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**Fatigue Behavior Of Additively Manufactured LPBF-M789 Maraging Steel
Under The Influence Of Discontinuities**

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DEDICATION

In loving memory of my late father, brother, and grandmother,

Sanni Suleiman, Nurudeen Sanni, and Amina.

Your memories remain a constant source of inspiration to me.

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RÉSUMÉ

Cette recherche doctorale s'inscrit dans le cadre du projet FatCo, un programme international de recherche sur la fatigue-corrosion réunissant Hydro-Québec, SACMI, Polytechnique Montréal, l'Université McGill, l'École de Technologie Supérieure (ETS) et Politecnico di Milano. Ce programme vise à améliorer la compréhension et la modélisation des mécanismes d'endommagement dans les roues de turbines hydrauliques et les outils de formage, en tenant compte de la fatigue, de la fatigue-corrosion et des hétérogénéités des matériaux dans des conditions de service.

L'objectif de cette thèse est d'étudier le comportement en fatigue de l'acier maraging M789 fabriqué par fusion laser sur lit de poudre (LPBF), en analysant l'influence des discontinuités intrinsèques et extrinsèques, en évaluant les modèles de prédiction de la résistance à la fatigue basés sur les discontinuités, en identifiant les facteurs gouvernant la sévérité des discontinuités et en caractérisant les mécanismes de rupture en fatigue de l'alliage.

L'approche expérimentale repose sur la qualification du matériau par analyse chimique, caractérisation microstructurale, essais de traction et mesures de dureté, afin de vérifier la conformité aux propriétés attendues de l'alliage. Des éprouvettes ont été produites dans différentes conditions de surface : polies (FP), telles qu'imprimées (FA), piquées (FC) et avec canaux internes (FIC), afin de représenter respectivement des discontinuités intrinsèques de fabrication, des vallées de rugosité, des piqûres de corrosion et des caractéristiques géométriques internes. Des piqûres extrinsèques ont été générées par polarisation potentiodynamique cyclique (CPP) et par une technique exploratoire de polarisation potentiostatique (PotPit) dans une solution de NaCl à 3,5 %. La caractérisation non destructive, la métallographie, les mesures de rugosité et l'analyse par éléments finis ont été réalisées avant les essais de fatigue.

Quel que soit l'état de surface, le comportement en fatigue est dominé par la présence de discontinuités. Aucun des échantillons n'a atteint la résistance à la fatigue sans discontinuité estimée à 550 MPa. La résistance maximale observée pour les éprouvettes polies (FP) correspond à environ 52 % de cette valeur idéale, tandis que les éprouvettes FA, FC et FIC présentent des résistances comprises entre 10 % et 30 % en raison de discontinuités de plus grande taille.

Les résultats expérimentaux ont été analysés à l'aide d'un diagramme de Kitagawa–Takahashi afin d'évaluer les prédicteurs de résistance à la fatigue en fonction de la taille des discontinuités. Les modèles d'El-Haddad et de la mécanique linéaire de la rupture (LEFM) ont permis de prédire avec précision la résistance à la fatigue, alors que le modèle de Murakami non. La taille des discontinuités à l'origine de la rupture varie entre 42 et 2860 μm , tandis que la taille critique est estimée à environ 15 μm . Les discontinuités extrinsèques, telles que les piqûres et les canaux internes, présentent un comportement de type entaille. La combinaison de techniques de caractérisation non destructives et de statistiques des valeurs extrêmes (EVS) permet d'obtenir des prédictions conservatrices pour les éprouvettes polies et piquées, mais reste limitée pour les éprouvettes telles qu'imprimées.

Au-delà de la taille des discontinuités, la géométrie et la proximité des discontinuités influencent la résistance à la fatigue. Une discontinuité intrinsèque plus petite mais plus aiguë (123 μm) peut dominer une discontinuité extrinsèque plus grande mais plus lisse (182 μm). Les effets d'interaction entre discontinuités voisines restent limités, la rupture étant principalement contrôlée par la discontinuité dominant, ce qui est cohérent la faible plasticité de l'alliage.

Bien que la LEFM décrive correctement le comportement des fissures longues, la microstructure de l'alliage contribue aux mécanismes de rupture en fatigue. Les facettes cristallographiques (52–67 μm) observées à l'origine de la rupture sont comparables à la taille des grains parents austénitiques. Lorsque les discontinuités géométriques et les hétérogénéités microstructurales coexistent à des échelles similaires, plusieurs mécanismes d'amorçage peuvent entrer en compétition.

Afin d'assurer des prédictions prudentes et de traiter les valeurs aberrantes, une analyse par éléments finis a été utilisée pour estimer les contraintes locales associées aux piqûres et aux canaux internes. L'approche basée sur les contraintes locales fournit une description plus représentative du comportement en fatigue que l'approche basée sur la contrainte nominale.

Enfin, un diagramme de Kitagawa–Takahashi proposé pour l'alliage M789, limitant la résistance maximale à la fatigue à 275 MPa, est en accord avec les résultats expérimentaux. Ce modèle intègre une taille de critique d'environ 17 μm , déterminée à partir de l'intersection entre la résistance à la fatigue projetée correspondant à un comportement d'entaille complète ($k_t = 2$) et la ligne de prédiction de la mécanique linéaire élastique de la rupture (LEFM). Cette approche permet de

s'affranchir de la résistance idéale et met en évidence le régime contrôlé par les discontinuités pour la prédiction de la résistance à la fatigue à 2.10^6 cycles.

Dans l'ensemble, ce travail contribue à une meilleure compréhension du rôle des discontinuités dans le comportement en fatigue de l'acier maraging M789 fabriqué par LPBF et propose un cadre prédictif pour l'évaluation de la résistance à la fatigue des matériaux sensibles aux discontinuités.

ABSTRACT

This doctoral research is part of the FatCo Project, an international research program on fatigue–corrosion involving Hydro-Québec, SACMI, Polytechnique Montréal, McGill University, École de Technologie Supérieure (ETS), Politecnico di Milano, and the University of Ferrara. The FatCo program aims to improve the understanding and modelling of damage mechanisms in hydraulic turbine runners and forming tools by accounting for fatigue, corrosion–fatigue, and material heterogeneities under demanding operating conditions. The objective of this thesis is to investigate the fatigue behaviour of LPBF-produced M789 maraging steel by examining the influence of intrinsic and extrinsic discontinuities, evaluating discontinuity-based fatigue strength prediction models, identifying the factors governing discontinuity severity, and characterising fatigue failure mechanisms.

The adopted methodology includes material qualification through chemical analysis, microstructural characterisation, tensile testing, and hardness measurements to confirm conformity with expected alloy properties. Specimens were produced in polished (FP), as-printed (FA), pitted (FC), and internally channelled (FIC) conditions to represent intrinsic manufacturing discontinuities, surface roughness valleys, corrosion-induced pits, and internal geometric features. Extrinsic pits were generated using cyclic potentiodynamic polarisation (CPP) and exploratory potentiostatic polarisation (PotPit) in 3.5% NaCl. Non-destructive characterisation, metallography, surface roughness measurements, and finite element analysis were conducted prior to fatigue testing.

High-cycle fatigue tests were performed at a stress ratio of 0.1 and aiming $2 \cdot 10^6$ cycles at failure using Maxwell and Nicholas step-loading protocols. Across all surface conditions, the **fatigue behaviour of the alloy is governed by discontinuities**. None of the specimens approached the discontinuity-free fatigue strength of 550 MPa, the highest fatigue strength for FP specimens was approximately 52% of this idealised value. FA, FC, and FIC specimens, containing larger discontinuities, exhibited fatigue strengths between 10% and 30% of the discontinuity-free value.

Experimental results were analysed using a Kitagawa–Takahashi diagram to evaluate fatigue strength predictors based on discontinuity size. **The El-Haddad and LEFM models accurately predicts the fatigue strength of the alloy in the long crack region, whereas the Murakami model did not.** Discontinuity sizes at failure origins ranged from 42 to 2860 μm , while the critical

crack size was estimated at approximately 15 μm . Extrinsic discontinuities such as pits and internal channels tend to follow notch-controlled behaviour. A combination of **non-destructive characterisation and extreme value statistics (EVS) enables conservative predict the fatigue strength of polished and pitted specimens, but not for as-printed specimens.**

Beyond discontinuity size, **geometry and proximity influence the fatigue strength.** A smaller, sharper intrinsic discontinuity (123 μm) may dominate over a larger but smoother extrinsic discontinuity (182 μm). Interaction effects between neighbouring discontinuities were limited, with failure governed by the dominant discontinuity, likely influenced by the limited plasticity of the alloy.

Although LEFM successfully describes long-crack behaviour, **the alloy microstructure contributes to its fatigue failure mechanisms.** Crystallographic facets (52–67 μm) observed at failure origins are comparable in size to the large prior austenite grain size. These results suggest that when geometric discontinuities and microstructural features coexist at similar length scales, multiple crack initiations compete, reinforcing the competing mechanisms for fatigue failure of the alloy.

To ensure conservative predictions and account for outliers, finite element analysis was used to estimate local stresses associated with pits and internal channels. **The local stress approach accurately models the fatigue behaviour of the alloy and conservatively accounts for the outlier data.**

Finally, a Kitagawa–Takahashi diagram proposed for M789, limiting the peak fatigue strength to 275 MPa, is consistent with the experimental results. This formulation incorporates an intrinsic transition size ($\sim 17 \mu\text{m}$) determined at the intersection of the fatigue strength at full notch behaviour (for $k_t = 2$) in the ALD and the LEFM prediction line. It avoids reliance on the discontinuity free fatigue strength of 550 MPa and emphasises the discontinuity-controlled region as a basis for fatigue strength prediction of LPBF M789 alloy at $2 \cdot 10^6$ cycles. Overall, this work advances the understanding of discontinuity-driven fatigue in LPBF M789 maraging steel and provides a practical framework for fatigue strength prediction in discontinuity-sensitive materials.

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LIST OF SYMBOLS AND ABBREVIATIONS

σ_a	Stress amplitude (MPa)
$\sigma_{b\#}$	Fatigue strength level at a hypothetical discontinuity size <i>b</i> (MPa)
σ_{FS}	Fatigue strength in the presence of discontinuities (MPa)
σ_0	Discontinuity-free fatigue strength (MPa)
$\sigma_{f,local}$	Local stress (MPa)
$\sigma_{f,nominal}$	Nominal stress (MPa)
$\sigma_{w,0}$	Fatigue strength of discontinuity-free material for a stress ratio of -1 (MPa)
$\sigma_{w,0.1}$	Fatigue strength of discontinuity-free material for a stress ratio of 0.1 (MPa)
$\sigma_{w,Nlife}$	Computed fatigue strength at N_{life} cycles (MPa)
$\sigma_{\theta\theta}$	Tangential stress (MPa)
$\Delta\sigma$	Stress range (MPa)
β	Parameter that is dependent on the notch opening angle
θ	Angular position around the crack tip (degrees)
γ	Stress field exponent
ρ	Notch tip radius (μm)
$\sqrt{area}$	Projected area of discontinuity perpendicular to the loading direction (μm)
K_I^V	Stress intensity of a notch ($\text{MPa}\sqrt{m}$)
$\Delta K_{I,max}$	Maximum stress intensity range ahead of an arbitrary crack front ($\text{MPa}\sqrt{m}$)
ΔK_{th}	Stress intensity threshold range for cracks ($\text{MPa}\sqrt{m}$)
$\Delta K_{th,LC}$	Stress intensity range threshold for long cracks ($\text{MPa}\sqrt{m}$)
$\Delta K_{th,sc}$	Stress intensity range threshold for short cracks ($\text{MPa}\sqrt{m}$)
$\Delta K_{I,th}^V$	Stress intensity threshold range for notch ($\text{MPa}\sqrt{m}$)

A_{ref}	Reference area (mm ²)
A_0	Inspection area (mm ²)
E_{pit}	Pitting potential (V)
I_p	Current density limit (mA)
I	Current density (mA)
k_t	Stress concentration factor
$k_{t,R}$	Stress concentration factor associated with roughness at the notch root
N_{life}	Number of cycles of interest
P	Load (kN)
R	Stress ratio
$R_{v,max}$	Maximum value of the depth of a trough on a roughness profile (μm)
R_a	Average roughness (μm)
R_v	Deepest valley from mean roughness line (μm)
R_z	Total profile height (μm)
S_u	Tensile strength MPa
S_{min}	Minimum separation distance (μm)
T	Return period
V_{ref}	Reference volume (mm ³)
V_0	Inspection volume (mm ³)
Y	Geometric factor
a	Crack length (μm)
a_0	Fictitious length (μm)
a_o^V	Material length parameter for a notched component (μm)

a^*	Critical size of discontinuity for notch to crack transition (μm)
a_{eff}	Effective crack size (μm)
a_{eff}^*	Effective transition size (μm)
b	Size of hypothetical discontinuity (μm)
b	Half width of roughness pitch (μm)
c	Crack depth of a shallow slender crack (μm)
l	Width of a shallow slender crack (μm)
r	Radial distance from crack tip (μm)
ALD	Atzori and Lazzarrin diagram
AM	Additive manufacturing
BD	Building direction
CAE	Complete Abaqus environment
CE	Counter electrode
CAD	Computer aided design
CPP	Cyclic potentiodynamic polarisation
DED	Direct energy deposition
EBM	Electron beam welding
EDS	Energy dispersive X-ray spectroscopy
ETS	Ecole de technologie superieure
EVS	Extreme value statistics
FA	Fatigue as-printed specimen
FC	Fatigue-corroded specimens
FCG	Fatigue crack growth

FE	Finite element
FEA	Finite element analysis
FIC	Fatigue-internal channel specimens
FIB	Focused ion beam
FP	Fatigue-polished specimen
FS	Fatigue strength
HCF	High-cycle fatigue
HIP	Hot isostatic pressing
HV	Vickers hardness
ID	Identification
KT	Kitagawa Takahashi
LC	Long crack
LEFM	Linear elastic fracture mechanics
LENS	Laser-engineered net shaping
LOF	Lack of fusion
LPBF	Laser powder bed fusion
LPBF	Laser powder bed fusion
MSC	Microstructural short crack
NA	Not available
ND	Natural discontinuities
NDC	Non-destructive characterisation
NMI	Non-metallic inclusions
OM	Optical microscope

PAGS	Prior austenite grain size
PD	Printing direction
PH	Precipitation hardened
PRI	Process-related influence
PSC	Physical short cracks
PSFO	post-surface finish operations
PotPit	Potentiostatic pitting
RE	Reference electrode
RIC	Roughness-induced crack closure
SEM	Scanning electron microscope
SLM	Selective laser melting
S-N	Stress versus number of cycles
WE	Working electrode
XCT	X-ray computed tomography

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CHAPTER 1 INTRODUCTION

Additive Manufacturing (AM) has become a mainstream technology across industries like biomedical, aerospace, sports, and energy sectors, to mention a few. Its adoption is driven by its ability to produce complex parts with unique structural properties, which cannot be achieved with conventional manufacturing processes [1]. This feature has also helped minimise the need for extra tooling during part production. Other advantages of AM are production flexibility, reduced lead times, reduced emissions, and the array of applicable materials. Gartner [2], in his review, advanced product development, prototyping, and innovation as the three most cited reasons for the adoption of AM technology.

The role of AM in product development is often understated despite its importance. Particularly in the aspect of research on new materials or alloys. Manufacturers of AM parts are increasingly engaging in leveraging the benefits of AM to introduce new metallic alloys with tailored properties to achieve targeted mechanical performance. This is achieved via the exploration of the processing, structure, and property relationships of materials typically used by researchers and manufacturers [3]. A correlation can be made to the welding process, often performed during the joining of materials, with a consideration of the resulting microstructural and mechanical properties. Laser-based AM processes operate similarly, as they use a localised heat source to facilitate production, accompanied by a high cooling rate. This breeds an opportunity for control over processing variables that result in alloys with properties attained via a one-phase production process, something not easily achieved with conventional manufacturing. Interestingly, one should note that it becomes a simultaneous process of part production and property enhancement, which may not require a post-heat treatment.

Based on these considerations, Böhler recently used the powder bed fusion variant of AM to develop a precipitation-hardened martensitic stainless steel, code-named M789. The alloy is described as having a good combination of strength (> 1700 MPa), hardness (50 HRC), toughness, and corrosion resistance.

Our industrial partner, SACMI, is into the manufacturing of plastic bottle caps, with high-throughput lines, where compression-moulding systems routinely reach 2,000 parts per minute [4]. At such production rates, the die insert assemblies are subjected to high-cycle loading that

makes them vulnerable to fatigue damage. Also, incorporated in the inserts are conformal cooling channels to circulate liquid coolant to cool parts faster and more evenly, improve the performance quality, and reduce the moulding cycle time. Figure 1-1 shows a coupled die insert in the press with the formation of a plastic cap, as shown in (a), whereas (b) and (c) respectively show the insert and the internal channels.

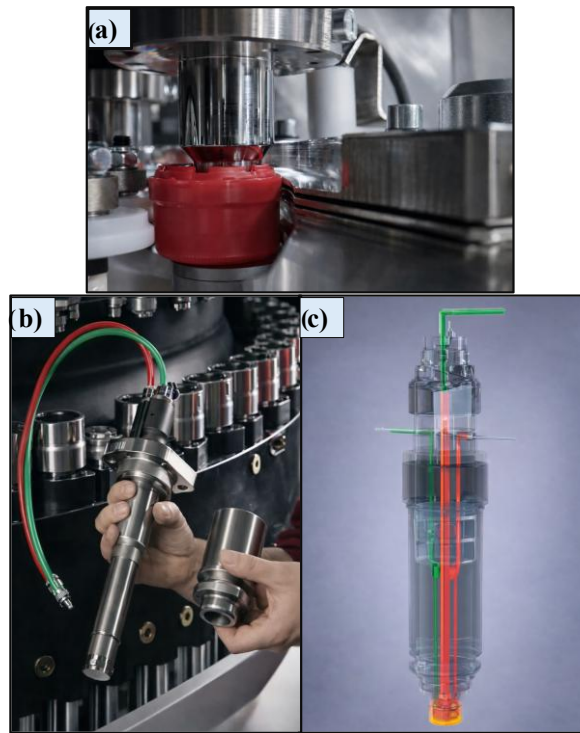


Figure 1-1. An assemblage of the die insert. (a) insert in contact with a plasticised cap polymer, (b) die insert tooling of a continuous compression moulding machine, and (c) a transparent CAD rendering of the insert showing internal cooling channel [4].

Coolant chemistry and flow can promote localised corrosion within the channels, leading to corrosion pits [5]. SACMI intends to use laser powder bed fusion (LPBF) for the production of die inserts; in this context, both manufacturing-induced discontinuities and service-induced discontinuities, such as corrosion pits, may act as stress concentrators. These features can accelerate crack initiation, reduce fatigue strength and life, and increase the susceptibility of the inserts to failure [6]. The mechanical behaviour of such discontinuities may be described using either linear elastic fracture mechanics (LEFM) or notch-based approaches. The selection of an appropriate design framework depends on understanding the behaviour of the discontinuities

present in the alloy, which in turn guides the choice of fatigue design methodology. Therefore, the fatigue behaviour of the M789 alloy, as a candidate die-insert material, needs to be investigated under the influence of the discontinuities likely to be encountered in service. This study is thus conceived to examine the response of both manufacturing-induced and externally induced discontinuities in the alloy under fatigue loading.

1.1 Document organisation

This thesis is organised into several chapters that describe the motivation, background, methods, results, and conclusions of the research.

Chapter 1 introduces the research topic and explains the context that led to this work, including the involvement of the industrial partners. The chapter ends with the research objectives that guide the study.

Chapter 2 presents the literature review. It begins with an overview of additive manufacturing and its limitations, followed by a detailed description of the M789 alloy, including its development, properties, and the limited research available on the material. The chapter also covers fatigue damage mechanisms, theories describing the effect of discontinuities on fatigue behaviour, and examples of the application of the Kitagawa–Takahashi (KT) diagram.

