducing the effect of the neutralisation procedure.

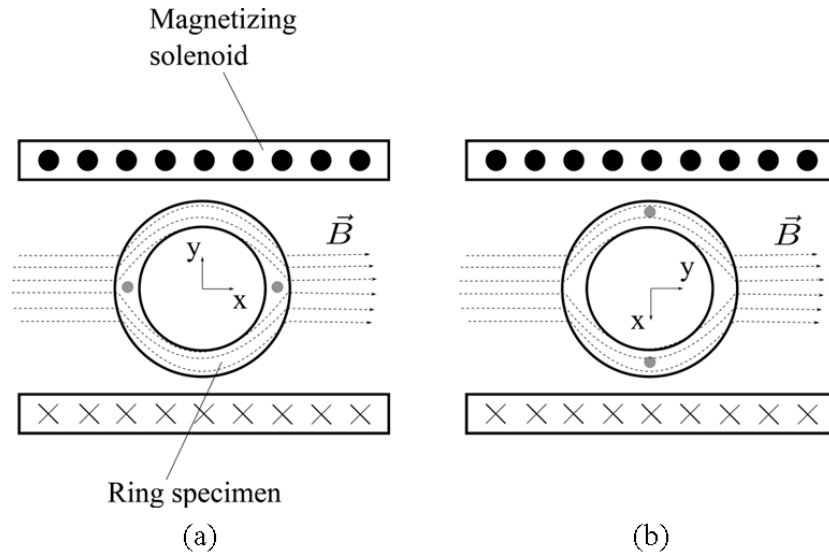


Figure 6-2: In order to assure homogenous demagnetisation, for each sample the procedure was repeated twice. With the first round of demagnetisation (a), the regions of the sample indicated by the grey dots cannot be fully demagnetised. In the second round (b), the sample was rotated by 90°, thus allowing for complete demagnetisation.

6.8 Build Quality and Microstructural Characterisation

6.8.1 Optical Microscopy Studies

The relative porosity was determined from the optical micrographs of the vertical cross-sections of the samples. An *Eclipse LV100ND* microscope (Nikon, Tokyo, Japan) was used to obtain the micrographs. To achieve repeatability, for each test sample a minimum of three to a maximum of six cross-sectional optical micrographs were acquired and analysed using the image processing software *ImageJ* (Ferreira and Rasb, 2012).

The magnitude of the SLM-induced residual stress was investigated through the analysis of the micro-cracks observed on the polished sample surfaces by means of optical microscopy. As a quantitative means of assessing crack formation, the cumulated crack length (c.c.l.) was used. This parameter was

defined as $c.c.l. = \frac{1}{n} \sum_i \sum_j l_{ij}$, where l_{ij} is the length of the j^{th} crack on the i^{th}

polished sample surface and n is the total number of sample surfaces considered for each sample.

6.8.2 Studies Employing Scanning Electron Microscopy

A TM3030 scanning electron microscope (SEM) by Hitachi, Tokyo, Japan, with a 20 kV accelerating voltage and backscattered electron detector was used to evaluate powder morphology and grain structure of the manufactured samples. In addition, the SEM was equipped with an Energy Dispersive X-ray (EDX) detector that was used to confirm chemical composition of the atomised powders corresponded to the supplier data and to assess the distribution of the main alloying elements in the as-built parts. For the elemental analysis using EDX, a 10 kV accelerating voltage was used, which produces an X-ray generation depth of approximately 1 μ m.

6.8.3 Micro-Hardness Measurement

In order to assess the effect of the processing parameters and heat-treatment on the mechanical properties, micro-hardness using a Vickers indenter was measured. For each sample, ten indentations were performed on the front and top planes. An indentation force of 1000g was employed to create the indents. Indentation results are expressed as mean $\pm$ standard error.

6.9 Solid Phase Determination and Micro-Stress

Measurement Using X-Ray Diffraction

6.9.1 X-Ray Diffraction Principles and System Description

X-ray diffraction (XRD) is an analytical technique that relies on the interaction between monochromatic X-ray radiation (normally the $K\alpha$ radiation) and an

atomic lattice to provide information on crystalline compounds, including phase identification, lattice parameter length, and residual stress.

The interaction of the monochromatic x-ray radiation with the valence electrons scatters the wave. If scattered x-rays interfere constructively (i.e., their associated waveforms are in-phase), then coherent scattering is formed. This occurs when the atomic planes from which the x-rays scatter satisfy Bragg's law. This is defined by $n\lambda = 2d \sin\theta$, where λ is the wavelength of the incident x-rays, n is a positive integer, d is the lattice spacing and θ is the x-ray's incident angle. Therefore, Bragg's law relates the wavelength of the incident radiation to the diffraction angle and the inter-planar distances. The diffracted x-rays are collected by a detector and counted. As the sample and detector are rotated, all angles satisfying Bragg's law generate a diffraction signal. Incoherently-scattered x-rays can also reach the detector, leading to the formation of background noise. The resulting diffraction spectrum is characteristic of the scattering material's specific crystalline structure and can therefore be used to characterise it.

In this research, solid phase determination, lattice parameter measurement and micro-residual stress assessment were carried out by XRD using a D500 diffractometer (Siemens, Munich, Germany). The employed anode was Cu, which produces a $K\alpha$ x-ray radiation characterised by a wavelength $\lambda=1.54 \text{ \AA}$. The use of Cu as an anode material is common for routine studies on ferrous materials (Cullity and Weymouth, 1957). The radiation produced by the X-ray tube is not perfectly monochromatic, since standard monochromators are not able to separate the two $K\alpha$ components (known as $K\alpha1$ and $K\alpha2$). Therefore, $K\alpha2$ radiation stripping was performed using Diffrac.EVA software (Bruker,

Billerica, Massachusetts). The latter was also employed for background noise removal.

6.9.2 *Use of X-Ray Diffraction for Solid Phase Determination*

Diffraction peaks in the range $2\theta = [25^\circ, 90^\circ]$ were analysed to determine the solid phase composition. The study was performed in one vertical plane of the cubic samples (the front view) and for the atomised powders. The acquisition conditions were step size $\Delta\theta = 0.05^\circ$ and step time $t = 8s$.

6.9.3 *Use of X-Ray Diffraction for Lattice Parameter Determination*

The lattice parameter was measured in order to understand how the SLM process parameters affect the order-disorder transformation in the studied Fe-Si alloy. Ordered phases were introduced in Section 3.2, where we mentioned their role in determining the mechanical properties of high-Si steel alloys. The lattice parameter was computed from the sixth diffraction peak (corresponding to diffraction plane (222)) of the XRD spectrum in order to achieve high accuracy. By knowing that the Fe-Si alloy under study is characterised by a cubic structure, the lattice parameter a could be calculated using Miller's indices (222) as $a = d\sqrt{12}$. The acquisition conditions were step size $\Delta\theta = 0.02^\circ$ and step time $t = 30s$.

6.9.4 *Use of X-Ray Diffraction for Micro-Stress Measurement*

The non-uniform lattice strain in the SLM samples and atomised powders was estimated from the broadening of the first six diffraction lines using the Williamson-Hall method (Williamson and Hall, 1953), in order to assess the stress-relieving effect of the heat treatment conditions considered in this study (these have been described in Section 6.6). The Williamson-Hall method

utilises the fact that diffraction peak broadening from different sources have different angular relationships. In particular, the total broadening β_t can be described by:

$$\beta_t^2(\theta) = \left(\frac{0.9\lambda}{D \cos \theta} \right)^2 + (4\epsilon \tan \theta)^2 + \beta_0^2 \quad (6.1)$$

where D is the average crystallite diameter and ϵ the lattice strain. In the above equation the first term, which follows a $1/\cos(\theta)$ relationship, corresponds to the Sherrer's equation. This describes the broadening due to crystallite size (Cullity and Weymouth, 1957). The second term, which is a function of $\tan(\theta)$, shows that the strain has a different angular relationship than crystallite size. Finally, β_0 is the broadening due to the instrumentation. In this work the broadening has been expressed using the Full Width at Half Maximum (FWHM) criterion under the assumption of Gaussian peak shape (Cullity and Weymouth, 1957). In the present study, the grain size effect (i.e., the Sherrer term in equation (6.1) can be neglected in light of the metallographic results, which show, as will be presented in Section 7.4 that the typical crystallite size in the SLM high-Si steel samples is well above 1000 Å. This crystallite size is generally considered the upper limit able to produce appreciable peak broadening (Cullity and Weymouth, 1957). Regarding the instrumentation broadening β_0 , this was determined experimentally using a standard LaB6 specimen, whose single-crystal and defect-free structure makes both the crystallite and strain contributions to broadening near zero. The goodness of fit between the experimental data and Williamson-Hall equation was assessed using the coefficient of determination R^2 .

While the strain values estimated using the Williamson-Hall method were not interpreted in an absolute sense, it was assumed that the atomised powder was stress-free, and therefore its measured strain value could be used as a reference for assessing residual stress in the manufactured samples. It is also noted that work in the SLM literature usually focuses on the analysis and removal of the macro-scale stresses, due to their higher magnitude and, therefore, importance in determining the strength of a structural part (Mercelis and Kruth, 2006). However, the magnetic performance of a ferromagnetic alloy can be also dramatically affected by the non-uniform stresses occurring at the grain scale, since local lattice distortions, such as linear defects (dislocations), represent pinning sites for the magnetic domain walls and thus hinder their motion during the magnetisation-demagnetisation processes (Hilzinger and Rodewald, 2013).

6.10 Crystallographic Texture Characterisation Using Electron Backscatter Diffraction

6.10.1 Electron Backscatter Diffraction Principles and System Description

Electron Backscatter Diffraction (EBSD) relies on the interaction between a beam of highly energetic electrons and the crystal grains at the surface of a polished crystalline material. As we discussed in the case of X-ray radiation, the interaction of the electrons with the material causes the electrons to be scattered away from the material. Electrons that backscatter both coherently and elastically produce so-called Kossel cones (Schwartz et al., 2009), that can be captured in the form of intersecting straight bands (known as Kikuchi bands) on a sensitive phosphor screen placed in close proximity to the surface of the sample. The diffraction patterns carry the following information:

- The luminance of each Kikuchi band is linked to the intensity of the diffraction, that is, to the distance of the diffracting plane from the detector;
- the width of the bands is related to the spacing of atoms in the diffracting crystallographic plane;
- the angles between the bands are related to the angles between the crystallographic planes.

Therefore, the EBSD pattern contains all of the angular information required to calculate the crystallographic structure and orientation of the investigated grain with respect to a fixed sample coordinate frame of reference. The interpretation (indexing) of the EBSD patterns is performed automatically and consists of three steps (Stojakovic, 2012):

- 1) **Band detection:** By means of a mathematical algorithm known as the Hough transform, the location of the Kikuchi bands in the EBSD pattern are detected;
- 2) **Phase identification/verification:** The crystal structure is deduced from the relative orientation of the identified bands in the pattern, by comparing the experimental data with the data in a large database. If the phases are known with sufficient confidence, then a simpler procedure of phase verification can be carried out, that involves the comparison of the experimental data with a pre-selected choice of database data;
- 3) **Grain orientation determination:** The inter-planar spacing and angles that derive from the width of the measured bands are compared to theoretical values to identify most probable crystal orientation.

In the present research work a Nova 600 Nanolab Dualbeam FIB/FEG-SEM system (FEI, Hillsboro, Oregon) was employed. The scanning step size was 2 μm . Grains were defined by a misorientation angle larger than 5° . Since the information on the phases present in the material was known with high confidence from previous EDX and XRD studies, the second step in the process of indexing of the EBSD data only consisted of phase verification. The average confidence index of the EBSD imaging was calculated to be around 0.70.

6.10.2 Presentation of the Crystallographic Texture Results

In the present research work the results from the EBSD scans are represented using pole figures (PF), inverse pole figures (IPFs) and orientation maps. PFs are useful for showing the orientations of specific crystallographic directions in the sample coordinate system. In practise, discrete PFs are built as stereographic projections of the crystal directions oriented with respect to the sample coordinate frame. Therefore, in discrete pole figures, each point represents the orientation of a grain with respect to the sample coordinates. Contour PFs are similar to discrete PFs, but instead of providing information regarding single grains, visualise using a colour scheme the texture strength of the measured sample with respect to a randomly oriented background. Depending on the specific crystal symmetry, in PFs single data points can appear several times (e.g., in a cubic system the [100] direction will appear three times, as [100], [010] and [001]). In this research work we adopt contour PFs.

IPFs are different from PFs in that they effectively provide information on the orientation of the sample coordinates with respect to the crystallographic

directions. The appearance of an IPF is determined by the symmetry of the crystal unit cell. As shown in Figure 6-3, in the case of cubic systems a triangular region is descriptive of the whole crystallographic system. Similarly to PFs, IPFs exist either in the form of discrete plots or contour plots. In this research work, contour IPFs are adopted. An advantage of IPF over PF representation is that it makes it more intuitive to identify the presence of fibre textures along a specific sample axis.

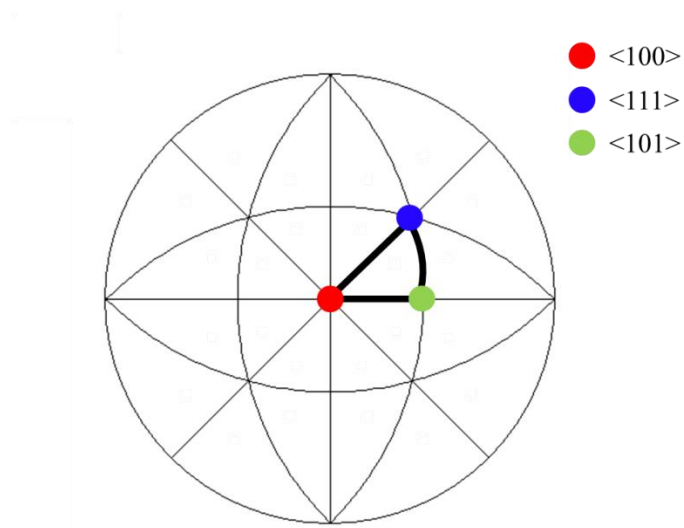


Figure 6-3: Derivation of the IPF as one of the symmetrically equivalent triangular regions in a cubic system. The crystallographic directions in correspondence of the triangle vertices are shown are indicated by the coloured circles.

EBSD orientation maps were used to obtain a spatially-resolved visualisation of the grain orientation with respect to specific sample directions. Orientation maps offer a way to visualise how the texture is spatially distributed within a sample. To do this, a colour is assigned to each crystallographic direction, according to a colour-key. Then, each measured sample pixel is coloured accordingly, depending on its orientation with respect to either of the sample axes. In order to visualise how several factors (contamination, sample preparation, grain boundaries, etc.) affect the quality of the EBSD

measurements, a pattern quality parameter can be created, which associate a number and a corresponding grey-scale level with how sharp or well defined the band positions are. An image quality map can then be superimposed to the orientation map in order to visualise the grain boundaries and other features. In this research work the EBSD orientation maps were coloured according to the IPF colour-key shown in Figure 6-4. The final EBSD images were obtained using microstructural data analysis software TSL OIM (Ametek Inc) by combining the orientation maps with the corresponding image quality maps.

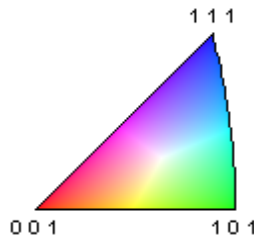


Figure 6-4: Inverse Pole Figure (IPF) colour key used to represent grain orientation with respect to crystallographic directions in the EBSD orientation maps.

6.11 Characterisation of the Magnetic Properties

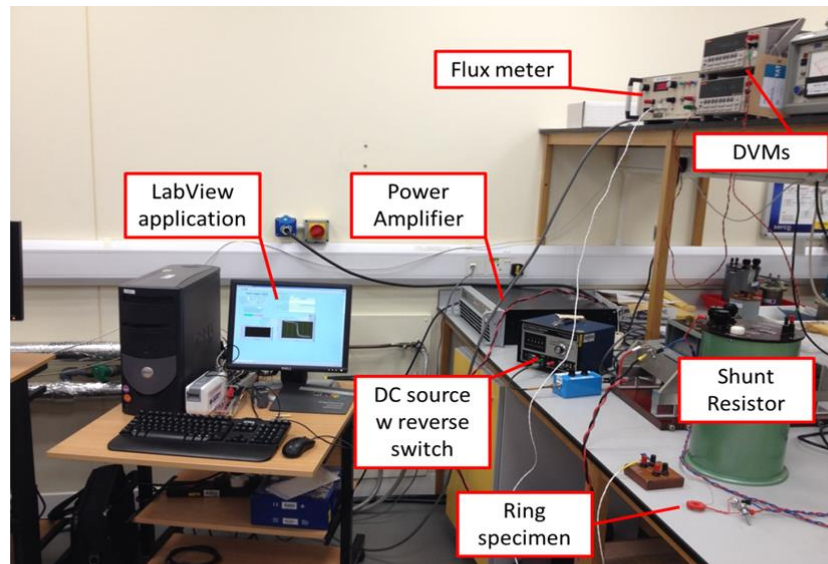
The materials' maximum relative permeability (μ_{max}), relative permeability at an applied field strength $H=40$ A/m (μ_{40}), as well as flux density at $H=5000$ A/m (B_{50}) were employed to assess the samples' magnetisation properties, while remanence (B_r) and coercivity (H_c) were determined from the B-H loop in order to assess loop shape and hysteresis loss, respectively. The quantity μ_{40} was used in place of the initial relative permeability, which is normally measured at $H=4$ A/m (Hilzinger and Rodewald, 2013), because of the difficulty encountered in achieving reliable readings at lower applied field values. Total power losses (P_{50-1}) at excitation frequency $f=50$ Hz and maximum flux density $B=1$ T were also determined, following the guidelines

presented in (IEC, 2003) for the measurement of the AC magnetic properties of soft magnetic materials by the use of ring samples. The magnetic measurements were repeated for each ring specimen until, for each of the magnetic properties considered, a level of uncertainty <3% could be achieved. Uncertainty was calculated using the standard error. Furthermore, for each of the manufacturing process parameters considered, four ring samples were manufactured in order to assess the repeatability of the manufacturing process. Therefore, the measured magnetic properties are expressed as mean $\pm$ standard error, and the standard error represents a measure of the among-sample variability.

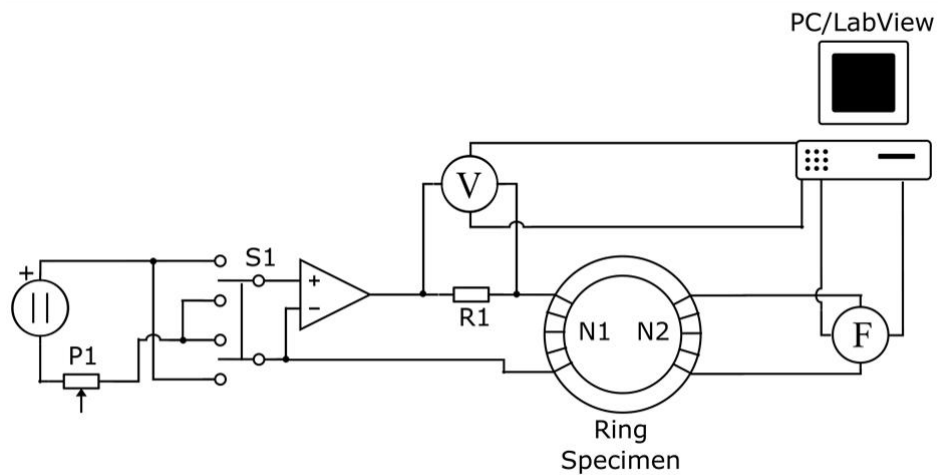
6.11.1 Measurement of the DC Magnetic Properties

The magnetic properties of the SLM Fe-Si alloy in quasi-static conditions were determined using the ring method outlined in the IEC standards (IEC, 2000). An overview of the experimental apparatus used for performing the DC measurements is shown in Figure 6-5(a). The apparatus was connected as shown in the circuit schematic in Figure 6-5 (b). Electric source E generates a DC voltage adjustable via potentiometer $P1$. The signal is amplified via a power amplifier, whose output current i_1 is calculated as $i_1 = V1/R1$, where $V1$ is the reading of the digital voltmeter V connected to the calibrated resistor $R1$. Voltage $V1$ was read on a Labview computer interface connected to the digital voltmeter. A magnetic field of intensity H originates inside the ring from the magnetomotive force $N_1 i_1$ and is calculated as $H = \frac{N_1 i_1}{l}$, where l is the average magnetic path length, which can be calculated from the ring's inner and outer diameters, whose values have been provided in Section 6.4. Magnetic

field density B induces an output voltage across the secondary winding. The induced voltage is then integrated by fluxmeter F, whose output was read on the Labview computer interface. Magnetic field density can then be calculated from fluxmeter output voltage V_2 as $B = \frac{kV_2}{AN_2} + B_0$, where k is the fluxmeter calibration constant in $[Wb/V]$, and B_0 the initial magnetisation of the ring. If specimen demagnetisation process has been carefully performed prior to the measurement then $B_0 = 0$.



(a)



(b)

Figure 6-5: The experimental setup used to perform the magnetic characterisation of the ring specimens. The actual set-up is in (a), the electrical circuit schematic is in (b).

The initial magnetisation curve was determined by steadily increasing the magnetising current from 0 up to the maximum value. The magnetising current was then reversed using switch S1 to the maximum negative value. The current was then reversed again and steadily increased from the negative to the positive maximum, to obtain the complete hysteresis loop. To avoid the formation of eddy currents – which would deteriorate the quality of the results by introducing Joule losses within the hysteresis curve of the material – while

limiting fluxmeter drift, magnetisation cycle time t_c was >110 s and <180 s for all of the tested samples. In order to verify the repeatability of the measurements, a minimum of 3 measurements were performed on each sample. From the measured initial and DC B-H loops the materials' maximum relative permeability (μ_{max}), relative permeability at an applied field strength $H=40$ A/m (μ_{40}), as well as flux density at $H=5000$ A/m (B_{50}) were determined to assess the samples' magnetisation properties, while remanence (B_r) and coercivity (H_c) were determined in order to assess loop shape and hysteresis loss, respectively. The quantity μ_{40} was used in place of the initial relative permeability, which is normally measured at $H=4$ A/m (Hilzinger and Rodewald, 2013), because of the difficulty encountered in achieving reliable readings at lower applied field values.

6.11.2 Measurement of the Dynamic Magnetic Properties

The AC measurements were outsourced to the Power Electronics, Machines and Control (PEMC) Group at the University of Nottingham. These were performed following the guidelines presented in the IEC standards for the measurement of the AC magnetic properties of soft magnetic materials by the use of ring samples (IEC, 2003). In particular, the property of interest in this study was the total specific power loss at fixed excitation frequency and output magnetic flux density. The total specific loss is the total loss per unit of sample mass and was calculated by means of the wattmeter method using the following equation:

$$P_c = \frac{N_1}{N_2} P_m - \frac{(1.111 V_1)^2}{R_i} \quad (6.2)$$

where P_m is the wattmeter reading, given the wattmeter connection to the system as shown in the simplified circuit schematic of Figure 6-6. The variable V_1 is the reading of voltmeter V1 (refer to Figure 6-6), and corresponds to the rectified average value of the secondary voltage. The constant 1.111 is the form factor of the output voltage wave. The form factor is the ratio of the RMS value to the half-period average value and here is assumed to take the value for a pure sine wave. This is because, in order to avoid introducing higher harmonics in the flux density waveform, which would in turn lead to the creation of extra losses thus distorting the total loss reading, the waveform of the secondary voltage has been maintained sinusoidal, with a form factor of $1,111 \pm 3\%$. The sinusoidal shape was maintained by means of closed-loop control on the magnetising current (Hilzinger and Rodewald, 2013). However, given the challenges encountered in the maintenance of the form factor variation within acceptable limits, the study of the dynamic properties could not be exhaustive. In particular, only the total specific power loss (P_{50-1}) at excitation frequency $f=50$ Hz and maximum flux density $B=1$ T was determined.

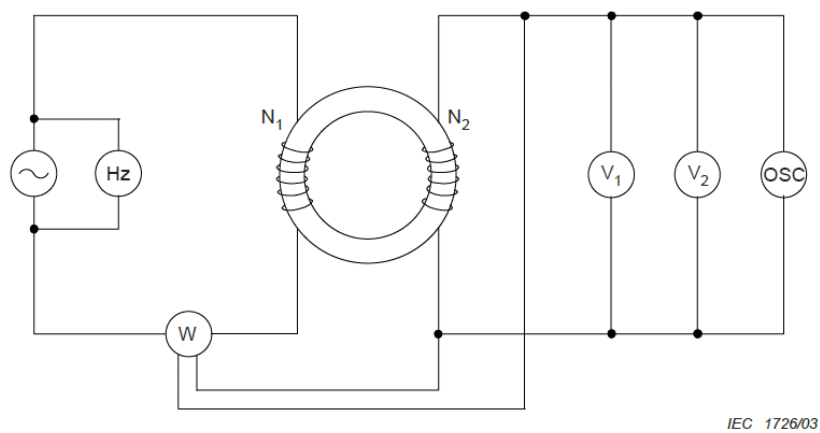


Figure 6-6: Simplified circuit schematic aimed at showing the connection of the wattmeter (W) to the measuring set-up. Reproduced from (IEC, 2003).

6.11.3 Post-Processing of the Acquired DC Magnetisation

Data

A drawback of using an integrator (fluxmeter) to measure the sample flux density is the drift affecting this instrument. Although care was taken to avoid excessively slow magnetization rates and the fluxmeter was set to zero prior to the start of the measurement, the acquired data needed to be corrected for drift. As a way to do so, integrator drift was modelled as a linear function of time and the acquired flux density reading was assumed to be a linear superposition of the real flux density and the drift, according to the following model:

$$\begin{cases} \hat{B}(t, H) = B(t, H) + d(t) \\ d(t) = mt + c \end{cases} \quad (6.3)$$

where:

- $\hat{B}$ is the magnetic flux density as obtained from the raw voltage readings;
- B is the real magnetic flux density, i.e., without the effect of drift;
- d is the drift, modelled as a linear function with slope m and intercept c

Assuming zero drift at the start of the measurement, we obtain a null value for the intercept, i.e.:

$$c = 0 \quad (6.4)$$

To estimate model parameter m , the methodology represented in Figure 6-7 was employed. We assumed that at saturation field strength H_{max} the values of the real field density are equal:

$$B_1(t_1, H_{max}) = B_2(t_2, H_{max}). \quad (6.5)$$

Therefore, combining equations (1)-(3) we can calculate slope m as:

$$m = \frac{\hat{B}_2 - \hat{B}_1}{t_2 - t_1} \quad (6.6)$$

If full material saturation is achieved, the estimation can be made more robust by considering the whole set of field density points at saturation (i.e., at field strengths $> H_{max}$) and take the average m value. In the present study, only the maximum field strength in the dataset was considered, since the magnetomotive force provided was not sufficiently high to enter the deep saturation region of the material.

Owing to the presence of asymmetries in the measuring system and possibly to incomplete demagnetisation of the specimen, the positive and negative branches of the B-H loop after drift correction were asymmetrical. To further improve the quality of the measurement, the average of the positive and negative branches was taken.

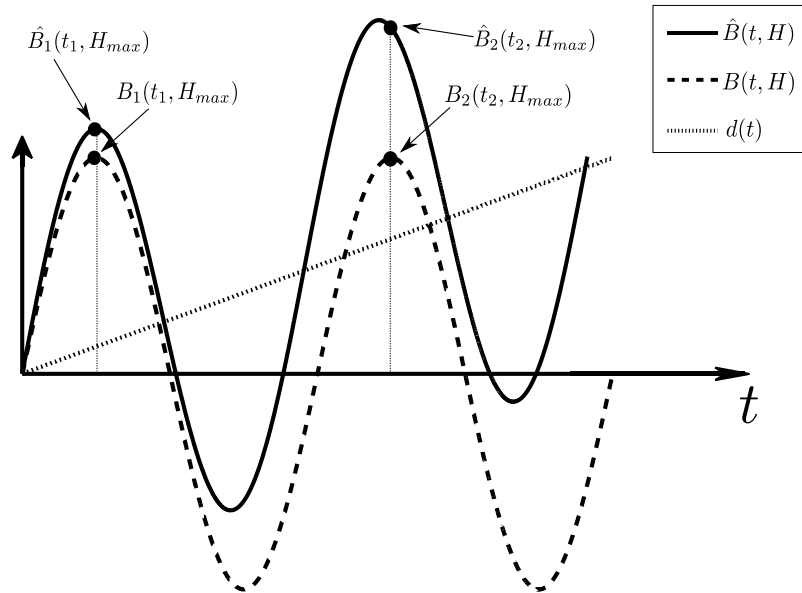


Figure 6-7: Integrator drift model. Integrator reading $\hat{B}$ is modelled as a superposition of the real induction waveform B and a linear function $d(t)$ with slope m and intercept $c=0$

6.12 Summary

The experimental methods used for the production, post-processing and characterisation of high-silicon steel by means of SLM technology have been described in this Chapter. The main experiments performed are summarised in the bullet points below:

- **Sample processing:** Fe-6.9wt.%Si gas-atomised powder was processed using a Realizer SLM50 system. Both cubic and ring samples were produced, to be employed for metallurgical and magnetic evaluation, respectively;
- **Sample post-processing:** as-built samples are annealed at different temperatures in the range 400°C – 1150°C to evaluate the effect of stress-relief, phase ordering and grain growth on the magnetic properties;
- **Defect analysis:** pores and cracks were analysed using optical microscopy after cross-sectioning and polishing the as-built cubic samples;
- **Microstructural analysis:** the following microstructural features were analysed for both the as-built and annealed samples:
 - Grain size and morphology using optical and electron microscopy;
 - Crystallographic texture using EBSD;
 - Phase distribution and residual stress using XRD;
- **Magnetic characterisation:** the quasi-static magnetic properties (hysteresis loop) and dynamic losses at 50 Hz and 1 T were measured for the as-built and annealed ring samples using the ring method.