Chapter 3 describes the scientific approach, publication strategy, and experimental methods used in this work. It includes material qualification, chemical composition, mechanical properties, microstructure, and residual stress measurements. The chapter also describes the preparation of fatigue specimens, including the production of electrochemical pits as extrinsic discontinuities, and the fatigue testing procedures. Some experimental details are presented in this chapter, while others are included in the associated research articles.

Chapter 4 presents Article 1. It investigates the influence of discontinuities on fatigue behaviour for three surface conditions: as-printed, polished, and electrochemically pitted specimens. This chapter mainly addresses Objectives 1, 2, and 4.

Chapters 5 and 6 present the experimental results for specimens containing internal channels, together with the analysis of extrinsic discontinuities, including the interpretation of outlier results. These chapters also include the construction of the Kitagawa-Takahashi diagram using the local stress approach.

Chapters 7 and 8 focus on the construction of the Atzori Lazzarin diagram for the alloy and a general discussion centred on the research objective.

Chapter 9 finally presents the overall conclusions of the thesis, recommendations for future work, and an outlook on potential industrial applications.

CHAPTER 2 LITERATURE REVIEW

AM, often referred to as “3D printing” [7], is defined by ISO/ASTM as “the process of joining materials to make objects from 3D model data, usually layer upon layer as opposed to subtractive manufacturing and formative manufacturing methodologies” [8]. This definition recognises AM as a family of manufacturing processes that build parts using a layer-wise fabrication approach. To buttress this, the ISO/ASTM categorises AM into seven process families: binder jetting, directed energy deposition (DED), material extrusion, material jetting, powder bed fusion (PBF), sheet lamination, and vat photopolymerization [9].

Each family is distinguished by how material is delivered and consolidated, for example, droplets in material jetting, filament or pellets in extrusion, or localised heat input to powders in PBF and DED. This classification helps researchers and manufacturers connect processing routes with the microstructures and properties. Table 2-1 presents the technologies commonly employed for the manufacture of parts from metallic alloy powders. Also reflected in the table are the respective industrial processes, energy source, and feedstock material.

Table 2-1. Classification of AM processes utilising metallic materials [10, 11].

Classification	Process	Energy source	Feedstock
Powder bed fusion	Direct Metal Laser Sintering (DMLS)	High-Powered Laser Beam	Atomised metallic powder or ceramic powder
	Selective Laser Melting (SLM) or Laser Powder Bed Fusion (LPBF)		
	Electron Beam Melting (EBM)	Electron Beam	
Directed Energy Deposition	Laser Engineered Net Shaping	Laser Beam	Metallic powder
	Electronic Beam Welding		
Binder Jetting	Indirect Inkjet Printing	Thermal Energy	Metallic powder with polymer binder

2.1 Laser powder bed fusion (LPBF)

The Laser variant of the PBF methods stands out for patronage in both industrial, research, and academic settings due to its versatility across different materials. Common LPBF methodologies

are selective laser sintering, direct metal laser sintering, electron beam melting, and selective laser melting.

All these processes involve the use of a feedstock alloy powder and high-energy sources (laser or electron) to build the desired part. The basic process involves spreading a thin layer of metal powder on a build platform, and a laser melts selected regions to form the part. The platform lowers, a new powder layer is added, and the process repeats until the component is completed.

In principle, the LPBF technology is suitable for application to any material that can be melted and resolidified, such as powders of steels, Ti, Al and polymers, amongst many others. One key requirement is the printability of the alloy's powder, similar to the weldability requirements of welding processes. A schematic representation of the LPBF setup is shown in Figure 2-1. The next section explores the development of the alloy.

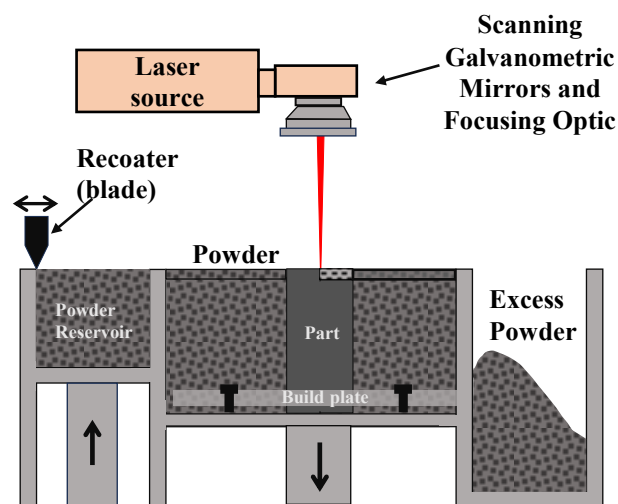


Figure 2-1. Schematic representation of the LPBF set-up [12].

2.2 The development of LPBF M789 alloy

The development of the M789 alloy incorporates the LPBF process of AM, leveraging the rapid melting and solidification features of the process to enhance the resulting mechanical properties. The alloy was designed to combine the high-strength characteristics of maraging steels with the corrosion resistance of precipitation-hardening (PH) stainless steels. Böhler developed the alloy based on a modified composition derived from PH 17-4 and 18Ni250 alloys, with carefully controlled additions of Ni, Ti, Si, and Cu. These elements enhance both the type and density of

precipitates, enabling strength levels comparable to maraging steels, while the increased Ni content promotes the formation of reversed austenite, contributing to improved toughness.

Tian et al. [13] performed atom probe tomography studies on the alloy and reported that the precipitate number densities are on the order of 10^{24} - 10^{25} m⁻³, achieved through age hardening heat treatment. This results in a balanced combination of high strength, toughness, and corrosion resistance. Table 2-2 shows a comparison of the chemical composition of the M789, PH 17-4, and 18Ni250. All three alloys maintain low carbon content (<0.05%), which supports ductility. However, M789 is distinguished by the absence of cobalt and the presence of approximately 12.2 wt.% chromium, significantly higher than in 18Ni250 and approaching the typical range of PH 17-4 stainless steel, which gives it corrosion resistance properties along with its mechanical properties.

Table 2-2. Chemical composition (wt.%) of M789 powder, 18Ni250 maraging steel and PH 17-4.

	Fe	C	Ni	Cr	Mo	Ti	Al	Mn	Si	Others
M789	Bal	< 0.02	10.00	12.20	1.00	0.50-1.50	0.60	0.10-0.50	0.40-0.80	0-0.50 Cu
18Ni250 [14]	Bal	≤ 0.03	17.00-19.00	≤ 0.50	4.60-5.20	0.30-0.50	0.05-0.15	-	-	7.00-8.50Co
PH 17-4 [15]	Bal	≤ 0.07	3.00-5.00	15.00-17.50	2.00-2.50	-	0.90-1.35	1.00	1.00	3.00-5.00 Cu 0.20-0.50 Nb

2.2.1 Production process

The tailoring of the M789's properties starts from the powder production phase up to the post-processing after the LPBF part building phase. The powder is produced using vacuum induction melting followed by gas atomisation under inert Ar gas, resulting in an optimum particle size range of 15 to 45 µm. This size range enhances the powder flowability of 15.84 s/50 g [13], packing density ≥ 3.5 g/cm³, and enhanced surface finish. Printing of parts with M789 powder is performed using the build parameters shown in Table 2-3.

Table 2-3. LPBF processing parameters of the M789 alloy [16].

Parameter	Value
Laser Power	160 W
Scan speed	763 mm/s
Hatch spacing	87 μm
Layer thickness	20 μm
Scan strategy	Zigzag stripe with 50 ° rotation
Inert atmosphere	Argon (<0.3 % O ₂)

2.2.2 Heat treatment and microstructure

The microstructure of the alloy after fabrication exhibits a fish-scale appearance observable using optical microscopy. This feature is attributed to melt-pool boundaries formed by laser-track overlap during the LPBF process (see Figure 2-2a). However, scanning electron microscopy (SEM) observations reveal dendritic features oriented along the building direction, forming a cellular network structure. The Cr and Mo contents of the alloy influence self-quenching properties, leading to the formation of martensite.

To optimise the properties of the alloy, heat treatment is typically performed after the build process. This involves solution annealing at 1000 °C for 1 h, followed by air cooling and age hardening at 500 °C for 3 h (see Figure 2-3). This treatment results in a clearly revealed martensitic microstructure with the absence of the scale-like structures observed in the as-printed condition (see Figure 2-2b). However, Palad et al. [17] reported that the scale-like features were still discernible in the grain orientation maps from EBSD. Solution annealing also relieves residual stresses and transforms any residual austenite favoured by the presence of Ni to martensite.

The ageing treatment facilitates the precipitation of nanoscale intermetallic precipitates in the microstructural matrix. Tian et al. [13] investigated the effects of ageing temperatures ranging from 400 °C to 600 °C on the mechanical properties of M789. They reported that an optimal ageing temperature of 500 °C, with a holding time of 2 h, promotes the formation of maximum hardening precipitates in the microstructure. This optimal ageing duration differs from the 3 h

recommended by the manufacturer. The variation suggests that the effective ageing time likely falls within a range of 2 to 3 h rather than being a fixed value.

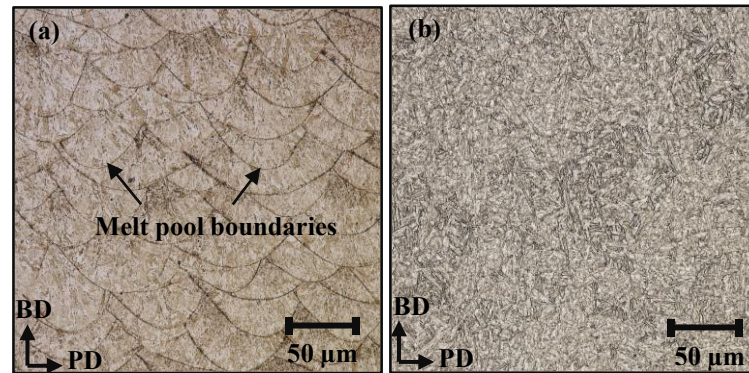


Figure 2-2. Optical micrographs showing: (a) as-printed M789 with visible melt-pool boundaries and (b) heat-treated M789 alloy without visible melt-pool boundaries.

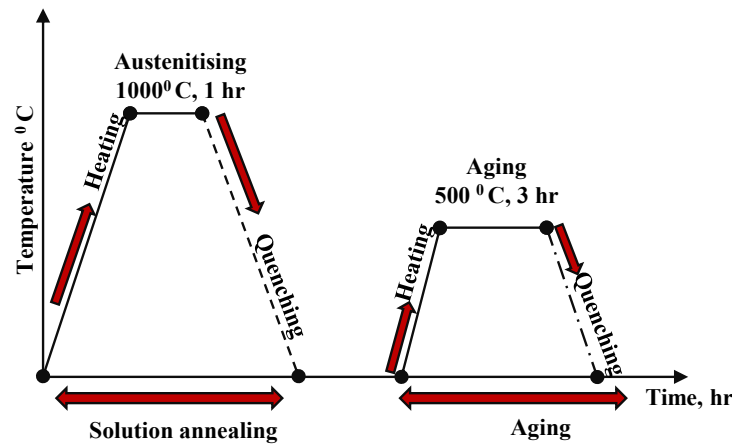


Figure 2-3. Heat treatment scheme for the LPBF M789 alloy after fabrication.

2.2.3 General mechanical properties

The mechanical properties of M789 alloy compared with those of 18Ni250, 18Ni300 maraging steels and 17-4 PH steels are shown in Table 2-4. Heat-treated M789 exhibits the highest mechanical performance, with a tensile strength ranging from 1770 to 1800 MPa and a yield strength between 1670 and 1770 MPa, surpassing both 18Ni 250 maraging steel and PH 17-4 stainless steel. Although its elongation is low (4.5–6.0%), which is typical of high-strength martensitic steels, its hardness (50–54 HRC) reflects substantial precipitation strengthening. In

contrast, 18Ni 250 offers greater ductility (17% elongation) but lower strength, and PH 17-4 stainless steel balances moderate strength (1448 MPa tensile) with improved corrosion resistance. Heat treatment of M789 enhances its strength and hardness, positioning it as a high-performance alloy for demanding structural applications.

As is typical of LPBF alloys, M789 is expected to contain discontinuities. Therefore, Section 2.3 reviews some common discontinuities observed in LPBF alloys.

Table 2-4. Mechanical properties of M789 alloy compared with other LPBF alloys.

Alloy	Tensile strength (MPa)	Yield strength (MPa)	Elongation	Hardness (HRC)
As-printed M789 (Not heat treated)	1006 – 1032	826-860	14.0-18.0	29.9-32.7
Heat-treated M789	1770-1800	1670-1770	4.5 -7.6	50.0-54.0
18Ni250 maraging steel [18]	1650-1700	1651-1684	6.0-9.0	52.0-55.0
18Ni 300 [19]	1981-995	1888-1970	1.7-3.3	55.0-57.0
PH 17-4 [20]	1385-1399	1241-1269	4.6-13.4	43.0-45.0

2.3 Discontinuities in LPBF-manufactured alloys

LPBF shares notable similarities with conventional welding processes because both rely on localised metal fusion to achieve material consolidation [21]. This similarity means that both processes are susceptible to similar inherent challenges, particularly the formation of discontinuities within the final part. In the context of LPBF, irrespective of the alloy type, these intrinsic discontinuities can compromise mechanical integrity, dimensional accuracy, and overall performance. Some common discontinuities in LPBF parts are discussed as follows:

Gas pores

Gas pores are discontinuities formed within LPBF-built parts due to gas entrapment during solidification. Their origin may arise from metal vapour or pre-existing pores within the alloy powder, especially those produced using gas atomisation [22]. The gases attempt to escape during solidification but eventually get trapped and appear as spherical discontinuities, thereby contributing to porosity within the final part. Their size, density, and distribution depend strongly on melt-pool dynamics. At high power or low scan speeds, deep vapour depressions (“keyholes”)

produce unstable oscillation and collapse seed bubbles that are subsequently frozen in, resulting in chains of relatively spherical pores trapped within the part.

Lack of fusion (LOF)

LOFs are one of the most observed discontinuities in LPBF-built parts. They arise due to insufficient melting and inadequate overlap between adjacent scan tracks and successive layers, resulting from insufficient energy input required for full consolidation [23, 24]. LOFs are often not associated with specific shapes, as they may adopt various geometries based on the prevailing condition. As such, they may appear irregular, spherical, elongated, and needle-like, with unmelted powder particles typically found inside them.

Laser power, hatch spacing, and scan speed are regarded as key parameters that influence the formation of LOFs. In the work of Yadroitsev et al. [25], they observed the existence of an optimum hatch spacing equivalent to the hatch-track width, which reduced porosity formation. Similarly, an analysis of the formation of discontinuities in LPBF-built Ti-6Al-4V parts conducted by Haijun et al. [24] unveiled that LOF formation is primarily caused by insufficient melt pool overlap. While noting the importance of energy density, they often emphasised that it was not a sufficient predictor of the formation of LOFs.

Non-metallic inclusions

These discontinuities are characteristic of metal-solidification processes such as casting, welding, and LPBF. They occur due to indigenous chemical reactions with melt impurities or exogenous reactions with the surroundings. Thus, incidents such as the transfer of pre-existing surface films or larger agglomerates from the powder, and the redeposition of oxidised spatter on the build surface due to residual oxygen or shielding gases in the atmosphere of the build chamber [26], promote the formation of non-metallic inclusions (NMIs). These discontinuities are typically oxygen-rich and associated with elements of high oxygen affinity (e.g., Al, Si, Mn in steels). Also, the presence of elements with high affinity for oxygen can influence the formation of oxide layers on previously processed layers, reducing interlayer bonding and consequently impacting the part densification [27].

Rough surface

LPBF-built parts in the as-printed condition are characterised by a rough surface due to troughs, valleys, attached and detached powder particles. Therefore, as-printed parts are often subjected to post-surface finish operations (PSFO) such as machining and chemical etching [28]. However, certain applications require the use of parts without PSFOs due to functional requirements like the internal walls of cooling channels, piping sections, or undercut surfaces in complex components. As the number of deposited layers increases, the surface roughness typically increases, a phenomenon commonly referred to as the staircase effect. Processing parameters, such as layer thickness and laser scan speed, often influence the roughness [29]. Also, surface oxidation due to atmospheric gases in the build chamber, partially melted and unmelted powder particles and the staircase effect have all been identified as additional causes of the surface roughness [30].

2.4 Methods for characterising discontinuities in LPBF parts

Detecting discontinuities in LPBF-built parts is challenging because they span a wide range of sizes, locations, and morphologies, all with their peculiar complexities. Several techniques are available for characterising discontinuities, but no single technique is universally applicable because each possesses inherent limitations. Some of the commonly used techniques are discussed in the following sections.

2.4.1 Optical and scanning electron microscopy (SEM)

Optical microscopy is employed to examine discontinuities, microstructural features, surface topography, and occasionally the feedstock powder. For effective observation of discontinuities, samples must undergo meticulous metallographic preparation to optimise light-optical performance and enable accurate feature analysis. Recent technological advancements have significantly expanded the capabilities of optical systems, allowing magnifications up to X2000, integration of 3D imaging, and acquisition of wide fields of view through image stitching. However, despite these improvements, the method remains fundamentally two-dimensional, constrained by the observable surface of the specimen. This inherent limitation can introduce bias, particularly in additively manufactured (AM) components, which often exhibit property heterogeneity across different build orientations and reference planes. Nevertheless, when

applied rigorously, this technique continues to be a dependable tool, especially valuable in fractographic investigations related to failure analysis.

Scanning Electron Microscopy (SEM), by contrast, offers significantly higher magnification reaching down to the nanometric scale along with a greater depth of field. These capabilities make it particularly effective for visualising small discontinuities, such as non-metallic inclusions, and for determining their chemical compositions using auxiliary analytical tools. SEM operates by analysing various electron-scattering interactions that occur when a focused beam of electrons interacts with the surface of the sample. From this interaction, SEM captures signals such as secondary electrons, backscattered electrons, and characteristic X-rays, each contributing valuable insights into the material's structure and composition. Thanks to these advanced imaging and analytical features, SEM provides substantially more detailed information than optical microscopy (OM), making it especially well-suited for fractographic analysis where pinpointing the origin of failure is critical, particularly in fatigue failure investigations. Sample preparation requirements for these techniques often require surface preparation procedures and, as a result, are classified as destructive techniques.

2.4.2 X-ray computed tomography (XCT)

X-ray computed tomography (XCT) is a widely used non-destructive characterisation technique that is capable of providing information on pore size distribution, sphericity, and clustering in the bulk volume of a manufactured part. A schematic representation of the XCT operating principle is shown in Figure 2-4. XCT operates by irradiating a rotating specimen with a beam of X-rays and acquiring image signals based on the X-ray absorption recorded. The penetration of the X-ray into the specimen helps in producing internal detail information. A series of such acquired information is reconstructed to generate a volumetric data set on the specimen. The resolution of an XCT system is strongly dependent on voxel size, which influences the voxels that make up the volumetric data set, typically on the order of a few microns [31]. This size bounds the minimum detectable discontinuity size in a given part volume [32]. An often-implemented rule of thumb is that the voxel size of an XCT machine is sufficiently small to capture the finest feature of interest, i.e., the feature size should be at least twice the voxel size.

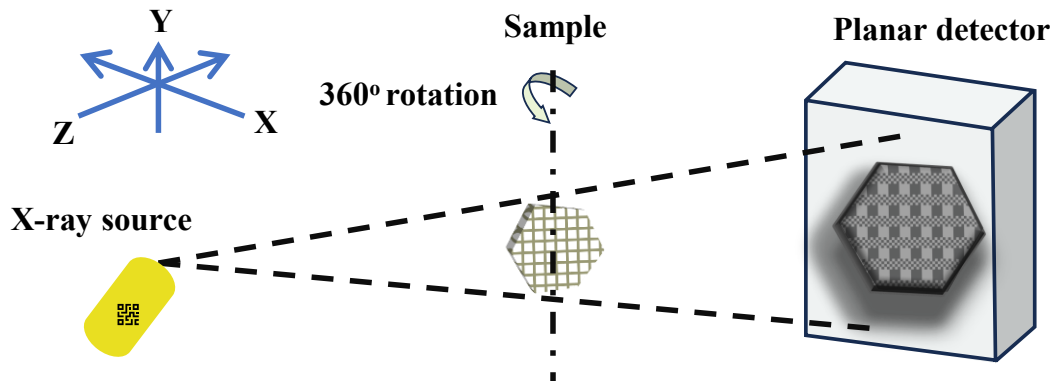


Figure 2-4. Schematic representation of an X-ray micro-computed tomography (micro-CT) scan [32].