7. Metallurgy of High-Silicon Steel Parts Produced by Selective Laser Melting

7.1 Introduction

The metallurgy of high-silicon steel (6.9%wt.Si) processed using Selective Laser Melting (SLM) is presented for the first time in this chapter. Specifically, the chapter aims to shed light on the processability and microstructure of the alloy, in order to lay the foundation for the interpretation of its magnetic properties and to perform a first assessment regarding its applicability in electromechanical applications. First, the effect of SLM-processing on the porosity and crack formation in test samples is investigated using optical microscopy and discussed in light of the relevant literature. Next, the effect of SLM-processing and post-thermal annealing on the solidification structures, grain morphology and crystallographic texture of the considered alloy is investigated using optical and electron microscopy, X-Ray Diffraction (XRD) and Electron Backscatter Diffraction (EBSD).

7.2 Results of Powder Characterisation

The results of the LPW Fe-Si powder characterisation analysis are presented in Figure 7-1. The SEM micrograph shows that the particles are approximately spherical in shape. The content of Si in the pre-alloyed particles is found to be comparable to the range declared by the supplier (6.2-7.0%wt.). The particle size distribution is found to be symmetrical and centred at 35 μm . Particle size ranges from approximately 10 μm to 60 μm . This is in reasonable agreement with the range declared by the supplier (15 μm -45 μm). The symmetry of the particle size distribution indicates the particles' high level of sphericity. This is

an important feature, since spherical powders are preferable in SLM as they possess good flowability, high packing density and therefore promote increased part densification (Simchi, 2004).

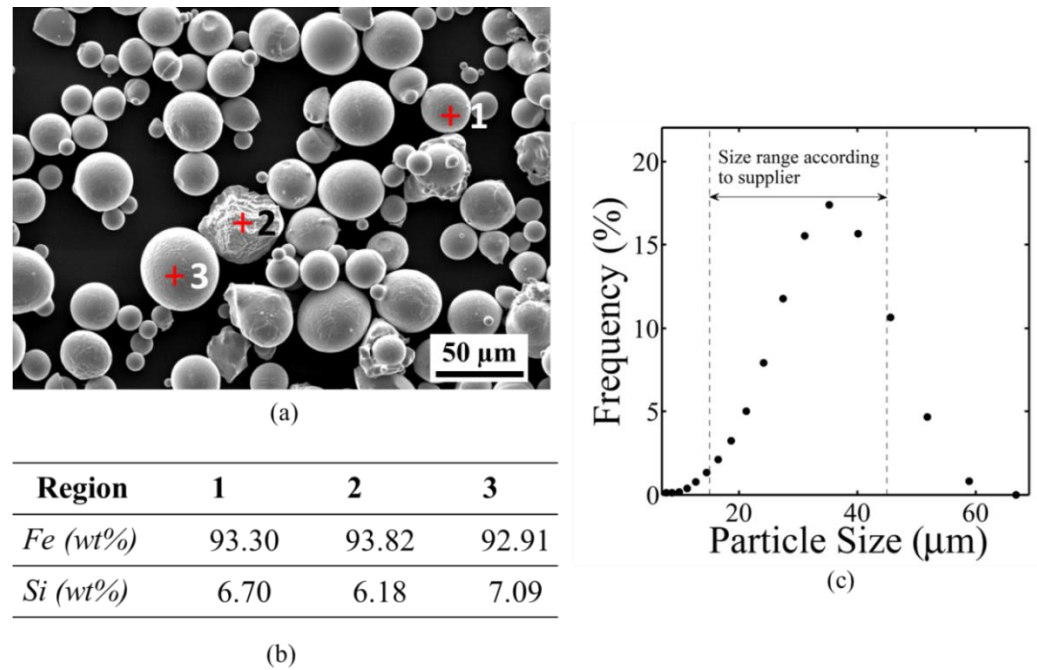


Figure 7-1: LPW Fe-Si powder size distribution and chemical composition. In (a) a SEM micrograph of the powder is shown. In (b) the chemical composition of the three powder particles marked by the “+” is provided, as measured from EDX analysis. In (c) the particle size distribution measured with the particle size analyser is shown.

7.3 Study of Porosity and Crack Formation

7.3.1 Pore and Crack Observation Results

The optical micrographs in Figure 7-2 present the evolution of porosity and pore morphology at different values of the laser power P and scan speed, v . Three main regions can be identified within the parameter window considered. The region of high porosity (indicated by the dotted background in Figure 7-2) is populated by the samples that are more than 1% porous. Large pores of irregular shape characterise the samples in this region. As the energy input of the processing laser is increased, a region of low porosity (identified by the

diamond background) can be achieved. As already found in the literature for other SLM-processed alloys, such as in (Aboulkhair et al., 2014) for SLM AlSi10Mg, increasing the energy input of the processing laser has the effect of reducing irregular pore formation. However, as the laser energy input increases further, a third region can be identified (dashed background in Figure 7-2). This is characterised by the formation of cracks whose average width is larger than 100 μm . These form due to the thermal gradient-induced residual stresses, which are known to affect SLM parts, as previously described in Section 4.3. Furthermore, the high temperature reached in the melt-pool promotes evaporation and, hence, the formation of small bubble-like voids, also referred to as spherical pores, which get trapped during melt-pool solidification. These disappear again when the highest laser energy input (720 J/m) is employed.

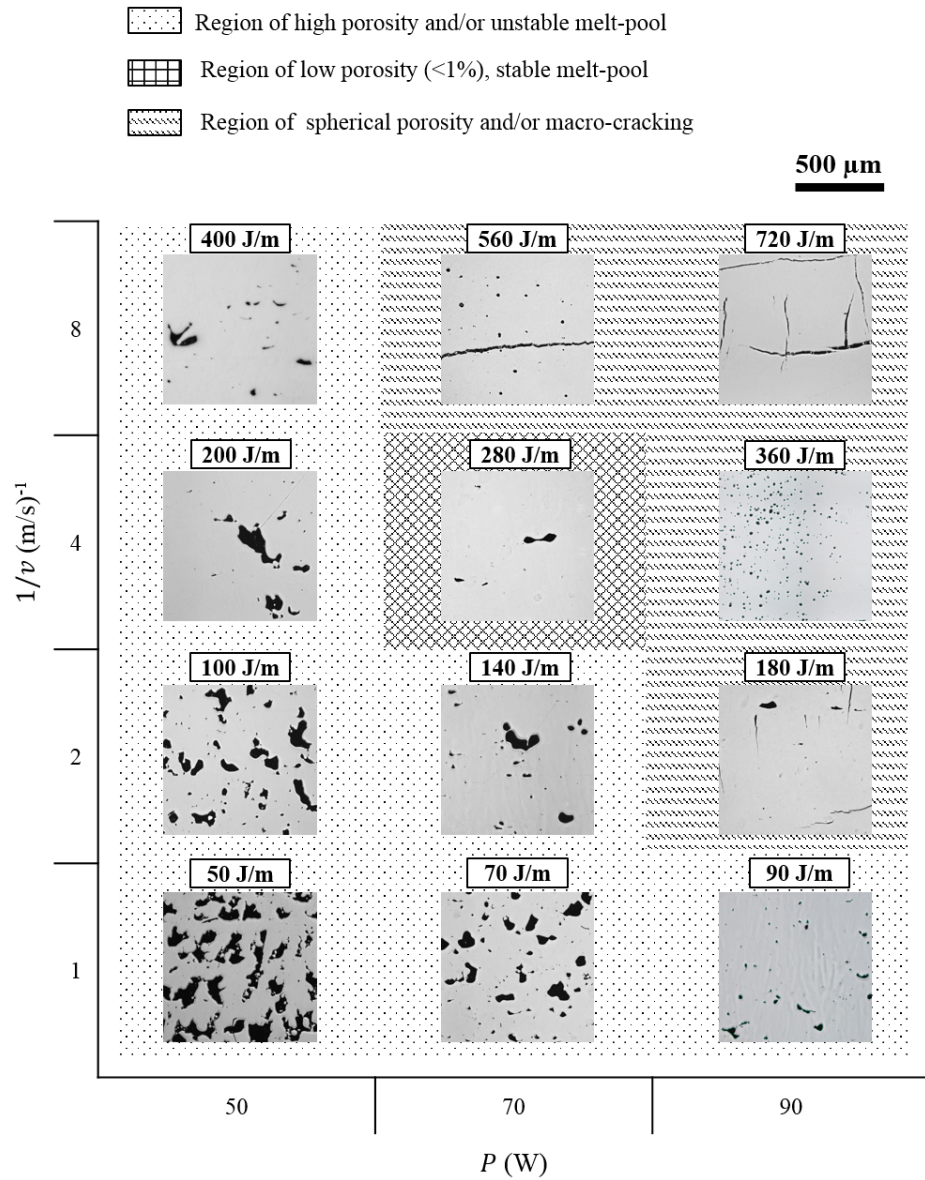


Figure 7-2: Optical micrographs showing the effect of increasing laser energy input $E=P/v$ on sample porosity and crack formation. The dark areas in the images are manufacturing defects.

The SEM micrographs in Figure 7-3 provide an insight into the two main types of porosities observed in the processed parts and their relationship with the qualitative shape of the melt-pool. Some of the melt-pool boundaries in Figure 7-3 are highlighted by the dashed and solid lines, depending on whether the corresponding laser scan tracks are aligned with the scan vector along the X-axis (SD_x) or Y-axis (SD_y), respectively. While the irregular pores are normally located at the intersection between wide, shallow melt-pools (left micrograph

in Figure 7-3), the spherical pores are observed at the bottom of deeper, protuberant melt-pools (right micrograph in Figure 7-3). Therefore, the phenomenon of spherical pore formation is attributed to a change from a conductive to a keyhole welding mode, as previously described in both the laser welding (Dahotre and Harimkar, 2008, Pastor et al., 1999) and SLM literatures (King et al., 2014), and has been justified by the unstable collapse of the keyhole during melt-pool cooling (also refer to Section 4.2 for a detailed description of the literature). The fact that no spherical pores could be observed for samples processed using the highest laser energy input (720 J/m) is attributed to a stabilisation of the keyhole, which had been previously observed to be possible, since a further increase in the heat-source energy input after keyhole formation allows gas entrapment to be prevented (Pastor et al., 1999).

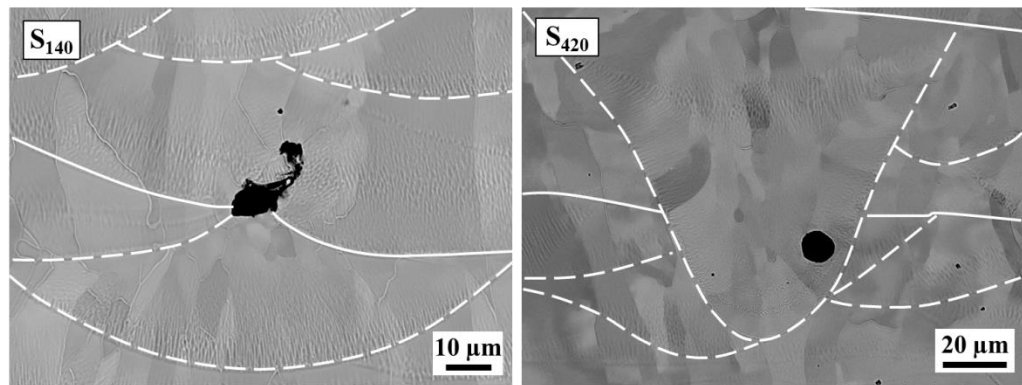


Figure 7-3: Front view SEM micrographs showing the relationship between melt-pool shape and pore morphology. On the left, an irregular pore in a low energy input sample ($E=140$ J/m) and, on the right, a spherical pore in a high energy input sample ($E=420$ J/m). The dashed and solid lines indicate the transversal and longitudinal melt-pool cross-sections, respectively.

In order to provide a more quantitative understanding of the roles of laser power and scan speed in pore and crack formation, Figure 7-4(a) shows the variation of part porosity with laser energy at different laser powers, while Figure 7-4(b) shows the evolution of micro-crack formation. At low energies,

porosity appears to decrease rapidly with increasing energy at all power levels. Within this range, increasing the energy input promotes densification of the part at all power levels considered, by improving powder melting, reducing melt-pool instabilities (Yadroitsev et al., 2007) and avoiding incomplete hole filling (Dahotre and Harimkar, 2008). For the 70 W samples, the mean porosity decreases monotonically with increasing laser energy input. However, as larger energy input values are approached, the porosity trend flattens out, suggesting that increasing the energy input beyond this does not promote part densification any further, owing to the formation of spherical porosities, as previously shown in Figure 7-2. For the 90W samples, the porosity curve reaches its minimum value of $0.22 \pm 0.05\%$ at a comparatively low energy input $E=180$ J/m, but reverses when the laser energy increases further. This trend is attributed to the formation of spherical pores, which is exacerbated as compared to the 70 W samples. While spherical pore formation reduces crack development in sample 90 W-0.250 m/s, decreasing the scan speed further to 0.125 m/s causes the spherical voids to disappear, but results in higher residual stress formation, followed by in-process part failure. At low power (50 W) the porosity decreases monotonically with increasing energy input and only reaches its minimum value of 0.73 ± 0.23 at $E=400$ J/m. As the energy input is increased further, the porosity trend also reverses, as observed in the 90 W case. However, for the $P=50$ W case the behaviour is interpreted in light of the enlargement of the heat-affected zone, which causes the surrounding powders and solid material from outside the scanning track to enlarge and acquire an irregular shape. This phenomenon, which had already been observed and interpreted by (Yadroitsev et al., 2007) for single tracks of Inco 904L, is

accompanied by the persistence of the irregular pores in the microstructure of the 50 W samples at high energy input levels, as the micrographs in Figure 7-2 have also shown.

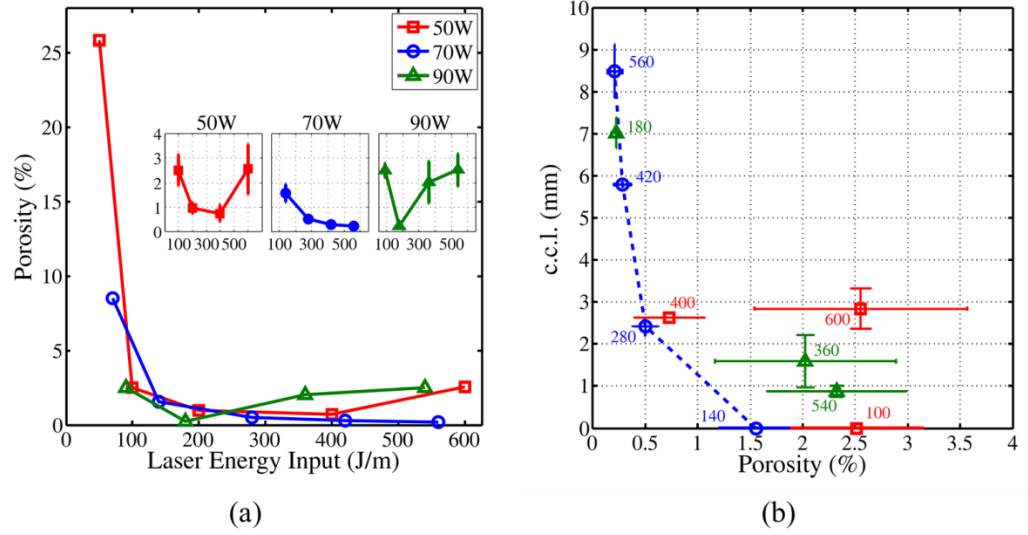
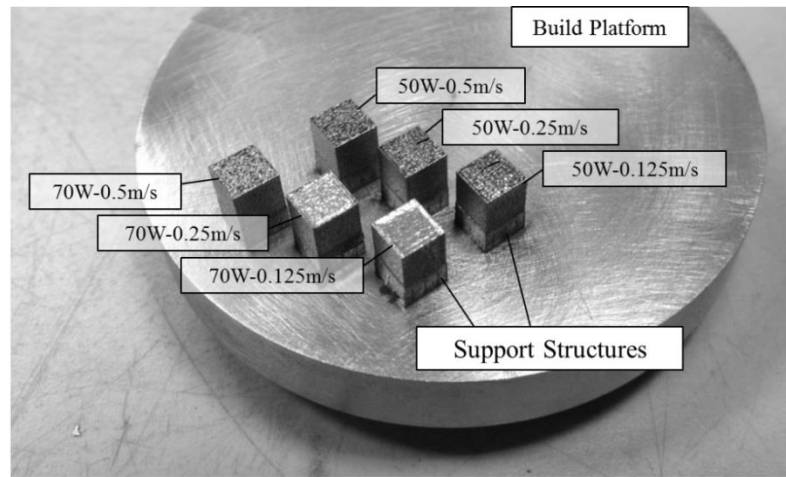


Figure 7-4: Effect of processing parameters on sample porosity and crack development. In (a) sample percentage porosity is plotted against laser energy input and for different laser power levels. In (b), the cumulated crack length (*c.c.l.*) is plotted against percentage porosity for the 70 W samples at different scan speeds, and for samples 50 W-0.125 m/s (400 W) and 90 W-0.5 m/s (180 J/m). The energy input values for each of the samples considered in plot (b) are reported next to the corresponding data point.

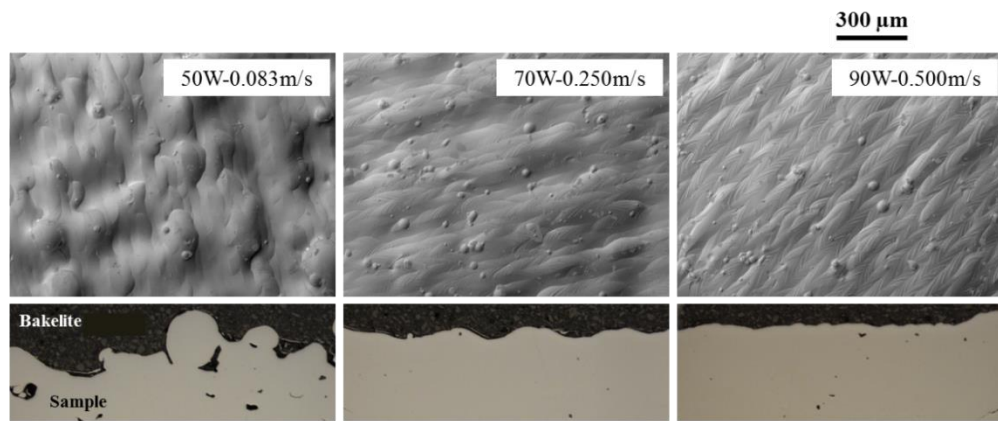
In Figure 7-4(b), crack formation is investigated for a selection of the samples using the cumulated crack length parameter (*c.c.l.*, introduced in Section 6.8.1). The trend suggests that crack formation is promoted with decreasing porosity, independently of the energy input, power or scan speed employed. This result is interpreted in light of the fact that a more porous sample contains more voids, which produce localised stress relief, as also suggested in (Merzelis and Kruth, 2006). However, Figure 7-4(b) also indicates that at fixed porosity, the samples processed using a laser power $P=70$ W are characterised by a lower *c.c.l.* Three among the samples processed at $P=70$ W are of particular interest.

The sample processed at $E=140$ J/m ($P=70$ W, $v=0.5$ m/s) is the one, among the crack-free samples ($c.c.l.=0$), characterised by the lowest porosity (1.5%). The sample processed at $E=280$ J/m ($P=70$ W, $v=0.250$ m/s) possesses the best compromise between porosity and $c.c.l.$, since its Euclidean distance from the origin of the axes is lower than for any other P - v combinations. Finally, $E=420$ J/m ($P=70$ W, $v=0.167$ m/s) produces the lowest porosity (0.2%) without causing build failure owing to excessive part warping/support structure failure. Therefore, the study of the microstructure in the following section, as well as the measurement of the magnetic properties in the next chapter, will be limited to the samples processed using $P=70$ W.

The photograph of some of the as-built samples (Figure 7-5a) provides a qualitative comparison of the surface roughness of samples obtained at different power levels and scan speeds. The image shows that increasing the energy input does not improve sample surface smoothness for the 50 W samples as much as it does for the 70 W samples. Furthermore, the SEM and optical micrographs in Figure 7-5(b) compare the surfaces of three samples, each obtained at a different laser power level, namely samples 50 W-0.083 m/s, 70 W-0.250 m/s and 90 W-0.5 m/s. Sample 50 W-0.083 m/s, processed using the highest energy input $E=600$ J/m, displays the roughest surface.



(a)



(b)

Figure 7-5: Effect of energy input on sample surface roughness. Photograph of the as-built cubic samples on the build platform are shown in (a). Surface roughness comparison for samples processed using similar energy input values, but different power-scan speed combinations are in (b). Top micrographs are SEM images of the top surfaces. The bottom micrographs are optical microscope images of the samples' vertical cross-section.

7.3.2 Origin of Defects

It is known from the literature of SLM alloys that increasing the energy input promotes densification of the part by improving powder melting and reducing melt-pool instabilities (e.g., balling) (Dahotre and Harimkar, 2008, Yadroitsev et al., 2007, Aboulkhair et al., 2014). However, it has been shown in Figure 7-3 that while the amount of irregular porosities is reduced with the increase in

energy input, the formation of the spherical pores and cracks with increasing energy input prevents the part from reaching full density. Spherical porosity formation is associated with the transition between different welding modes. The qualitative change of the melt-pool shape from flat to protuberant at high energy inputs, as shown in Figure 7-3B, suggests that conduction mode welding is replaced by key-hole welding (Dahotre and Harimkar, 2008). The formation of a key-hole occurs by surface vaporization of material, thus involving the creation of gas bubbles that can remain trapped inside the melt-pool during solidification. Furthermore, the dynamics of the Marangoni flow increases with laser energy input, thus increasing the probability of any gas (either newly formed or originally trapped within the highly packed powder particles) being dragged towards the bottom of the melt-pool (Ng et al., 2009). The formation of cracks can be attributed to the thermal stresses induced during SLM processing. As presented in the literature (Pohl et al., 2001), SLM introduces tensile residual stresses in the part owing to the localised application of high energy over a very short time interval (in the order of the μs) which, alongside the poor heat conductivity of the powder bed, develops a high thermal gradient around the laser spot. Differential expansion takes place and eventually the heat affected zone, whose tendency to expand is inhibited by the surrounding material, is plastically compressed. During cooling, the upper layer's tendency to contract is again counteracted by the underlying layers, thus creating tensile stresses. In the present study, cracks propagate more or less parallel to the scanning plane (i.e., the horizontal plane), as observed in Figure 7-3A. On the other hand, inter-granular crack propagation has not been

observed and this is probably linked to the absence of significant macro-segregation phenomena at the grain boundaries.

7.3.3 Effect of Different Power-Speed Combinations on Build Quality

We aim to understand better the underlying phenomena explaining why, given a similar energy input, different P - v combinations lead to different defect distributions in SLM high-silicon steel, as the results in Section 7.3.1 have shown. Given the same energy input, at lower laser power two mechanisms promote pore formation. First, the net heat input (i.e., the difference between the total laser energy absorbed by the powder bed and the heat dissipated through the substrate by conduction) is expected to be lower, the lower the rate at which the laser energy is delivered. Second, at a lower laser power, a longer time interval elapses before two adjacent melt-pools are processed, thus causing the initial temperature in the substrate to be lower. Both mechanisms cause the melt-pool temperature to decrease, thus decreasing the vaporisation rate and depth, and consequently preventing the creation of a recoil pressure, which has been shown to help flatten the melt and improve surface quality, both in SLM (Kruth et al., 2004) and laser welding applications (Dahotre and Harimkar, 2008). Irregular molten track formation results in a rough specimen surface, which in turn prevents regular layer deposition and thus promotes the formation of porosities. It is important to note that the extent of substrate temperature decrease with laser power at fixed energy input depends on the thermal conductivity of the material processed. In particular, a high thermal conductivity is expected to cause a larger effect. The thermal conductivity of silicon steel is around 30 W/K·m (Kowal et al., 2013), which is relatively high

compared, for instance, to that of Ti-6Al4V (7 W/K·m (Parry et al., 2016)). Therefore, the importance of the laser power-scan speed combination on build quality is expected to be higher for high-silicon steel.

7.4 Study of Solid Phase Formation and Microstructures

7.4.1 XRD Studies

The results obtained from XRD analysis (Figure 7-6) suggest that the microstructure of both the powder and SLM-processed samples is characterised by a single bcc-phase (i.e., α -ferrite), owing to the high silicon content that suppresses γ -phase formation. The SEM micrograph and EDX analysis reported in Figure 7-7 for sample S₂₈₀ support the XRD data in that the compositional range for iron and silicon corresponds to that expected for the solid solution.

With regards to the structural state of the ferritic phase, the amplified versions of the XRD spectrogram in Figure 7-6(a) show no superstructure lines. This suggests that coarse areas of ordered phases B₂ and DO₃ are not present in the microstructure. Further evidence of the absence of significant phase ordering in the SLM samples is provided by the analysis of the bcc lattice parameter reported in Figure 7-6(a). It is well known from the literature (Bozorth, 1993b) that the lattice parameter of the Fe-Si system decreases linearly with increasing Si content. The slope of the function describing the linear relationship changes abruptly at 5%Si, owing to the onset of atom ordering.

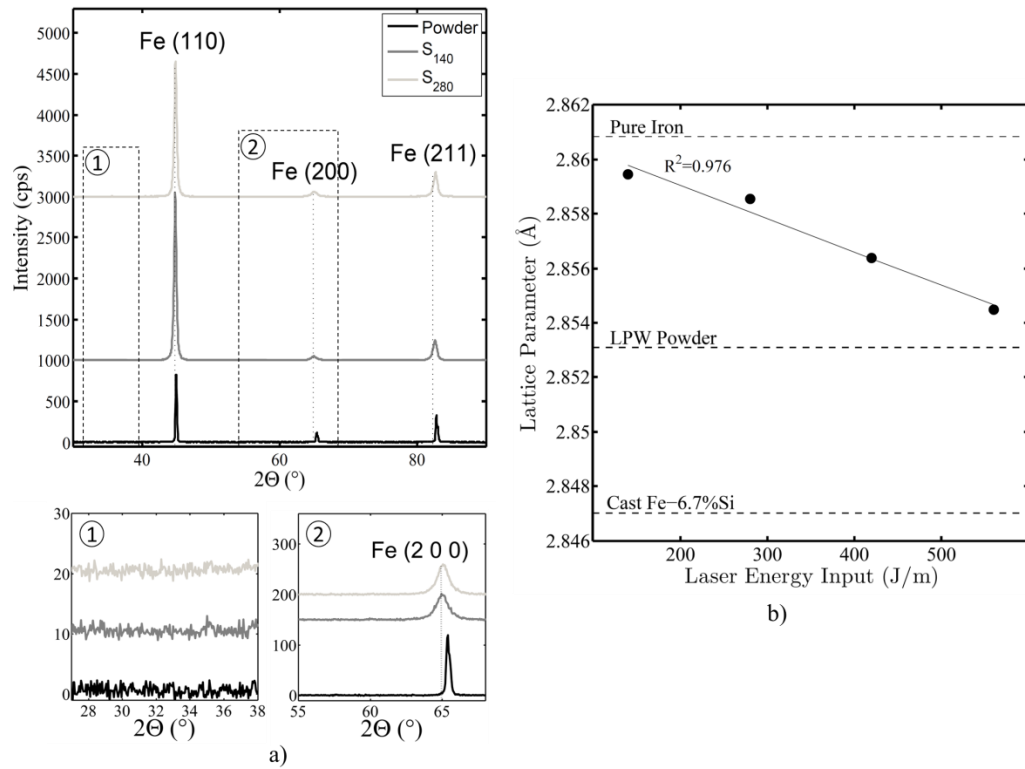


Figure 7-6: XRD analysis of SLM high-silicon steel. In a), the main and amplified plots show the XRD spectra for powder and samples S₁₄₀ and S₂₈₀. In b) the lattice parameters are shown for four samples processed using different energy inputs, the powder and the reference values of pure iron and cast Fe-6.7%Si from (Bozorth, 1993b).

The lattice parameter for the SLM high-silicon steel samples considered in Figure 7-6(a) lie below the reference value for pure bcc-Iron (upper limit) and well above that for cast Fe-6.7%Si, where superstructure lines in the XRD spectrum are known to appear due to ordering initiation (Bozorth, 1993b). The value for the unprocessed atomised powders is found to be smaller than for any of the processed samples, but still well above the value for cast Fe-6.7%Si. However, it is found that the lattice parameter of the SLM samples linearly decrease with decreasing laser scan speed (i.e., with increasing energy input), suggesting that phase reorganisation, although probably limited, might take place and is influenced by the thermal history of the samples during SLM processing.