2.4.3 3D Surface profilometry and laser scanning

Three-dimensional surface profilometry is a non-destructive characterisation technique used to evaluate surface-related discontinuities, such as surface roughness in the as-built condition. Mason et al. [33] demonstrated that surface roughness and geometrical irregularities resulting from the layer-by-layer AM process necessitate the use of three-dimensional (3D) measurements in their surface characterisation. The laser-based surface-scanning variant of the technique is often used to capture the surface features and even help to detect discontinuities. The capability to reconstruct surface features of AM parts helps to better capture the discontinuity to facilitate the investigation of their effects on the mechanical behaviour of a built part. Lucà et al. [34] investigated optical methods for surface reconstruction and found that 3D techniques provide better accuracy than traditional 2D methods. Compared to contact-based profilometry, laser scanning is non-contact, rapid, and can measure large areas with high resolution. These advantages make laser scanning a useful tool for identifying surface discontinuities and understanding their effect on the performance of additively manufactured parts.

The previously discussed techniques, although not exhaustive, can be grouped into destructive and non-destructive techniques. Table 2-5 presents a classification of some common techniques into the two groups.

Table 2-5. Classification of characterisation techniques used for discontinuity investigation in AM parts.

Non-destructive techniques	Destructive techniques
X-ray computed tomography	Scanning electron microscopy
Optical microscopy (surface inspection)	Fractography
3D surface profilometry/laser scanning	Metallography
Thermography	Energy dispersive X-ray spectroscopy
Ultrasonic testing	-

2.5 Discontinuity characteristics relevant to fatigue strength

LPBF discontinuities possess characteristics that influence their severity on the fatigue strength of built parts in which they are present. The following subsections discuss three key discontinuity characteristics.

2.5.1 Size of discontinuity

Discontinuity size is recognised as one of the primary factors influencing fatigue strength in LPBF alloys. Numerous studies have reported a reduction in fatigue strength with increasing discontinuity size. This effect becomes increasingly significant in the high-cycle fatigue regime, where stress localisation phenomenon predominates, making the discontinuity the primary crack-initiation site.

The measurement protocol adopted for a discontinuity becomes a critical concern as it is linked to the resulting morphology and size parameter adopted. In Section 2.4, non-destructive techniques reviewed facilitate the detection of discontinuity, but the size descriptor remains a separate challenge. The irregularity of LPBF pores makes it challenging to adopt a single size descriptor. A brief review of size parameters used in recent literature is as follows.

Murakami size descriptor $\sqrt{area}$, for discontinuities

A commonly used parameter to represent discontinuity size is the $\sqrt{area}$ parameter proposed by Murakami et al.[35]. They proposed tracing contours (see the red traced dots in Figure 2-6) around the discontinuity found at the fracture surface to determine the crack size represented by the $\sqrt{area}$ parameter of the discontinuity. Specifically, $\sqrt{area}$ is defined as the square root of the projected area of the discontinuity on a plane normal to the maximum principal stress. A schematic presentation is given in Figure 2-5. This approach has been successfully applied for additively manufactured alloys. For instance, Meneghetti et al. [36] successfully applied this method to measure discontinuity size at the failure origin in additively manufactured maraging steel produced by laser beam melting. Nevertheless, other researchers have instead adopted tracing the actual contour of the discontinuity without smoothing it out, as seen in Figure 2-6.

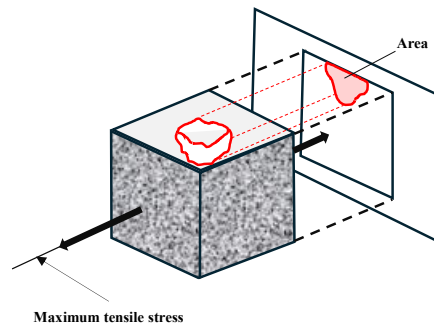


Figure 2-5. Schematic representation of the projected area of a discontinuity [37].

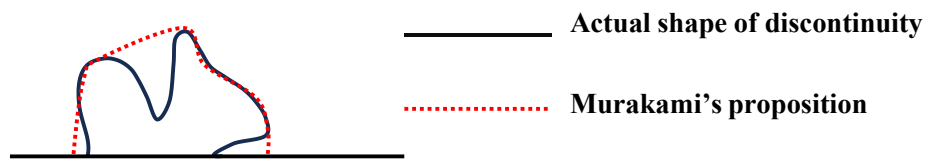


Figure 2-6. Schematic representation of discontinuity size measurement.

In applying Murakami's parameter, researchers have also proposed that the successful application requires consideration of the orientation of the discontinuity and the projection plane, especially when the discontinuity is elongated or irregular, as is typical of LOFs. On a rather modified approach, Hu et al. [38] proposed a low-cycle fatigue life prediction model for AM Ti-6Al-4V alloy, extending Murakami's $\sqrt{area}$ concept by representing discontinuity size with projected

equivalent ellipses and incorporating aspect ratio ($\frac{\sqrt{area}}{aspect\ ratio}$) as an additional geometric feature. According to Hu et al. [38], their approach is particularly suitable for characterising the influence of volumetric discontinuities on fatigue performance in low-cycle fatigue modelling.

I. Near discontinuities and size estimation

It is a common occurrence in LPBF alloys to find more than a single discontinuity at the failure origin. This incident raises the question of which discontinuity size should be considered representative of the failure origin. According to Murakami [35], two near discontinuities will interact when the space between them is smaller than the size of the smallest discontinuity. However, if the spacing exceeds the size of the smallest discontinuity, then they may be treated in isolation. In Figure 2-7, the two discontinuities ($\sqrt{area}_A$ and $\sqrt{area}_B$) spaced d apart will interact if $d < \sqrt{area}_A$. Thus, for application in the KT diagram, the size used should encompass both the discontinuity and the connecting ligament.

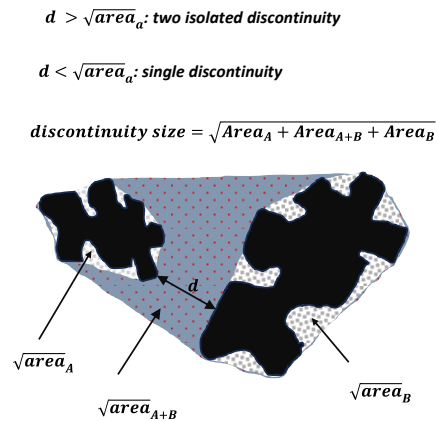


Figure 2-7. Estimation of the size of two closely spaced discontinuities [39].

II. Surface features of as-printed surfaces

As-printed LPBF-built parts are characterised by rough surface features, including troughs, valleys, and unmelted powder particles. Consequently, Murakami's early work on the application of fracture mechanics to rough surfaces has been similarly applied to LPBF parts. The following subsections discuss several approaches proposed by Murakami.

a. Treatment of roughness troughs as shallow, slender cracks: 2D approach

Roughness troughs on as-printed surfaces are treated as surface cracks, and the crack is considered shallow when its width exceeds 10 times its depth and deep when its depth exceeds 5 times its width. A schematic description of the crack is shown in Figure 2-8. Therefore, the $\sqrt{area}$ is given computed using either of Equation 2-1 or Equation 2-2.

For shallow, surface cracks:

$$\sqrt{area} = \sqrt{10c} \quad \text{Equation 2-1}$$

For deep surface cracks:

$$\sqrt{area} = \sqrt{10b} \quad \text{Equation 2-2}$$

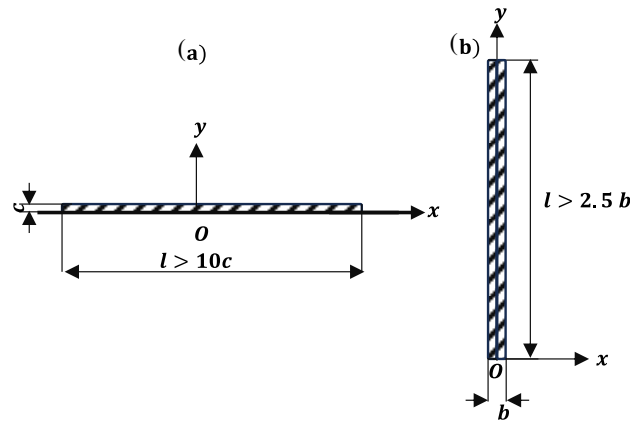


Figure 2-8. Schematic representation of (a) a shallow surface crack and (b) a deep surface crack [35].

This approach was successfully employed by Beretta et al. [40] to assess the fatigue strength of selectively laser melted AlSi10Mg, in which Equation 2-1 was used to calculate discontinuity size at the failure origin. Sausto et al. [41] similarly used the approach for the same alloy to quantify failure origin on as-printed surfaces.

More recently, Straub et al. [42] used the approach to evaluate $\sqrt{area}$ for failures of from rough surface of LPBF produced AlSi10Mg alloy where the maximum profile height R_z was used as the depth. However, researchers have also used roughness parameters such as the maximum total height of the profile (R_z) [42], and maximum valley depth from the mean profile line (R_v) [43]

for the variables of ‘ b ’ or ‘ c ’ in the equations. A schematic representation of these roughness parameters is shown in Figure 2-10 depicting the mentioned roughness profile parameters including the average roughness R_a . However, detailed descriptions of these parameters are provided in the work of Gadelmawla et al. [44] and in the recently released ASTM standard [45] for AM alloys. This approach is beneficial for non-destructive fatigue-strength prediction, but the challenge remains how to acquire the roughness parameter associated with a failure origin of as-printed surfaces, as different researchers often adopt different methodologies.

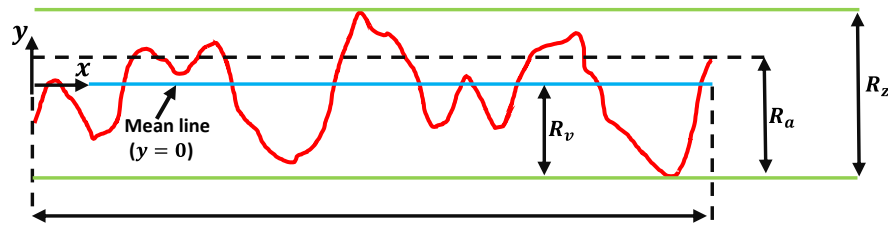


Figure 2-9. Schematic representation of roughness profile showing parameters of R_a , R_v and R_z .

b. Shielding effects of surface roughness

In general, it is known that rough surfaces can exhibit either periodic or non-periodic surface profiles depending on the industrial or manufacturing process that has been performed on them. According to Murakami et al [35], a shielding effect emerges in the case of periodic surfaces, referring to the regular spacing of valleys in terms of depth and pitch.

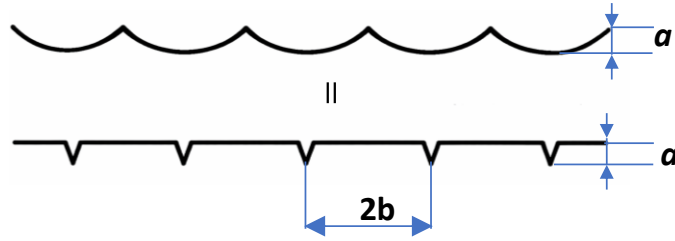


Figure 2-10. Schematic depiction of a periodic rough surface and its equivalent crack representation [35].

To estimate the $\sqrt{area}$ for a periodic rough surface having a trough depth, a , and a pitch, $2b$, as schematically shown in Figure 2-10, they proposed the relationships in Equation 2-3 and Equation 2-4.

$$\frac{a}{2b} < 0.195, \quad \frac{\sqrt{area}}{2b} \cong 2.97 \left(\frac{a}{2b} \right) - 3.51 \left(\frac{a}{2b} \right)^2 - 9.74 \left(\frac{a}{2b} \right)^3 \quad \text{Equation 2-3}$$

$$\frac{a}{2b} > 0.195, \quad \frac{\sqrt{area}}{2b} \cong 0.38 \quad \text{Equation 2-4}$$

Several researchers have adopted this periodic surface assumption for additively manufactured alloys and used it to estimate the $\sqrt{area}$ at the origin after fatigue fracturing. For instance, Jamal et al. [46] applied this approach to as-built Ti6Al4V surfaces produced by additive manufacturing, where they assumed periodicity of the surface and applied Equation 2-3 and Equation 2-4 to estimate $\sqrt{area}$. However, Collini and Meneghetti [47] mentioned that real surface-roughness profiles in AM alloys are generally non-periodic and are characterised by varying valley depths and peak heights. Thus, they proposed replacing the roughness parameter R_v and R_{sm} with a and b , respectively, in Equation 2-3 and Equation 2-4. Irrespective of the approach adopted in the estimation of the $\sqrt{area}$ at the failure origin, it appears that for AM alloys and by extension LPBF alloys, the selection or adaptation of an appropriate approach remains subjective.

III. Feret diameter (max/min) and aspect ratio of discontinuities

The feret diameter, where the feret maximum diameter represents the longest calliper distance across a discontinuity and the feret minimum the shortest, their ratio characterises elongation and has been considered an appropriate discontinuity-size descriptor. This has found relevance particularly for LOF discontinuities and elongated pores. Recent studies increasingly consider

elongation as part of discontinuity morphology, but it often appears alongside size because elongated discontinuities tend to exhibit more crack-like behaviour and reduce fatigue strength more than round pores for similar area/volume. Kousoulas et al. [48] used the feret diameter within an extreme-value statistical framework to predict the maximum discontinuity size. They held the opinion that the feret diameter was an appropriate discontinuity size metric standard because it provides a comparable metric for both 2D and 3D discontinuities.

Using a 3D approach, Buffiere [49] proposed to characterise size by using the diameter of a sphere of the same volume as the real discontinuity in combination with its tortuosity represented by the surface to volume ratio. Noteworthy is that this approach was intended to characterise discontinuity populations rather than individual discontinuities. The goal of the approach is to realise a measurable representation of the discontinuity sizes from 3D images often acquired from X-ray CT images. Despite its relevance, concerns have been raised regarding the potential for non-conservative representation of the discontinuity size while trivialising the influence of sharp edges or curvature of the discontinuity that could control crack initiation.

The array of approaches used to represent discontinuity size supports the concerns raised by researchers to have a universally standardised approach. Today, the $\sqrt{area}$ remains widely adopted and adapted, but concerns remain. In the work of Chen et al. [50], they demonstrated that discontinuity-fatigue correlation modelling becomes substantially more reliable when discontinuity quantification is standardised, particularly for surface-connected ones, where morphology and local stress concentration differ from spherical pores.

2.5.2 Location of discontinuity

The location of a discontinuity refers to its position within the bulk material relative to the free surface and is typically classified as a surface discontinuity when it is exposed to the surface, a subsurface when it lies very close to the surface and an internal when it is deep into the material. A schematic representation is presented in Figure 2-11, where a min distance, s_{min} , below which it is considered a surface or subsurface and above an internal discontinuity. LPBF alloys are known to exhibit fatigue behaviour that is sensitive to the position discontinuity. Recent studies on LPBF ALSi10Mg conducted by Afroz et al. [51] unveiled that surface-connected discontinuities and surface roughness dominate failure despite the existence of internal pores of

the same size. This was also highlighted in the review by Li and Affolter for LPBF Ti-6Al-4V [52], where they mentioned that discontinuities located within one grain diameter from the surface act as surface flaws and accelerate crack initiation, thereby reducing high-cycle fatigue strength. Conversely, internally located discontinuities have less impact on fatigue strength compared to surface discontinuities of similar size. In general, surface discontinuities are considered to govern the lower bound of fatigue strength, while internal discontinuities influence more of the scatter of the results and upper bound. A controlled study on cast AlSi alloy conducted by Serrano-Munoz et al. [53] introduced identical 1-mm spherical discontinuities either at the surface or at the centre of fatigue specimens. The results showed that internal discontinuities behaved similarly to discontinuity-free samples, indicating that crack initiation from internal discontinuities was significantly delayed. In contrast, the same discontinuity located at the surface caused a substantial reduction in fatigue strength, but this reduction disappeared when testing was performed under vacuum conditions. Since the mechanical stress concentration and stress-intensity factors for internal and surface spherical discontinuities are identical, the study concludes that the dominant reason surface discontinuities are more harmful is environmental exposure rather than purely mechanical factors. Although the authors mentioned that their findings should be extrapolated with caution to other alloys, nevertheless it provides insight into the possible origin of the differences observed in fatigue strength associated with LPBF surface discontinuities and internal discontinuities.

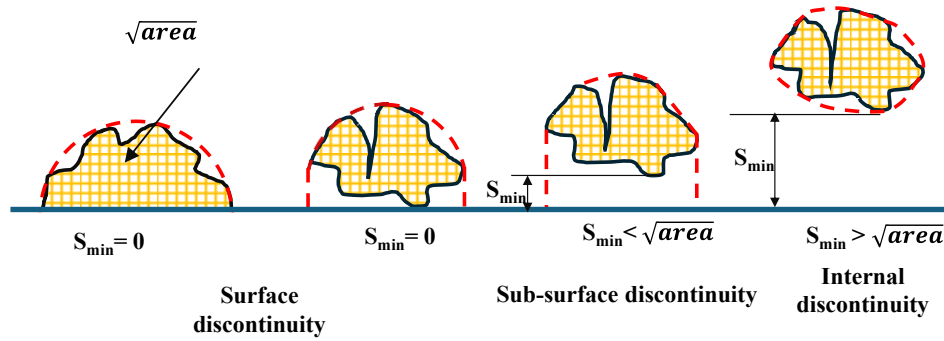


Figure 2-11. Schematic showing the $\sqrt{area}$ for discontinuities of different sizes and locations based on $\sqrt{area}$ and s_{min} [54].

2.5.3 Morphology of discontinuity

The morphology of LPBF discontinuities is a critical factor governing their severity on fatigue strength. The shape, sharpness, and overall geometry of a discontinuity dictate the local stress concentration that influences their behaviour as a benign pore or an effective crack. Irregular or elongated LOF discontinuities exhibit crack-like acuity, promoting elevated stress intensity and early crack initiation. In contrast, gas-entrapped pores, typically spherical and smooth, generate much lower stress concentrations and are therefore far less detrimental to fatigue performance than crack-like LOF discontinuities of comparable size.

The crack-like behaviour of LOF discontinuities is particularly important because their irregularity accelerates the transition from microstructurally small cracks to long cracks. This explains why even small irregular LOF discontinuities can govern fatigue life in LPBF alloys. Ngo et al. [55] showed that post-processing routes such as forging can significantly improve fatigue properties in LPBF AlSi10Mg by blunting elongated discontinuities, effectively reducing their severity and delaying crack initiation. These findings reinforce the importance of morphology and its contribution to the severity of a discontinuity's influence on fatigue behaviour of LPBF alloys.

The next section will focus on the use of statistical approaches for the prediction of the critical discontinuity expected to influence fatigue strength.

2.6 Statistical prediction of discontinuity size

LPBF-built parts exhibit stochastic discontinuity populations with varying size, morphology, and spatial distribution within and between specimens. Such distributions can influence fatigue properties as they influence scatter in fatigue behaviour. For fatigue critical applications, it is expected that the largest discontinuity is the weakest link for a built part. It is therefore important to estimate the most severe discontinuity within a given material volume or part. Commonly, the parent distribution of these discontinuities may follow exponential, normal or lognormal distributions [56] [57]. Since only a finite region of a component can be examined at finite resolution, a robust statistical framework is required to estimate the maximum expected discontinuity size in a part from limited inspection data of the parent distribution.