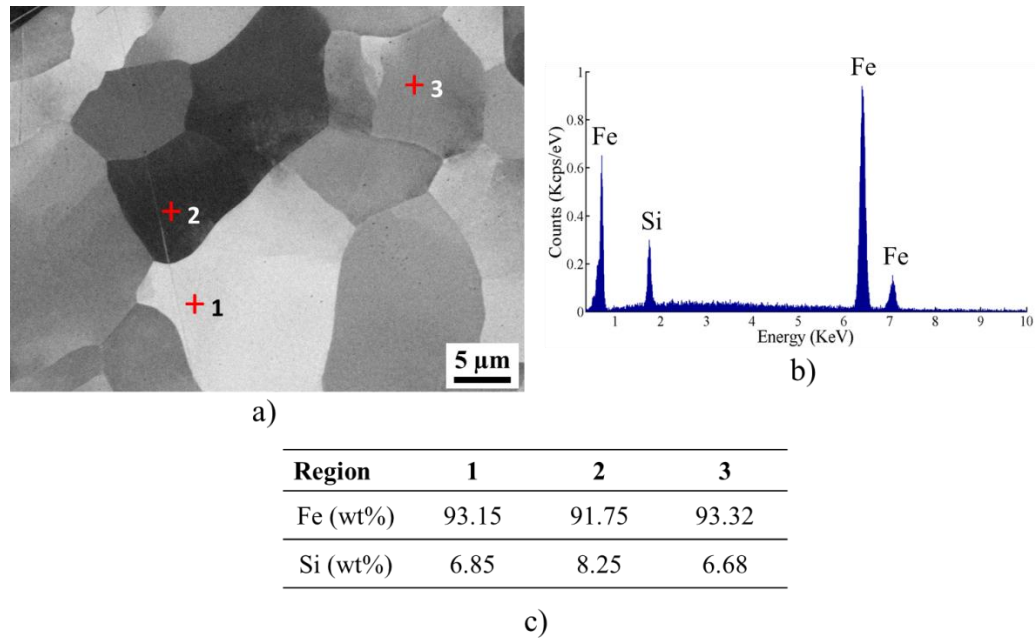


Figure 7-7: Grain composition of sample S₂₈₀ measured by EDX: SEM image of the investigated region (a), EDX spectrum corresponding to region 1 (b), and chemical composition in the three regions under exam (c).

7.4.2 Metallographic Studies

The SEM micrographs in Figure 7-8 and Figure 7-9 provide insights into the microstructural features of samples S₁₄₀ and S₂₈₀ for the top and front views (please note that the microstructures for the front view of the samples were also found for the side view and are therefore not shown here). Figure 7-8 shows that the grain boundaries of the processed samples are not always clearly identifiable. However, in the vertical plane the grains appear to present an overall columnar morphology and are more or less oriented along the build direction (BD), suggesting that epitaxial growth of the grains takes place, according to a grain selection process in which grains whose “easy growth” directions (the $\langle 001 \rangle$ family for the bcc system) are oriented along the direction of solidification outgrow the less optimally oriented grains. For

sample S₂₈₀ the grains appear more elongated and in some cases span several hundred microns along the BD.

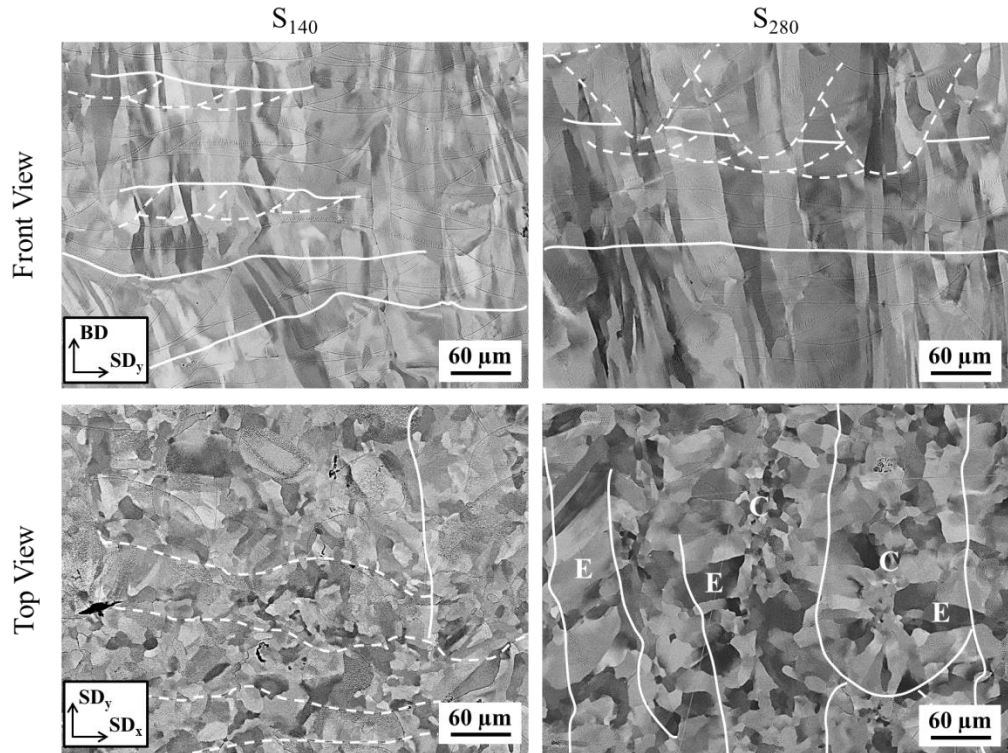


Figure 7-8: SEM micrographs for the top and front views of samples S₁₄₀ and S₂₈₀. The boundaries of some of the melt-pool cross-sections are highlighted by the dashed and solid lines, depending on the orientation. In the top view of sample S₂₈₀ the letter E and C indicate regions of elongated and equiaxed grains, respectively.

As regards the horizontal plane of the samples, the two micrographs at bottom of Figure 7-8 show that the grain structure of sample S₂₈₀ is coarser than that of sample S₁₄₀. Furthermore, the grains are equiaxed for sample S₁₄₀, whereas the grain structure of sample S₂₈₀ is more heterogeneous. In particular, two grain morphologies are of interest in the higher energy input sample. Regions of fine, equiaxed grains (indicated by the letter C in Figure 7-8) are found in correspondence of the scan track centreline. Adjacent to the centrelines (which are not always visible owing to melt-pool overlap), regions of coarser, elongated grains (identified by the letter E in Figure 7-8) are observed.

From the SEM images in Figure 7-8 the melt-pool boundaries are also visible and appear to significantly affect the grain structures of the two samples considered. In the vertical plane, the transversal melt-pool cross-sections of sample S₁₄₀ are shallow and flat, whereas in S₂₈₀ both shallow and protuberant melt-pools are observed, in similar proportions. Since melt-pool shape determines the orientation of the heat flow and consequently the solidification patterns, it significantly affects grain orientation. In this regard, the front view of the samples in Figure 7-8 suggest that in sample S₁₄₀ the longitudinal melt-pool cross-sections are more irregular than in sample S₂₈₀ and this causes the formation of a more heterogeneous grain orientation in the former sample.

In order to understand how shallow and protuberant melt-pools affect the grain structures, the SEM micrographs in Figure 7-9 offer a closer look at the two types of melt-pool shapes for the front view of the samples. Inside both melt-pool types, the grains are formed by cellular-dendritic solidification substructures that appear as thin stringy features stretching parallel to the longitudinal axis of the grains. The fact that the cellular substructures become visible upon etching suggests that excess-solute partitioning into the cellular spacings has occurred during solidification. The orientations of the solidification substructures in different regions of the melt-pools are indicated by the arrows in Figure 7-9. Inside the shallow melt-pools (Figure 7-9(c) and (d)) the solidification substructures grow mainly parallel to the BD, regardless of the location within the melt-pool, owing to the almost vertical orientation of the solidification front normal.

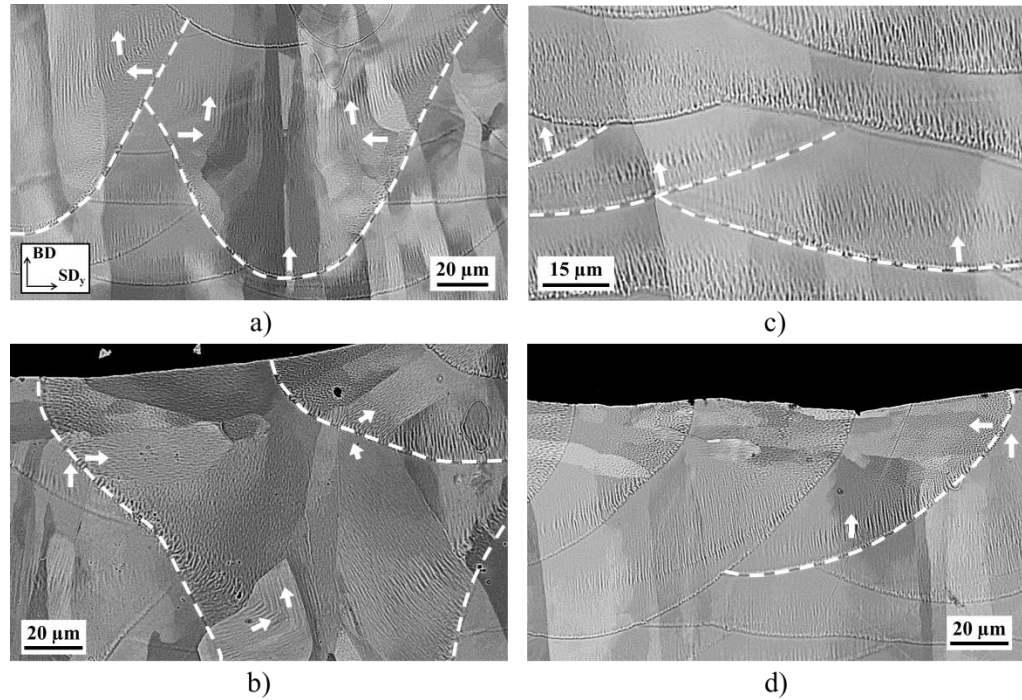


Figure 7-9: Cross-sections of deep and shallow melt-pools in the sample vertical plane. The grain and solidification structures inside deep and shallow melt-pools are visible for partially remelted layers (a and c, respectively) and for the top layer (b and d, respectively). The local growth direction of the solidification structures is indicated by the arrows. The dashed lines highlight the melt-pool boundaries.

In the protuberant melt-pools (Figure 7-9(a) and (b)) the growth path orientation depends on the location inside the melt-pool. In correspondence of the melt-pool centreline the cellular substructures and corresponding grains are oriented along the BD and grow epitaxially. In the regions adjacent to the centreline, growth has mainly occurred towards the centreline according to the inclination of the melt-pool boundaries. The grains in this region of the melt-pool correspond, in the horizontal plane of the sample, to the elongated structures identified by the letter E in the image at the bottom-right of Figure 7-8. In these regions epitaxial growth along the BD should be hindered by the inclination of the melt-pool boundary with respect to the BD. However, since the lateral melt-pool boundaries are oriented more or less along the BD in the

protuberant portion of the melt-pool, the cellular-dendrites growing direction in Figure 7-9A and B is often observed to shift by angles close to 90° in the vertical plane. In these situations the shift does not prompt the nucleation of a new grain since the grain orientation is not affected, owing to the orthogonality of the lattice vectors in the bcc crystal structure. Therefore, the existing grains can still grow across multiple melt-pools. The same mechanism is observed as the solidification structures growing from the lateral melt-pool boundaries approach the melt-pool centreline. Here, some grains are observed to shift their direction of growth abruptly by nearly 90° , as the images in Figure 7-9A and B also show.

At the top layer of the printed parts (Figure 7-9B and D), the upper portion of both the deep and shallow melt-pools is characterised by grains that have grown almost horizontally, following the direction of the solidification front, which is almost vertical in that region of the melt-pool. However, for the previous layers the upper part of the melt-pools is re-melted when a new layer is deposited and therefore will disappear from the structure.

In the presence of processing defects, such as a pore, the microstructure can be strongly distorted, as the example in Figure 7-10 shows. In this example, the presence of a large pore prevents the heat from dissipating through the normal conduction route, thus increasing the temperature reached in the melt-pool. The anomalous temperature rise leads to the formation of two regions of equiaxed grains. The first forms within the melt-pool tip. Here, the surrounding airgap blocks heat conduction and thus extends the cooling time of the melt-pool. This promotes grain growth, thus leading to the formation of a comparatively coarse microstructure. The second sub-region of equiaxed grains forms just above the

former. Here, the combination of the anomalously high temperature and the presence of the surrounding metal (which provides a conducting route for heat dissipation) leads to very fast cooling rates and thus to the formation of a very fine microstructure (i.e., high nucleation rates).

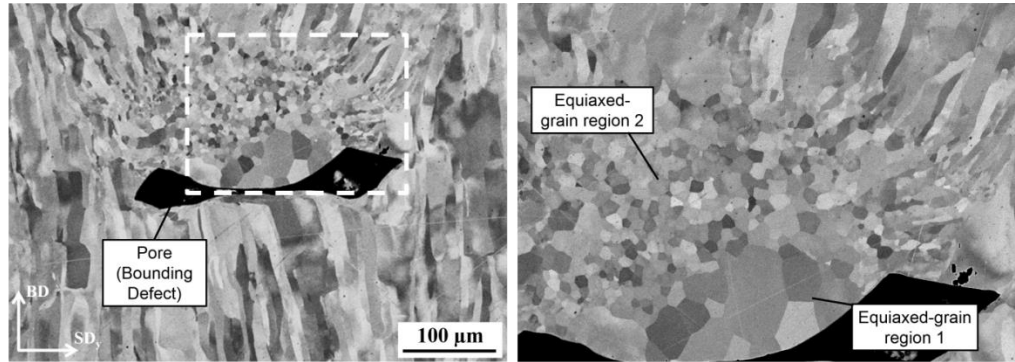


Figure 7-10: Example of the microstructural distortion caused by the presence of a large pore in the processed part. The micrograph on the right is a magnified version of the dashed inset.

7.4.3 Origin of Solid Phase Formation

The results from the XRD analysis presented in Figure 7-6 can be explained by considering the extremely high cooling rates (normally varying from 10^3 up to 10^6 °K/s in laser welding (David and Vitek, 1989)) at which the material solidifies during SLM processing. Such fast cooling, often results in non-equilibrium phase formation, i.e., the binary Fe-Si system is “frozen” into the disordered solid solution. The order of magnitude of the cooling rates during gas-atomisation is comparable to those reached during SLM, thus explaining the relatively high lattice parameter value of the LPW powder with respect to cast Fe-6.7%Si. Literature results (Shin et al., 2005) suggest that to suppress the B₂ ordered phase when the concentration of Si is 5.6%, an average cooling rate of 340 K/s is necessary. Therefore, the high average cooling rates achieved

during SLM processing can reasonably be expected to lie beyond the ordering suppression threshold.

The suppression of the ordered phases B_2 and DO_3 can be beneficial, since it makes silicon steels less brittle and electrically conductive (Shin et al., 2005, Jang et al., 2008). The at least partial suppression of phase ordering might be another advantageous aspect of using SLM for processing high-silicon steel. However, it should be noted that XRD alone is generally not sufficient to characterise phase ordering. As discussed in section 7.4, the processed parts are characterised by fine cellular grain substructures and by visible melt-pool boundaries. These features might indicate micro-segregation sites where intermetallic Fe_3Si could have potentially precipitated..

7.5 Study of Crystallographic Texture

7.5.1 EBSD Studies

EBSD data analysis shows a crystallographic texture in the grains of SLM high-silicon steel. Figure 7-11 shows the resulting PFs of two different samples. These are samples S_{140} (left-hand plots) and S_{280} (right-hand plots). The PF plots quantify the texture intensities for the three crystallographic direction families $\langle 001 \rangle$, $\langle 101 \rangle$ and $\langle 111 \rangle$ with respect to the front view of the sample. Both samples present a $\langle 001 \rangle$ texture along the BD, which is consistent with the morphological orientation of the columnar grains discussed in section 7.4. Figure 7-11 also shows that the fibre-texture in sample S_{140} is altered into a cube-texture by increasing the energy input to 280 J/m. In the cube-textured sample, the $\langle 001 \rangle$ axes are aligned with the BD and, in the scanning plane of the sample, with the directions running approximately

halfway between the SD_x and SD_y . However, a larger spread of the texture is observed around these directions than around the BD.

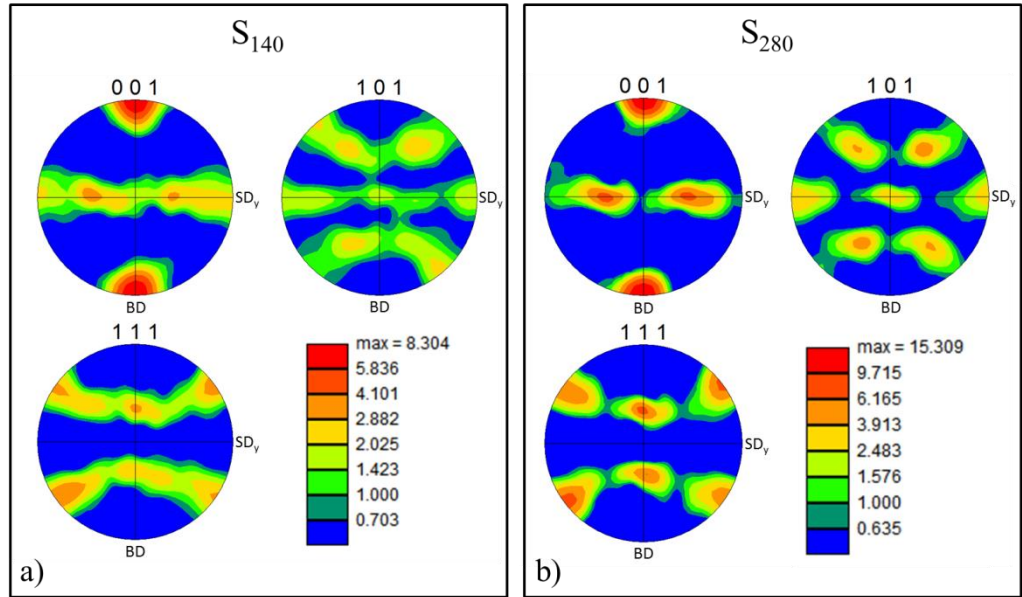


Figure 7-11: Pole figures (PF) showing the preferred orientation of the three crystallographic directions of easy, medium and hard magnetisation (i.e., $\langle 001 \rangle$, $\langle 101 \rangle$ and $\langle 111 \rangle$) for samples S_{140} (a) and S_{280} (b) with respect to the front view of the samples.

The maximum texture intensity of S_{280} is found to be almost twice larger than the maximum intensity of S_{140} . This suggests that the laser energy input largely affects the intensity of the texture, with a larger energy input producing a more textured sample. The EBSD maps of sample S_{280} in Figure 7-12 aim at providing an overview of the correlation between grain morphology and their $\langle 001 \rangle$ texture, both along the BD and, in the scanning plane, along the bisector of the angle between the SD_x and the SD_y . The colour-coded EBSD maps for the front view of the sample (Figure 7-12A) show that the majority of the long columnar grains of sample S_{280} preferentially orient their $\langle 001 \rangle$ axes along the BD (left image in Figure 7-12A). This result corroborates the hypothesis of epitaxial grain growth along the BD. Along the SD_x - SD_y bisector (right image in Figure 7-12A), a smaller portion of the grains is red coloured.

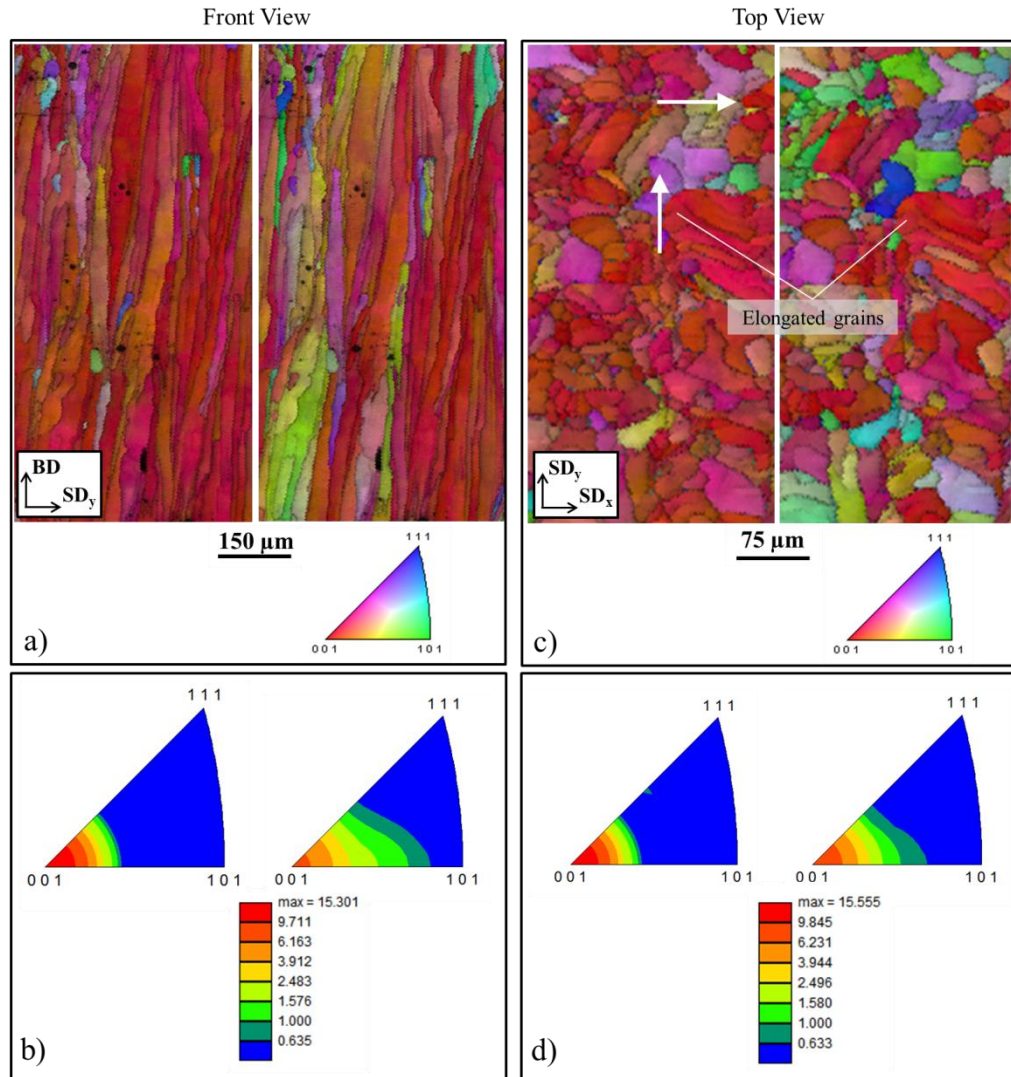


Figure 7-12: EBSD maps (a and c) and IPFs (b and d) for the front and top views of sample S₂₈₀. The maps/IPFs on the left of each frame show the texture found along the BD. The maps and IPFs on the right of each frame show the texture along the bisector of the angle formed by SD_x and SD_y. The white arrows in the EBSD image for the top view indicate the directions of two laser scan tracks that are still visible.

In the horizontal plane (Figure 7-12C) the majority of the grains are coloured red when the orientation of the <001> directions is considered along the BD (left image in Figure 7-12C), regardless of grain morphology. Conversely, the analysis along the SD_x-SD_y bisector direction (right image in Figure 7-12C) reveals that there is a correlation between the morphology of the grains and their crystallographic texture. In particular, the grains that present an equiaxed

morphology in the horizontal plane tend to be characterised by a $\langle 001 \rangle$ texture along the BD only (fiber-texture), whereas a large portion of the elongated grains present a $\langle 001 \rangle$ texture both along the BD and approximately along the SD_x - SD_y bisector directions (cube-texture). The latter more or less correspond to the orientation directions of the elongated grains. The IPFs (Figure 7-12B and D) are consistent with the EBSD maps in that in the scanning plane the crystallographic texture is characterised by a lower peak intensity and larger spread than along the BD.

7.5.2 *Origin of Texture*

Comparison between samples S_{140} and S_{280} has shown that doubling the laser energy input promotes columnar grain formation (Figure 7-8) and intensifies the crystallographic texture along the BD (Figure 7-11). The dependency of morphological and crystallographic texture intensities on laser energy input in SLM-processed parts had already been reported by the authors of (Niendorf et al., 2013), who found that by replacing a 400W laser beam source with a 1000W one, the morphological and crystallographic textures of an SLM austenitic stainless steel could be remarkably increased. The relationship can be explained by the increased amount of partial remelting and, as shown in Figure 7-8, by the improved smoothness of the deposited layers. These phenomena both occur when the energy input of the scanning laser is increased. Partial remelting of the grains whose crystallographic directions of “easy growth” (the $\langle 001 \rangle$ family in bcc systems) orient along the BD provides a substrate for the newly processed material to solidify epitaxially, as well-known from weld metal solidification theory (David and Vitek, 1989). Clearly,

a greater amount of optimally oriented grains will result in stronger texture intensities.

Figure 7-11 shows that increasing the energy input from 140 J/m to 280 J/m also alters the $\langle 001 \rangle$ fibre-texture into a cubic-texture, characterised by the tendency of the $\langle 001 \rangle$ axes to orient both along the BD and more or less along the direction of the SD_x - SD_y bisector. This has been associated to the appearance of deeper, protuberant melt-pools, whose solidification front is almost horizontal at the melt-pool centreline, but shifts its direction towards the BD at the melt-pool sides. The protuberant melt-pool shape determines a shift in the growth direction of the cellular-dendritic solidification structures of nearly 90° in the vertical plane with respect to the BD, as the images in Figure 7-9(a) and (b) have shown, owing to the orthogonality of the crystallographic axes in the bcc system, which causes the axis of the $\langle 001 \rangle$ family closer to the thermal gradient direction to be selected for growth.

In the scanning plane, the solidification structures that have grown almost horizontally have produced grains characterised by an elongated cross-section and oriented following the solidification front normal (refer to the grains identified by the letter E in Figure 7-8). Importantly, the morphological orientation has also prompted a crystallographic texture, and the EBSD results (Figure 7-11 and Figure 7-12) have shown that in the scanning plane the $\langle 001 \rangle$ crystallographic axes tend to align along a range of directions close to the SD_x - SD_y bisector, following the morphological orientation of the grains. The formation of a cube-texture for an SLM alloy is reported in the literature (Thijs et al., 2013a), where it is attributed to the employment of a bidirectional X-Y scan strategy. While in the present study the formation of the cube-texture is

attributed to the morphology of the melt-pool, understanding the effect of the scanning strategy on the development of the crystallographic texture would be of great interest. Unfortunately, a more extensive exploration of the effect of the SLM-process design space on the texture of the samples has been, to this date, limited by the alloy's brittle nature, which makes the creation of near-fully dense parts a challenge.

It is finally noteworthy that the nature of the specific alloy under investigation promotes texture formation, since the high silicon content suppresses γ -phase formation from the phase diagram (refer to the FeSi phase diagram shown in Figure 3-1) thus leaving no secondary phase to hinder ferritic grain development.

7.6 Effect of Annealing

7.6.1 Metallographic Studies

The optical and SEM micrographs in Figure 7-13 show the grain structure in the horizontal plane of the samples in the as-built condition as well as after annealing at 700 °C, 900 °C and 1150 °C. Grain size is within the 10-30 μm range for the as-built sample (Figure 7-13(a)), and remains unchanged after annealing at 700 °C (Figure 7-13(b)). Although not visible in Figure 7-13, along the BD the grains are columnar and can span several hundred microns owing to the grain-remelting/epitaxial growth mechanism previously discussed in Section 7.4. In Figure 7-13(c), the microstructure appears partially recrystallized after annealing at 900 °C, i.e., above the order-disorder transition temperature. The recrystallization is thought to be driven by the relief of the SLM-induced residual stresses, as suggested in the literature for other SLM-processed alloys, including iron and Inconel 718 (Sames et al., 2016). Besides

recrystallization, the SEM micrograph at the right of Figure 7-13(c) shows that annealing at 900 °C also leads to the complete disappearance of the melt-pool boundary lines, likely due to the dissolution of Si in the solid. However, the microstructure is not homogenised and Si-rich regions alternate with Si-depleted regions, as visible in the SEM micrograph. The micrographs of Figure 7-13(d) show the optical and SEM microstructures of a sample annealed at 1150 °C. The sample is characterised by a homogenised microstructure, characterised by a non-uniform grain size. In particular, while the largest grains are predominantly equi-axed and present sizes up to 300 µm, much smaller grains with sizes below 30 µm are also visible, suggesting that the recrystallization process is not complete.

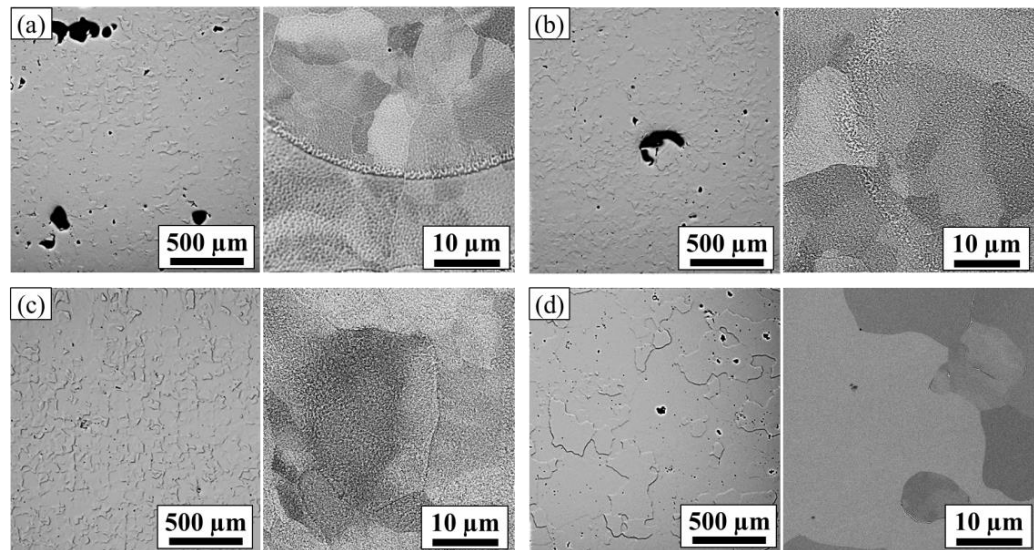


Figure 7-13: Optical and SEM micrographs of SLM Fe-6.9%wt.Si in the as-built condition (a) and after annealing at 700 °C (b), 900 °C (c) and 1150 °C (d). The micrographs reveal the microstructural features in the horizontal plane of the samples. The black features in the optical micrographs are pores.

7.6.2 XRD Studies

The XRD studies in Figure 7-14 shed light on the effect of annealing on the phase distribution and micro-stress relief. The XRD spectra in Figure 7-14(a)

indicate that the microstructures of the atomised powder and both the as-built and annealed samples are characterised by a single bcc-phase. However, the amplified version of the spectra in Figure 7-14(b) reveals that a super-lattice diffraction line associated with phase ordering of type DO_3 appears following annealing at 900 °C and 1150 °C. According to the phase diagram of the binary alloy Fe-Si, ordered phases of type $DO_3/B2$ in Fe-6.9%wt. Si are thermodynamically stable below approximately 830 °C (Swann et al., 1975). Therefore, the result that diffraction lines associated with ordering only appear after annealing at high temperatures suggests that the fast cooling rates experienced by the material during SLM partially suppress diffusion-driven ordering reactions, or at least prevent their coarsening. The appearance of an iron oxide (Fe_2O_3) diffraction peak in the spectrum of the sample annealed at 1150 °C is attributed to the leak of oxygen into the argon furnace from the surrounding environment during heat-treatment.

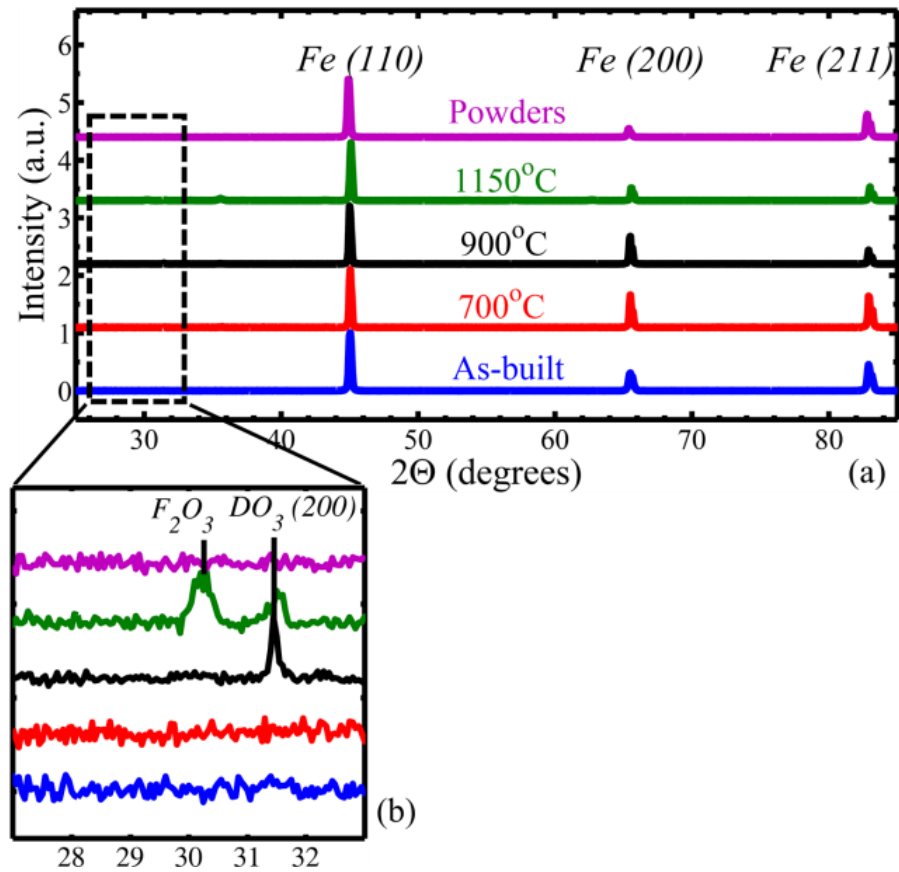


Figure 7-14: XRD studies performed on Fe-6.9%Si atomised powder and SLM samples, before and after annealing at various temperatures. The complete XRD spectra are in (a), while the amplified extracts of the spectra are in (b).

7.6.3 EBSD Studies

The EBSD maps in Figure 7-15 show the crystallographic texture of the grains along the BD, while the PFs show the preferred orientations of the $\langle 001 \rangle$ crystallographic directions with respect to the sample axes' directions. The EBSD maps and PF are only reported for the samples that underwent annealing at 700 °C, in Figure 7-15(a) and (b), and 1150 °C, in Figure 7-15(c) and (d). The EBSD analysis indicates that annealing does not alter significantly the $\langle 001 \rangle$ texture induced by SLM along the BD. In particular, the following texture strength index values were measured along the BD after various annealing conditions: 3.68 ± 0.82 for the as-built samples, 3.99 ± 0.86 for the samples annealed at 700 °C, and 4.78 ± 1.02 for the samples annealed at 1150 °C. This is an important finding, since high-temperature annealing can be performed

on the as-built samples to induce stress-relief and produce grain growth without weakening the beneficial crystallographic texture induced by SLM. As observed in the inset of Figure 7-15(b), which zooms in a region of fine microstructure, the finer grains are characterised by a random crystallographic orientation. Therefore, while it is not fully clear why recrystallization does not alter the as-built texture, it is believed that the as-built grains with a preferential orientation are more likely to grow by consuming the randomly oriented ones, due to the typically smaller size of the latter. Furthermore, it is possible that the recrystallization partially proceeds via the growth of nuclei that are preferentially oriented according to the as-built texture, in agreement to the oriented nucleation theory of recrystallization (Park and Szpunar, 2003).

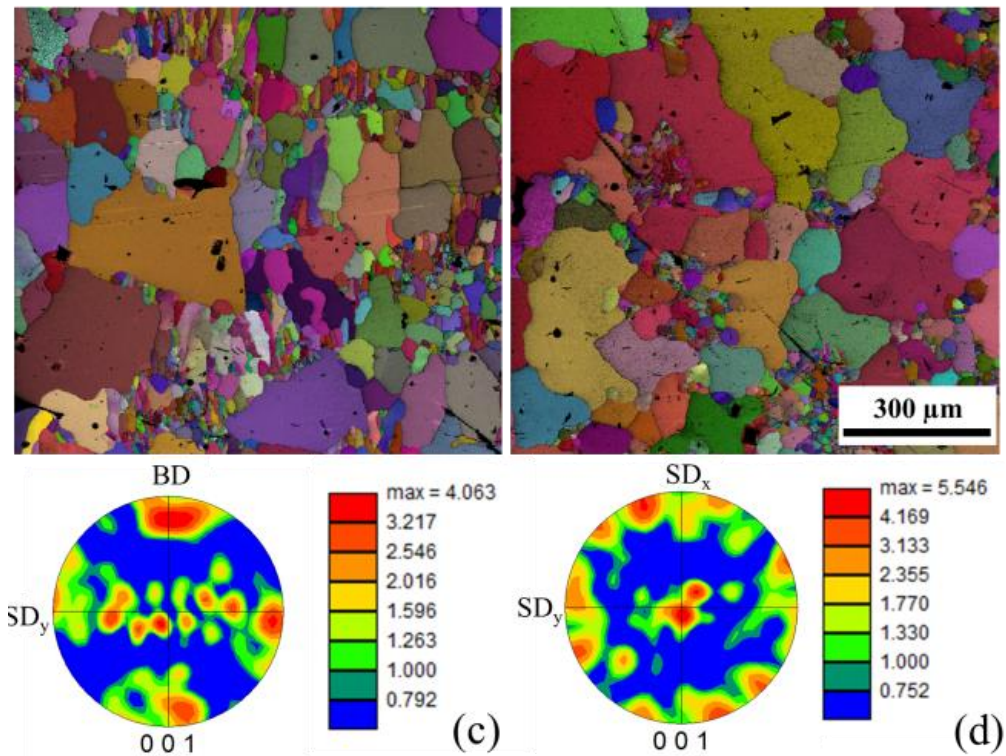
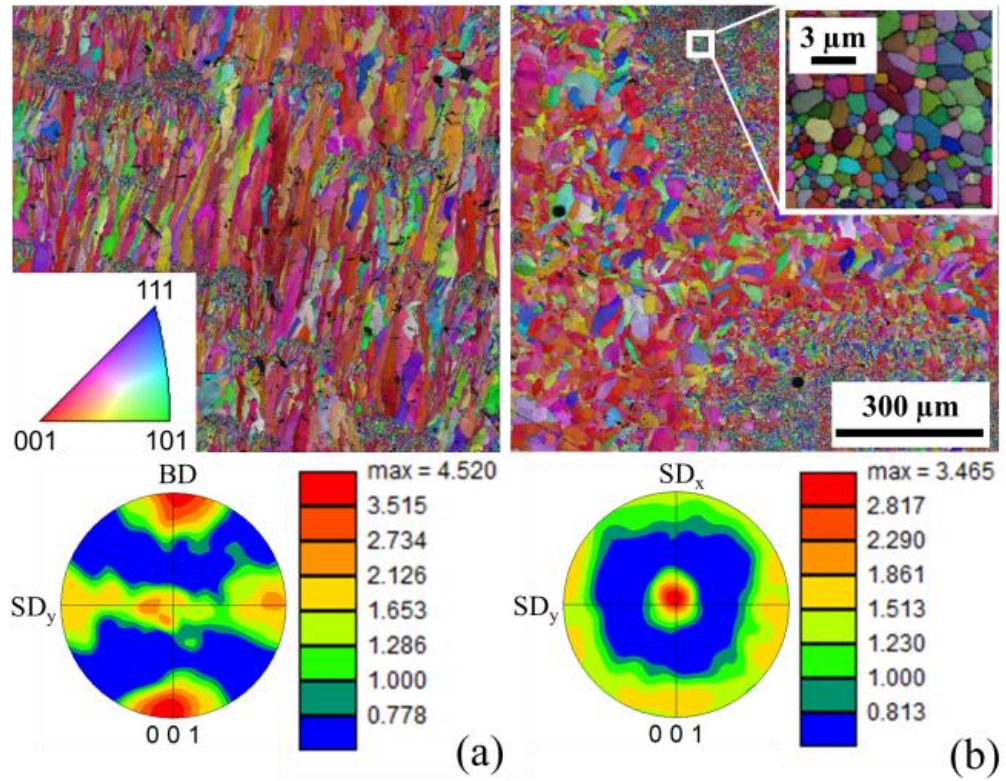


Figure 7-15: EBSD maps and PFs for samples annealed at 700 °C (a and b) and 1150 °C (c and d). The front and top views of the samples are presented on the left- and right-hand sides, respectively.

7.7 Summary

The main findings of this research can be summarised in the following points:

- Near-fully dense high-silicon steel three-dimensional parts were manufactured using SLM for the first time in this work, starting from pre-alloyed powder;
- While the overall effect of increasing the energy input of the processing laser is to reduce bonding defect formation, at fixed energy input a high laser power is more effective than a low scan speed in fostering powder melting and reducing the presence of bonding defects in the manufactured part. However, a higher laser power level prompts crack and spherical pore formation. Therefore, a medium laser power level (70 W on a Realizer SLM50 system) is recommended for the processing of high-Si steel;
- The SLM high-silicon steel is a single ferritic phase and superstructure lines could not be observed from the XRD measurements, suggesting the fast cooling rates of SLM retard phase ordering formation;
- The microstructure of the samples is characterised by columnar grains oriented towards the BD. Increasing the processing laser energy input seems to promote grain elongation both along the BD and in the scanning plane of the sample;
- By increasing the laser energy input, a crystallographic <001> fibre-texture can be altered into a cube-texture. This has been associated with a change in the qualitative shape of the melt-pool from shallow to protuberant.

- Annealing at 1150 °C for 1 hour produces a recrystallized microstructure characterised by predominantly equi-axed grains with sizes of up to 300 µm. Importantly, annealing coarsens the grain size while retaining the SLM-induced crystallographic <001> texture along the BD.

In conclusion, this research has shown that SLM might potentially open the path to an alternative route for the manufacture of grain-oriented high-silicon steels for magnetic applications, since the <001> directions are the so-called “easy axes” of magnetisation and the grain size obtained after annealing is in general desirable for magnetic applications, as previously described in Section 3.2 and shown in Figure 3-4. How the metallurgy affects the magnetic properties is the subject of the next chapter.

8. Magnetic Properties of High-Silicon Steel Produced by Selective Laser Melting: Effect of Laser Energy Input and Annealing Temperature

8.1 Introduction

In this chapter the effect of laser energy input and post-processing annealing on the magnetic properties of high-silicon steel ring samples are studied. The magnetic properties are first measured and discussed for 4 mm-high rings obtained using three laser energy inputs that provide the best compromise in terms of porosity and crack formation. These were determined in Section 7.3. For the study of the effect of the process parameters, a post heat-treatment was performed that releases most of the SLM-induced stresses while keeping the grain structure unchanged. This post heat-treatment step makes sure that the measured magnetic properties can be interpreted in terms of the microstructures, rather than in terms of the SLM-induced residual stress.

Since the relatively small grain size obtained with SLM is thought to determine the hysteresis losses and limit the magnetic permeability, for one of the laser energy input levels considered, the magnetic properties of 2 mm-high rings are studied with respect to the temperature of post-process annealing, whose effect on the microstructure was presented in Section 7.6.

8.2 Stress-Relief of SLM Cubic Samples

The histogram in Figure 8-1 shows the effect of stress-relieving the SLM sample at a temperature of 700 °C for one and five hours on the magnitude of the non-uniform lattice strain of the SLM high-Si steel samples, as estimated

using XRD and the Williamson-Hall method (Section 6.9.3). The values are reported for a cubic sample processed using laser power $P=70$ W and $v=0.5$ m/s in the as-built condition, after heat-treatments at 700 °C for 1 hour, and after heat-treatment at 700 °C for 5 hours. The reference strain value for the atomised powder, which is assumed to be stress-free, is also shown. The observed trend shows that after 1 hour at 700 °C about half of the non-uniform lattice strain has been recovered, while after 5 hours the strain has dropped to the value measured for the stress-free powder. This result indicates that crystal lattice recovery by residual stress relief is complete after annealing the samples at 700 °C for 5 hours. It is noted that the high values of the coefficients of determination R^2 , also reported in Figure 8-1 above the corresponding bars, indicates the good fit between the measured XRD peak broadening and the Williamson-Hall model.

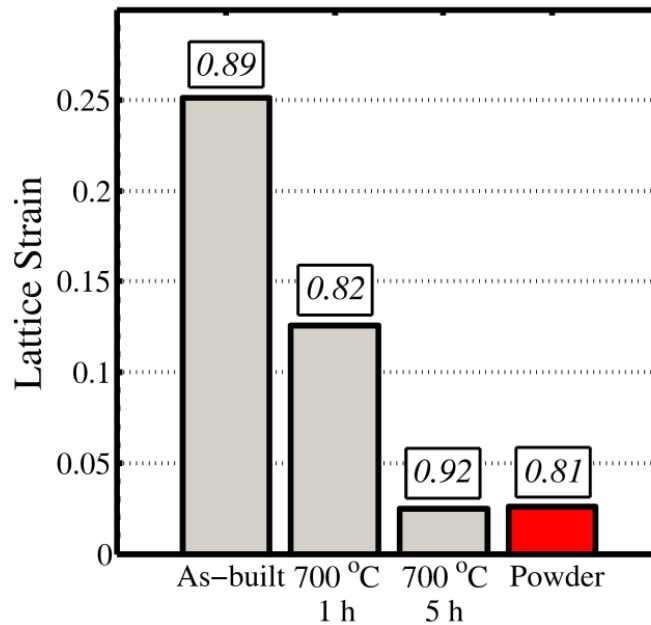


Figure 8-1: Effect of heat-treatments on the non-uniform lattice strain in SLM high-Si steel and atomised powder. The numbers above the bars are the values of the coefficient of determination R^2 . The values reported refer to the sample processed using laser power $P=70$ W and scan speed $v=0.5$ m/s.

The optical micrographs for the front cross-section of a sample processed using $E=140 \text{ J/m}$ ($P=70 \text{ W}$, $v=0.5 \text{ m/s}$) before and after stress-relief are shown in Figure 8-2(a) and (b) respectively. The images suggest that the performed heat-treatment does not affect the grain size, in agreement with the results previously reported in (Faudot et al., 1989) for high-Si steel microcrystalline ribbons. While the stress-relief treatment does not prompt grain-growth, it causes the microstructure to homogenise, as the SEM micrographs in Figure 8-2(c) and (d) indicate. The homogenisation effect was already visible after 1 hour at $700 \text{ }^{\circ}\text{C}$, as previously shown in Section 7.6.1, where the effect of the annealing temperature on the microstructures of SLM high-silicon steel samples was studied in greater detail. In particular, the solidification structures associated with the micro-segregation of Si at the melt-pool and cell boundaries are clearly visible in the as-built microstructure, but appear significantly attenuated in the stress-relieved microstructure of Figure 8-2(d). The observed homogenisation is attributed to the diffusion-driven dispersion of the micro-segregates into the solid solution taking place during stress-relief.

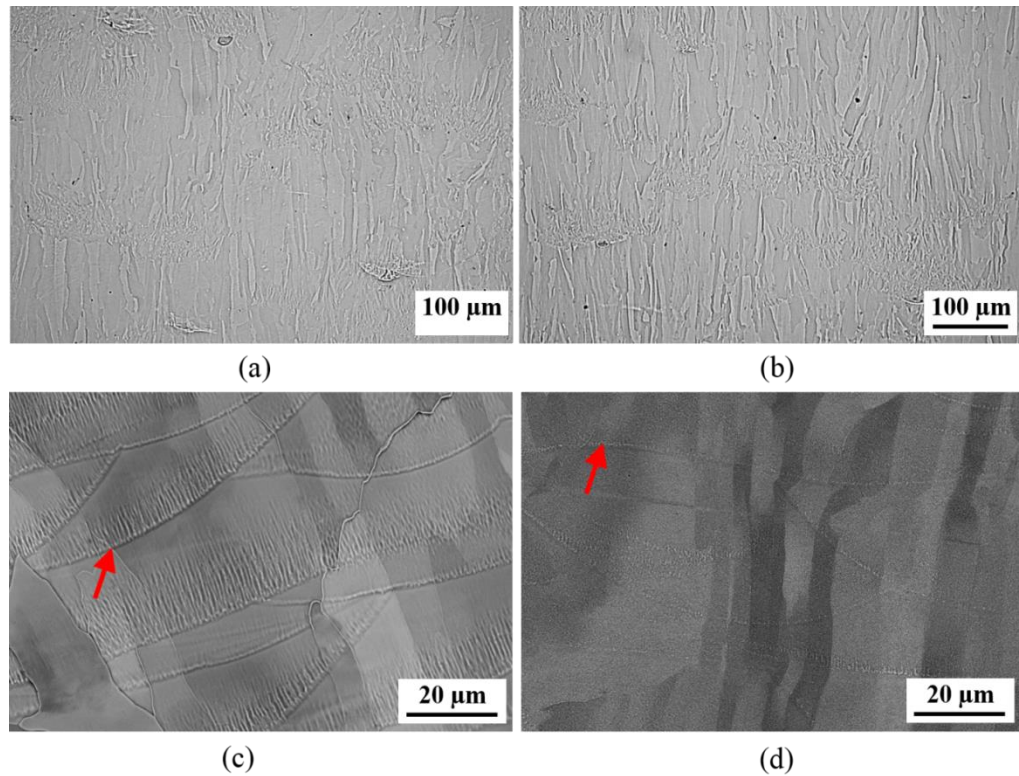


Figure 8-2: Optical and SEM micrographs of a sample processed using a laser energy input $E=140$ J/m before ((a) and (c)) and after 5 hours of stress relief at 700 °C ((b) and (d)). The red arrows in the SEM micrographs indicate the fusion lines (melt-pool boundaries). Both the optical and SEM micrographs reveal the microstructural features in the vertical plane of the samples (front view).

Figure 8-3 shows the evolution of the Vickers micro-hardness (HV) with increasing laser energy input at fixed laser power $P=70$ W, before and after stress-relief. While HV does not seem to be affected by the processing parameters, the stress-relief step causes a significant increase in the HV of the high-Si samples. The observed increase in HV following the stress-relief treatment is attributed to the redistribution of the Si-rich micro-segregates shown in the SEM micrographs of Figure 8-2(d). While phase-ordering is known to increase the HV of Fe-Si alloys (Faudot et al., 1989), whether this also contributes to the increase in HV following stress-relief is not clear, since no direct evidence of phase-ordering reactions could be provided using the employed techniques.

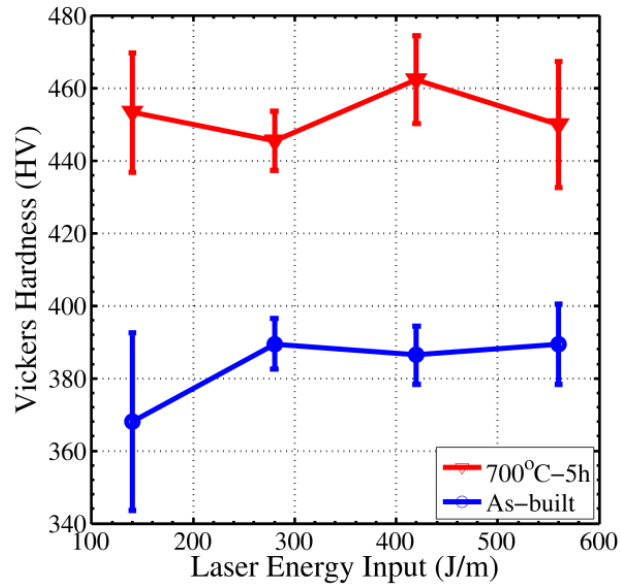


Figure 8-3: Dependence of Vickers micro-Hardness (HV) on processing laser energy input, before stress-relief (blue line) and after stress-relief (red line).

8.3 Effect of Laser Energy Input on the Magnetic Properties

In Figure 8-4 the measured initial magnetisation properties in quasi-static conditions are plotted against laser energy input E . In Figure 8-4(a), μ_{max} does not appear to change significantly for the two samples at lower E , but appears to decrease only slightly at $E=420$ J/m. A different behaviour is found for μ_{40} in Figure 8-4(b), which increases significantly (by around 50%) from $E=140$ J/m to $E=280$ J/m, before decreasing again slightly at $E=420$ J/m. As shown in Figure 8-4(c), the value of B_{50} is significantly different for the three laser energy inputs under study, and $E=280$ J/m provides again the best property.

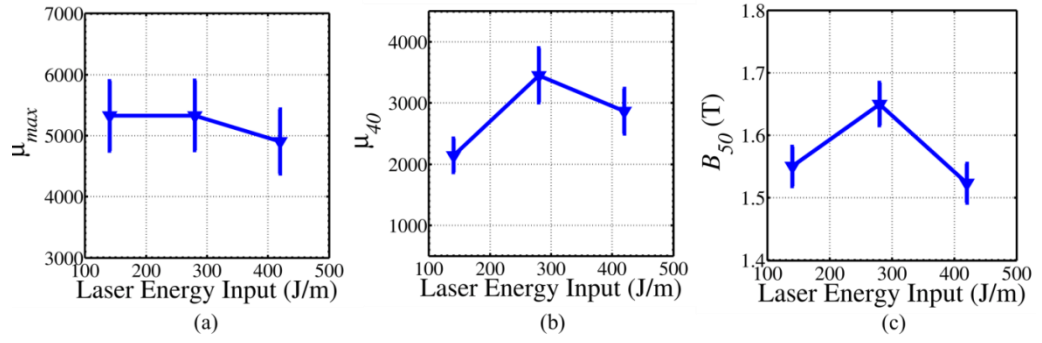


Figure 8-4: Dependence of the initial magnetisation properties of the ring samples on laser energy input. The properties shown are μ_{max} , i.e., the maximum permeability (a), the initial μ_{40} , i.e., the permeability at $H=40$ A/m (b), and B_{50} , i.e., the flux densities at $H=5000$ A/m (c).

Regarding the loss performance of the ring samples, Figure 8-5 shows the dependence of B_r and H_c on the laser energy input in quasi-static conditions (Figure 8-5(a) and (b), respectively) and of the total power loss P_{50-1} (Figure 8-5(c)). The results show a decrease in H_c from 140 J/m to 280 J/m. A further increase in laser energy input does not seem to affect H_c . The remanence, B_r , shows a slowly decreasing behaviour throughout the laser energy input range considered. In Figure 8-5(c), P_{50-1} shows a strong non-monotonous trend, with a clear minimum value observed at $E=280$ J/m.

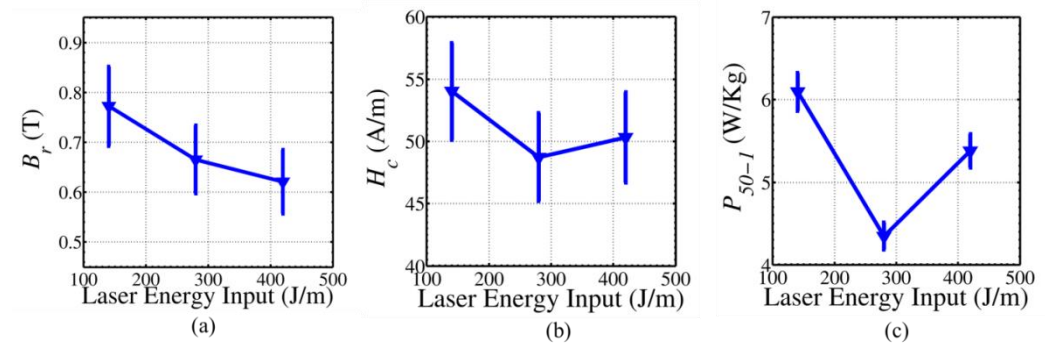


Figure 8-5: Loss performance of ring samples at increasing laser energy input. The values of B_r and H_c in quasi-static conditions are in (a) and (b), respectively, while the total power losses P_{50-1} are in (c).

The presented trends are explained in the remainder of this section in terms of the various interacting metallurgical properties of grain size, defects and crystallographic texture. In soft magnetic materials a finer grain structure leads to poorer quasi-static magnetisation properties and higher hysteresis losses owing to the higher concentration of grain boundaries, which act as pinning sites for the domain walls, thus impeding their motion during the magnetisation/demagnetisation process (Hilzinger and Rodewald, 2013). However, since large grains cause eddy current losses to increase, grain size has to be designed to match the working conditions at which the material is exposed, especially the frequency of the magnetic field (De Campos et al., 2006). However, it is well-established that at 50 Hz the grain size can increase up to 5-7 mm without inducing significant dynamic loss (Cullity and Graham, 2011). Manufacturing defects, such as pores and cracks, represent pinning sites impairing the movement of the magnetic domain walls, thus decreasing the permeability and increasing the coercivity and remanence, as well as the hysteresis loss. If large enough, pores and cracks represent airgaps from which demagnetisation fields emanate, thus resulting in sheared B-H curves. With an increased degree of shearing, the measured permeability and remanence are diminished and do not correspond to the material's intrinsic properties. On the other hand, shearing does not affect the coercivity H_c (Hilzinger and Rodewald, 2013). With regards to the crystallographic texture, this affects the preferred mode of magnetisation/demagnetisation and eventually the shape of the quasi-static B-H loop. In particular, in a polycrystalline sample whose grains have a preferential crystallographic orientation (strong crystallographic texture), if the external field is applied in the direction of the grains' easy axis of

magnetisation (the $\langle 001 \rangle$ axis family in Fe-Si) fast magnetisation through domain wall motion will result in a B-H loop characterised by a square shape, with high maximum permeability, but typically low initial permeability, low hysteresis loss and a high remanence. On the other hand, if the applied field is aligned with the medium or hard magnetisation axes (the $\langle 101 \rangle$ and $\langle 111 \rangle$ crystallographic directions in FeSi, respectively), the material's B-H loop assumes a round shape (i.e., the remanence is low), and is characterised by a low, constant permeability (Hilzinger and Rodewald, 2013) and, importantly, by a high hysteresis loss (Bozorth, 1993a), especially at high induction values (Landgraf et al., 1999). The above notions provide an insight into the complex metallurgical effects contributing to the observed magnetic behaviour of the material, and are used as a basis to interpret the results of the present research.

In Figure 7-2 we have shown that increasing the energy input, E , of the processing laser prompts part densification. In particular, a decrease in porosity by approximately 1% can be achieved by increasing the energy input from 140 J/m to 280 J/m. The observation that the maximum relative permeability μ_{max} does not change between these two laser energy input levels, while the initial permeability μ_{40} shows a drastic improvement, can be interpreted in light of the fact that the pinning forces introduced by the defects in the lattice of the part can be overcome if enough magnetomotive force is provided, thus only affecting the properties at low applied field strengths. The decrease in porosity with increasing laser energy input from 140 J/m to 280 J/m also seems to justify the decrease in remanence B_r , coercivity H_c , as well as the drop in the power losses P_{50-1} . In Figure 7-8 we showed that the grain size tends to coarsen with increasing E . However, the coarsening is non-uniform across the

horizontal plane of the samples and therefore it does not seem to affect their Vickers micro-hardness, as was shown in Figure 8-3. Therefore, while both grain coarsening and decrease in porosity could contribute to the magnetic behaviour of the samples, the latter is expected to represent the biggest contribution.