Extreme value statistics (EVS) provide a framework that is widely used to estimate the largest discontinuity within a given LPBF part. According to extreme value theory, if the underlying discontinuity population follows a parent distribution with exponentially decaying tails (such as normal or exponential), then the distribution of the largest discontinuity converges to a Gumbel type. The theory has been widely applied to LPBF alloys as it facilitates the prediction of fatigue strength based on the expected largest discontinuity size. Wang et al. [58] emphasised the need to model extreme discontinuities in LPBF alloys since the likelihood of fatigue failure can be linked to the variable discontinuity sizes in the built part. Most importantly, the discontinuity type does not appear to limit its general applicability, as several researchers have applied it to roughness, LOF, NMI and pores. For instance, Fox and Pintar [59] applied EVS to LPBF surface by estimating extreme areal parameters such as maximum valley depth from limited fields-of-view, which is directly relevant because surface extremes are often correlated with fatigue crack initiation in as-built conditions.

A commonly adopted EVS implementation is the block maxima method, illustrated schematically in Figure 2-12a. In this approach, the inspected surface or volume is partitioned into several equal-sized blocks, the largest discontinuity within each block is extracted, and these maxima are collectively analysed using EVS. In practical terms, the inspected region could be the gauge volume of a tensile specimen, a coupon volume or highly stressed regions in a built part, and the maximum represents the discontinuity most likely to trigger crack initiation. Nikfar et al. [60] applied the block maxima approach to surface measurements of LPBF parts, extracting the

maximum valley depth per block, and using the block-maxima distribution to extrapolate surface-tail characteristics across surfaces and measurement techniques.

Another commonly used approach is the peak over threshold method, illustrated schematically in Figure 2-12b. The method considers data values exceeding a specified threshold to avoid loss of information. The challenge with this technique is the selection of an appropriate threshold that ensures reliable statistical behaviour for accurate analysis. This challenge becomes more significant when the dataset contains multiple discontinuity types, like LOF and NMI, which may limit the reliability of the selected threshold.

Both approaches have found application for LPBF alloys, but the block-maxima approach is often preferred for LPBF parts because: (i) it aligns with the weakest-link concept associated with fatigue failure, and (ii) it is relatively simple to implement while avoiding threshold-selection challenges associated with the peak-over-threshold method. A detailed treatment of statistical variability and size effects in fatigue, including the application of extreme value concepts, is well documented in “Metal Fatigue in Engineering” by Stephens et al. [61]. For brevity, the fundamental principles are not reproduced here, and the reader is referred to this established text for a comprehensive discussion.

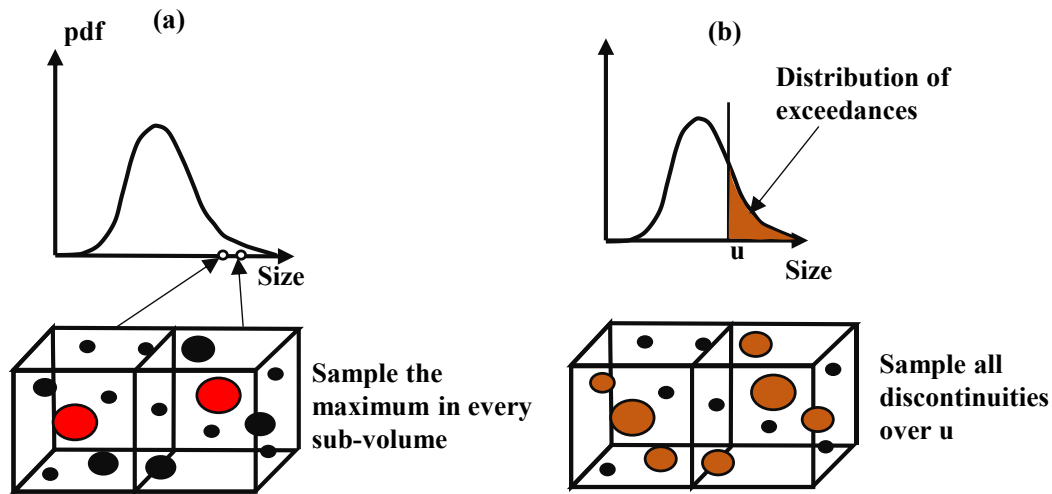


Figure 2-12. Sampling of maxima using the (a) block maxima method with red balls as block maxima and (b) peak over threshold method with all brown balls $> u$ [62].

Following the preceding discussions on discontinuities, Section 2.7 reviews the influence of discontinuities on the fatigue behaviour of materials.

2.7 Discontinuity-controlled fatigue behaviour

This section introduces the mechanics governing discontinuity-controlled fatigue, setting the foundation for tools such as the Kitagawa-Takahashi (KT) diagram, which provides a practical framework for relating discontinuity size to fatigue strength.

2.7.1 The KT diagram as a tool for modelling fatigue strength

Considering the existence of discontinuities is inevitable in LPBF-built parts, researchers and industry practitioners increasingly seek to predict the fatigue strength of their components. Effective implementation requires bridging the discontinuity-free fatigue strength and the long crack behaviour. The KT diagram (see Figure 2-15 in the theoretical background section) is an indispensable tool for this purpose. However, the fact that discontinuities in the LPBF alloys are not absolute cracks introduces an additional uncertainty and complexity to the use of the KT diagram. To simplify, researchers often assume that discontinuities behave as pre-existing cracks whose severity depends on size, morphology and location. Murakami's parameter (see Section 2.5.1) has been particularly handy for LPBF alloys because of its usefulness in correlating discontinuity morphology with stress-intensity factors. Tenkamp et al. [63] used the Murakami parameter for AM AlSi10Mg across high and very high cycle fatigue for a mechanical assessment of discontinuities and showed that fatigue strength can be rationalised based on discontinuity size and stress intensity. Other works, like that of Uematsu et al. [64], used an electron beam-melted Ti-6Al-4V alloy with artificial discontinuities as reference material. They proposed design curves based on Murakami's model and, through the construction of a KT-type diagram, demonstrated a practical framework for discontinuity-tolerant fatigue design in additively manufactured components.

2.7.2 The crack and notch behaviour of LPBF discontinuities

LPBF discontinuities are sometimes treated as notches, governed by stress concentration and local damage mechanics. So, broadly speaking, discontinuities in LPBF-built parts can be regarded as features within the bounds of cracks and notches, where discontinuity morphology can tilt the adopted behaviour towards any of the extremes. The governing parameters, therefore,

involve either stress intensity or stress concentration, and $\sqrt{area}$ parameter or notch geometry. Afroz et al. [65] reported in their review that, on the fracture surfaces of LPBF Ti-6Al-4V and AlSi10Mg, crack initiation commonly occurs at notch-like discontinuities. The crack front initially follows the discontinuity geometry, before transitioning into a more conventional semi-elliptical shape during subsequent growth. Furthermore, Ahmed [66] investigated the effect of discontinuity on the fatigue performance of LPBF AlSi10Mg and showed that internal and surface discontinuities can be treated as notches with an effective root radius, and that fatigue life correlates strongly with the local stress concentration factor derived from finite element analysis (FEA).

LPBF discontinuities can be regarded as analogous to non-propagating cracks irrespective of their adopted notch or crack behaviour. In a recent work by Chapetti [67], the author analysed notch sensitivity and size effects in metallic materials and showed that at the fatigue limit, small cracks emanating from notches can be arrested when the local driving force is less than a threshold, leading to non-propagation. Their work was not performed in LPBF alloys, but parallels can be drawn as short cracks emerging from discontinuities may be arrested if the threshold is below the short crack growth threshold. The reverse is the case for larger cracks that overcome their threshold. This framework helps explain why some LPBF specimens with visible discontinuities can still survive long fatigue lives, while others fail prematurely.

Post-processing operations such as machining and hot isostatic pressing, often carried out on LPBF alloys, can alter the behaviour of intrinsic discontinuities through elimination or even modification of the morphology. Li and Affolter [52] emphasised that for LPBF Ti-6Al-4V, machining the surface to remove roughness and near-surface discontinuities can shift the governing crack initiation mechanism from surface notches to internal discontinuities like cracks.

In the next section, a brief theoretical background is provided for the discussion of the influence of discontinuities on fatigue strength from the viewpoint of design philosophies adopted in fatigue.

2.8 Fatigue design philosophies

Different fatigue design philosophies have been developed, each based on distinct assumptions regarding discontinuity tolerance and failure prevention. The most widely adopted approaches

include the safe-life design philosophy, which assumes a discontinuity-free material and aims to prevent crack initiation within a prescribed life, and the damage-tolerant philosophy, which accounts for the presence and growth of cracks during service. The following subsections discuss these philosophies.

2.8.1 The safe-life design philosophy

This approach is based on the principle that failure is deemed to occur when fatigue cracks are nucleated in a component subjected to cyclic loading. In the context of practical application, the component is withdrawn from service before crack initiation or propagation occurs. Therefore, the implementation of this approach relies on fatigue data obtained from extensive tests carried out to establish the stress amplitude versus number of cycles plots (Wöhler S-N curves) on the alloy of interest. A typical schematic plot of the curve is shown in Figure 2-13, where the trend for ferrous and nonferrous alloys is depicted. The plot shows a stress threshold ' σ_o ' below which no fatigue failure is expected to occur. Classical fatigue literature refers to this stress level as the fatigue limit, and it is often observable for ferrous metals. However, the applicability of this terminology is now questioned as specimens are thought to eventually fail under very high cycle (10^8 - 10^{10}) fatigue conditions using ultrasonic fatigue testing. Thus, fatigue strength corresponding to a specified number of cycles is considered more appropriate. The conventional fatigue limit is often estimated between 10^6 - 10^7 cycles, with tests typically performed in accordance with ASTM E466 [68].

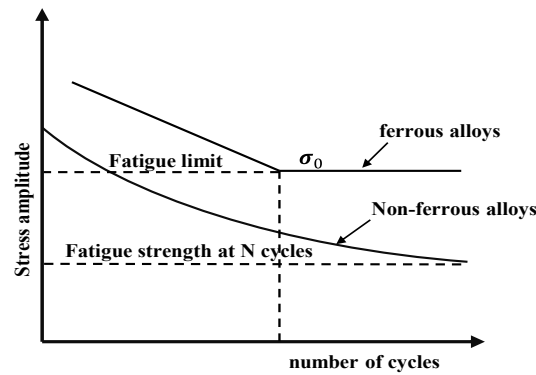


Figure 2-13. Schematic plot of stress amplitude versus number of cycles curves.

2.8.2 Damage-tolerant design philosophy

This approach is based on the assumption that a flaw already exists in the alloy and may grow to a critical size capable of causing failure. A damage-tolerant design strategy is particularly suitable for LPBF parts due to the presence of discontinuities. The approach relies on the stress intensity threshold, $\Delta K_{th,LC}$, a linear elastic fracture mechanics parameter below which crack propagation is not expected to occur. This parameter is typically determined using standardised methods such as ASTM E647, which produce fatigue crack growth (FCG) curves (see Figure 2-14). The curve is a plot of the crack growth rate, da/dN , against the stress intensity factor range, ΔK , and describe crack propagation behaviour under cyclic loading, from the threshold, $\Delta K_{th,LC}$ to final fracture. The emphasis of the philosophy leverages the ΔK of cracks as an appropriate parameter for crack evolution.

In the context of LPBF alloys, several studies have studied the discontinuities using the stress intensity as the driving force of their fatigue behaviour. Javidrad et al. [69] synthesised results from LPBF alloys and reported that the transition of cracks from non-propagating to propagating behaviour in LPBF alloys aligns with the condition $\Delta K > \Delta K_{th}$. Similarly, Benedetti et al. (2023) showed that the fatigue strength of LPBF Ti-6Al-4V is well predicted by comparing the stress intensity of a discontinuity to the ΔK_{th} of the material.

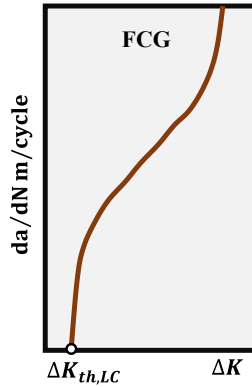


Figure 2-14. Schematic representation of the FCG behaviour in materials.

The discussed design philosophies individually focus on crack initiation and propagation, respectively. The following section discusses the integration of both approaches into a fatigue design framework that relies on discontinuity severity.

2.8.3 The two fatigue thresholds

The discussions on fatigue design philosophies in Sections 2.8.1 and 2.8.2 highlight their association with two distinct thresholds characterised by different failure mechanisms. The safe life approach utilises the fatigue strength threshold, σ_0 , governed by crack initiation, while the damage-tolerant approach employs the use of the varying stress intensity threshold, $\Delta K_{th,LC}$, governed by crack propagation.

Kitagawa and Takahashi [70], in their effort to evaluate the performance of material under cyclic conditions, observed non-propagating surface cracks exceeding 0.5 mm in length to be associated with a constant ΔK_{th} . As a result, they proposed the diagram shown in Figure 2-15, which is a log-log plot of the stress amplitude versus the crack size at a given number of cycles. The diagram illustrates the evolution of the fatigue strength with discontinuity sizes and subsequently defines bounds of finite and infinite component life. The diagram will be referenced in subsequent discussions.

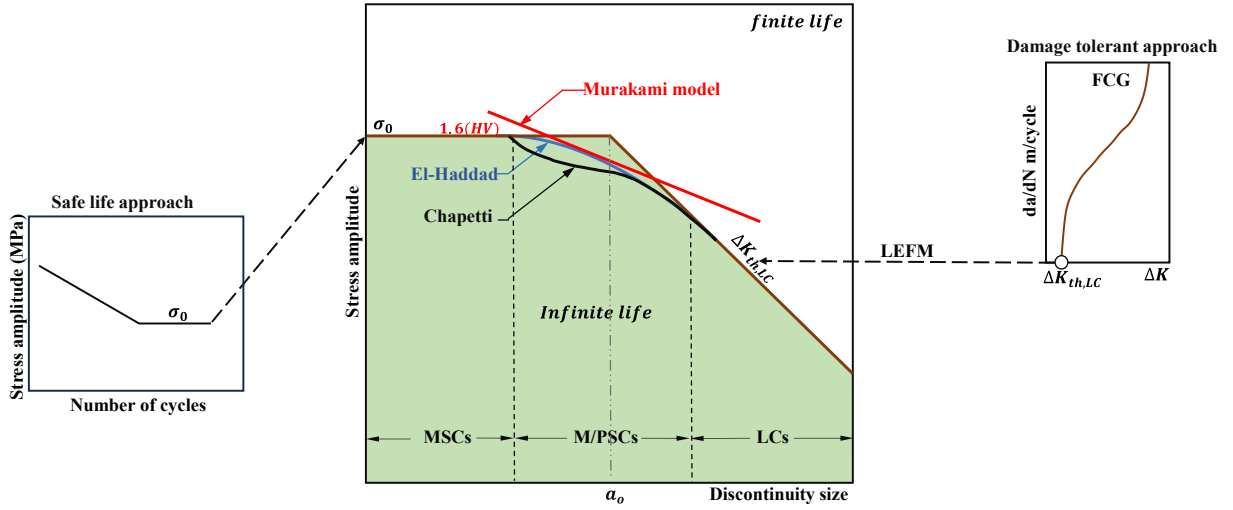


Figure 2-15. Schematic representation of the Kitagawa-Takahashi diagram illustrating fatigue behaviour as a function of crack size.

Horizontal branch of the KT diagram

The horizontal line of the KT diagram represents the discontinuity-free fatigue strength, σ_0 , of the material. The value of σ_0 is consistent with the safe-life fatigue threshold, where fatigue failure is governed by crack initiation and is described by Equation 2-5.

$$\sigma_a = \sigma_0 \quad \text{Equation 2-5}$$

For many steels and metallic alloys, σ_0 is commonly estimated using the empirical hardness-fatigue strength correlation originally proposed by Garwood et al. [71], expressed in Equation 2-6. Their study examined six steel alloys and demonstrated a strong linear relationship between Vickers hardness and fatigue strength for low-hardness alloys ($HV \leq 400$), where microstructural slip behaviour governs fatigue resistance. The fatigue strengths reported in their study correspond to the high-cycle regime of approximately 10^6 - 10^7 cycles. As a result, the empirical relationship between fatigue limit and hardness is given by Equation 2-6.

$$\sigma_0 \approx 1.6 (HV) \pm 0.1 HV (HV \leq 400) \quad \text{Equation 2-6}$$

However, at $HV > 400$, this linear correlation deteriorates, and the scatter in fatigue strength becomes more pronounced, as illustrated in Figure 2-16. This increased scatter is attributed to the

higher sensitivity of high-strength materials to non-metallic inclusions, which act as crack-like discontinuities and reduce the reliability of hardness-based fatigue predictions. A similar behaviour is expected in LPBF alloys, where intrinsic discontinuities introduce variability in fatigue behaviour.

Although Equation 2-6 is sometimes extended to high-strength steels under the idealised assumption of a discontinuity-free material, it is also used for LPBF alloys that have undergone post-processing treatments like hot isostatic pressing intended to close internal discontinuities and restore near-intrinsic fatigue strength.

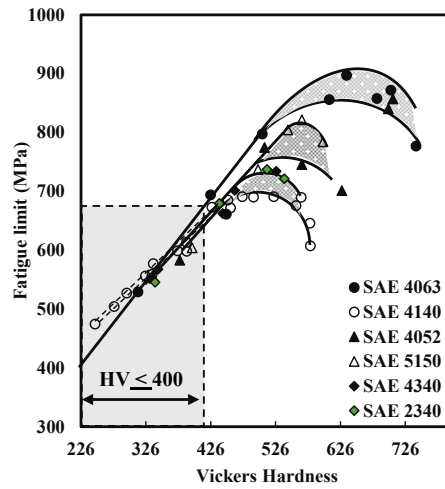


Figure 2-16. Fatigue strength as a function of Vickers hardness. Adapted from Murakami et al. [72].

Declining branch of the KT diagram

The declining branch of the KT diagram is governed by linear elastic fracture mechanics (LEFM). In this regime, the fatigue limit is controlled by $\Delta K_{th,LC}$. Crack propagation occurs when the applied stress intensity exceeds this threshold. The governing equation for this behaviour is given in Equation 2-7.

$$\Delta K_{th,LC} = Y \Delta \sigma \sqrt{\pi a} \quad \text{Equation 2-7}$$

where Y is the geometry factor, $\Delta \sigma$ is the applied stress range, and a is the discontinuity or crack size.

2.8.4 Crack classification in the KT diagram

The diagram presented in Figure 2-15 shows cracks classification across different regimes according to their propagation characteristics, and is briefly discussed as follows.

I. Microstructurally short cracks (MSCs)

These cracks characterise the initial stage of crack nucleation from the microstructure or discontinuity. These cracks or discontinuities are typically comparable in size to microstructural features, such as the grain size. Their propagation is characterised by intermittent arrest, where they may accelerate or decelerate due to the crack arrest capability of micromechanical barriers such as grain boundaries [73]. They establish the condition governing the plain-specimen fatigue strength, σ_0 , which is influenced by the largest MSC capable of arrest.

II. Mechanically/physically short cracks (M/PSCs)

These are cracks/discontinuities that are either larger than the characteristic microstructural feature size or have propagated to overcome the microstructural barriers. The cracks are deemed to be mechanically short when their size is on the order of the plastic zone size. Thus, as the MSCs propagate into this crack growth regime, the microstructural influence diminishes, and the mechanical plastic zone size begins to govern the crack behaviour.

Overlapping with the size range of the mechanical short cracks are those referred to as physically short cracks. These cracks are influenced by crack-closure phenomena first brought to light by Elber et al. [74, 75]. Studies have shown that crack closure is not a single mechanism but a family of shielding mechanisms. These include plasticity-induced crack closure arising from the plastic wake, roughness-induced closure caused by asperity interlocking, and oxide-induced closure associated with debris/film formation.

Long cracks (LCs)

These cracks or discontinuities are larger than their MSC and M/PSC counterparts and exhibit a quasi-stabilised crack-closure behaviour. Their characteristic dimensions typically range from a few tenths of a millimetre to several millimetres. Combined with the relatively small plastic zone at the crack tip compared with the overall crack length, this justifies the use of LEFM long-crack propagation criteria, specifically the threshold stress-intensity factor range $\Delta K_{th,LC}$. Although

$\Delta K_{th,LC}$ is considered a material property, it may be influenced by crack-closure phenomena. Kumar et al. [76] demonstrated that crack closure can suppress near-threshold crack-growth rates, highlighting the sensitivity of this regime to closure artefacts.