Deterioration of the total power losses P_{50-1} observed at the highest laser energy input considered (420 J/m) suggests that the beneficial effect of porosity reduction and, possibly, of grain coarsening, is counteracted by one or more opposing effects. We have shown in Figure 7-2 that crack formation is promoted at high laser energy inputs as sample porosity decreases. Cracks could contribute to the shearing of the B-H loop, i.e., the modification of the intrinsic magnetic properties of a material owing to the introduction of an airgap (Hilzinger and Rodewald, 2013). However, since the observed cracks primarily run parallel to the horizontal plane and thus to the magnetisation direction, they are expected to contribute only partially to the shape of the B-H loop. Furthermore, it seems unlikely that the pinning effect of the crack surface irregularities has a visible effect, as this would result in an increase in remanence B_r .

A metallurgical condition that could help to explain the magnetic behaviour of the material processed at high laser energy input is the crystallographic texture. In Section 7.5 we showed that the $\langle 001 \rangle$ fibrous texture obtained at low laser energy inputs (around 140 J/m) can be changed into a stronger cubic texture with higher laser energy, due to epitaxial grain growth and the change in melt-pool shape from shallow to protuberant. In particular, in the observed cubic texture the $\langle 001 \rangle$ axes in the horizontal plane of the samples orient at around

45° from the laser scanning directions (i.e., the X- and Y-axes). The diagrams in Figure 8-6 provide a schematic representation of the expected orientation of one $\langle 001 \rangle$ crystallographic direction with respect to the local direction of the magnetic field during magnetic testing of ring samples built at low and high laser energy inputs (Figure 8-6(a) & (b) respectively), based on the EBSD results presented in Section 7.5. The diagrams suggest that as the texture becomes markedly cubic (Figure 8-6(b)), the easy axis of magnetisation is aligned with the magnetic field direction only at around 45° from the X- and Y-axes, thus causing the horizontal plane to lose its isotropic character. As previously mentioned, an increase in misalignment between the magnetic field and the $\langle 001 \rangle$ axes is well-known to produce a decrease in permeability, an increase in hysteresis loss, as well as a decrease in remanence B_r , while the coercivity H_c is barely affected (Hilzinger and Rodewald, 2013, Bozorth, 1993a). Therefore, while increasing the energy input of the processing laser can be initially beneficial in that it strengthens the preferential orientation of one of the $\langle 001 \rangle$ axes along the BD, thus decreasing the number of $\langle 111 \rangle$ fibres (i.e., the axes of hard magnetisation) in the horizontal plane, too much laser energy proves detrimental for the magnetic properties due to the creation of a cubic texture.

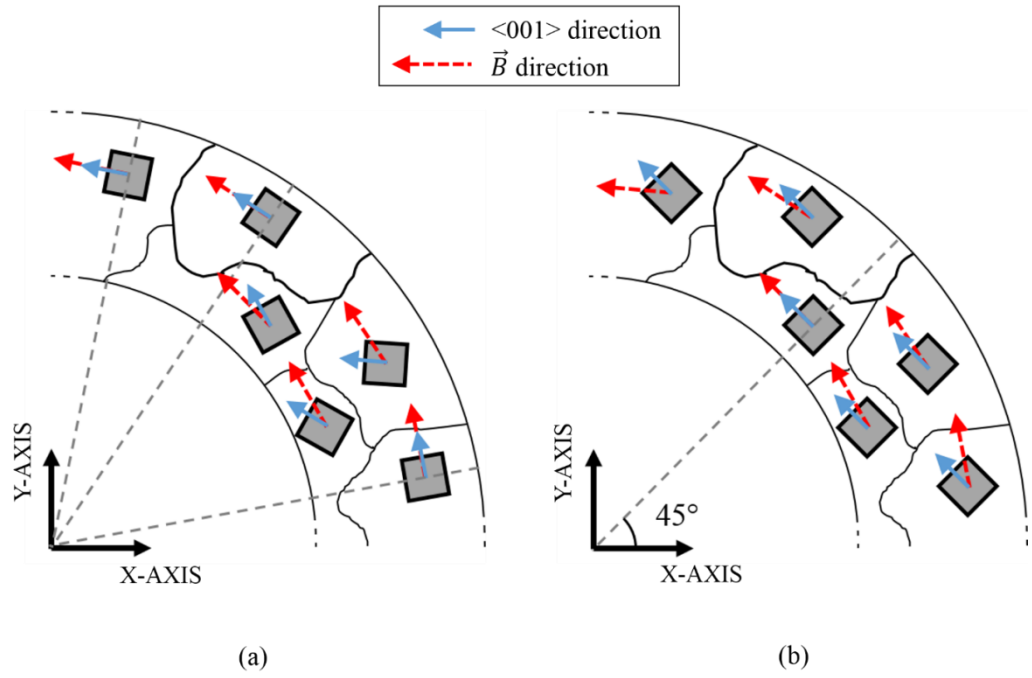


Figure 8-6: Expected orientation of one <001> crystallographic direction (solid arrows) in the horizontal plane with respect to the laser scanning directions (X and Y) and to the local orientation of the magnetic field during magnetic testing (dashed arrows), based on the EBSD results presented in Section 7.5. Samples produced using low (a) and high laser energy input (b) produce different crystallographic textures. Only one quarter of the ring samples is shown in these diagrams.

Finally, it is important to observe that the overall magnetic performance of the SLM high-Si steel presented in this manuscript is still somewhat distant from that of commercially available high-Si steel products, such as *JNEX* grade electrical steel (Namikawa et al., 2005). This is mainly attributed to the comparatively small size of the grains obtained using SLM. While the latter was observed to be approximately in the order of 20 μm , the optimal size for soft magnetic applications has been suggested to be of approximately 150 μm (De Campos et al., 2006). Therefore, heat-treatments aimed at producing substantial grain growth are investigated in the next section.

8.4 Effect of Annealing

The quasi-static and dynamic magnetic properties of the as-built and annealed ring samples are shown in Figure 8-7. For comparison, the magnetic properties of a commercial grade of 0.1 mm-thick high-Si electrical steel, namely the JNEX Super Core by JFE Steel (JFE, 2015) are also reported in Figure 8-7. In Figure 4(a) it can be seen that the maximum permeability, μ_{max} , of the SLM steel increases with increasing annealing temperature from a value of approximately 2000 in the as-built condition to more than 24000 after annealing at 1150 °C. Until 900 °C the curve increases with constant slope under the effect of stress relief, which can be reasonably expected to be completed after annealing at 900 °C, based on the data reported for low-alloy steels in the welding literature (Welding Handbook, 1976). Between 900 °C and 1150 °C the slope of the curve increases, and this is attributed to grain growth, which has been shown in Figure 7-13 to occur within this temperature range. Both stress relief and grain growth reduce the density of lattice defects, which are known to have a detrimental effect on the permeability by acting as pinning sites for the magnetic domain walls and therefore hindering their motion (Hilzinger and Rodewald, 2013).

The plots in Figure 8-7(b) show that induction B_{10} increases with annealing temperature from 0.95 T for the as-built samples to 1.3 T after annealing at 700 °C. Increasing the annealing temperature beyond 700 °C does not improve B_{10} further, suggesting that most of the residual obstacles hindering domain wall motion can be overcome at the considered applied field. From Figure 8-7(b) we also observe that B_{50} is only moderately affected by annealing, since it increases from 1.5 T for the as-built sample to 1.6 T after annealing at 1150 °C.

This indicates that the considered applied field of 5000 A/m is strong enough to overcome most of the manufacturing-induced obstacles to magnetisation. The hysteresis loss in quasi-static conditions, expressed in terms of H_c (Figure 8-7(c)) decreases with constant slope until 900 °C. Increasing the annealing temperature further causes the slope to become even more negative and the curve reaches a minimum of 16 A/m at 1150 °C. It is noteworthy that while permeability μ_{max} and coercivity H_c vary over one order of magnitude within the annealing temperature range considered, their product is approximately constant, because both properties are determined by the extent of the pinning force that the lattice defects and stress centres exert on domain wall motion (Hilzinger and Rodewald, 2013). The remanence B_r (Figure 8-7(c)) presents a non-monotonic dependence with annealing temperature. Specifically, B_r increases up to 700 °C but, as the annealing temperature increases further it decreases to 0.8 T after annealing at 1150 °C. The initial increase in B_r is an effect of the residual stress relief, which reduces the “shearing” of the hysteresis curve introduced by the lattice deformation during SLM (Landgraf and Emura, 2002). On the other hand, grain enlargement reduces the defect density (grain boundaries), which hinder demagnetisation (Landgraf et al., 1999), thus leading to a reduction in B_r at high annealing temperatures.

The total power loss in dynamic conditions is shown in Figure 8-7(d). The loss decreases with increasing annealing temperature from approximately 4.3 W/Kg for the as-built sample to 2.2 W/Kg upon annealing at 1150 °C. The trend of the total power loss is qualitatively similar to that of H_c , suggesting that the eddy current loss component does not increase significantly with grain growth. This finding is in agreement with the well-established notion that at 50 Hz the

grain size of electrical steels can grow up to 5–7 mm without prompting significant loss associated with increased domain wall displacement speed (Cullity and Graham, 2011). Therefore, the considered annealing temperature-time combination of 1150 °C – 1 hour reduces the hysteresis loss component without inducing extra dynamic loss. Importantly, the plots in Figure 8-7(a)-(c) have shown that the quasi-static magnetic properties of the SLM material after annealing at 1150 °C for 1 hour compare favourably with those of the JNEX Super Core. The total loss for the SLM steel at $f=50$ Hz is approximately four times higher than that for the JNEX steel, which is due to the much higher thickness of the SLM rings (2 mm) compared to the laminated steel of the JNEX (0.1 mm). In Figure 8-8, a concise comparative view of the magnetic properties of SLM high-silicon steel after optimal annealing (1150 °C for 1 hour) and those of the reference JNEX Super Core is provided.

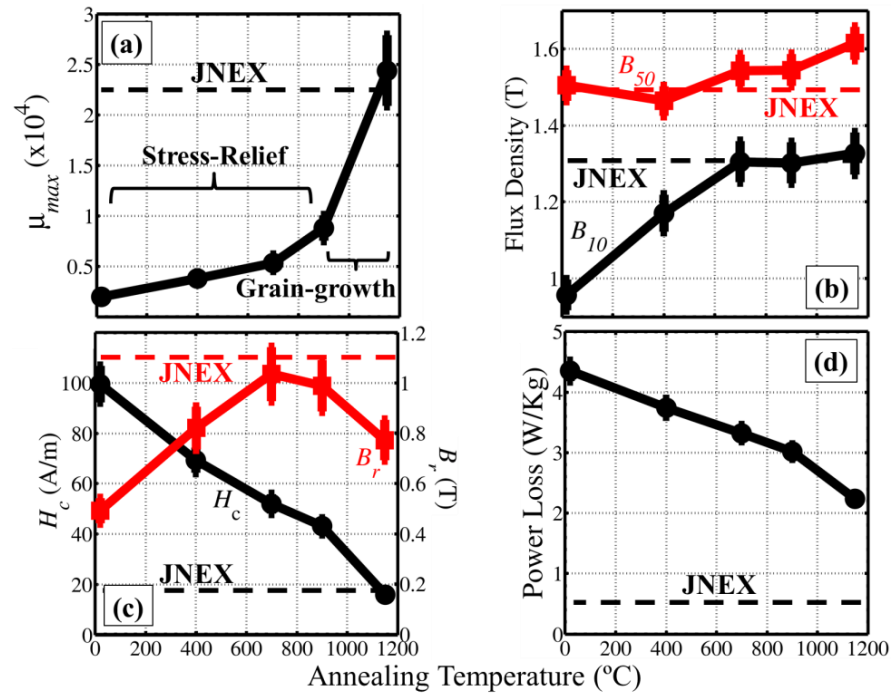


Figure 8-7: Evolution of the magnetic properties of SLM Fe-6.9%wt.Si with annealing temperature: μ_{max} (a), B_{10} and B_{50} (b), H_c and B_r (c) and power loss at $B_{max} = 1$ T and $f = 50$ Hz (d). The dashed lines indicate the magnetic properties of the JNEX Super Core, from (JFE, 2015).

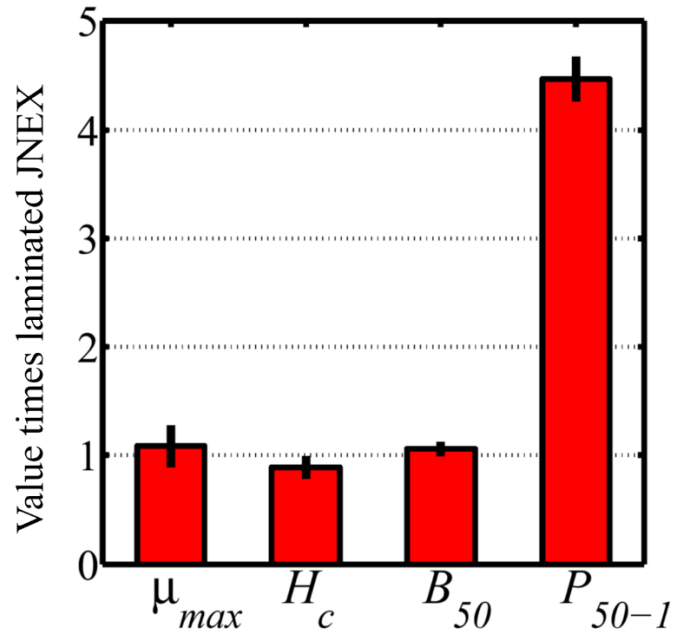


Figure 8-8: Comparison between the magnetic properties of SLM Fe-6.9%wt.Si after annealing at 1050 °C and laminated counterpart JNEX. The magnetic properties considered are the maximum permeability (μ_{max}), coercivity (H_c), flux density at $H=5000$ A/m (B_{50}), remanence (B_r) and total power losses at excitation frequency $f=50$ Hz and maximum flux density $B=1$ T (P_{50-1}). The value of each property is expressed times the reference value reported for JNEX Super Core.

8.5 Summary

The best magnetic properties were observed for the stress-relieved samples processed using a laser power $P=70$ W and scan speed $v=0.250$ m/s. A maximum relative permeability $\mu_{max}=5300$, a coercivity $H_c=49$ A/m and power losses $P_{50-1}=4$ W/Kg were found for this sample. While the interpretation of the relationship between magnetic properties and laser energy input is complicated by the large number of metallurgical effects at play (including defects, grain size distribution and crystallographic texture), we propose that changes in porosity and crystallographic texture are the main factors involved.

The samples obtained using $P=70$ W and $v=140$ m/s have been heat-treated at increasing annealing temperatures in order to release the residual stress and induce grain growth. Results have shown that annealing at 1150 °C for 1 hour

produces a recrystallized microstructure characterised by predominantly equiaxed grains with sizes of up to 300 μm . Importantly, annealing retains the SLM-induced crystallographic $\langle 001 \rangle$ texture along the BD of the sample. The quasi-static magnetic properties achieved upon the straight-forward and inexpensive annealing procedure (maximum relative permeability of 24000 and coercivity of 16 A/m) compare favourably with those reported for commercial high-Si steel laminations. Therefore, the annealed material has potential for the development of novel core geometries, especially in applications where complex geometrical core design is advantageous, while the employment of a laminated core is not viable due to structural integrity concerns, or is not strictly necessary.

9. Methodology for Electrical Machine Rotor Core Design for Production Using Laser Additive Manufacturing

9.1 Introduction

This chapter describes the Finite Element Analysis-based, topology optimization (TO) methodology for the combined magnetostatic and structural design of a Permanent Magnet Motor (PMM) rotor core. The proposed methodology uses the Bi-directional Evolutionary Structural Optimization (BESO) heuristics (see Section 5.6 for a description of the method in the context of structural TO) in order to remove the inefficient elements from a meshed model, in order to improve the average torque/power density. The BESO method is chosen over other TO methods because of its simplicity to implement and computational efficiency. Furthermore, the magnetic and structural objectives are expressed in terms of magnetostatic and strain energies, respectively. In this way, the elemental sensitivities are readily available from the FEA solution and computation of the gradient information is not necessary, in favour of speed and simplicity.

In this chapter the overall design workflow, including the interrelationship among the different software tools employed, is provided first. Next, we describe the electrical machine chosen as a case study for this research, followed by the description of the FEA model setup. Finally, the BESO implementations for structural, magnetostatic and combined structural – magnetostatic TO are presented.

9.2 Design Workflow

The design procedure implemented in this work started with the generation of a digital model of a pre-existing electrical machine design using the CAD software *AutoCAD* (Autodesk Inc., San Rafael, California). The CAD model has then been imported into the MSC pre-processing software *Mentat* (MSC Software Corp., Newport Beach, California). In *Mentat* the CAD geometry was meshed to create a Finite Element (FE) model. The pre-processor was also employed to assign the materials properties, loads and boundary conditions to the meshed FE model. The FE solver *Marc* (MSC Software Corp., Newport Beach, California) was then used to solve the equilibrium equation (for the structural load-cases) and the Poisson's equation (for the magnetostatic load-cases). Only linear material properties were considered for the structural simulations, while both linear and nonlinear magnetic properties were considered in the magnetostatic case.

The Topology Optimisation (TO) code for the free-form design of the machine's rotor was implemented using *Python* scripting. At each new iteration, the FE analysis results served as input to the TO code, which then updated the FE model, before a new FE run was performed. The TO scheme was stopped when the prescribed volume was reached and convergence achieved. The library *PyMentat*, provided by MSC Software as part of the *Marc-Mentat* package, was used for the communication of data between *Marc-Mentat* and the TO code. After a TO solution was found, the resulting FE model was converted into .stl format, in order for it to be exported to the post-processing freeware *MeshLab* (Visual Computing Lab, ISTI, CNR, <http://meshlab.sourceforge.net/>), where the topology's surface was smoothened

for manufacturability using a Laplacian smoothing tool. The Laplacian smoothing tool neither changes the topology of the mesh nor adds/removes vertices or triangles, but only moves each vertex of the mesh to the average of the adjacent vertices. The .STL file corresponding to the smoothed topology was then imported into the *Magics* environment (Materialise NV, Leuven, Belgium) for preparation to manufacturing using SLM. Specifically, *Magics* was used to set the position of the part on the build platform and to assign the auxiliary support structures. At this stage the digital file was ready to be sent to the SLM system's proprietary software for manufacturing. An overview of the steps described above is shown in the flowchart of Figure 9-1.

All of the FEA simulations and TO have been carried out using a 12-core Intel Xeon 3.33GHz with 48GB RAM. Matrix solving and assembly operations have been processed in parallel on all of the 12 available cores using the MSC Marc parallel computing capabilities.

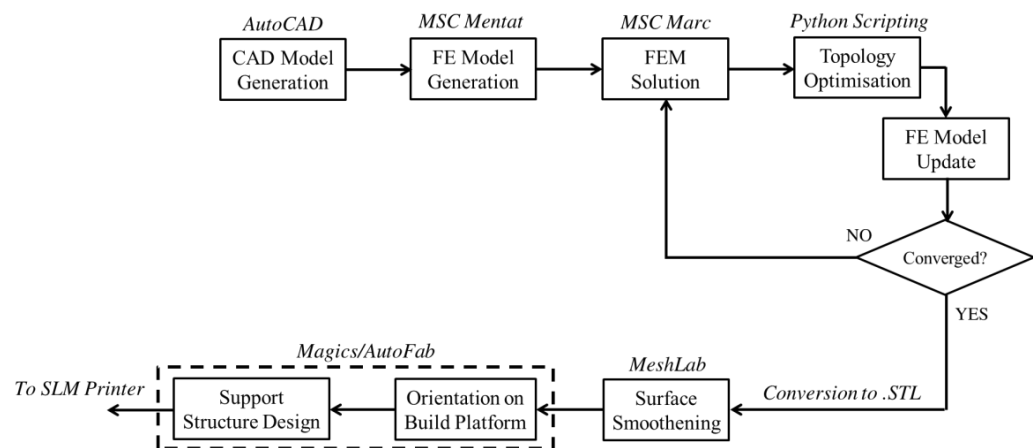


Figure 9-1: Flowchart illustrating the design workflow implemented in this research.

9.3 Electrical Machine Analysis Using the Finite Element

Method

9.3.1 Machine Specifications and Initial Design

A synchronous AC Permanent Magnet (PM) motor was used as a case study to investigate the impact of AM/SLM and TO on the development of machine rotor cores for high torque density. The PM synchronous motor has a classic three-phase stator with distributed windings and the rotor has surface-mounted permanent magnets, as shown in Figure 9-2. The specific PM motor under study is a 2-pole aerospace starter developed as part of an aero-engine system for a typical regional passenger more-electric-aircraft (Sanhueza et al., 2016). A rotating reference frame is fixed to the rotor and defined as a system of three orthogonal axes, known as the d -, q - and z -axes, as shown in Figure 9-2. In particular, the d -axis is aligned along the permanent magnet poles, while the q -axis is oriented along the inter-polar direction (i.e., perpendicular to the permanent magnet poles). The stator-rotor relative position can therefore be expressed in terms of the angle θ_r , also shown in Figure 9-2. The motor is driven by 3-phase sinusoidal excitation currents fed so that they are aligned with the rotor q -axis. The main dimensions of the PM machine were imposed by the constraints of the specific application and are summarized in Table 9-1. The machine design specifications are shown in Table 9-2. Since the preliminary design of the machine was not performed as part of this research, the analytical procedure for determining the machine parameters starting from the design specifications is not presented.

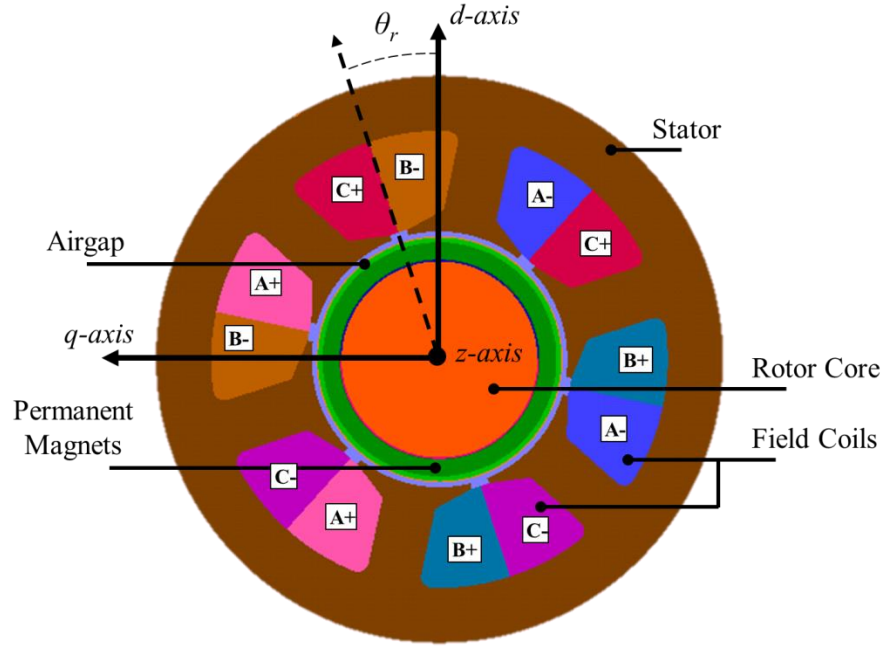


Figure 9-2: Cross-sectional view of the 2-pole PM motor used as a case study to test the proposed magnetostatic-structural TO methodology.

Table 9-1: Main motor dimensions. N_s is the number of slots, m the number of phases, p the number of pole pairs, D_s the stator diameter, L the motor axial length, D_r the rotor diameter, t_m the permanent magnets thickness and l_a the airgap length.

Parameter	N_s	m	p	D_s (mm)	L (mm)	D_r (mm)	t_m (mm)	l_a (mm)
Value	6	3	1	104	62	37	3	2

Table 9-2: Motor Specifications.

Rated Torque (N·m)	Maximum Speed (rpm)	Rated Current Density (A/mm ²)	Current Angle (°)
1	10000	1	90

9.3.2 Material Properties

The material properties defined are those of air, the copper windings, the stator and rotor cores, as well as the permanent magnets. Furthermore, for the 3D

optimisation, a stainless steel shaft was introduced, to which an axial moment was applied. Regarding the magnetic properties, the magnetic permeability of vacuum $\mu_0=4\pi\times10^{-7}$ N/A² was assigned to air, copper and stainless steel. The permanent magnets are made with sintered neodymium-iron-boron (Nd₂Fe₁₄B). The permanent magnet was modelled as an orthotropic material. The following material model for simulating the B-H relationship in the permanent magnet along the d -, q - and z -axes was used:

$$\begin{bmatrix} B_d \\ B_q \\ B_z \end{bmatrix} = \begin{bmatrix} \mu_r \mu_0 H + B_r \\ \mu_r \mu_0 H \\ \mu_r \mu_0 H \end{bmatrix}, \quad (9.1)$$

where μ_r and B_r are the (constant) relative permeability and residual magnetisation of the permanent magnet, respectively. These two properties were set to 1.045 and 1.11 T, respectively, according to the data presented in (Hilzinger and Rodewald, 2013).

For the 2D optimisation, the rotor and stator cores was assumed to be made of lamination material. Both linear and non-linear magnetisation curves were considered for the core materials. Two different materials for the stator and rotor were adopted. The stator was realized using low specific loss electrical steel *M15* while for the rotor lamination material *M250* was adopted, which is characterised by high mechanical strength at the expense of an increased specific iron loss. The relative permeability value at field strength $H=2000$ A/m were employed for the linear simulations/optimisation since these are the closest to the operating point of the electrical machine under study. The non-linear magnetisation and corresponding relative permeability curves, as well as the linearized relative permeability values used for the rotor and stator lamination materials are shown in Figure 9-3 (a) and (b), respectively.

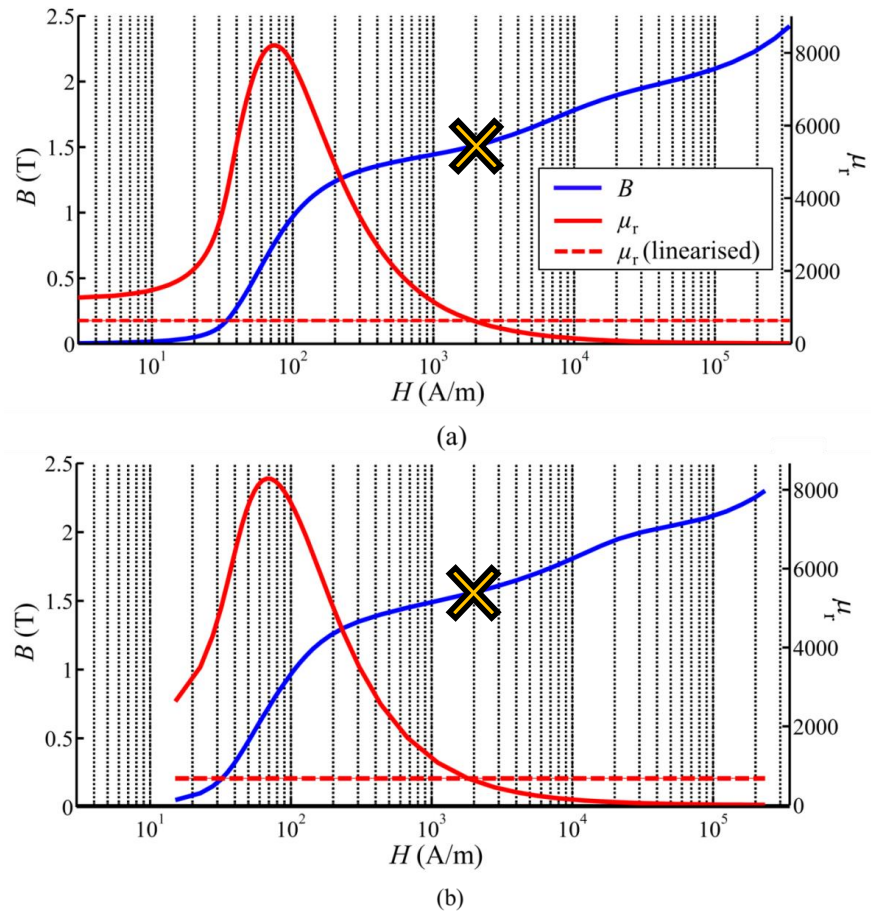


Figure 9-3: Magnetisation and relative permeability curves for rotor material M250 (a) and stator material M15 (b). In both graphs, the dashed lines show the relative permeability values used for linear analysis/optimisation. The crosses on the $B(H)$ plots indicate the approximate operating point of the material.

For the 3D simulations, the same material was used for the stator laminations. For the rotor core the magnetic properties of Fe6.9%Si processed using SLM were employed. The B - H curve of the material was measured on ring samples according to the procedure described in Section 6.11. The measured magnetisation curve and corresponding relative permeability curve, as well as the linearized relative permeability value are shown in Figure 9-4, alongside a photo of one the additively manufactured Fe-6.9%Si rings. A detailed presentation and discussion of the magnetic properties of SLM Fe6.9%Si is provided in Chapter 8. Owing to the high computational burden of the

nonlinear simulations in 3D, only the linear properties were considered for 3D TO. However, the torque of the optimised 3D topology was calculated considering the nonlinear magnetisation curve of the SLM material.