Recently, researchers have expressed reservations about the ASTM E647 load reduction approach due to the non-conservative value of the obtained $\Delta K_{th,LC}$. Mote et al. [77] showed that the method can artificially elevate crack-opening loads during load shedding due to load-history effects and closure, rather than representing the intrinsic crack-growth resistance of the material. These issues may be further amplified in AM alloys, where surface roughness, crack-path tortuosity, and heterogeneous discontinuity populations promote roughness-induced closure. For example, Alfred et al. [78] reported that although ASTM E647 can be applied to determine $\Delta K_{th,LC}$ for AM Ti-6Al-4V, the resulting values may include crack-closure contributions, leading to increased scatter near threshold.

In response to these limitations, far-field-compression methods have gained attention as an alternative for determining $\Delta K_{th,LC}$. These approaches involve subjecting the specimen to cyclic compressive loading away from the crack tip, typically under compressive loading, before switching to tension for constant-amplitude crack-growth measurements. This preconditioning ensures that the starter crack remains fully open during subsequent loading, thereby minimising load-history and closure artefacts while yielding a more conservative and reliable estimate of $\Delta K_{th,LC}$.

2.8.5 Evolution of the Kitagawa-Takahashi diagram

While the work of Kitagawa and Takahashi was revolutionary, it is also associated with non-conservative predictions, as experimental data show the occurrence of fatigue failure below the transition zone. To this end, El-Haddad et al. [79, 80] introduced a fictitious intrinsic crack length parameter " a_0 " to describe the transition zone to improve consistency with experimental observations. The value of " a_0 ", is obtained at the intersection between the horizontal branch (Equation 2-5) and the declining branch (Equation 2-7) of the KT diagram, thus leading to the following relationship;

$$a_0 = \frac{1}{\pi} \left(\frac{\Delta K_{th,LC}}{Y \Delta \sigma_0} \right)^2 \quad \text{Equation 2-8}$$

Where the σ_0 is the fatigue strength of the material without discontinuity in MPa, $\Delta K_{th,LC}$ is the stress intensity threshold of long cracks in $\text{MPa}\sqrt{m}$, Y is the shape factor, a_0 is the El-Haddad parameter. Thus, Equation 2-8 is used to modify Equation 2-7 in accordance with El-Haddad to give;

$$\Delta K_{th} = Y \Delta \sigma \sqrt{\pi(a + a_0)} \quad \text{Equation 2-9}$$

Several authors have expressed differing views regarding the physical meaning of a_0 , reflecting the lack of a common consensus. Its mathematical derivation suggests that it behaves as a fitting parameter to align with experimental data. Nevertheless, El-Haddad et al. [79, 80] mentioned that it is linearly related to the material's grain size and can be viewed as a measure of a material's reduced flow resistance at the surface due to the lower constraint imposed by the surrounding material. The lack of a clear physical meaning for the modifying parameter motivated the development of alternative approaches. Additionally, interest in accounting for the progressive buildup of crack closure triggered the emergence of other models discussed as follows.

Murakami model

Murakami introduced the $\sqrt{area}$ parameter to represent discontinuity size and correlate well with the maximum stress intensity factor at the front of irregularly shaped cracks, as defined by Equation 2-10 [81].

$$K_{I_{max}} = Y \sigma \sqrt{\pi \sqrt{area}} \quad \text{Equation 2-10}$$

Where σ is the applied stress, Y is 0.65 for surface discontinuities and 0.5 for internal discontinuities. Figure 2-17 presents data for various materials with the ΔK_{th} plotted against $\sqrt{area}$ of discontinuities and demonstrates a consistent relationship between both parameters. These results support the assumption that discontinuities in materials can be considered as preexisting cracks whose size can be estimated using $\sqrt{area}$.

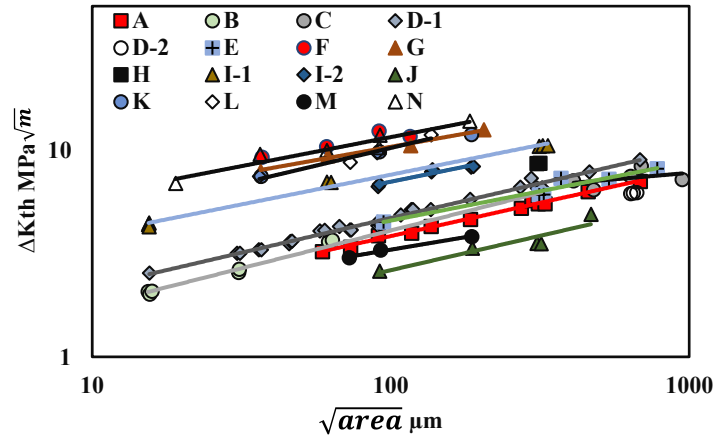


Figure 2-17. Relationship between ΔK_{th} and $\sqrt{area}$ of discontinuities and cracks adapted for different materials, adapted from the work of Murakami et al. [72].

As a result, an empirical relationship between $\sqrt{area}$ and ΔK_{th} was proposed, as expressed in Equation 2-11.

$$\Delta K_{th} = y \cdot (HV + 120)(\sqrt{area})^{1/3} \quad \text{Equation 2-11}$$

Where y equals 3.3×10^{-3} for surface discontinuities and 2.77×10^{-3} for internal discontinuities, respectively.

The fatigue strength can thus be estimated by combining Equation 2-10 and Equation 2-11 under the conditions $\Delta K_{max} = \Delta K_{th,sc}$ and $\Delta\sigma = 2\sigma_w$, leading to Equation 2-12.

$$\sigma_w = \frac{k \cdot (HV + 120)}{(\sqrt{area})^{1/6}} \cdot \left[\frac{1 - R}{2} \right]^\alpha \quad \text{Equation 2-12}$$

$k = 1.43$ and 1.56 for the surface and internal discontinuities, $\sqrt{area}$ in μm , σ_w is in MPa, R is the stress ratio, and $\alpha = 0.226 + HV \times 10^{-4}$. Equation 2-12 is often called Murakami's fatigue strength prediction model. The superimposition of the Murakami prediction on the KT diagram is depicted in Figure 2-15. The model adequately captures the transition behaviour between the short-crack regime and the fatigue-strength limit.

Therefore, the KT diagram provides a complementary perspective to the S–N approach, particularly for discontinuity-dominated components such as AM parts. The diagram integrates both stress-based and fracture mechanics concepts, thereby bridging the gap between safe life (σ_0) and damage-tolerant ($\Delta K_{th,LC}$) philosophies. In contrast, the S–N focuses on correlating stress amplitude with fatigue life for parts not plagued with discontinuity. The effectiveness of the bridge is reliant on El-Haddad's a_0 or other crack closure accounting models discussed. Recent work of Sunder [82] highlighted this linkage with the construction of the KT diagram from S–N data and crack-growth threshold towards design and inspection planning.

Chapetti model

The Chapetti model [83], just like the El-Haddad model, is one that was developed to account for the crack closure phenomenon occurring in the short crack region and subsequently improve the conservative attributes of the diagram. Chapetti's model modifies ΔK_{th} by accounting for crack closure to give Equation 2-13.

$$\Delta K_{th} = \Delta K_{th,int} + (\Delta K_{th,LC} - \Delta K_{th,int}) \left[1 - e^{-a/a_0} \right] \quad \text{Equation 2-13}$$

Where, $\Delta K_{th,int}$ is the material's intrinsic threshold range.

The fatigue strength can be computed subsequently using Equation 2-14.

$$\Delta \sigma_{th} = \frac{\Delta K_{th,int} + (\Delta K_{th,LC} - \Delta K_{th,int}) \left[1 - e^{-a/a_0} \right]}{Y\sqrt{\pi a}} \quad \text{Equation 2-14}$$

The model thus accounts for crack closure with the progressive growth of the crack and the evolution of the threshold as it gradually approaches $\Delta K_{th,LC}$. The resulting evolution of fatigue strength according to this model is reflected in Figure 2-15.

The foregoing discussions and models have focused on treating discontinuities as preexisting cracks in materials. The next section focuses on the theoretical framework applicable to discontinuities exhibiting notch-like behaviour.

2.8.6 Notch-like behaviour of discontinuities

Peterson [84] classically defined notches as geometric discontinuities in a material that give rise to localised stress concentration. This behaviour is particularly relevant when the discontinuity geometry can be characterised using scalar parameters such as tip radius and depth. Within the KT diagram framework, discontinuities may exhibit crack-like behaviour when evaluating their influence on fatigue limits, particularly through the threshold stress-intensity factor range $\Delta K_{th,LC}$. As the size of these discontinuities increases, their geometry becomes more distinct and can be described by measurable parameters such as depth and tip radius, which gives rise to a stress concentration factor, k_t . It should be noted that k_t primarily represents the geometric amplification of local stress caused by the notch. Tabulations of k_t values for various notch geometries are available in standard design handbooks [85]. Since discontinuities in engineering components can range from blunt notches to extremely sharp ones, a natural overlap exists between notch-like and crack-like behaviour. This transition was explored by Smith and Miller, who proposed that sufficiently sharp notches with high k_t values may be treated as cracks [86]. Their schematic diagram, shown in Figure 2-18, illustrates the evolution of fatigue strength and stress concentration for a notch of constant depth, a , and varying tip radius ρ .

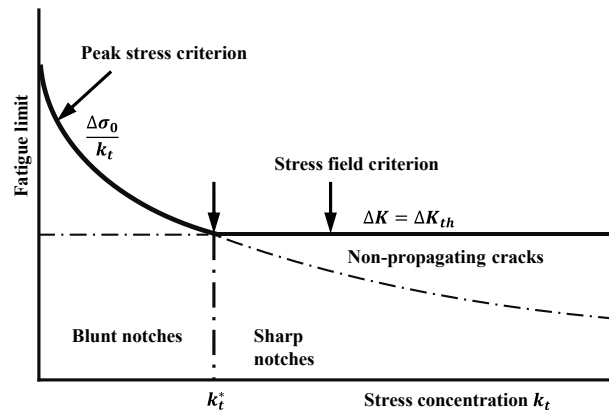


Figure 2-18. Stress concentration of a notch of constant depth and varying radius [86].

The diagram shows two curves intersecting at a critical stress concentration factor, k_t^* . Below k_t^* , fatigue strength is governed by the elastic peak-stress criterion associated with notch effects. Conversely, above k_t^* , the fatigue strength becomes independent of k_t , indicating a transition to crack-like behaviour governed by the LEFM threshold ΔK_{th} . Thus, for a discontinuity of fixed

size, there exists a critical k_t^* beyond which the fatigue response is no longer controlled by notch effects, and the discontinuity behaves effectively as a crack.

The Atzori and Lazzarin diagram

Atzori and Lazzarin (AL) [87] proposed a KT diagram (see Figure 2-19) to assess the fatigue limit of short and long cracks, crack-like notches, and blunt notches with parallel flanks centred in an infinite plate and subjected to a remotely applied stress. In their work, the notch size was scaled while keeping the notch acuity constant so that the stress-concentration factor k_t remained unchanged. The resulting diagram shows that, as the notch dimension increases, the fatigue behaviour transitions at a characteristic size, a^* , toward a constant fatigue strength described by $\Delta\sigma_0/k_t$, corresponding to full notch behaviour. Conversely, as the notch dimension decreases, the behaviour transitions toward LEFM-controlled behaviour. Thus, for a given notch geometry, k_t governs the fatigue behaviour for large discontinuities ($a > a^*$), where the peak-stress criterion is dominant. For smaller discontinuities ($a_0 < a < a^*$), the fatigue limit is instead governed by the stress-field criterion. The a_0 is the El-Haddad-Smith-Topper parameter. It is also noteworthy that, in the limiting case of a perfectly sharp crack, both k_t and a^* tend to infinity, implying that notch effects become negligible.

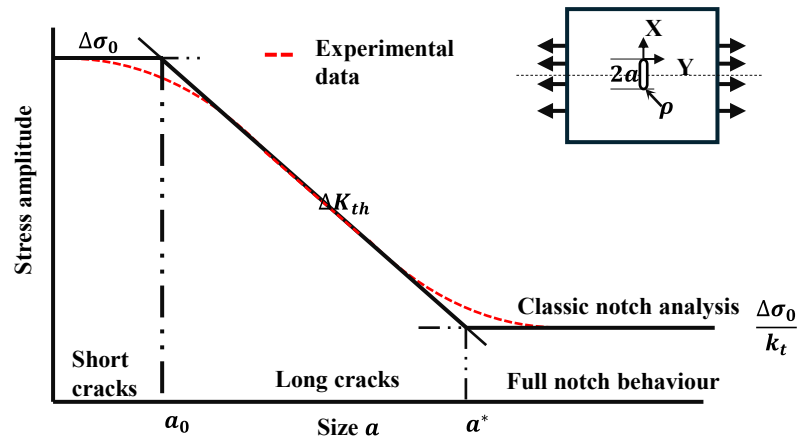


Figure 2-19. Fatigue behaviour of materials weakened by notches or cracks proposed by Atzori and Lazzarin [88].

From a geometrical perspective, a crack may be considered as a limiting notch configuration with an opening angle approaching 0° , represented by 2α . Atzori et al. [88] generalised the diagram to

account for different opening angles and geometrical configurations, enabling the framework to describe the transition between crack-like ($2\alpha = 0^\circ$) and notch-like ($2\alpha > 0^\circ$) fatigue behaviour more comprehensively. As such, Gross and Mendelson [89] defined the notch stress intensity factor as expressed in Equation 2-15.

$$K_I^V = \sqrt{2\pi} \cdot \lim_{r \rightarrow 0} [\sigma_{\theta\theta}(r, \theta = 0) \cdot r^{1-\lambda}] \quad \text{Equation 2-15}$$

Where r , is radial distance from the notch tip and θ , angular position around the notch tip, give the polar coordinates centred at the notch tip, $\sigma_{\theta\theta}$ is the opening stress distribution evaluated along the notch bisector line ($\theta = 0$) direction, and $1 - \lambda$ depends on the notch opening angle, 2α , and represents the stress singularity exponent. Consequently, El-Haddad's formulation describing the evolution of fatigue strength with discontinuity size may be rewritten as Equation 2-16.

$$\Delta\sigma_{th} = \frac{\Delta K_{I,th}^V}{2Y\sqrt{\pi}(a + a_o^V)^\gamma} \quad \text{Equation 2-16}$$

Where a is the characteristic dimension of the discontinuity, a_o^V is the material length parameter for a notched component, and $\gamma = 1 - \lambda$ is the singularity exponent. Furthermore, $\Delta\sigma_{th}$ represents the fatigue strength amplitude of the notched component, while $\Delta K_{I,th}^V$ is the threshold range of the mode I notch stress intensity factor. Equation 2-16 may alternatively be reformulated to account for different shape factors.

$$\Delta\sigma_{th} = \frac{\Delta K_{I,th}^V}{2\sqrt{\pi}(a_{eff} + a_o^V)^\gamma} \quad \text{Equation 2-17}$$

Note that a_o^V , like a_0 , can be obtained using Equation 2-18.

$$a_o^V = \left(\frac{\Delta K_{I,th}^V}{\sqrt{\pi} \Delta\sigma_0} \right)^{1/\gamma} \quad \text{Equation 2-18}$$

In terms of significance, $\Delta K_{I,th}^V$ is analogous to the threshold stress intensity factor for long cracks, $\Delta K_{th,LC}$. Hence, it represents the threshold stress intensity factor for the propagation of a crack

from a notch and may be determined experimentally for the notch geometry of interest. However, Atzori et al. [90] proposed relationships of the form given in Equation 2-19 to estimate $\Delta K_{I,th}^V$ for a notch using the $\Delta K_{th,LC}$ of cracks and the plain fatigue strength at the desired number of cycles.

$$\Delta K_{I,th}^V = \beta \cdot \Delta \sigma_0^{(1-2\gamma)} \cdot \Delta K_{th,LC}^{2\gamma} \quad \text{Equation 2-19}$$

Where β is a parameter that is dependent on the notch opening angle 2α , and values can be found in the work of Atzori et al. [90]. The evolution of fatigue strength for different notch opening angles, ranging from the crack-like case ($2\alpha = 0^\circ$) to notch-like configurations ($2\alpha > 0^\circ$), is schematically illustrated in Figure 2-20. The illustration presents several sloped lines, each associated with a different notch opening angle. From the diagram, a specimen having a hypothetical discontinuity of size, b , will have a corresponding fatigue strength level, $\sigma_{b\#}$, associated with its notch opening angle. Consequently, the notch opening angle influences the resulting fatigue strength at a given discontinuity size.

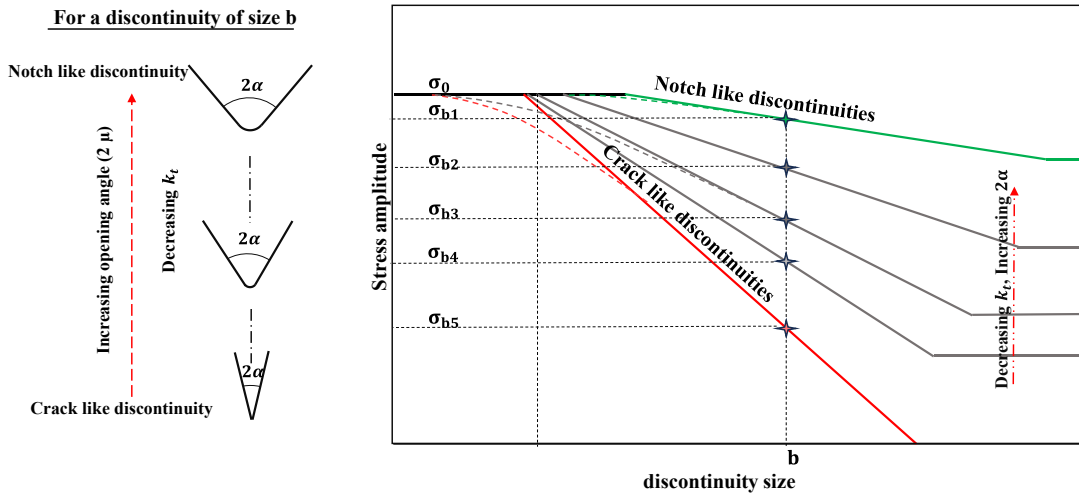


Figure 2-20. Schematic illustration of fatigue strength evolution for different notch geometries with slope trend decreasing from crack-like ($2\alpha=0$) to notch-like ($2\alpha>0$) discontinuities.

The foregoing discussion of notch effects becomes particularly relevant in additively manufactured components, where there can be variability in the dimensions and severity of discontinuities. External discontinuities arising from corrosion or geometric irregularities may

also exhibit behaviours that are not adequately captured by crack-like modelling approaches. This limitation becomes particularly important in engineering design applications requiring optimised component utilisation.

Effective crack-size representation of different discontinuity types

The application of the KT and Atzori–Lazzarin diagrams (ALD) to different discontinuity sizes, irrespective of shape and location, has motivated the use of an effective discontinuity size parameter, a_{eff} , defined in Equation 2-20.

$$a_{eff} = a.Y^{1/\gamma} \begin{cases} a = \sqrt{area} & \text{For cracks} \\ a = depth & \text{For notches} \end{cases} \quad \text{Equation 2-20}$$

In the work of Syed et al.[91], they studied the location effect by seeding gas pores either near-surface or at the centre in LPBF Ti-6Al-4V, followed by the construction of a modified KT diagram using a_{eff} to represent the discontinuity size. This approach is widely credited to Atzori-Lazzarin [87] and also extends to ALDs. Replacing the discontinuity descriptor with a_{eff} enables discontinuities of different locations to be represented within a unified framework.