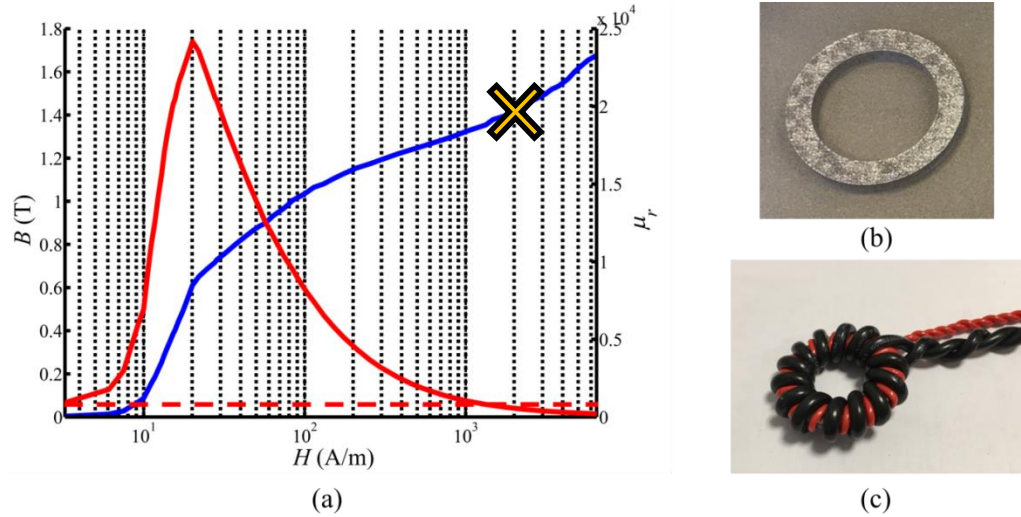


Figure 9-4: Magnetic properties of Fe-6.9%Si alloy processed using SLM (a) and example of the additively manufactured ring in the as-built condition (b) and during magnetic testing (c). The cross on the B(H) plot in (a) indicates the approximate operating point of the material.

With regard to the mechanical properties, it was assumed that only the rotor is subjected to mechanical loads. Therefore, the airgap and stator elements were excluded from the structural simulation pass. For the rotor, it was assumed that all of materials are used within their linear range. The mechanical properties data of the M250 electrical steel and neodymium permanent magnets were taken from the Infolytica MagNet database (Magnet, 2006) and (Hilzinger and Rodewald, 2013), respectively. For the annealed Fe-6.9%Si manufactured using SLM, it was assumed that the mechanical properties are the same as for the cast material (Bozorth, 1993a). The mechanical properties of the materials of interest are summarised in Table 9-3.

Table 9-3: Mechanical properties of the rotor materials

	Mass Density (Kg/m ³)	Young's Modulus (GPa)	Poisson's Ratio	Yield Stress (MPa)
M250	7.5×10^3	200	0.3	360
SLM Fe-6.9%Si	7.5×10^3	210	0.3	100
Nd ₂ Fe ₁₄ B	7.5×10^3	140	0.24	8
Stainless Steel (AISI Type S20910)	7.9×10^3	200	0.3	730

9.3.3 Finite Element Model Setup: 2D Case

The 2D model of the PM motor was meshed using parabolic triangular elements. For the magnetostatic analysis, heat-transfer elements were used, while for the structural and coupled magnetostatic-structural analysis plane-stress elements were employed. In order to reduce the computational burden, a non-uniform mesh was used, wherein the element size was set according to the expected magnitude of the magnetic-strain field gradients. The number of elements used to mesh each motor component and corresponding element size range is reported in Table 9-4. The rotor was divided into the four distinct meshed regions shown in Figure 9-5. The first region represents the permanent magnets (purple elements in Figure 9-5), which are excluded from the design space (i.e., their topology does not change in the optimisation process). The remaining three regions make up the rotor core. The pink elements in Figure 9-5 represent the design space. The number of elements used to mesh this region (approximately 28000) is higher than that required for solution accuracy, with the fine mesh being employed to make sure that the final topology had smooth boundaries. The yellow and green regions of the rotor

cores do not belong to the design space, since they provide the surface on which the permanent magnets are bonded to the rotor core.

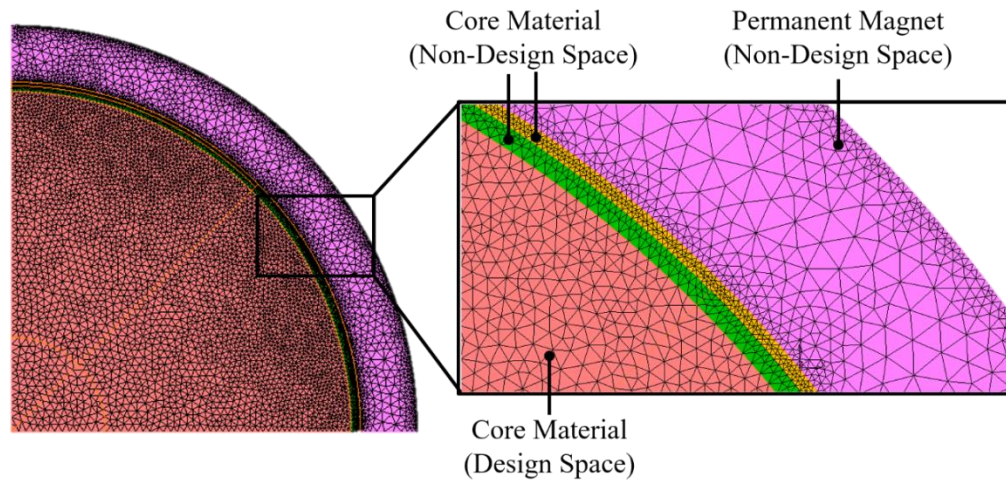


Figure 9-5: PM rotor meshing strategy. In the amplified version, the design and non-design regions of the rotors are indicated. Because of the structure's symmetry, only one quarter of the rotor is represented.

Table 9-4: Number of triangular elements and corresponding element size range in the 2D model of the studied PM motor

	Number of Elements	Element Size Range (mm)
Rotor core (Design Space)	28368	0.1 – 0.2
Rotor core (Non-design Space)	13224	0.1 – 0.2
Permanent Magnets	16912	0.1 – 0.3
Airgap	10664	0.1
Armature Coils	2473	0.3 – 2.0
Stator	7481	0.1 – 2.0
Exterior Region	1024	2.0 – 10
Total	87230	0.1 – 10

9.3.4 Finite Element Model Setup: 3D Case

The 3D model was derived from a 2D model by extruding the 2D elements. However, a different 2D-mesh than the one used for the 2D simulations was

employed, in order to reduce the computational burden of the 3D simulations to a minimum. In particular, a mix of triangular and quad elements was used to mesh the preliminary 2D model. The 2D elements were extruded 0.5 mm along the z-axis and the resulting penta-hedral and hexahedral elements were then repeated along the z-axis to achieve the final motor axial length, $L=62$ mm. In order to further reduce the computational time, 3D FE analyses and optimisation were performed at rotor angle position $\theta_r=0^\circ$, in order to exploit the symmetry of the magnetic field. Therefore, only one half of the stator, airgap and field coils had to be modelled. On the other hand, the whole rotor had to be modelled, in order to calculate the strain field due to the torsional load applied at the two ends of the shaft. However, each iteration step of the TO procedure was performed on 1/8 of the rotor (shown in Figure 9-6), and the resulting topology was mirrored onto the remaining 7 portions. The final number of 3D elements and corresponding approximate elemental size ranges (in the cross-sectional plane) of the different model regions are summarised in Table 9-5.

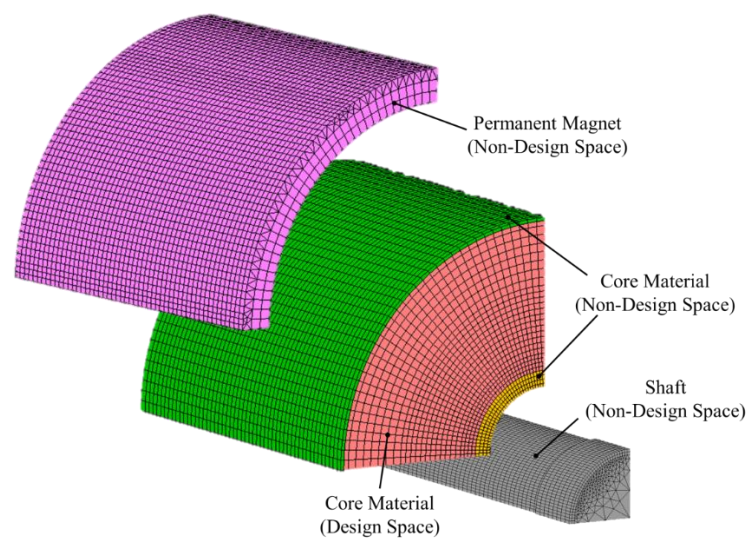


Figure 9-6: Exploded-view of the 3D rotor FE model, including permanent magnets, rotor core regions and shaft. Only 1/8 of the model is represented in this drawing.

Table 9-5: Number of triangular elements and corresponding element size range in the 2D model of the studied PM motor

	Number of Elements	Element Size Range (mm)
Rotor core (Design Space)	52268 ($\times 8$)	0.2 – 0.6
Rotor core (Non-design Space)	11216	0.2 – 0.6
Permanent Magnets	73216	0.3 – 0.6
Airgap	30888	0.2 – 0.3
Armature Coils	33172	0.3 – 3.0
Stator (half)	55016	0.3 – 4.0
Exterior Region	38645	2.0 – 10
Rotor Shaft	156416	0.2 – 2.0
Total	816713	0.1 – 10

9.4 Rotor Core TO using the BESO Method

9.4.1 Structural Problem

The structural TO problem in equations (5.1)-(5.4) for rotor design consists of optimising the rotor's geometry by maximising the rotor stiffness against centrifugal and torsional loads. In this study we use the strain energy defined in equation (5.5) as a measure of the structural stiffness. In 2D, the applied load is represented by the centrifugal force, i.e., $\mathbf{f}_e = x_e \rho_s r_e \omega^2$, where $\mathbf{f}_e$ is the elemental force, r_e the distance of the e^{th} element from the centre of rotation and ω the angular rotation. In 3D, torsional loading is also included by applying an axial moment to the rotor's shaft.

We solved the structural TO problem using the “hard-kill” BESO method (introduced in Section 5.6), since this allows us to employ the same equations

to calculate the sensitivity numbers in the case of design-dependent load (i.e., the centrifugal load) and design-independent load (i.e., the torsional load). Therefore, the structural elemental sensitivity numbers were determined using equation (5.19). The structural sensitivity number α_e^s was filtered using equation (5.22) in order to avoid the problems of mesh-dependence and checkerboard, and stabilised using the stabilisation scheme given in equation (5.23) in order to facilitate convergence.

9.4.2 Magnetostatic Problem

At each magnetostatic step, each phase of a rotating electrical machine can be modeled using the magnetic circuit shown in Figure 9-7 (the permanent magnets, which act as magnetomotive force generators have been excluded for simplicity). The magnetic flux ϕ passing through the circuit can be related to the magnetomotive force NI using basic circuit theory, i.e.:

$$\phi = NI / (R_a + R_r + R_s) \quad (9.2)$$

where N is the number of turns, I is the input phase current and R_a , R_r and R_s are the reluctances of the airgap, rotor and stator, respectively. The total magnetic energy W_T stored in the circuit is the sum of the energies stored in the circuit reluctances and is thus function of the flux and circuit reluctances according to:

$$W_T = W_a + W_r + W_s = \phi^2 \cdot (R_a + R_r + R_s) \quad (9.3)$$

Since the magnetic permeability of air is several orders of magnitude lower than that of ferromagnetic materials, normally we can assume that:

$$R_a \gg R_r, R_s. \quad (9.4)$$

Therefore, (9.2) can be re-written as:

$$\phi \cong NI/R_a. \quad (9.5)$$

Accordingly, $W_T \cong W_a$, i.e., the airgap reluctance determines the amount of magnetic energy stored in the circuit. Removing ferromagnetic material from the rotor causes its reluctance to increase. If condition (9.4) is met and flux leakage outside the rotor can be neglected, the change in magnetic energy W_r stored in the rotor due to a change in reluctance R_r is obtained by combining $W_r = \phi^2 R_r$ and (9.5) and taking the derivative of the resulting expression, to give:

$$\frac{\delta W_r}{\delta R_r} = \left(\frac{N^2 I^2}{R_a^2} \right) \cdot \frac{\delta R_r}{\delta R_r} = \frac{N^2 I^2}{R_a^2}, \quad (9.6)$$

which shows that, in the simple case of a linear magnetic circuit like the one shown in Figure 9-7, W_r increases linearly with rotor reluctance. Clearly, as the reluctance of the rotor increases beyond the point where condition (9.4) is no more met, $\frac{\delta W_r}{\delta R_r}$ will assume a more complex, non-monotonic behavior, owing to the reduction in flux ϕ and the increase in flux leakage.

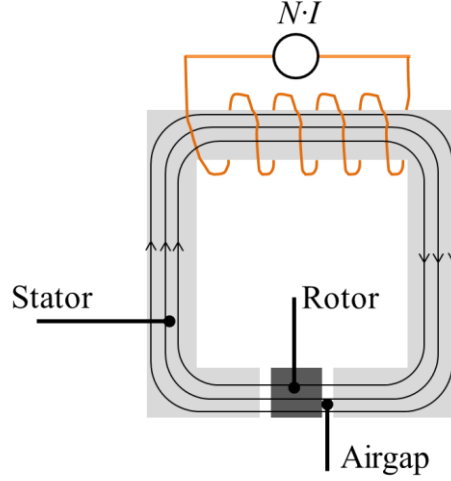


Figure 9-7: sketch of a simple magnetostatic circuit, showing the magnetomotive force NI and main motor components. The flux lines across the circuit are also shown.

The simple example proposed above has shown that, in a real world scenario, given a constant magnetomotive force and approximatively constant flux, the capability of a rotor core to guide magnetic flux (and, thus, of the motor to generate torque) is maximized by minimizing its energy content W_r . Therefore, the magnetostatic optimization problem given a volume constraint can be expressed in finite element form as:

$$\min_{\mathbf{x}} f = W_r = \sum_{e=1}^n w_e v_e, \quad \mathbf{x} = [x_1, \dots, x_e, \dots, x_n] \quad (9.7)$$

$$\text{Subject to: } V_0 - \sum_{e=1}^n x_e v_e = 0 \quad (9.8)$$

$$x_e = 0 \text{ or } 1 \quad (9.9)$$

$$\nabla \times \left(\frac{1}{\mu} \nabla \times \mathbf{A} \right) = \mathbf{J}. \quad (9.10)$$

In (9.7) w_e is the elemental magnetic energy density. Constraint (9.10) is the equation describing the magnetostatic problem (Lee et al., 2010), to be solved

using the FE method. In (9.10), $\mathbf{A}$ is the magnetic vector potential and $\mathbf{J}$ the source current vector. It is important to note that in the linear case the magnetic energy is numerically equivalent to the magnetic co-energy, since the magnetic energy density can be written as:

$$w = \int H(B)dB \quad (9.11)$$

whereas the magnetic co-energy density reads:

$$w_{co} = \int B(H)dH \quad (9.12)$$

In (9.11) and (9.12), H is the magnetic field strength and $B = \mu H$ the magnetic flux density, with μ being the material's magnetic permeability. The two equations indicate that in the linear case the following expression holds:

$$w = w_{co} = \frac{1}{2} \mu H^2. \quad (9.13)$$

In order to derive the sensitivity information of the objective function with respect to the design variable, we express the magnetic permeability of the material using the material interpolation scheme (5.29):

$$\mu(x_e) = \mu_0 \cdot (\mu_s \cdot x_e^p + 1 - x_e^p)$$

where μ_s is the relative permeability of the solid (ferromagnetic) material and μ_0 the permeability of vacuum. Using this material interpolation scheme, in the linear case the sensitivity of the magnetostatic objective function W_r with respect to x_e can be obtained as:

$$\frac{dW_r}{dx_e} = \frac{1}{2} p x_e^{p-1} (\mu_s \mu_0 - \mu_0) H_e^2 = p x_e^{p-1} w_{co}^0 (1 - \frac{1}{\mu_s}) \cong p x_e^{p-1} w_{co}^0 \quad (9.14)$$

where w_{co}^0 is the co-energy density of the solid element. The term $\frac{1}{\mu_s}$ in the above equation is negligible because $\mu_s \gg 1$ in the linear case. The BESO sensitivity number can then be obtained from the sensitivity equation as:

$$\alpha_e^m = -\frac{1}{p} \frac{dW_r}{dx_e} = \begin{cases} w_{co}^0, & x_e = 1 \\ x_{min}^{p-1} w_{co}^0, & x_e = x_{min} \end{cases} \quad (9.15)$$

The above “soft-kill” sensitivity number can be changed into a ‘hard-kill’ sensitivity number by setting p to infinity, which leads to:

$$\alpha_e^m = -\frac{1}{p} \frac{dW_r}{dx_e} = \begin{cases} w_{co}^0, & x_e = 1 \\ 0, & x_e = x_{min} \end{cases} \quad (9.16)$$

In the nonlinear case, we use the total magnetic co-energy density as the objective function:

$$f = \lim_{n \rightarrow \infty} \frac{1}{2} \sum_{t=1}^I (\mathbf{H}_t^T - \mathbf{H}_{t-1}^T) (\mathbf{B}_t + \mathbf{B}_{t-1}) \quad (9.17)$$

where I is the total number of increments used to solve the nonlinear problem by means of the Newton-Raphson method. By analogy with the nonlinear structural problem in equations (5.24)-(5.26), the sensitivity of the objective function can be expressed by:

$$\frac{df}{dx_e} = \lim_{n \rightarrow \infty} \frac{1}{2} \sum_{t=1}^I (\mathbf{H}_t^T - \mathbf{H}_{t-1}^T) \left(\frac{d\mathbf{B}_t}{dx_e} + \frac{d\mathbf{B}_{t-1}}{dx_e} \right) \quad (9.18)$$

By using the interpolation scheme (5.29) the elemental sensitivity number is found as:

$$\alpha_e = -\frac{1}{p} \frac{df}{dx_e} = x_e^{p-1} \lim_{n \rightarrow \infty} \frac{1}{2} \sum_{l=1}^I (\mathbf{H}_l^T - \mathbf{H}_{l-1}^T) [\mu_0(\mu_l^0 - 1)\mathbf{H}_l + \mu_0(\mu_{l-1}^0 - 1)\mathbf{H}_{l-1}] \quad (9.19)$$

$$= x_e^{p-1} \lim_{n \rightarrow \infty} [\sum_{l=1}^I (w_{co,l}^0 - w_{co,l-1}^0)] - \frac{1}{2} \mu_0 H_{e,I}^2 = x_e^{p-1} (w_{co,I}^0 - \frac{1}{2} \mu_0 H_{e,I}^2)$$

where $w_{co,I}^0$ is the final co-energy density in the solid element. From the above equations, the hard-kill sensitivity number can be expressed as:

$$\alpha_e^m = -\frac{1}{p} \frac{dW_r}{dx_e} = \begin{cases} w_{co,I}^0 - \frac{1}{2} \mu_0 H_{e,I}^2, & x_e = 1 \\ 0, & x_e = x_{min} \end{cases} \quad (9.20)$$

As in structural BESO, the magnetostatic sensitivity number α_e^m has been filtered and stabilised using the filtering and stabilising schemes in equations (5.22) and (5.23), respectively.

9.4.3 Combining Structural and Magnetostatic Sensitivities

In order to optimise the structural and magnetostatic performances of the electromechanical component simultaneously, a combined objective function E_{comb} is derived as a weighted sum of the structural (E_s) and magnetostatic (E_m), i.e.:

$$E_{comb} = \omega_s \cdot \widehat{E}_s(x) + \omega_m \cdot \widehat{E}_m(x) \quad (9.21)$$

where $\widehat{E}_s(x)$ and $\widehat{E}_m(x)$ are the normalised structural and magnetostatic objective functions, respectively, and ω_s and ω_m the structural and magnetostatic weights, respectively. A standard normalisation strategy is employed (Arora, 2004). For the structural objective function this normalisation strategy can be written as:

$$\widehat{E}_s = \frac{E_s - E_s^0}{E_s^{max} - E_s^0} \quad (9.22)$$

where E_s^0 is the minimum value of E_s obtained while solving the single-objective TO problem, and E_s^{max} is the value of E_s corresponding to the design obtained by minimising the magnetostatic objective function E_s . This normalisation approach guarantees that all objective functions are bound in the interval $[0,1]$ during multi-objective optimisation.

The weighted sum method has been employed in most of the works that attempted to tackle a multi-functional TO problem (Guest and Prévost, 2006, de Kruijf et al., 2007, Lee and Kikuchi, 2010, Panesar et al., 2015). In the present research, two approaches for the determination of the structural and magnostatic weights are employed. The first consists in the manual assignment of fixed weights. By changing the values of the weights, we create a Pareto front from which the determination of the optimal combination is possible. Since this approach requires numerous runs of the optimisation, it is generally computationally-intensive and time-consuming. Therefore, an adaptive approach to weight determination was also proposed and implemented in this work. In this method, at each iterative step a different weighting combination was assigned to each element in the design space, based on the value of the structural sensitivity number α_e^S after filtering. The structural weighting coefficient was calculated as the ratio between the current sensitivity number and a prescribed upper limit. Defining λ as the Young's modulus of the material, we set the upper limit to be the elemental strain energy density value $\epsilon_e = \frac{1}{2} \frac{\sigma_{th}^2}{\lambda}$, corresponding to an equivalent Von Mises stress $\sigma_{th} = SF \cdot \sigma_Y$, where σ_Y is the material's yield stress and SF a safety factor. Therefore, the proposed adaptive weight assignment scheme can be written as:

$$\omega_s^e = \begin{cases} \frac{\alpha_e^s}{\epsilon_e} & \text{if } \frac{\alpha_e^s}{\epsilon_e} < 1 \\ 1 & \text{if } \frac{\alpha_e^s}{\epsilon_e} \geq 1 \end{cases} \quad (9.23)$$

$$\omega_m^e = \begin{cases} 1 - \frac{\alpha_e^s}{\epsilon_e} & \text{if } \frac{\alpha_e^s}{\epsilon_e} < 1 \\ 0 & \text{if } \frac{\alpha_e^s}{\epsilon_e} \geq 1 \end{cases} \quad (9.24)$$

where ω_s^e and ω_m^e are, respectively, the structural and magnetostatic weights for element e . Finally, a general expression for the combined sensitivity number can be derived as:

$$\alpha_e^{comb} = \omega_s^e \cdot \frac{\alpha_e^s}{E_s^{max} - E_s^0} + \omega_m^e \cdot \frac{\alpha_e^m}{E_m^{max} - E_m^0}. \quad (9.25)$$

When the adaptive approach is used, the overall weights ω_s and ω_m were obtained as an average of the elemental weights ω_s^e and ω_m^e , respectively. In the case of fixed weighting, elemental and overall weights correspond.

9.4.4 Adopted Evolutionary Parameters

The complete set of BESO parameter values used for the structural and magnetostatic TO problems are listed in Table 9-6. The values of ER , AR_{max} , x_{min} and τ were set based on the recommendations of (Huang and Xie, 2010). The value of the filter scheme radius r_{min} was chosen in order include a sufficiently high number of elements, given the element size, as also recommended in (Huang and Xie, 2010). With regards to the historical sensitivity information, w_{i-1} , this was set to the typical value of 0.5 in the structural and linear magnetostatic case. In the nonlinear magnetostatic case, the saturation of the ferromagnetic material can generate flux leakage (Chapman, 2005), which in turn can make it difficult to achieve convergence. In an effort to foster convergence by stabilising the evolutionary process, the importance of the historical sensitivity information was increased. In particular,

various w_{i-1} values were considered to help understand how these affect the algorithm stability in non-linear magnetostatic conditions.

Table 9-6: Evolutionary parameters used to implement the BESO TO. The list includes the Evolutionary Ratio (ER), the maximum Addition Ratio (AR_{max}), the filter radius (r_{min}), the historical weight (w_{i-1}), the density of void elements (x_{min}) and the convergence tolerance (τ).

	BESO Parameter					
	ER	AR_{max}	r_{min} (mm)	w_{i-1}	x_{min}	τ
Structural	0.01	0.05	2.0	0.5	10^{-3}	10^{-4}
Magnetostatic						
Linear	0.01	0.05	2.0	0.5	10^{-3}	10^{-4}
Non-linear	0.01	0.05	2.0	0.5, 0.8, 0.95	10^{-3}	10^{-3}

9.5 Summary

This chapter has presented the methodology adopted to implement the structural and magnetostatic design of a PMM rotor core by means of BESO-based TO. The structural problem consists in the minimisation of the rotor compliance (or strain energy) against centrifugal and torsional loads given a relative mass constraint of 50%. The BESO theory for design-dependent loads as previously presented in Section 5.6 could be employed to solve the structural problem. The magnetostatic problem consists in minimising the magnetic co-energy inside the rotor core while respecting a mass constraint. It has been shown that minimising the magnetic co-energy inside the rotor is equivalent to maximising the magnetic flux flowing through the rotor, and therefore the latter's capability to produce a torque. The advantage of using the magnetic co-energy is that the sensitivity number is obtained directly from the FEA results. Finally, a coupled magnetostatic-structural objective has been derived as a weighted sum of the structural and magnetostatic objectives.

10. Electric Machine Rotor Core Topology Optimisation and Manufacturing by Selective Laser Melting

10.1 Introduction

This chapter presents the results obtained using the BESO-based TO methodology described in Chapter 9 for the combined magnetostatic and structural design of a Permanent Magnet Motor (PMM) rotor core, aimed at maximizing the motor's torque-to-weight ratio. The methodology is first tested on linear and nonlinear two-dimensional problems. For the two-dimensional problem the starting rotor design does not include a fixed central hole for the shaft, in order to leave the door open to alternative shaft designs (i.e., shaft designs characterized by non-circular and axially-varying cross-section). Next, we show that the proposed optimization scheme can be easily extended to three-dimensional problems, and we show that the resulting designs are suitable for manufacturing using AM. An attempt to produce the final design by means of SLM using high-silicon steel is finally presented at the end of the chapter.

10.2 Magnetostatic Topology Optimisation

10.2.1 Initial Motor Performance

The distribution of the elemental co-energy density w_{co} in the full rotor (initial design) of the analysed PMM is shown in Figure 10-1 for rotor positions $\theta_r=0^\circ$ (Figure 10-1(a)) and $\theta_r=30^\circ$ (Figure 10-1(b)). The colour-maps show that the magnitude and distribution of w_{co} change significantly between the considered rotor positions. The plots showing the variation of the total magnetic co-energy in the initial rotor core with rotor angle position is shown in Figure 10-2. The

angle range considered for these plots corresponds to one slot pitch (i.e., 60° mechanical). The change in the magnetic co-energy of the rotor with rotor position is due to the shape of the stator, which causes the latter's reluctance to change with orientation. Next, the plots in Figure 10-3 show the variation with rotor position of the rotor's magnetic co-energy (Figure 10-3(a)), rated torque (Figure 10-3(b)) and cogging torque (Figure 10-3(c)). The angle range considered in this case is 180° , which corresponds to one pole pitch. A comparison of the three plots suggests that the variation of rated torque with rotor position is the result of both the change in magnetic co-energy and of the cogging torque. The latter is the attraction-repulsion between the rotor's permanent magnets and the stator teeth when the armature coils are unenergised.

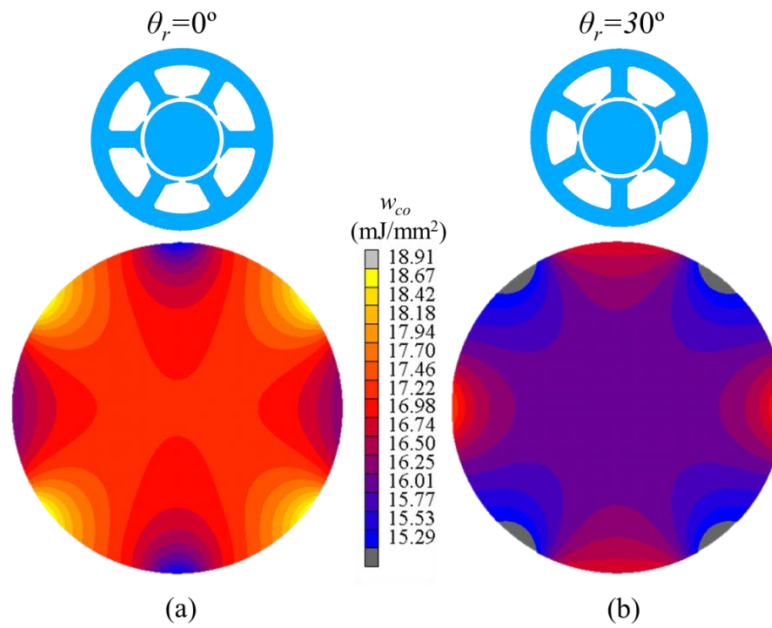


Figure 10-1: FE simulation results showing the elemental co-energy density (w_{co}) in the initial motor design (full rotor). Simulations run using the rated current values and at rotor positions $\theta_r=0^\circ$ (a) and $\theta_r=30^\circ$ (b).

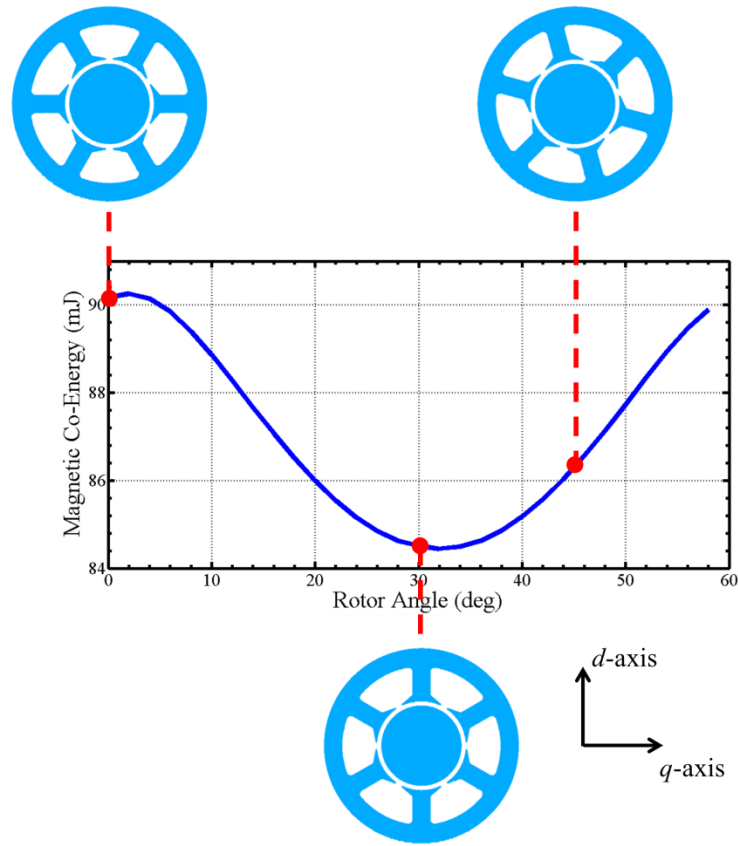


Figure 10-2: Variation of the magnetic co-energy in the rotor at different rotor positions, over one slot pitch.

The results presented in this section suggest that carrying out the TO at different rotor positions can be expected to produce different optimal topologies. Therefore, the TO is carried out at various rotor positions along a slot pitch in order to determine how these affect the final motor's rated torque profile, especially in terms of average torque and torque ripple (expressed in terms of the standard deviation). In particular, we have decided to consider rotor positions corresponding to angles $\theta_r=0^\circ$, 15° , 30° and 45° , since these correspond to peaks or troughs in the co-energy and cogging torque profiles, as the plots in Figure 10-3 show. It is important to note that the TO procedure presented in this work has the only objective to maximise torque, and not to minimise the cogging torque/torque ripple. Furthermore, the selected

application does not have any significant requirement for torque ripple, and the results in Figure 10-3(b) show a torque ripple of approximately 2%, which is insignificant for most applications. If the torque ripple were to be reduced, a conventional method such as rotor skewing can be applied after the preliminary TO design, or a different TO strategy should be adopted, which is beyond the scope of this work.

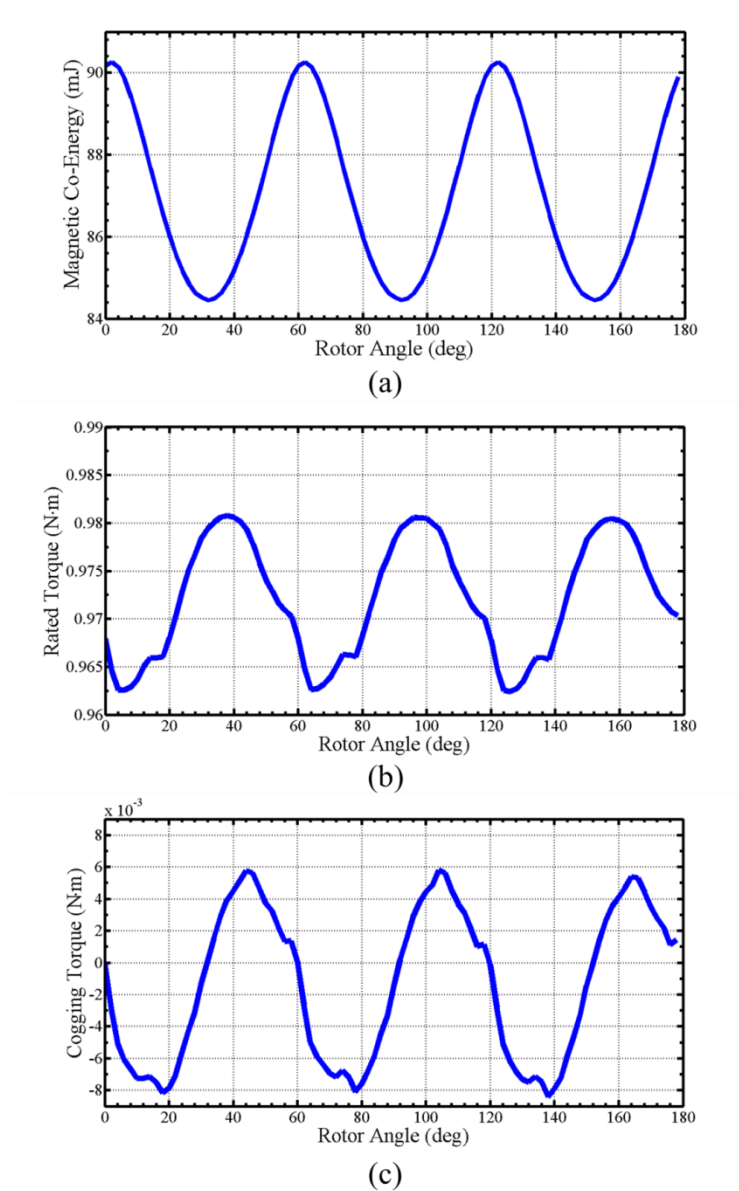


Figure 10-3: Initial performance of PM motor along one pole pitch (180°). The three plots show the rotor magnetic co-energy (a), the rated torque (b) and the cogging torque (c).

10.2.2 Linear Magnetic Properties

The plots in Figure 10-4 show the evolution histories of the objective function (i.e., the mean magnetic co-energy), the volume fraction and the topology when soft magnetic material with constant permeability is considered. The TO in this example has been performed at rotor angle $\theta_r=30^\circ$. The plots show that as the elements are removed from the design space (i.e., as the volume of the rotor gradually decreases), the magnetic co-energy increases steadily. After slightly over 70 iterations the prescribed volume fraction is reached, and the objective function stops increasing. Using a convergence tolerance $\tau=10^{-4}$, convergence is reached after around 113 iterations, which corresponded to a computational time of approximately 80 min. However, the value of the objective function (and corresponding topology) has very little change after the prescribed volume fraction is reached. This suggests that using a higher convergence tolerance could have led to convergence after fewer iteration steps, without significant effect on the final rotor performance.

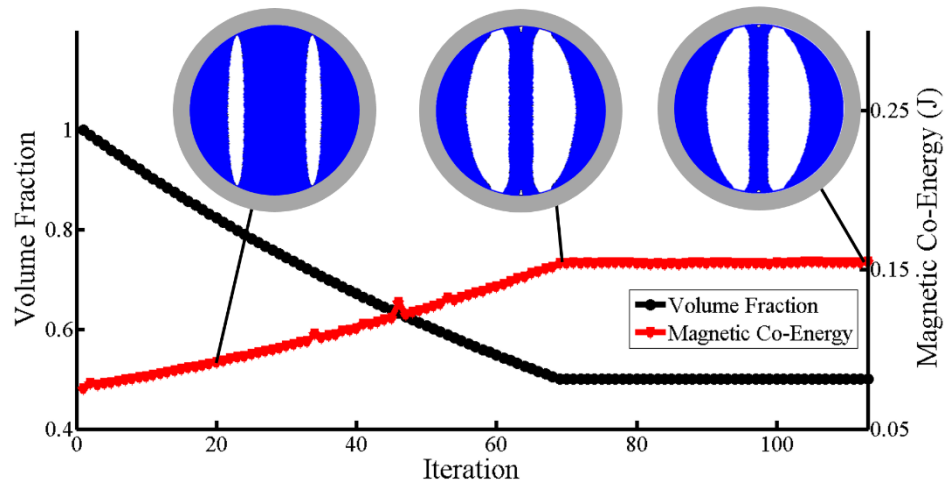


Figure 10-4: Evolution histories of mean magnetic co-energy, volume fraction and topology in the case of linear magnetostatic TO at rotor angle position $\theta_r=30^\circ$. The permanent magnets mounted on the surface of the rotor core are coloured in grey as they are excluded from the design space.

The final topologies in Figure 10-5 show that the result of the TO changes with rotor position. In particular, Figure 10-5 shows the resulting topologies when the BESO algorithm is run at rotor positions $\theta_r=0^\circ$ (Figure 10-5(a)) and $\theta_r=30^\circ$ (Figure 10-5(b)), respectively. A comparison between the initial design (full rotor) and the two final topologies in terms of the rated torque and cogging torque along one pole pitch is shown in Figure 10-5(c) and Figure 10-5(d), respectively. Importantly, the plots in Figure 10-5(c) show that the performed material removal does not affect the average rated torque of the machine. This means that the rotor's torque density can be improved by 50% thanks to the proposed magnetostatic TO methodology. Figure 10-5(c) shows that while the average rated torque is not significantly affected by the rotor position at which the optimisation is performed, the torque ripple appears to vary slightly, as expected given the variation of net effective reluctance with position. In particular, the torque ripple is lowest if the optimisation is carried out at $\theta_r=30^\circ$, where the stator reluctance for the considered motor is at its maximum, as the plot of magnetic co-energy in the rotor (Figure 10-3 (a)) has shown. Once again, it is noted that the minimisation of the torque ripple is beyond the scope of this work, and that a different objective function should be designed for that purpose. Finally, Figure 10-5 (b) suggests that from a magnetostatic perspective, having a central shaft of circular cross-section is not optimal. Different shaft topologies could be considered that exploit the empty spaces around the rotor centre. Such complex shaft topologies can be conveniently manufactured by means of AM/SLM technology, but this is out of the scope of this thesis and is left for future research.

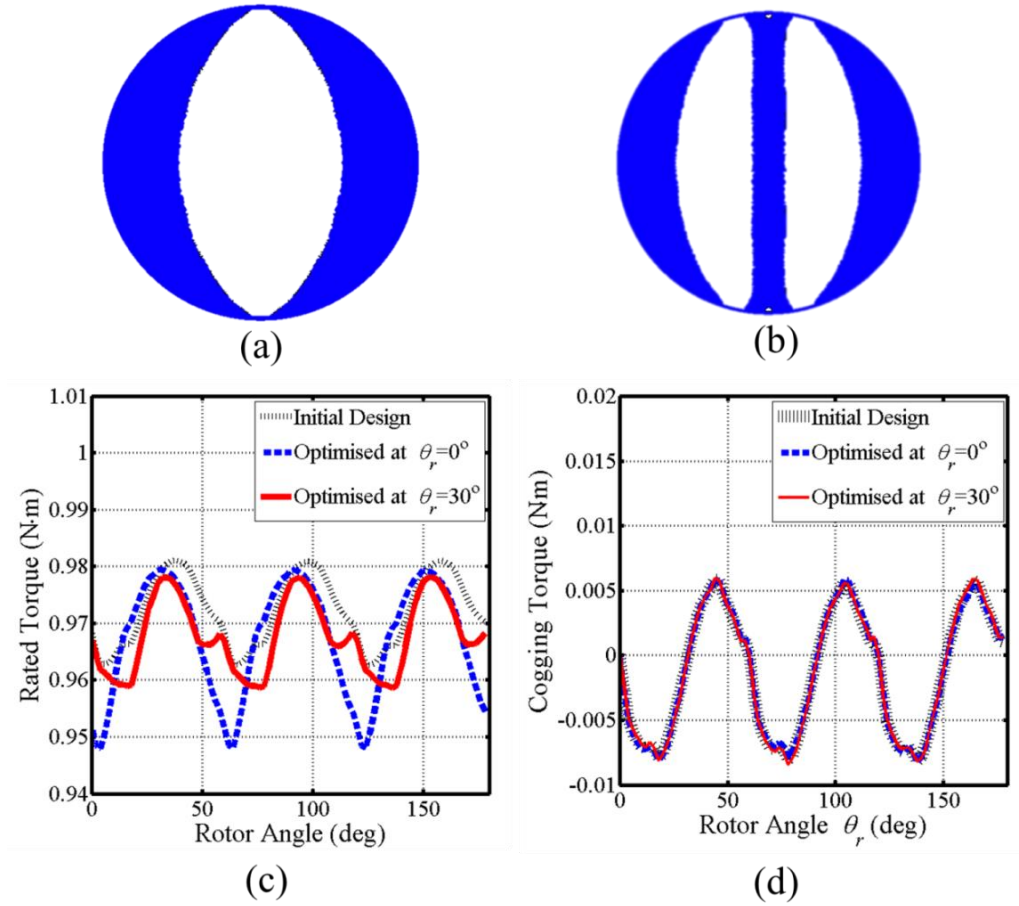


Figure 10-5: Results of the linear magnetostatic TO at rotor angles $\theta_r=0^\circ$ and $\theta_r=30^\circ$. The resulting topologies are in (a) for $\theta_r=0^\circ$ and in (b) for $\theta_r=30^\circ$. The performance of the two resulting topologies is presented in terms of the rated (c) and cogging torque profiles (d). The torque profiles for the initial rotor design are also reported in (c) and (d). The permanent magnets mounted on the surface of the rotor core are not shown.

10.2.3 Nonlinear Magnetic Properties

Convergence of the TO scheme in case of nonlinear magnetic properties has been found difficult to achieve. The instability of the nonlinear solution is attributed to the fact that under a certain volume fraction, most of the elements in the design space enter the nonlinear region, thus causing the solution to become very sensitive to small changes in the topology. In order to stabilise the objective function when the nonlinear magnetisation properties of the core material are considered, the weight of the historical information in w_{i-1} in the

stabilisation scheme has been increased. The effect of different stabilisation weight values on the convergence history of the objective function is shown by the three plots in Figure 10-6. The plots show that increasing the weight associated with w_{i-1} helps to damp the oscillations of the objective function and therefore stabilises the scheme. In the example of Figure 10-6 a convergent solution has been found with tolerance $\tau=10^{-3}$ after less than 120 iterations when $w_{i-1}=0.95$ and $w_i=0.05$ have been used. The convergent solution has been achieved within 110 min.

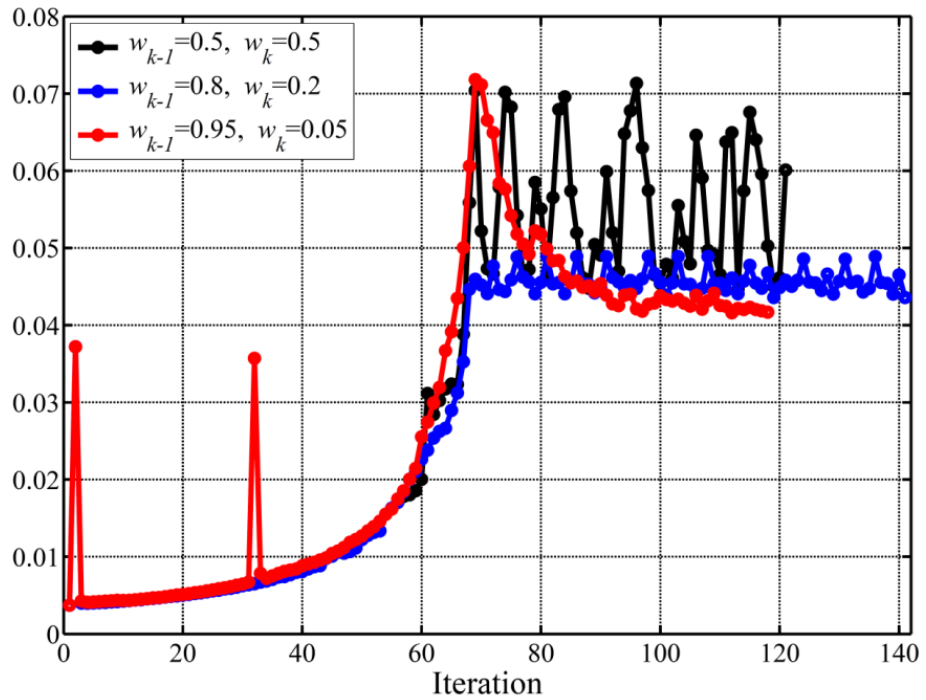


Figure 10-6: Effect of the stabilization scheme weights on convergence history in the case of nonlinear magnetic properties. Convergence with tolerance $\tau=10^{-3}$ was achieved by fixing $w_{i-1}=0.95$ and $w_i=0.05$.

Figure 10-7 shows an example of final rotor topology and the corresponding rated torque profile after nonlinear BESO has been performed using $w_{i-1}=0.95$ at rotor position $\theta_r=30^\circ$. A comparison between the topologies in Figure 10-7(a) and in Figure 10-5(b) shows that the nonlinear and linear designs are

different. The torque profiles produced by the two designs are also different, as the comparison between plots in Figure 10-7(b) and Figure 10-5(c) show. However, overall the linear and nonlinear solutions produce similar average rated torque and torque ripple, as shown in Figure 10-8, where the average and standard deviation of the rated torque are plotted against the rotor angle at which the TO has been performed, for materially linear and nonlinear rotors. Therefore, for the present case study no significant advantages derive from considering the nonlinear properties of the material. This is not true in general, and the nonlinear approach can be expected to produce better and more robust solutions in applications where saturation plays an important role in determining the device's performance. In this regard, it is worth mentioning that the overall torque of a motor is made up of two components: the interaction of the currents with the magnet's field and the interaction of the currents with the variation of rotor reluctance, i.e., the difference between the d - and q -axis inductances. For machines which rely on the second torque component such as interior permanent magnet machines and PM-assisted SyR machines, material non-linearity would need to be considered.

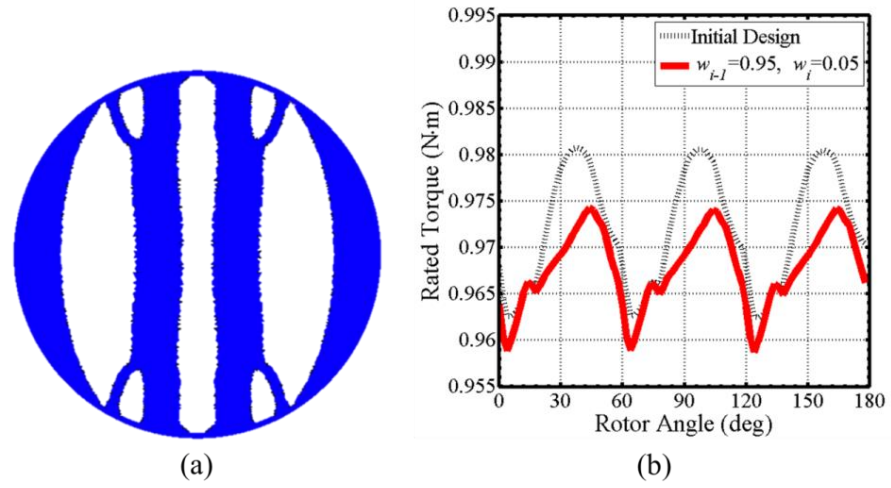


Figure 10-7: Rotor TO considering a material with nonlinear magnetic properties. The final rotor topology for historical weight value 0.95 is shown in (a). In (b) the final rated torque profile is compared to that of full rotor. Optimisation carried out at rotor position $\theta_r=30^\circ$. The permanent magnets mounted on the surface of the rotor core are not shown.

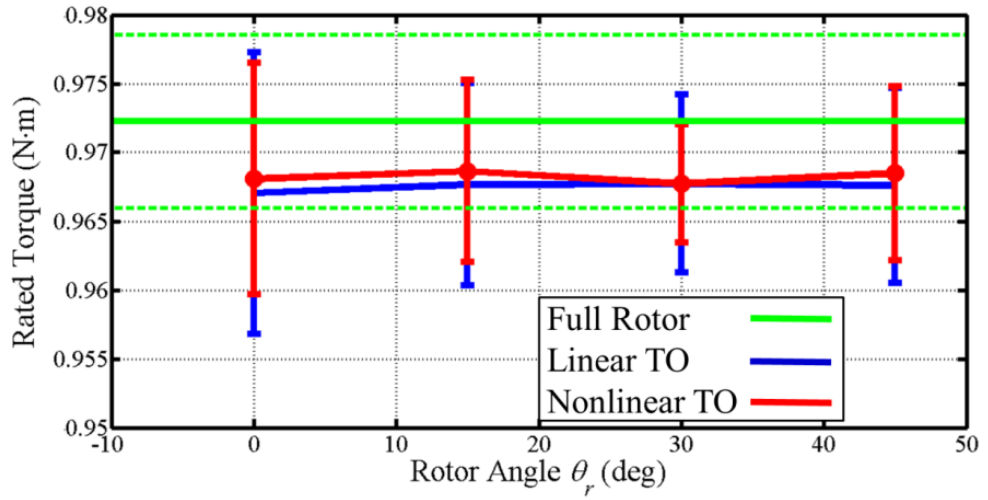


Figure 10-8: Variation of the final rated torque with the rotor angle considered during TO. The plots compare the results in case of linear and nonlinear material properties. Furthermore, the average and standard deviation of the rated torque for the full rotor are also plotted for comparison.

10.3 Joint Structural-Magnetostatic TO

10.3.1 Two-Dimensional Studies

The Pareto front derived by solving the TO problem using structural-magnetostatic BESO with various combinations of fixed and adaptive structural and magnetostatic weights is shown in Figure 10-9. All optimisation runs presented in this section have been performed at fixed rotor position $\theta_r=45^\circ$. The Pareto plot shows that decreasing the fixed structural weight ω_s from 1 to 0.1 does not affect the structural performance of the rotor as measured using the mean strain energy, which represents the structural objective. This indicates that the structural and magnetostatic functions do not represent conflicting objectives. The Pareto plot also shows how the combined efficiency of the final topology changes when the weights' values are determined automatically based on the yield strength of the material, according to the proposed adaptive strategy. The two Pareto points represented by the round markers were obtained using the adaptive strategy for different yield strength values, namely, 100 and 400 MPa. A safety factor (SF) of 4 has been used in both cases. The two points are close to the origin of the Pareto plot, suggesting that the adaptive method allows the designer to automatically make optimal use of the material from both the mechanical and magnetostatic standpoints.

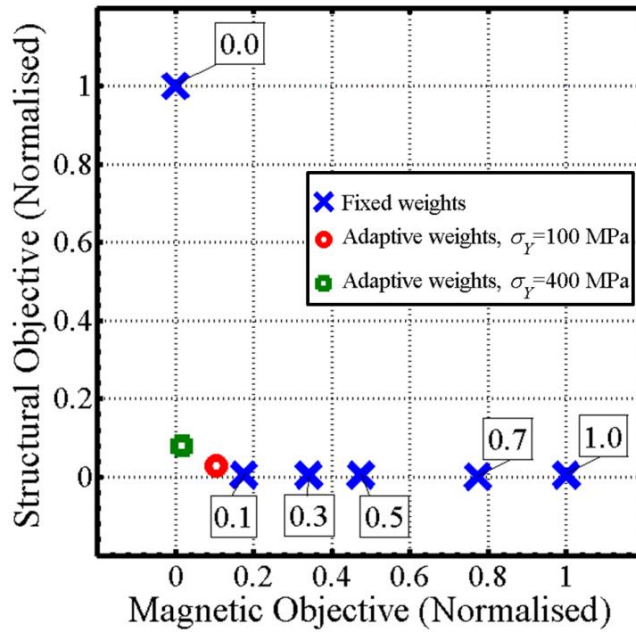


Figure 10-9: Pareto front from structural-magnetostatic BESO with nonlinear magnetic properties. The numbers next to the data points are the values of the associated structural weight. All optimisation runs performed at fixed rotor position ($\theta_r=45^\circ$).

The effect of various structural-magnetostatic weight combinations on the final topology is shown in Figure 10-10. The images show that by decreasing the weighting of the structural term of the objective function, the optimal topology evolves from one recalling a wheel with spokes, to one that resembles the solutions of the magnetostatic optimisation presented in the previous sections. The structural performance of the rotor in terms of the Von Mises stress and the distribution of the flux density within the rotor are also shown in Figure 10-10. In particular, the contour maps show that decreasing the structural weight does not necessarily result in an increase in the maximum Von Mises stress, as a result of the fact that the structural and magnetostatic functions do not represent opposing objectives. On the other hand, the peaks of flux density inside the rotor decrease with the structural weight, thus suggesting that a more efficient use of the material is produced. When an adaptive strategy is used to calculate the weights of structural and magnetostatic terms in the objective

function, the most efficient use of the soft magnetic material is made, as corroborated by the uniform distribution and comparatively small peak value of the flux density. At the same time, for both the core and permanent magnets the maximum Von Mises stresses obtained (6.70 and 3.38 MPa, respectively) are well-within the maximum values allowed, which equal 180 and 4 MPa, respectively.

Finally, the rated torque plots are shown in Figure 10-11. These confirm the above results, in that the motor torque performance improves with decreasing structural weight. As expected from the analysis of the objective function (Figure 10-9) and flux density (Figure 10-10), the adaptive strategy to objective function weighting produces the best torque performance.

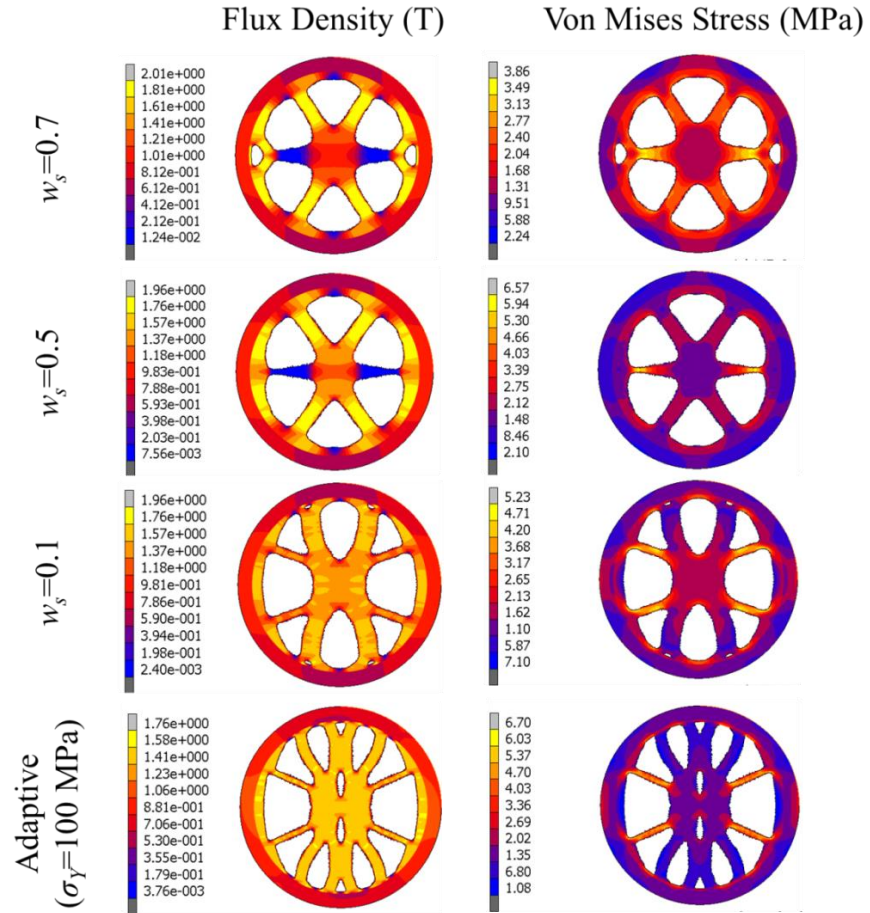


Figure 10-10: Various optimal rotor topologies obtained using the structural-magnetostatic BESO method for different combinations of structural (w_s) and magnetostatic (w_m) weights. The color maps show the distributions of the flux density (left column) and Von Mises equivalent stress (right column). Optimisation run at rotor position $\theta_r=45^\circ$.

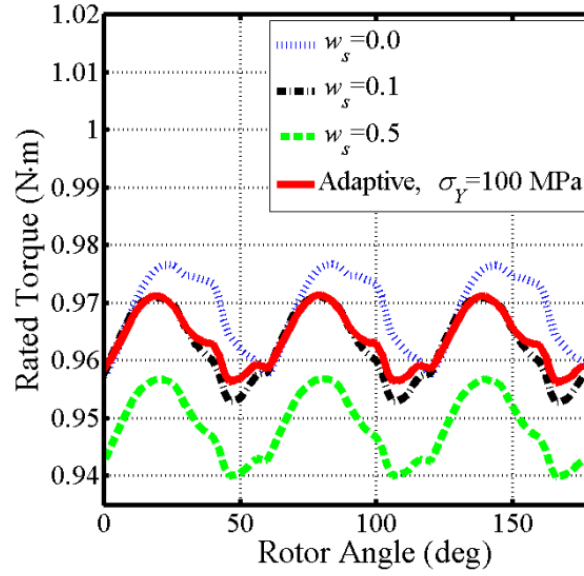


Figure 10-11: Rated torque profiles produced by various optimal rotor topologies, each obtained using a different magnetostatic-structural weight combination in the structural-magnetostatic BESO implementation. Optimisation run at rotor position $\theta_r=45^\circ$.