2.8.7 Applications of the KT diagram and ALD to LPBF alloys

The background discussions and the fatigue philosophies unveil that the fatigue strength of an alloy is controlled by multiple factors. These factors, though not exhaustive, are schematically shown in the fishbone diagram of Figure 2-21. The effect of the manufacturing discontinuities is particularly relevant for the fatigue response of additively manufactured alloys. Additional discontinuities may also arise from external contributions associated with design features or service environments. Consequently, fatigue strength arises from the combined interaction of multiple contributing factors rather than a single controlling parameter.

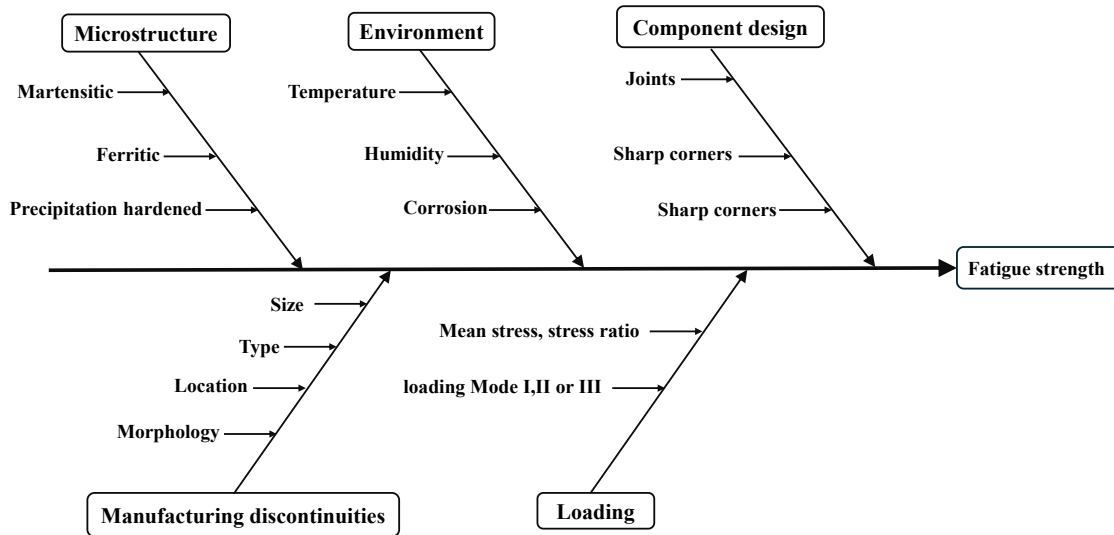


Figure 2-21. A fishbone diagram illustrating the factors affecting the fatigue behaviour of a component.

For LPBF alloys, the presence of discontinuities and the associated scatter in fatigue strength necessitate careful interpretation of fatigue behaviour. Conventional S–N curves are limited, as they do not account for discontinuity severity. In contrast, the KT and AL diagrams provide effective frameworks for relating discontinuity severity to fatigue strength. By incorporating some of the key factors illustrated in the fishbone diagram (Figure 2-21), these approaches present fatigue behaviour as maps linking allowable stress to flaw size, while also capturing the transition from microstructure-controlled crack initiation to LEFM dominated crack propagation.

Applications of KT-based approaches to additively manufactured alloys

Several studies have investigated the application of the Kitagawa-Takahashi (KT) diagram in understanding the fatigue behaviour of LPBF alloys. For example, Andreau et al. [92], in their study on LPBF-manufactured 316L alloy, reported significant scatter in fatigue strength at 10^6 cycles, indicating substantial variability in fatigue response. This observation motivated a more focused investigation to better understand the behaviour at 10^6 cycles and a stress ratio $R = 0.1$. Subsequently, they examined the harmfulness of artificial discontinuities with controlled spherical geometries. Fatigue testing was then conducted to assess the influence of discontinuity size and location, with the results compared against optimised specimens exhibiting minimal discontinuities. This approach enabled a detailed evaluation of the competing effects of surface

and internal discontinuities in LPBF 316L steel. Their fatigue tests showed that specimens failing from the microstructure exhibited an average fatigue strength of 568 ± 12 MPa. According to the authors, this represented one of the highest reported fatigue strengths in the literature at $R = 0.1$. On the other hand, specimens failing from surface and internal discontinuities exhibited decreasing fatigue strength with increasing discontinuity size, measured using $\sqrt{area}$. Finally, they used the KT diagram in Figure 2-22 to show their results to aid in examining the interrelationship between the data generated and the observed sizes of discontinuities at the failure origin.

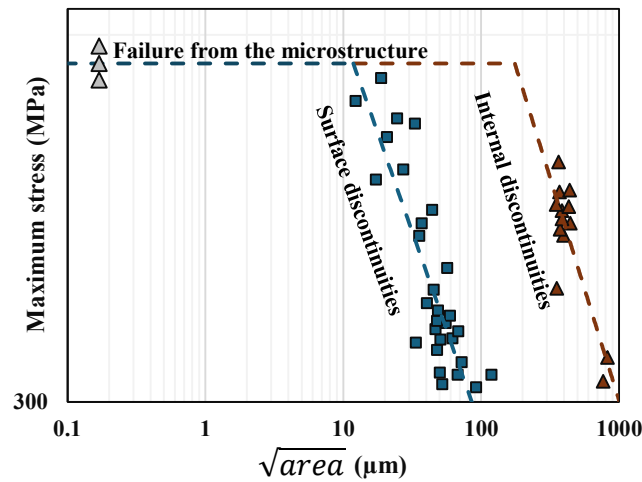


Figure 2-22. The KT diagram for all the characterised samples. The discontinuity size associated with microstructural crack initiation was arbitrarily fixed at $0.5 \mu\text{m}$ (the range of the cell size) [92].

The KT diagram for the 316L LPBF-built alloy successfully distinguished fatigue behaviours according to failure origin.

For specimens that failed from discontinuities, they were all aligned with the prediction trend of the LEFM, and thus, the authors found no need to incorporate the models of El-Hadad or Murakami on the KT diagram for the alloy. Nevertheless, the results from specimens failing from internal discontinuities (triangled brown symbol) were distinct from those that failed from the surface (squared blue symbol). In fact, the KT diagram unveiled that the critical internal discontinuity size was approximately one order of magnitude greater than that of surface discontinuities.

It is interesting to note that the absence of data to describe the short crack region was attributed to the specimen sizes chosen. The authors chose small specimen sizes of 5×3 mm to trigger failure from the microstructure, while the other specimens' sizes were 10×6 mm. Obtaining results that allow failure from the microstructure must be under conditions of reduced presence of discontinuities, which can be achieved through highly optimised printing parameters, along with small-sized specimens.

In a related study on the same alloy, by Roirand et al. [93], towards exploring the competition between the microstructure and the discontinuities in 316L LPBF produced steel, they reported a possible microstructural influence associated with grain size, when the discontinuity size is small ($\sqrt{area} < 100 \mu\text{m}$). Nevertheless, they concluded that despite the reduced size of discontinuities, the microstructural control to the fatigue behaviour of the alloy was minimal.

Another application is seen in the work of Poulin et al. [94], where they leveraged on the KT diagram, the Murakami model and the El-Haddad model to evaluate the predictability of fatigue strength. Their approach was to introduce pores of known sizes into LPBF Inconel 625 and test if the fatigue strength correlated well with predictions.

More recent LPBF papers also use the KT-based framework to compare discontinuity-tolerant approaches (Murakami, El-Haddad, and LEFM-based routes) and to justify model choice for practical components. Salvò et al. [95] studied LPBF CuCrZr with intrinsic discontinuity and benchmarked multiple discontinuity-tolerant methods, including Murakami's $\sqrt{area}$ model and an extended KT approach. They showed that KT-type models can provide reliable fatigue-threshold predictions when they are properly validated with experiments. This is particularly relevant, where small changes in LPBF processing parameters can significantly change the size and density of discontinuities and therefore the fatigue performance. Similarly, Perghem et al. [96] compared different fatigue life and strength models for LPBF Scalmalloy, including the El-Haddad model based on the long-crack threshold. Their objective was to evaluate the predictive capability of these models under varying surface conditions and build orientations, which is particularly relevant for aerospace applications.

Overall, these reviewed works highlight that newly developed LPBF alloys often require dedicated experimental calibration to validate their behaviour against the KT diagram. Such

investigations promote a broader understanding of fatigue behaviour. This is particularly important because discontinuity size, morphology, microstructure, and surface finish may vary significantly between alloys and even between builds. The KT diagram for new alloys should be carefully calibrated using real fatigue data and realistic discontinuity measurements to improve predictive understanding of alloy fatigue behaviour before confidently employing them for fatigue-critical components.

Applications of the Atzori–Lazzarin diagram to additively manufactured alloys

A representative example of notch behaviour is provided in the work of Collini and Meneghetti [97], who investigated the fatigue response of severely notched LPBF-produced AlSi10Mg specimens. Their results, presented in the Atzori-Lazzarin diagram in Figure 2-23, included plain specimens, single-edge-notch specimens, and sharp V-notch specimens with opening angles of $2\alpha = 45^\circ$ and 110° . The results for the plain specimens (represented by empty markers) were controlled by manufacturing discontinuities and aligned with the LEFM trend. In contrast, increasing notch size and opening angle reduced the stress concentration effect, causing the results to progressively converge toward the behaviour associated with the 110° V-notch condition.

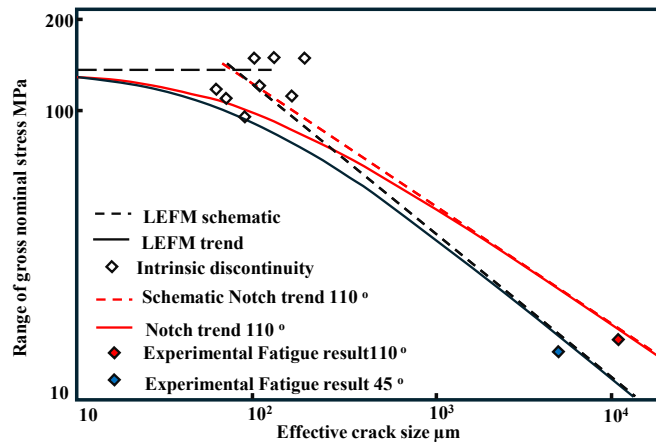


Figure 2-23. Plot of stress range versus effective discontinuity size for crack-like discontinuities (black lines) and notched discontinuities (red line) in LPBF AlSi10Mg alloy, including experimental data [97].

It is noteworthy that the specimens with a 45° opening angle exhibited behaviour closer to the LEFM prediction, indicating that decreasing the notch opening angle promotes a more crack-like response. Furthermore, for this work, the full notch-behaviour plateau identified in Figure 2-19 was not fully explored. Their work showed that in components with extrinsic discontinuities, like in the form of geometric stress raisers (corrosion pits, holes, sharp transitions), the design problem can be notch-controlled fatigue rather than solely by a discontinuity behaving as a small crack. Consequently, for newly developed LPBF alloys, calibration of the AL diagram using representative surface conditions, build orientations, and discontinuity characteristics is essential. This is particularly important for applications where notches are unavoidable.

Another example was reported in the work of Razavi et al. [98] on selective laser-melted Co-Cr-Mo that was tested in both plain and V-notched conditions. The authors used the AL diagram to interpret how internal discontinuities and notch geometry influence fatigue strength for specimens subjected to hot isostatic pressing and vacuum heat-treated conditions. Their findings demonstrated the usefulness of AL-type diagrams for LPBF alloys, particularly when geometric notches are intentionally introduced as design features. The fatigue strength may no longer be controlled by the largest discontinuity but instead by notch-induced stress concentration effects. The AL provides a clear way to place these effects on a single diagram for engineering decisions.

Having discussed discontinuity- and notch-based fatigue assessment approaches, the following sections focus on experimental methodologies commonly employed in determining fatigue strength for KT-based analysis.

2.9 High-cycle fatigue (HCF) testing

The development of the KT diagram for an alloy requires validation using experimental data. Often, the diagram is established under high-cycle conditions for a given alloy, as fatigue failure in this regime is strongly influenced by local stress amplifications associated with discontinuities. Most studies on discontinuities and their effect on fatigue strength are thus performed in the region of $\sim 10^5$ – 10^7 cycles. Pessard et al. [99] showed with HCF tests that fatigue strength of Ti-6Al-4V and scatter are strongly governed by the size of the highly stressed volume, which then alters crack initiation from discontinuities or microstructural features and constructed a KT diagram conveying the regions of dominance. Similarly, Moquin et al. [100] demonstrated that, in LPBF Ti-6Al-4V, discontinuity size and proximity to the free surface can dominate HCF strength. Standard methods often employed in the determination of the fatigue behaviour of an alloy require the use of several specimens, which results in the generation of a large amount of data for statistical analysis. Reducing the experimental burden associated with such extensive testing without a compromise of the quality of data was a concern for early researchers. To overcome this, accelerated fatigue testing methods emerged to reduce the time required to characterise a material's fatigue behaviour without neglecting the incidence of damage accumulation. Several of the methods are focused on accelerating the frequency of testing, exploiting physical quantities like temperature dissipation, energy dissipation, and acoustic emission. However, in the context of this research work, the next subsection offers a brief discussion on the methods focused on the loading sequence applied to specimens to unveil their fatigue response.

2.9.1 Locati method

The Locati [101] method is an accelerated fatigue testing technique that emerged in 1955, primarily used for the estimation of the fatigue strength of an alloy based on a single sample. The rapid test nature of this method was leveraged by Schroeder et al. [102], who used it as a check to facilitate comparison of post-treatments for LPBF Ti-6Al-4V hip-stem-type components.

Similarly, Ji et al. [103] used the Locati step-loading method to obtain fatigue-strength estimates while studying LPBF Hastelloy X, demonstrating that the method is particularly effective when the objective is to compare processing routes rather than generate full S–N curves. The application of this method requires prior knowledge of the specimen's S–N curve, which is often needed to provide an idea of its slope. The S–N curve of the material is drawn first, and parallel sloped lines (same slope as the material's S–N curve) are drawn to each of its sides. The test begins at a stress level below the fatigue strength corresponding to the target number of cycles. Thereafter, the stress level is increased if the specimen does not fail at the number of cycles and continues till failure. Miner's rule is employed to estimate the fatigue strength in relation to the initially assumed S–N curve of the material, using Equation 2-21.

$$\sum_{i=1}^m \left(\frac{n_i}{N_i} \right) = 1 \quad \text{Equation 2-21}$$

Where m is the number of stress steps till the sample fails, n is the number of cycles the specimen experiences at each stress level of the test, and N is the number of cycles obtained using the assumed S–N curves corresponding to the test stress. Based on the calculated values of $\sum(n/N)$ From the assumed curves, the fatigue strength at the pre-specified cycles is determined by interpolation. This methodology becomes limited when a prior knowledge of the material's S–N curve is not available, especially in the case of newly developed alloys. Nevertheless, researchers like Zhang et al.[104] have proposed modifications to it by eliminating the prior requirement.

2.9.2 Step loading technique

In a modified accelerated testing approach by Nicholas and Maxwell [105], they modified the accelerated testing technique of Locati by testing the specimen at a stress level below the expected fatigue strength for a given number of cycles. If the specimen does not fail, the specimen is loaded to a higher stress and then subjected to the cycles of interest. This cycle is continued until the specimen fails. A schematic representation of a typical step loading protocol of five tested specimens, where the target number of cycles is 1.10^7 is shown in Figure 2-3. The fatigue strength of each specimen is computed using Equation 2-22.

$$FS = \sigma_0 + \Delta\sigma \left(\frac{N_{fail}}{N_{life}} \right) \quad \text{Equation 2-22}$$

Where FS is the fatigue strength corresponding to the cycle of interest, σ_0 is the last level of test stress before failure, $\Delta\sigma$ is the step increase in the stress level, N_{fail} is the number of cycles to failure and N_{life} is the target number of cycles. The technique enables the estimation of the fatigue strength of a particular specimen, especially when the influence of discontinuity is to be studied. The technique also offers significant cost and time advantages, with the elimination of the need for many samples. This modified method has also found its way into AM materials and is widely used. For instance, it was used in the previously mentioned work of Pessard [99] for the high-cycle fatigue studies on Ti-6Al-4V. Gupta et al. [106] used a step-loading procedure to reveal large fatigue-strength differences between smooth and notched LPBF Ti-6Al-4V brackets, illustrating a practical engineering use case rapidly quantifying how design features amplify LPBF's inherent discontinuity sensitivity. In addition, step-loading-derived fatigue limits can be paired with microstructural/discontinuity analysis to interpret why some LPBF parameter sets perform better; for instance, Esmailizadeh et al. [107] reported fatigue-strength differences obtained via step-loading while connecting performance changes to discontinuity microstructure interactions in LPBF Hastelloy X.

The main caution for LPBF is that Locati/step-loading results can be biased by load-sequence effects and by damage accumulated in earlier blocks, which may be more pronounced when crack initiation is controlled by surface roughness or near-surface lack-of-fusion.

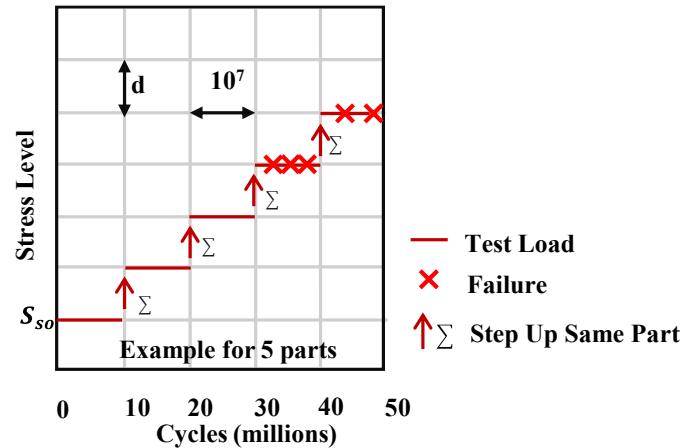


Figure 2-24. Schematic representation of a typical step-loading fatigue-testing protocol.

2.9.3 Staircase loading technique

This is another accelerated testing technique, also called the up and down method. The technique was originally proposed by Dixon and Mood in 1948 [108] and differs from the step loading technique in terms of the loading sequence. In this method, the first specimen is first tested at an estimated stress level of the fatigue limit. If the specimen survives the step, the stress level is raised for the next specimen to be tested. However, if the specimen fails, the next specimen is tested at a lower stress level. The increment between the stress levels is equal. Therefore, the stress level is seen to fluctuate around the mean value of the fatigue strength with an accuracy that is dependent on the number of tested specimens. A schematic depiction of the staircase loading protocol is shown in Figure 2-25.

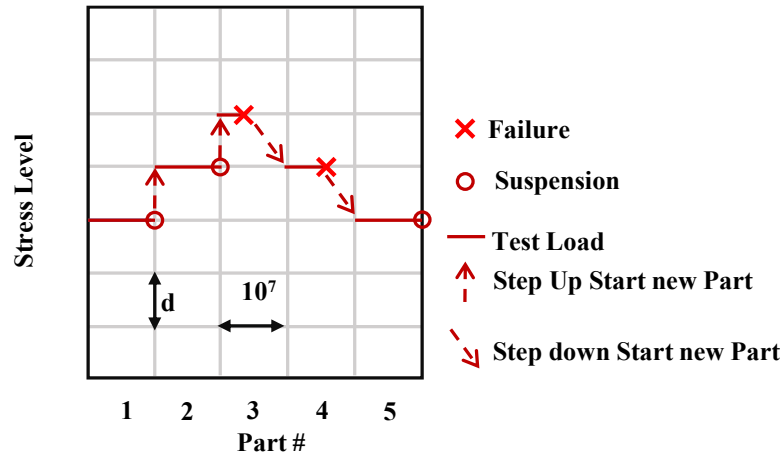


Figure 2-25. Schematic representation of a staircase fatigue-testing protocol.

Several studies have shown that the results obtained from any of the above-discussed testing protocols are reliable. The choice of any of these techniques will depend on the intended goal, the available resources, which include cost, time, and number of specimens. For instance, Thompson et al [109], in a comparative analysis, noted that the step loading technique is better suited for understanding the discontinuity tolerance of an alloy. However, for instances where the dominance of discontinuity is absent, it provides a better estimation of the mean fatigue strength.