10.3.2 Three-Dimensional Studies

The structural-magnetostatic BESO method was also applied to a three-dimensional model of the PM motor with the rotor at position $\theta_r=0^\circ$. The proposed adaptive strategy to structural and magnetostatic weight assignment was employed. Since results in two-dimensions have shown that linear and nonlinear magnetic properties lead to similar final electromechanical performances, only linear magnetic properties were considered in the 3D study in order to decrease the computational burden. Given these settings, the algorithm could converge to a solution in less than 130 iterations, which correspond to approximately 24 hours of process time. The optimal topology produced by the algorithm is shown in Figure 10-12. Importantly, the optimised topology of the 3D geometry varies along the axial direction, thus effectively producing a 3D design. While in the central section of the geometry the optimised topology is similar to that obtained in the 2D case, (see Figure

10-5(a)), the topology of the two end sections is different. This result is attributed to the fact that in correspondence of the two ends the rotor core is connected to the shaft. Therefore, in order for the torque to be transferred from the shaft to the core, the adaptive TO scheme has caused an increase of the structural weight in correspondence of the two ends, thus explaining the different topology.

The resulting magnitude of the Von Mises stress under torsional and centrifugal loads is mapped on the model in Figure 10-12. The contour plot shows that the highest stress is generated close to the interface between the rotor core and the shaft, suggesting that the torsional load transferred from the shaft represents the main source of stress in the core. Importantly, the maximum Von Mises stress is below 50 MPa, which was set as the maximum equivalent stress allowed in the rotor core, as shown in Table 9-3. The maximum Von Mises stress in the permanent magnets, shown in Figure 10-12(a), was found to be 4.82 MPa, which is only slightly higher than the maximum value allowed of 4 MPa. In this regard, including the permanent magnet elements in the calculation of the filtered sensitivity numbers of the elements in the design space (equation (5.22)) could provide a tighter control of the maximum stresses in the permanent magnets, without having to increase the value of the structural weight, nor the core's target volume. The rated torque produced by the optimal topology at rotor position $\theta_r=0^\circ$ is 0.955 N·m, which is very close to the target torque of 1 N·m. Finally, although the complete rated torque profile over one pole pitch was not calculated, it is expected that similar plots to those obtained in the two-dimensional case would be obtained.

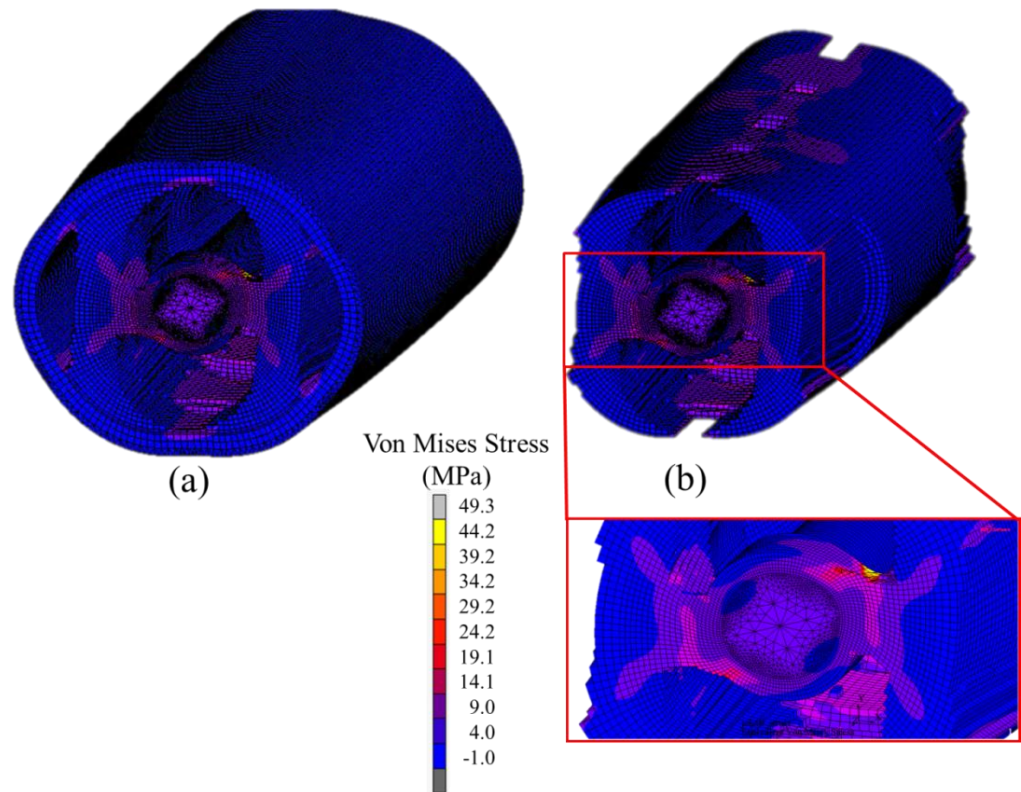


Figure 10-12: Optimized topology using structural-magnetostatic BESO, and Von Mises stress distribution under operational loading conditions. The deformed rotor is shown with (a) and without (b) permanent magnets. The zoom over the highly stressed region provides a better view of the stress distribution.

As the model plots in Figure 10-13 suggest, the optimized topology is not suitable for manufacturing, owing to its relatively coarse surface. Therefore, after optimisation the .stl representation of the final topology has been exported to freeware *MeshLab* (Visual Computing Lab, IST-CNR, Italy) for surface smoothing, in order to make the topology suitable for manufacturing using SLM technology. The final rotor topology before and after smoothing is shown in Figure 10-14(a) and (b), respectively.

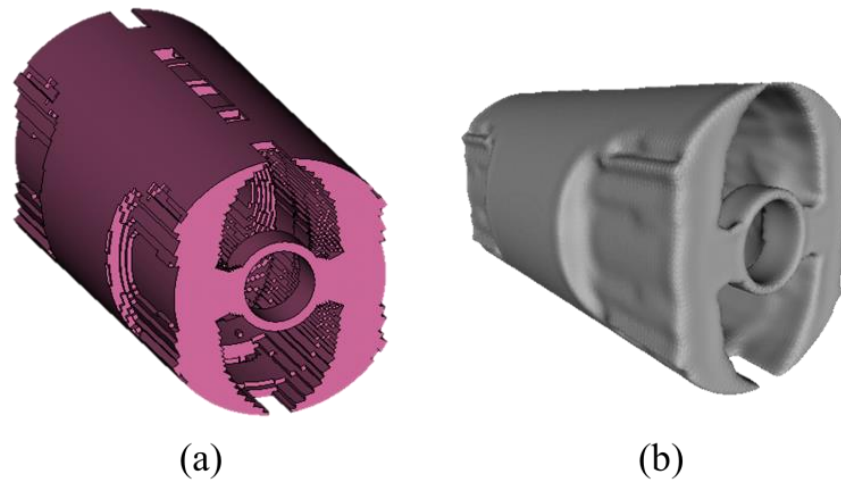


Figure 10-14: Optimized three-dimensional rotor topology, before (a) and after surface smoothening (b) using Laplacian filtering in *MeshLab* software.

10.4 Rotor Manufacturing Using SLM: Preliminary Results

10.4.1 Process Parameter Optimisation

While the production of near-full density test specimens could be achieved using SLM, removing residual porosity and cracks has the potential to further improve the magnetic properties of the parts. Furthermore, in this work the cubes and rings considered are characterised by small sizes compared to typical real world components, such as the PM rotor core optimised in Chapter 9. In order to achieve high part quality for large, real-world components, a systematic approach to process parameter optimisation should be envisaged. In this regards, the Design of Experiments (DoE) theory (Montgomery, 2017) provides a spectrum of statistically-robust methodologies for the analysis of the effect of multiple processing parameters on the quality of the build (e.g., porosity and crack formation) while minimising the number of necessary experimental runs.

A preliminary attempt to optimise the SLM process parameters for large component production has been carried out in this research using the Box-Behnken design, an experimental design strategy for implementing the Response Surface Methodology (RSM) (Montgomery, 2017). The test samples were manufactured using a Renishaw *AM250* SLM system, which was preferred over the Realizer *SLM50* because of the bigger building volume, which in turn allows for the production of more samples within the same building process. The employed Renishaw *AM250* system uses a pulsed laser with peak power $P=200$ W. The factors (i.e., the processing parameters) considered in our preliminary study are laser power, laser point distance, laser exposure time and hatch distance. The layer thickness has been fixed to $50\text{ }\mu\text{m}$ in order to maximise productivity, given our ultimate goal to produce relatively large components. The scan strategy has also been fixed to the checker-board pattern since this is widely considered as the most suitable strategy for minimising the introduction of residual stress (refer to Table 4-1 in Section 4.5). For each process parameter combination, four replicated samples were created, in order to achieve statistical robustness. The test samples used for this process parameter optimisation attempt were 1-track thick walls of dimensions $10\times 10\text{ mm}$ and $10\times 10\times 10\text{ mm}$ cubes. The thin wall samples were built in order to obtain a window of feasible process parameters. Clearly, the hatch spacing was not considered for thin wall production. We considered two objective functions: the wall thickness, which represents an index of wall densification, as well as the fraction of cracked/warped walls over the total number of replicates. For the first objective function linear, interaction and second order terms were included in the response model. Since the value of the second

objective function is bound between 0 and 4, the logistic regression is used to fit the data and only the linear terms of the logistic regression equation are considered (i.e., the factors interactions and the higher order effects are excluded). The process parameter values used for thin-wall construction are listed in Table 10-1. Table 10-2 shows the significance of the different process parameters in describing variations in wall-thickness and in cracked sample fraction, as expressed by the p -value. A p -value <0.05 is an indicator of statistical significance. Based on the statistical analysis results it was established that the laser power does not significantly affect crack formation. Therefore, this variable was fixed to its maximum (i.e., 200 W) for the subsequent step of cubic sample production, in order to improve part density as well as the productivity of the process. Furthermore, a new, narrower window in the exposure time-point distance design space was identified.

Table 10-1: Summary of the Box-Behnken experiment for thin wall production

Factor	Low Value	High Value
Laser Power (W)	150	200
Exposure Time (μ s)	20	80
Point Distance (μ m)	20	60
Hatch Spacing (μ m)	-	
Layer Thickness (μ m)	50	

Table 10-2: Factor significance in describing variation of the objective functions, in terms of the p -value. All interaction and second order terms have $p>0.05$ and are therefore not reported.

Factor	Wall-Thickness	Cracked sampled fraction
Laser Power	0.025	0.5
Exposure Time	<0.001	<0.001
Point Distance	<0.001	<0.001

For the production of cubic samples the objective function is represented by the *c.c.l.* (refer to Section 6.8.1 for a definition of this parameter). Sample density was not directly measured, and instead it was assumed that part density increases with laser energy density. Since laser energy density does not require any measurements, but can be directly calculated from the process parameters, this approach allows accelerating the parameter optimisation procedure. The process parameter values used for cubic sample production are listed in Table 10-3, while factor significance in describing response variations is reported in Table 10-4 for the significant factors (i.e., for the terms with $p < 0.05$). The results show that the exposure time has the larger effect on crack formation, while the hatch spacing does not seem to affect it. Interestingly, the results also suggest that the point distance has a significant non-linear effect on crack formation, as shown by the effect plot in Figure 10-15. While the reason for this behaviour is not clear and might be linked with the choice of using the *c.c.l.* as a measure of crack formation, it is possible that the explanation has physical grounds, e.g., that a very low laser point distance favours stress relaxation through substrate remelting.

Table 10-3: Summary of the Box-Behnken experiment for cubic sample production

Factor	Low Value	High Value
Laser Power (W)	200	
Exposure Time (μs)	40	100
Point Distance (μm)	20	60
Hatch Spacing (μm)	40	100
Layer Thickness (μm)	50	

Table 10-4: Factor significance in describing variation of the *c.c.l.*, in terms of the *p*-value (only significant terms are reported)

Factor	<i>c.c.l.</i>
Exposure Time	<0.001
Point Distance*Point Distance	0.01

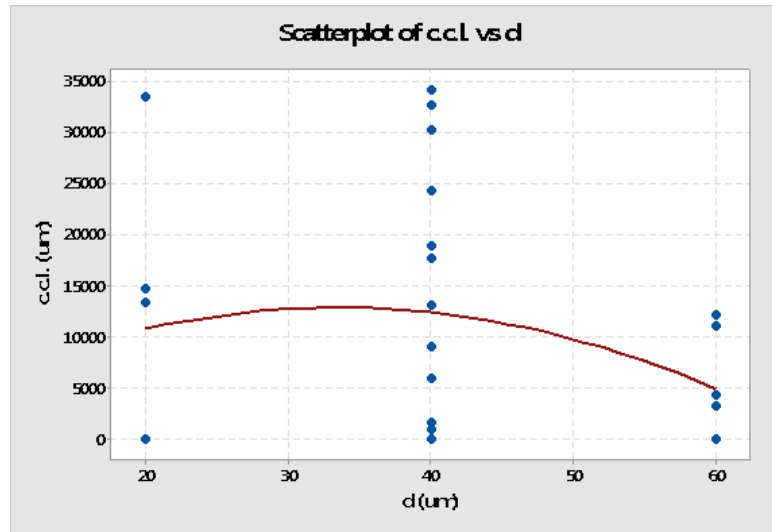


Figure 10-15: *c.c.l.* vs Point Distance (*d*) effect plot

10.4.2 Rotor Manufacturing

Based on the preliminary results presented above the topologically optimised rotor in Figure 10-14 has been manufactured using the process parameters in Table 10-5. It can be seen that the exposure time was fixed to a low value to hinder crack formation. Based on the nonlinear *c.c.l.* vs. point distance relationship shown in Figure 10-15, point distance was fixed to the highest value. On the other hand, the hatch spacing was fixed to its minimum, in order to increase laser energy density without prompting crack formation. After cross-sectioning of the cubic samples, a metallographic analysis revealed that the selected optimised process parameters in Table 10-5 could only achieve around 95% part density (i.e., 5% porosity). As shown in Figure 10-16, using the process parameters listed in Table 10-5 rotor manufacturing via SLM could

be completed successfully in around 48h of print-time, but the final part is affected by cracks in correspondence of the smallest cross-sectional features.

The results presented above suggest that further effort is needed in order to build a defect-free rotor in high-silicon steel by means of SLM technology.

Some possible developments left for future research are suggested below:

- **Component design for AM (DfAM):** in the present research TO has been employed for the design of the component with the goal of optimising its operational performance. Considerations regarding the specific constraints of the manufacturing technology have not been included in the design phase. Component DfAM should be targeted at the removal/modification of the geometrical features more susceptible to cracking, especially the geometrical features characterised by a small cross-sectional area. DfAM can be implemented as a manual step to be performed in CAD environment after the TO step, or introduced in the form of constraint into the formulation of the TO problem. The first implementation approach is expected to impose a number of practical challenges, because robust commercial CAD tools allowing designers to perform operations on faceted models issued from TO software are still in their infancy to this day;
- **Local modification of the process parameters:** parameter optimisation has been performed in this research using cubic specimens of fixed size. Therefore, the selected parameters are only optimal when used to scan layers whose representative length is approximately constant and similar to that of the cubic specimens used for the optimization. Real components, such as the rotor considered in this

research are characterised by complex cross-sectional shapes that cause the laser scan vector length to change across a single layer, thus making the selected process parameters suboptimal when the scan vector length presents significant variations. This issue can be tackled by changing the scanning laser parameters (especially the power and speed) in order to compensate for the reduced or increased time-delay between adjacent tracks. The drivers and controllers of SLM machines currently available on the market do not support the local adjustment of the process parameters, and it is therefore foreseen that such a research task would require more effort than the other development steps presented in this list;

- **Alloy modifications:** the result that the manufactured component is affected by both crack formation and high porosity suggests that the brittle nature of the considered Fe-6.9%Si alloy makes its processability for the production of large components a challenging task. Since the brittle nature of high-silicon steel is due to the high silicon content, as thoroughly discussed in Section 3.3, it is of interest to study the effect of gradual silicon removal on the processability of the alloy by SLM. An obstacle to this task is that Fe-Si powders are hard to source from commercial atomised powder suppliers in pre-alloyed form. Therefore, in-situ alloying of existing Fe-6.9%Si and pure Fe powders could allow researchers to perform parametric studies aimed at finding the optimal Fe-Si composition in terms of electromagnetic losses and processability.

Table 10-5: Process parameter values employed for rotor manufacturing via SLM

SLM Parameter	Value
Laser Power (W)	200
Exposure Time (μ s)	40
Point Distance (μ m)	60
Hatch Spacing (μ m)	40
Layer Thickness (μ m)	50

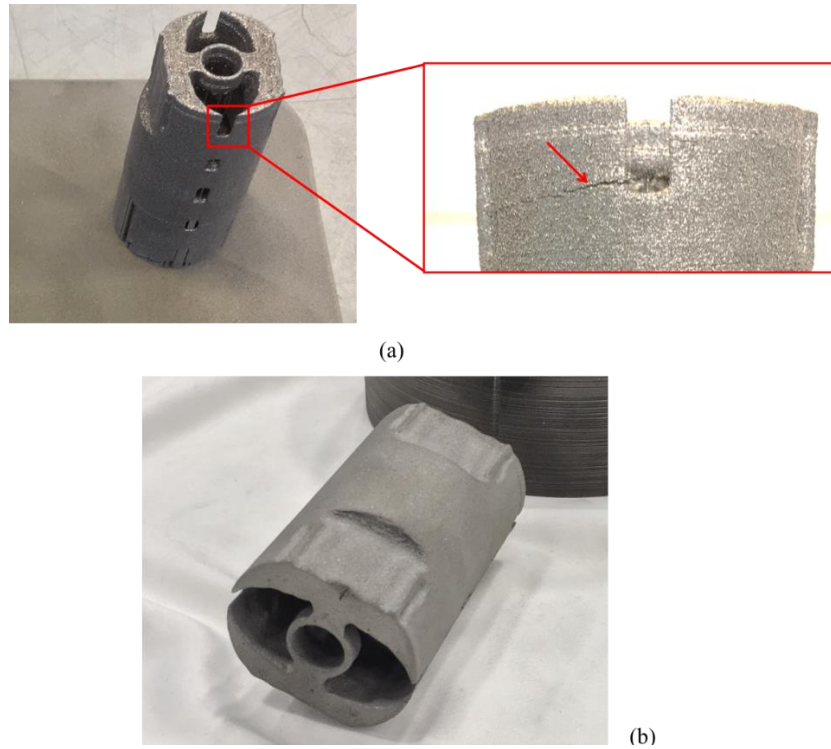


Figure 10-16: PM rotor manufacturing via SLM. In (a), the as-built component, still anchored to the build platform by means of support structures. The zoomed region shows the presence of the crack. In (b) the rotor after support structure removal and sand blasting.

10.5 Summary

This chapter has presented the applicability of a multi-objective BESO methodology for the magnetostatic and structural TO of a PMM rotor core, aimed at improving the motor's torque-to-weight ratio (torque density). The methodology was first tested on a two-dimensional model of the PMM

considering both linear and nonlinear magnetic properties. Next, the methodology was applied to a three-dimensional model, for which a solution could be found within an acceptable computational timeframe of around 24 hours. Using the proposed multi-functional TO approach, both 2D and 3D rotor designs weighing 50% less than the full design could be produced that generate the required average torque while maintaining the structural integrity under the static loads (torsional and centrifugal).

When the proposed structural-magnetostatic TO scheme is extended to the three dimensions, the optimal design is found to be three-dimensional in nature, suggesting that the traditional, 2D-based approaches to rotor design are suboptimal, as they assume that the structural and magnetic loadings do not vary along the rotor longitudinal axis. Therefore, we proposed that a suitable way of manufacturing the obtained 3D geometry is by means of laser AM technology (SLM), using high-silicon steel as the core material.

The production of the optimised rotor core by means of SLM using high-silicon steel was finally attempted, but despite the systematic parameter optimisation performed using a DOE-based approach, the final component was characterised by the presence of porosities and cracks. Future work is therefore necessary to improve the processability of the alloy, either by means of more extensive parameter optimisation efforts, through the modification of the alloy chemistry, and via re-design of the rotor for AM.

11. General Discussion

The results presented in the previous chapters have revealed the twofold novelty of our research work. First, the production of soft magnetic material for electromechanical applications by means of AM has been successfully achieved for the first time. Importantly, it could be shown that the quasi-static magnetic properties of the produced samples are at least comparable to those of the commercial laminated material. The considered high-silicon steel is a soft ferromagnetic material of great interest in the electromechanical industry, thus these results could potentially hold a commercial interest. The second novel aspect of this work has been the use of TO for the structural-magnetostatic design of real-world rotor cores in three-dimensions, which has also never been attempted in the literature before. While TO approaches have been employed for the magnetostatic design of magnetic cores, multi-physical TO for the design of real-world components is still unexplored to the authors' knowledge. Importantly, the above mentioned advancements have allowed us to achieve the goal initially set for this thesis, i.e., to demonstrate that AM has enormous potential in the field of electrical machine development. Indeed, the potential of AM to electrical machine design and fabrication is not limited to the aspects considered in this research work. As a matter of fact, the AM technologies could potentially affect the whole life-cycle of an electrical machine, starting from the acceleration of the prototyping stage, to the design and manufacture of highly-efficient heat-sinks for the control systems (power electronics). In the following sections of this chapter we summarize the main contributions of this work, and discuss them with regard to the objectives initially set in Section 2.6.4.

11.1 Main Contributions in Material Processing and Characterisation

The first objective set for this research regarded the selection of process- and post-process parameters capable of producing high-silicon steel samples characterised by good microstructural and functional properties (objective 1.1 in Section 2.6.4). This objective has been achieved, as in Chapter 7 we showed that, at least for small components (cubes and rings), near-full density could be achieved. Unfortunately, due to its brittle nature, the high-silicon iron alloy tended to develop in-process cracks induced by the localised application of the laser heat source. However, for cubes and ring samples, a density of around 98.5% could be achieved without causing crack formation. Importantly, it was shown that, given sufficiently high laser energy input, the layer-by-layer nature of the SLM process produced a strong fibrous $\langle 001 \rangle$ crystallographic texture along the build direction. Since the $\langle 001 \rangle$ directions in a b.c.c. crystal are easy axes of magnetisation, while the $\langle 111 \rangle$ are hard axes of magnetisation, the observed crystallographic texture is desirable for magnetic applications, because the $\langle 111 \rangle$ is kept out of both horizontal and vertical planes of a sample. While the long columnar grains in the as-built samples were characterised by a small cross section (with a diameter of approximately 30 μm), which is not ideal for magnetic applications since a large grain boundary area increases the hysteresis loss and limits the maximum permeability, in Chapter 8 we showed that after a simple one-hour annealing step at a temperature of 1150 °C aimed at enlarging the grain size and relaxing the process-induced residual stress, a dramatic improvement in the magnetic properties could be achieved. This result leads to the second objective set at the

beginning of this work, i.e., to compare the functional properties of the processed material with the traditionally-manufactured counterpart (see objective 1.2 in Section 2.6.4). In this respect, the final quasi-static magnetic properties of the annealed samples, with maximum permeability $\mu_{max}=24000$ and coercivity $H_c=16$ A/m, were found to be comparable to those of commercial laminated grade JNEX. The power loss at 50 Hz was found to be approximately four times higher than for JNEX, but this can be attributed to the thickness of the produced ring, which is twenty times higher than for the lamination sheet (2 mm and 0.1 mm, respectively). In this respect, it seems that given the same sample thickness the SLM material could produce lower power loss than the laminated grade. Besides the achievement of the initially set objectives, the relationship between process conditions, observed microscopy (phase distribution, grain morphology, crystallographic texture) and magnetic properties has also been established. This in turn sheds light on the benefits and challenges of the application of SLM to the fabrication of soft magnetic materials. An open challenge regards the fabrication of large, complex components, such as the rotor core considered for re-design as part of this work. The aspect of real-world component production has been addressed as part of the third objective of this research, which regarded the manufacture of the optimised rotor model by means of SLM using the developed high-silicon steel. While the manufacturing process could be accomplished, as shown in Section 10.4, the formation of cracks in the build could not be avoided. In correspondence of certain geometrical features, especially tips and overhangs, overheating causes the magnitude of the residual stress to raise considerably compared to other regions where the thermal history is closer to that

experienced by the test samples on which the process parameter optimisation had been performed. The brittle nature of the high-silicon steel, which is caused by the high silicon content, makes the alloy incapable of resisting the strong residual stress and cracks are formed.

11.2 Main Contributions in Component Design

After showing that soft magnetic material with good magnetic properties could be produced using SLM technology, in Chapter 9 and Chapter 10 we have provided an example of how the developed material could be employed in an electromechanical application. In Chapter 9 we have proposed a methodology for the coupled structural-magnetostatic TO of a general electromechanical component, aimed at jointly minimising the magnetostatic and strain energies while satisfying a volume constraint. The developed methodology satisfies objective 2.1. In Chapter 10, the proposed TO methodology has been applied to a real-world machine rotor core design, in an effort to satisfy the objective 2.2. Specifically, Chapter 10 has focussed on the free-form design of a synchronous machine rotor core, in order to show that AM technology can indeed lead to the production of lighter electrical machine components. The TO-based approach employed for the magnetostatic and mechanical optimisation of the rotor structure achieved a decrease in component weight by 50% without affecting its electromagnetic performance, nor its structural integrity under torsional and centrifugal static loading. The improvement of the torque density of an electrical motor is particularly desirable for applications where weight saving is a priority, such as in the aerospace and automotive industries. While the resulting rotor construction as a solid component is expected to produce more eddy current losses than the laminated counterpart, for applications like that

presented in this study, characterised by a synchronous rotation of the rotor, and given the relatively low operational rotation speed (10000 rpm), it is expected that the increase in losses is not significant.

An important objective of this work regarded the use of the measured magnetisation curve of selective laser melted high-silicon steel in the optimisation process (see objective 2.3 in Section 2.6.4). This objective was achieved in Section 10.3.2, in which the 3D TO scheme for the magnetostatic-structural design of the rotor was carried out. While only the linearized B-H curve was employed, in order to reduce the computational effort, the scheme could be easily extended to the non-linear case. Therefore, the technology developed in this work allowed us to assess how both the design strategy and material properties affect the performance of the device as compared to the traditionally-design and produced part.

12. Conclusions and Further Work

Demonstrating the impact of AM on the development of electrical machines from both materials and design perspectives was the main goal of this work. The outcome is represented by the following advancements in the fields of material development and electromechanical device design:

- For the first time, it has been shown that near-fully dense high-silicon steel cubes and rings with good magnetic properties can be produced by means of SLM technology and simple post-annealing;
- The metallurgy of high-silicon steel parts produced by SLM have been explained for the first time, by relating the observed microstructural features (grain size and morphology, defects, crystallographic texture, and phase distribution) to the SLM and post-annealing parameters;
- The quasi-static and dynamic magnetic properties of SLM high-silicon steel could be measured and explained in light of the observed microstructural properties of the samples;
- From a component design perspective, we proposed a novel TO approach for the structural and magnetostatic design of the rotor core of a synchronous machine. Besides the computational attractiveness, the proposed TO methodology was capable to reduce the weight of the rotor component by 50% while maintaining its structural integrity under centrifugal and torsional static loads, as well as delivering the target torque.

The described outcomes have the potential to pave the path towards the employment of AM in the development of electrical machinery for the creation of lighter and hence more efficient devices.

12.1 Further Work

12.1.1 Material Processing

A number of further actions are proposed to address the issue of silicon steel processability by means of SLM technology, namely:

- 1) Modify the composition of the alloy, in order to improve ductility, and therefore counteract its tendency to form cracks under the effect of residual stress. Since the high Si content is the main responsible for alloy embrittlement (refer to Section 3.3 for a description of the metallurgical and functional effects of Si addition to Fe), it seems reasonable that a first step should consist in gradually reducing the Si content. This could be achieved, for example, by mixing the pre-alloyed Fe-6.9%Si powder with high-purity Fe powder and attempt in-process alloying;
- 2) As an alternative to silicon content reduction, elements that increase the ductility of the alloy could be added. A straight-forward candidate is Al, which is often added to Fe-Si alloys to improve their ductility, thus making the production of both laminated and powder cores less prone to failure. Furthermore, Al also increases the resistivity of Fe-Si. A drawback of Al addition is the decrease in permeability and saturation polarisation;
- 3) Modify the component design for AM, especially in correspondence of geometrical features that are more susceptible to cracking. This aspect is discussed in the next section.

12.1.2 Component Design

The proposed approach to rotor design using TO has only been tested on one case study, which regards one of the simplest types of rotor core. Extending the developed design strategy to other PM motor designs, especially those involving a higher number of poles, is seen as a natural development of this research. Furthermore, it is not clear whether the proposed approach could be extended to the case of other rotor topologies, especially the SyR machine. Since SyR performance relies on the difference in inductance (and hence in magnetic co-energy) between the d- and q-axis, the objective function (9.7) could be extended to the case of SyR rotors as follows:

$$\max_x f = |W_d - W_q|, \quad (12.1)$$

where W_d is the co-energy of the motor when the stator is aligned with the rotor d-axis, while W_q is the co-energy of the motor when the stator is aligned with the rotor q-axis. Since the above objective function could assume non-positive values, it is important that the rotor initial design is “close enough” to optimality, in order to guarantee the convergence of the BESO algorithm.

While the approach to rotor design proposed in this research was shown to improve the torque density, the implemented BESO methodology is only suitable for magnetic energy minimisation. Extending the optimisation to more complex objective functions and introducing arbitrary constraints (regarding, for instance, the manufacturing constraints and electrical performances) requires the employment of a gradient-based optimisation scheme, such as the SLP-based approach proposed in (Lee et al., 2010). A comparative assessment

of the two methodologies is therefore of interest, especially from a computational burden stand-point.

In the present research the topologically optimised rotor did not undergo a step of “manual” design after the TO step. In particular, electrical machine-specific considerations have the potential to improve both the electromechanical performance and manufacturing feasibility. One example is the skewing of the rotor core. This is a typical strategy adopted in laminated core production for torque ripple minimisation (Bomela and Kamper, 2002). In SLM, skewing could also decrease the accumulation of residual stress, by reducing the amount of overlap between subsequent layers.

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