2.10 Anticipated fatigue properties of the LPBF M789 alloy

In view of the peculiarity of discontinuities in LPBF alloys, it is important to anticipate the fatigue response of the alloy in comparison to alloys of the same classification. A recent work by Ordnung et al. [110] that focused on the fatigue properties of the M789 alloy in the as-built surface condition reported improved fatigue performance following surface treatment and the introduction of compressive residual stresses. A comparison of the fatigue data for the alloy extracted from the S-N curve generated in their work and other LPBF alloys for a stress ratio of -1 is given in Figure 2-26. A scatter in the fatigue strength across the number of cycles can be observed across all the alloys and is attributed to the influence of discontinuities, with limited failures attributed solely to the microstructure. This influence of discontinuities is also expected in the M789 alloy under cyclic loading conditions, thereby reinforcing the need to unveil its fatigue behaviour.

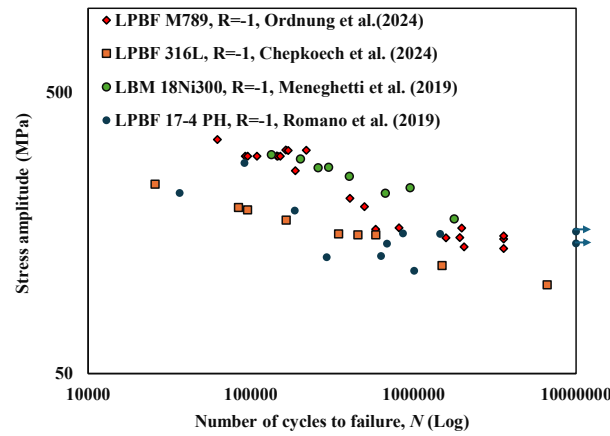


Figure 2-26. Comparison of S-N data extracted from the literature of axial fatigue of LPBF M789 [110], 316L [111], 18Ni300 [36] and 17-4 PH [112].

2.11 Literature summary

The literature and background theory show that fatigue resistance remains one of the main challenges limiting the wider use of Laser Powder Bed Fusion (LPBF) alloys in structural applications. Although LPBF alloys offer important advantages, their fatigue strength is often lower and more scattered than that of conventionally manufactured materials. This behaviour is mainly controlled by discontinuities present in LPBF parts. The severity of these discontinuities depends on their size, type, location, and shape. In many cases, fatigue failure is governed not by the average discontinuity population, but by the largest discontinuity present in the material. As a result, fatigue assessment requires both a fracture-mechanics approach and a statistical description of discontinuity populations. EVS, often using Gumbel distributions, are increasingly applied to estimate the maximum discontinuity size within a stressed volume of a built part.

The literature review also identifies the Kitagawa-Takahashi (KT) diagram as an essential tool for understanding how fatigue strength changes with discontinuity size. In LPBF materials, discontinuities are often treated as pre-existing cracks, making the KT diagram particularly suitable for fatigue assessment. However, the classical KT diagram can sometimes lead to non-conservative predictions. This limitation has led to the development of modified approaches such as the El-Haddad model, which introduces the parameter a_0 , and Murakami's model, which is

based on the $\sqrt{area}$ parameter. The concept has also been extended to the AL diagram, which is used when discontinuities behave more like notches than cracks.

In addition, fatigue testing methods commonly used to evaluate the fatigue strength of materials containing discontinuities are reviewed briefly. The methods considered include the Locati method, the step-loading method, and the staircase method. These methods offer important advantages, including faster testing, improved time efficiency, and suitability when only a limited number of specimens are available. However, they also have limitations, particularly the possibility of cumulative damage effects in materials that are sensitive to loading history.

Finally, the review rounds up with a comparison of the limited fatigue information of the M789 alloy from a recent publication with that of alloys in a similar classification. This helped in the anticipation of the alloy's behaviour, particularly regarding the scatter of its results in the high cycle regime.

2.12 Research gaps

The foregoing discussions and literature review reveal the significant importance of unveiling the fatigue behaviour of newly developed discontinuity-prone LPBF alloys like M789. Despite its promising properties, the alloy's full potential cannot be realised in service without designing against its shortfalls. The alloy's recency has limited the number of literature studies conducted on its fatigue behaviour. A bibliometric survey was conducted using the "Web of Science" database to evaluate the relative research publications on fatigue behaviour of commonly studied LPBF alloys. The search was performed using the Boolean query:

("Alloy name") AND ("laser powder bed fusion" OR LPBF OR "additive manufacturing" OR AM) AND (fatigue OR "fatigue life" OR "fatigue crack" OR "S-N").

The number of publications returned for each alloy was used to generate Figure 2-27. As shown, alloys such as Ti-6Al-4V, 316L, and AlSi10Mg exhibit a substantial volume of fatigue-related publications, whereas M789 shows a markedly lower number of studies. This limited literature is an indication of the limited knowledge available on the alloy despite its potential.

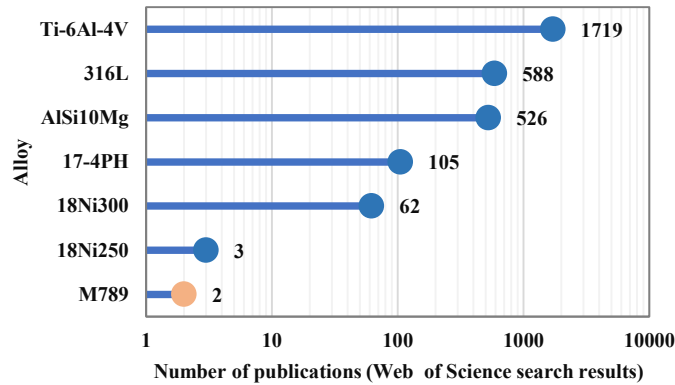


Figure 2-27. Log-scale publication counts for fatigue studies on selected LPBF alloys, highlighting the limited literature on M789.

Beyond the limited volume of research, several critical scientific gaps remain. First, no calibrated KT diagram currently exists for LPBF M789, leaving the transition between intrinsic fatigue limit and discontinuity-controlled behaviour not understood. Second, the intrinsic short-crack parameter (a_0), central to the El-Haddad modification of the KT framework, has not been experimentally identified for the alloy. Meaning its sensitivity to discontinuities is yet to be ascertained. Third, although fracture-mechanics-based discontinuity modelling and EVS approaches have been independently applied to LPBF alloys, their integrated implementation for LPBF M789 has not been done. The absence of an experimentally validated and statistically informed framework limits predictive reliability and introduces uncertainty in the use of the alloy for fatigue-critical applications. Addressing these gaps will enable the development of a physically grounded, discontinuity-informed, and reliable fatigue strength prediction methodology, supporting the safe and optimised deployment of LPBF-produced M789 in practical engineering environments.

2.13 Research objectives

This thesis is undertaken to address this knowledge gap concerning the LPBF M789 response to fatigue phenomenon, which can serve as a critical input in designing either against or counter fatigue failure of the die inserts for which the alloy is intended. On this note, the underlisted objectives become imperative towards the fulfilment of this research work.

- I. Unveil the influence of intrinsic (LOFs, NMI, Roughness asperities, pores) and extrinsically induced discontinuities (electrochemical pits and internally designed channels) on the fatigue strength of the M789 alloy.
- II. Ascertaining which of the fatigue strength prediction models of El-Haddad, LEFM, and Murakami is most accurate for the prediction of the fatigue strength of the M789.
- III. Assess the critical factors influencing discontinuity severity in M789 beyond discontinuity size through fractographic analysis and KT-diagram interpretation.
- IV. Identifying the fatigue failure mechanisms of the alloy and evaluating the influence and potential contribution of its microstructural characteristics.

The experimental plans for each of the identified objectives are shown in Table 3-1. The objectives are listed in sequential order in each column, while the corresponding experimental activation defines the rows. Each box tick represents an intended experiment for the referred column.

CHAPTER 3 SCIENTIFIC APPROACH ADOPTED FOR THIS RESEARCH

3.1 Organisation of the FATCO research project

This doctoral research is part of FATCO Project: Fatigue-Corrosion International Research Project. This project is a collaboration between industrial partners Hydro-Quebec (H.-Q), SACMI, Polytechnique Montreal, McGill, Ecole de Technologie Supérieure (ETS), Politecnico Milano, and University of Ferrara.

The FATCO project aims to improve the understanding and modelling of damage mechanisms in hydraulic turbines and ceramic forming equipment by accounting for fatigue, corrosion-fatigue, and material heterogeneities under demanding operating conditions. The program is divided into several milestones:

Milestone 1: Electrochemical Characterisation of Corrosive Environments for 13Cr-4Ni Steel

Milestone 2: Development of Corrosion Assessment Methods for CT Crack Growth Specimens

Milestone 3: Influence of Corrosive Environments on Fatigue Crack Growth Behaviour

Milestone 4: Effect of Hydrogen on Intergranular Crack Propagation Mechanisms

Milestone 5: Influence of Microstructure and Impurity Levels on Mechanical Behaviour

Milestone 6: Correlation Between 3D Discontinuities, Stress Fields and Crack Propagation

Milestone 7: Mechanical Behaviour of Different Martensitic Morphologies

Milestone 8: Influence of Austenite Grain Structure on Crack Propagation Thresholds

Milestone 9: Probabilistic Fatigue Failure Prediction Using Kitagawa-Based Modelling

The work presented in this thesis constitutes part of milestones 6 and 9.

3.2 Purpose and structure of scientific approach

This chapter describes the scientific approach adopted to achieve the objectives of this research work. The work leverages intrinsic and extrinsically induced discontinuities to investigate the fatigue strength of the alloy, ascertain the suitability of established fatigue-strength prediction models, identify the factors governing discontinuity severity, and elucidate the alloy's fatigue-failure mechanisms in relation to its microstructure. Consequently, a combined approach of experimental fatigue testing, discontinuity characterisation, fractographic analysis, and analytical modelling of the experimental findings was adopted. The objectives of this research work have been given previously in Section 2.13.

To illustrate how each objective activity contributes to the research objectives, Table 3-1 maps the experimental and analytical tasks to the specific objectives they support. This ensures that the scientific approach remains coherent, targeted, and aligned with the intended outcomes of the study.

Table 3-1. Illustration of the contribution of each activity to the corresponding research objectives.

Scientific activity	Objective 1	Objective II	Objective III	Objective IV
Sample manufacturing and supply	Samples provided for and received from SACMI/Bohler. ✓			
Material characterisation (OM, SEM, chemical analysis, hardness, and tensile test)	Material characterisation ✓			
Cyclic potentiodynamic polarisation (CPP) and PotPit	✓			
High-cycle axial fatigue test	✓			
Fractographic analysis: SEM, OM and FIB		✓	✓	✓
Non-destructive characterisation, Metallography and Roughness measurement, FEA		✓	✓	
Extreme value statistics (EVS)		✓		
Determination of $\Delta K_{th,LC}$		✓	✓	
Finite element analysis of channels and pits			✓	



✓ indicates that the activity contributes directly to the corresponding research objective

3.3 Overview of the research workflow

The scientific approach adopted in this thesis is structured to address the objective identified in Chapter 2. Detailed descriptions of some experimental procedures are provided in the accompanying publications associated with this thesis. This section outlines the logical sequence of activities and the role each component plays in achieving the intended outcomes. The workflow integrates the various experimental and analytical activities into a coherent research strategy.

The workflow begins with material qualification and is then followed by specimen selection and preparation to generate representative intrinsic and extrinsic discontinuities. Next, we have the production of pits as an extrinsic discontinuity of the polished fatigue specimen to introduce. Then we have non-destructive characterisation to estimate the largest discontinuity sizes and support fatigue-strength prediction for each specimen. Thereafter, all fatigue specimens undergo high-cycle fatigue testing. The resulting data, combined with fractographic observations, feed into analytical and numerical models used to evaluate the predictive capability of discontinuity-based fatigue-strength models and to identify the factors governing discontinuity severity. In addition, predicted results are compared with experimental results using parity plots. The overall workflow is illustrated in Figure 3-1, while the specific role of each specimen category is discussed in the following subsections.

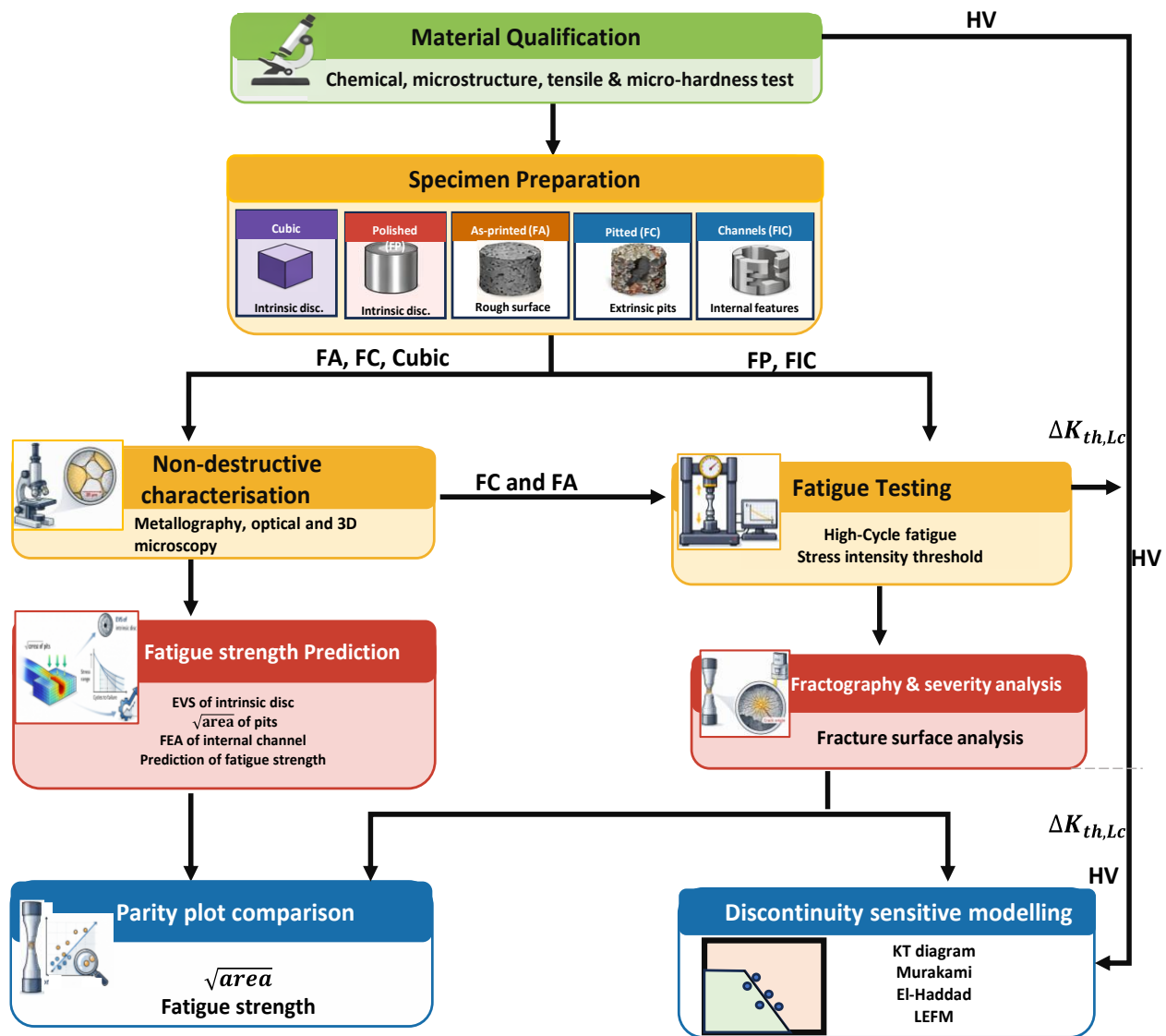


Figure 3-1. Structured research workflow illustrating the integration of specimen types, characterisation techniques, fatigue testing, and discontinuity-sensitive modelling within the scientific approach

3.4 Publication strategy

The research carried out during these studies led to the submission of an article to a peer-reviewed journal. The article (Article 1) forms Chapter 4 of this thesis and is briefly described below. All fatigue results of FA, FP, and FC are used in the publication of Article 1, except that of FIC, which is discussed in Chapter 5 and shall be used for future publication. However, the tabulated results and fractography of all specimens are given in Section A.3.

Article 1: Discontinuity-controlled fatigue performance and model assessment for LPBF-M789 maraging stainless steel with different surface conditions

This first article addresses Objectives I, II and partly IV. Three specimen groups of FA, FP and FC are investigated to unveil the discontinuity (intrinsic and extrinsic) behaviour on the alloy. Murakami, El-Haddad and LEFM-based fatigue strength prediction models are evaluated for accuracy within the framework of experimentally constructed KT diagram of the alloy.

This article was submitted for publication in the International Journal of Fatigue on December 19, 2025 (manuscript **IJFATIGUE-D-25-01980R2**) under the title:

Discontinuity-controlled fatigue performance and model assessment for LPBF-M789 maraging stainless steel with different surface conditions (K. Sanni, A. D’Andrea, M. Cova, L. Patriarca, M. Brochu, 2026).

3.5 Material qualification

Material qualification forms the foundation of the scientific approach adopted in this thesis. Before commencement of the experimental activities, it is essential to establish the baseline properties of the LPBF-produced M789 alloy. This ensures that the findings of this work are conducted with specimen materials that conform to the alloy's specification. The qualification activities of chemical analysis, hardness and tensile testing, microstructural examination, and intrinsic discontinuity assessment are presented in detail in Section 4.3.1 of Article 1. They collectively define the reference material state against which the fatigue performance of the various specimen groups (FP, FA, FC, FIC) is evaluated throughout the remainder of the thesis.

3.6 Specimen design and selection

To address the research objectives, specimens were chosen to isolate the influence of different discontinuity types on the fatigue behaviour of the alloy. The selected specimen categories are described below.

Cubic specimens

Cubic specimens were used for microstructural qualification and intrinsic discontinuity characterisation. These specimens enabled the quantification of pores, LOFs, and NMIs, and provided the statistical data for predicting the largest intrinsic discontinuity expected in a polished specimen. Their role in this study is for material qualification and the prediction of the largest discontinuity in polished specimens.

Tensile specimen

Tensile specimens were employed to establish the baseline mechanical properties of the LPBF M789 alloy. These include yield strength, ultimate tensile strength, and elongation in vertically and horizontally built directions. Results from these specimens are given in Section 4.3.2.2 of Article 1.

As-printed (FA) specimens

As-printed specimens were tested in the as-built condition to evaluate the fatigue behaviour associated with intrinsic LPBF discontinuities. These specimens retained the surface roughness and near-surface discontinuities generated during the manufacturing process. The discontinuities

targeted in these specimens are mainly the rough surface asperities together with subsurface discontinuities like the pores, LOFs and NMIs that could enhance the severity. The results will convey the extent to which the as-printed surface reduces the fatigue strength of the M789 alloy in comparison with the polished specimens.

Polished (FP) specimens

Polished specimens were applied to remove surface asperities generated during the LPBF process and to produce a smooth surface finish. This will allow the intrinsic discontinuities in the alloy to be explored. It is expected that the results from this specimen will be dominated by intrinsic discontinuities like pores, LOFs, and NMIs. The fatigue results obtained from polished specimens will be used as inputs in the construction of the KT diagram for the alloy.

Polished and electrochemically pitted Specimens (FC)

Polished specimens containing electrochemically produced pits were used to investigate the influence of controlled surface discontinuities on fatigue behaviour. Corrosion pits were generated using cyclic potentiodynamic polarisation to produce discontinuities with controlled geometry and size. Pits as an extrinsic discontinuity were considered because the intended service application environment is expected to predispose the alloy to pitting corrosion. Consequently, the specimen is predisposed to both intrinsic discontinuities of pores, LOFs, and NMIs and extrinsic discontinuities of Pits.

Specimens with internal channels (FIC)

Specimens containing internally designed channels were used to investigate the fatigue behaviour associated with extrinsic discontinuities in the form of the channels. The channels were intentionally introduced during the LPBF manufacturing process and represent the expected stress concentration from mechanically designed channels likely to be encountered in the service component for which the alloy is intended. Hence, they are also referred to as a technological sample.

A breakdown of the sample types is given in Table 3-2. and a summary of their purpose in Table 3-3. In addition, the dimensional drawings for the FA, FP, FC specimens are given in Article 1 of Chapter 4, and that of the FIC specimens is shown in Figure 3-2.

Table 3-2: Breakdown of samples used for this research.

Sample type	Dimension mm	Type and description	Number	Purpose
Cubic sample	17 x 17 x 17		5	Microstructural characterisation and hardness test
Tensile sample	2.6 x 6 x 32	Vertical Horizontal	2 2	Tensile test
Fatigue sample	2.6 x 26 x 11.6	As-printed (FA)	5	Fatigue test
		Polished (FP)	5	
		Polished and pitted (FC)	5	
		Specimens' internal channels (FIC)	5	
Rectangular bars	12.7 x 110 x 24	All vertical and polished	3	Estimation of $\Delta K_{th,LC}$ of the alloy

Table 3-3: Discontinuities expected to be the competing triggers to failure in each specimen.

Type of fatigue specimen	Discontinuity classification	Expected competing discontinuity triggers
As-print specimen (FA)	Intrinsic	Roughness and internal discontinuities of (LOF, NMIs, Pores)
Polished specimen (FP)		Internal discontinuities (LOF, NMIs, and Pores)
Specimen with pit (FC)	Combined Intrinsic and extrinsic	Internal discontinuities (LOF, NMIs, and Pores) and electrochemical pits
Specimens with internal channels (FIC)		Internal discontinuities (LOFs, NMIs, and Pores), internal channels)

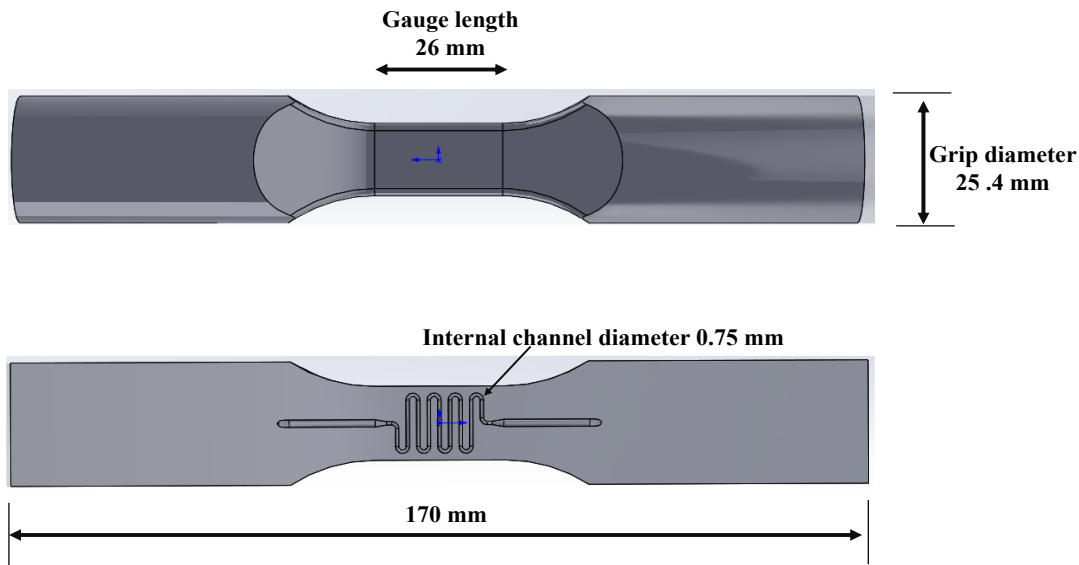


Figure 3-2. Drawing of the fatigue specimen containing an internal channel (FIC). Dimensions are given in mm.

3.7 Specimen preparation

The principal specimen-preparation procedure employed in this work was the polishing of cubic specimens for microstructural and discontinuity analysis and fatigue specimens (FP, FC, and FIC). No additional surface preparation was performed on the as-printed fatigue specimens (FA). The detailed polishing procedure is presented in Section 4.3 of Article 1.

3.7.1 Generation of extrinsic discontinuities (pits) on FC specimens

Critical to this work is the consideration of extrinsic discontinuities, which arise from external factors such as service environment and design-related conditions not associated with the manufacturing process. On this premise, “electrochemical pitting techniques were employed to generate pits on the specimens to obtain discontinuities representative of those expected during service exposure of the M789 alloy. The objective was not to investigate the electrochemical properties of the alloy, but rather to employ the technique solely for pit generation. Consequently, potentiodynamic polarisation (CPP) was used and complemented by a potentiostatic pitting (PotPit) technique. This supplementary technique was employed to ensure variability in pit size and geometry and expand the range of discontinuity morphologies included in the fatigue

assessment of the M789 alloy. The CPP procedure is already described in Section 4.3.5.3 of Article 1. A summary of the PotPit procedure is presented here.

Potentiostatic pitting (PotPit)

In this technique, a constant potential was applied to the specimen while the current response was monitored over time. The electrochemical scheme of the PotPit technique is shown in Figure 3-3.. A potential higher than the pitting potential was applied and maintained for a selected holding time. The test was stopped either when the current density, I , exceeded a predefined limit, I_p , ($|I| > I_p$) or when the specified holding time was reached.

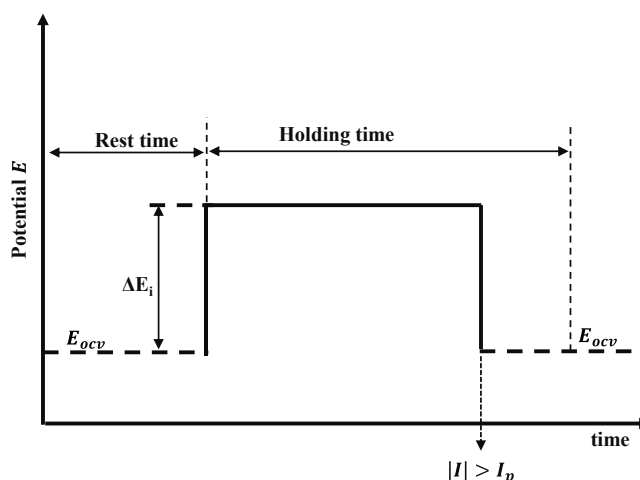


Figure 3-3. Schematic representation of the potentiostatic pitting (PotPit) procedure.

To determine the appropriate potentiostatic potential, a stepwise potential-increase method was used (see Figure 3-4a). The specimen was first polarised at a potential slightly below the experimentally determined pitting potential (E_{pit}) obtained from cyclic potentiodynamic polarisation (CPP). This potential was held for 120 seconds to allow possible pit formation. If no pitting was detected, indicated by a stable current response, the potential was increased by 0.2 V versus Ag/AgCl, and the test was repeated. This process continued until pitting occurred. The onset of pitting was identified by a clear increase or fluctuation in the anodic current density, as illustrated in Figure 3-4b. After the corrosion process, the specimens were removed, rinsed thoroughly and air dried. A typical example of a pitted surface is shown in Figure 3-5. Overall, the use of electrochemical pitting within the scientific approach served a single, targeted purpose: to

introduce surface discontinuities that enable the mechanistic evaluation of discontinuity-driven fatigue behaviour in the M789 alloy.

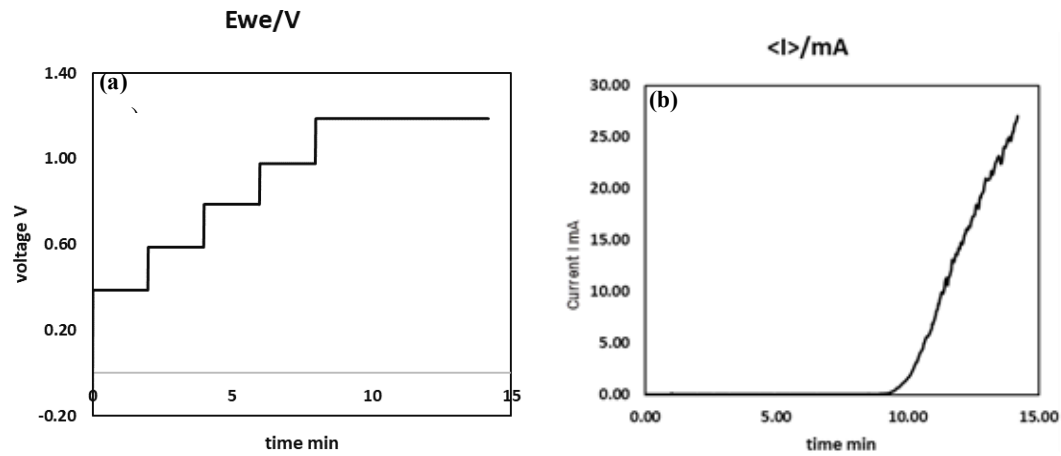


Figure 3-4. Results obtained from the PotPit technique: (a) stepwise increment of the pitting voltage and (b) evolution of the current response with time.



Figure 3-5. Representative FC specimen after electrochemical pitting.

3.8 Non-destructive characterisation strategy

Non-destructive characterisation (NDC) is relevant to the research approach as the different surface conditions of the alloy are associated with distinct discontinuity populations. The objective of this step is to help obtain the discontinuity population and statistical prediction of the largest discontinuity in FA (roughness, $R_{v,max}$) asperity and FP specimens using cubic samples. As for the FC specimens, measurement of the largest pit can be performed directly on the surface. The primary characterisation techniques included metallography and optical microscopy used for the characterisation, roughness measurement and pit characterisation. The procedure used for the NDC is detailed in 4.3.5 of Article 1, including the employed imaging modalities and data-processing

methodology. The emphasis here is on the characterisation of the most severe discontinuity, which is critical towards the prediction of the fatigue strength of the alloy discussed in Section 3.10.

3.9 Finite element analysis (FEA) of the FIC specimen

The channels produced in the FIC specimen can be classified as geometrical discontinuities or notches since they possess a defined shape with a radius and depth. This specimen is not included in the published article, as such, it is necessary to determine the stress concentration associated with the channel geometry due to its potential notch effect on fatigue strength. The specimen was not discussed in Article 1; as such, the simulation methodology is briefly described below.

3.9.1 Simulation procedure for determining the stress concentration factor k_t of the internal channel

Finite element analysis (FEA) was carried out using Abaqus CAE software to simulate and obtain the stress concentration factor of the channels in the fatigue specimen. The material properties used for the simulation are Young's modulus of 200 GPa and a Poisson ratio (ν) = 0.3. A half-symmetry model across the specimen thickness was employed for the simulation to reduce computational cost and simulation time. Figure 3-6 shows the model used for the FEA simulation. Y-symmetry boundary conditions were imposed on the model, while displacement in the Z-direction was constrained, and a nominal stress of 1 MPa was applied on the top Z-plane of the specimen. Quadratic 10-node tetrahedral elements (C3D10) were employed in the model. The total number of elements in the model was 66368, and the number of nodes was 104418 with a global size of 2 μm and a fine mesh of 0.15 μm . A convergence study was performed to ensure that the results are consistently optimised, considering time and computing requirements, so that the results are unaffected.

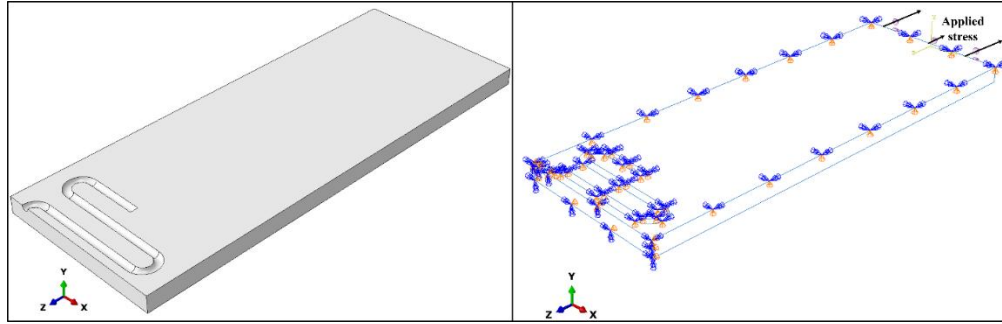


Figure 3-6. Finite element simulation model of the gauge section with a cooling channel.

3.9.2 Stress distribution from the notch

The maximum stress concentration factor obtained from the FEA simulation process was 2.89, as shown in Figure 3-7. Also, the stress gradient in the Y-direction is shown in Figure 3-8. It can be seen that stress concentration peaks at the base of the notch relative to the Y-axis. It, however, reduces with a power law of the form shown in Equation 3-1 as the distance from the notch root increases towards far regions of the specimen until it is equivalent to the far field applied stress of 1 MPa.

$$y = Nx^{-m} \quad \text{Equation 3-1}$$

Where x is the distance from the maximum stress concentration. From Figure 3-8, the value of N is 0.9661, while m is 0.256 which is different from the value of 0.5 often obtained for crack-like discontinuities [113].

This result from the simulation provides information that can be used for post-fatigue test analysis for the FIC specimens, particularly if their behaviour is different from that associated with cracks.

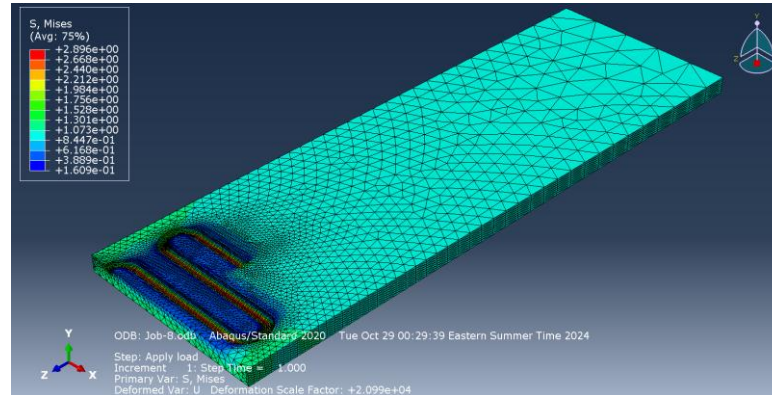


Figure 3-7. FEA simulation of the channels in the fatigue specimen.

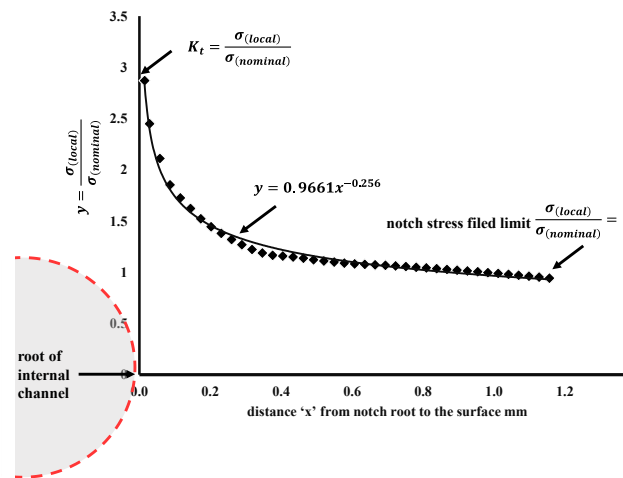


Figure 3-8 Stress distribution in the y-direction of the specimen with a channel.

3.10 Strategy for fatigue strength prediction

The fatigue strength prediction strategy is hinged on predicted largest discontinuity sizes described in Section 3.8 for all groups of specimens. The prediction strategy is adapted for each group since they have a different type of dominant discontinuity. For FA and FP specimens, EVS are used for the predictions on internal discontinuities and the deepest valley values of roughness measurements $R_{v,max}$ respectively. In the case of FC specimens, the prediction is based on the largest $\sqrt{area}$ electrochemically produced pit. For FIC specimens, the prediction is based on the $\sqrt{area}$ of the channel where the failure is expected to occur. Consequently, the predicted sizes are used to predict the fatigue strength. Except for FIC specimens. The details of the predictions for FA, FP and FC specimens are given in article 1 of Chapter 4.

3.11 Fatigue testing strategy

Constant amplitude loading

The fatigue testing strategy adopted in this work is designed to isolate the influence of discontinuities on the high-cycle fatigue (HCF) behaviour of LPBF M789 at 2.10^6 cycles. The selected number of cycles is representative of the expected service life conditions. Furthermore, the selected fatigue life lies within the high-cycle fatigue regime, beyond the knee point of the S–N curve, where variations in fatigue behaviour become significant, particularly in the presence of discontinuities.

Axial loading was chosen to ensure a uniform stress state within the gauge section, thereby avoiding the stress gradients commonly associated with bending configurations. This uniformity is essential for isolating the role of discontinuities across all the specimens. The adopted loading protocol was based on the step-loading methodology proposed by Maxwell and Nicholas [114] for the targeted 2.10^6 cycles. The methodology is regarded as a rapid testing technique suitable for situations involving limited specimen availability, while also enabling the evaluation of discontinuity effects on an individual specimen basis. The implementation of the protocol is detailed in Article 1, Chapter 4.

Fatigue test of FIC specimens

The test for the FIC specimen follows the same protocol as other specimen groups, except for the choice of the starting stress amplitude. To determine the starting stress amplitude, it was assumed that failure would occur at a lower stress amplitude than that observed for the pitted specimens. Consequently, the starting stress amplitude was selected to be 20 MPa lower than that used for the pitted specimen group. All other testing conditions remained the same as described for the other specimens.

Stress intensity threshold test

The test procedure was adapted for estimating the $\Delta K_{th,LC}$ of the alloy is given in Section 4.3.3.1 of article 1.

3.12 Fractographic analysis of failed fatigue specimens

Fractographic analysis constitutes the final analytical stage of the experimental methodology, through a post-mortem examination of fracture surfaces of the mechanisms governing fatigue failure across the different specimen groups.

Primarily, the strategy involves identifying and measuring the $\sqrt{area}$ parameter of the discontinuity at the failure origin, and examining any other microstructural feature around it. For FA and FP specimens, fractography verifies whether pores, LOFs, surface roughness, or a combination of discontinuities acted as the fatigue crack initiation site. For FC specimens, the analysis confirms whether the electrochemically generated pits served as the initiation sites and whether their geometry is consistent with the assumptions adopted in the modelling framework described in Section 3.13. For FIC specimens, fractography information on the internal channel geometry and associated stress concentration influenced the failure origin. The detailed protocols adopted during fractography to measure $\sqrt{area}$ and detailed procedures are provided in Article 1 of Chapter 4.

This approach enables the integration of fracture surface evidence, discontinuity predictions, and fatigue results as useful inputs towards model validation and unveiling the behaviour of discontinuities in the alloy.

3.13 Modelling and data integration strategy

Model construction

The modelling strategy adopted in this study is based on discontinuity-sensitive fatigue frameworks capable of capturing the influence of discontinuities on the fatigue behaviour of the M789 alloy. The fatigue strength was analysed using LEFM (see Equation 2-7), El-Haddad (see Equation 2-9), and Murakami (Equation 2-12) within the framework of the Kitagawa-Takahashi diagram. All models were constructed using the material parameters of stress intensity threshold and the plain specimen fatigue strength estimated from the alloy's microhardness. The assumption is made that

failures will arise from the surface; only surface-related modelling is carried out with the models. This is supported by the fractographic results that showed all failures emerging from the surface and subsurface locations of the specimen.

Model validation

Fatigue test results, together with fractographic measurements of the $\sqrt{area}$ parameter at the fatigue crack initiation site, provides the primary validation for the proposed modelling approach. The experimentally measured fatigue strengths are compared with the predicted values to evaluate the predictive capability of the models within the KT framework. Agreement between predicted and experimental results supports the validity of the modelling approach (El-Haddad, Murakami and LEFM), while discrepancies prompt further investigation into the discontinuity behaviour governing fatigue response in the alloy. In addition, the validation process helps identify the limitations of the adopted models.

Furthermore, the KT framework employed in this study is approached primarily from a mechanistic perspective for interpreting the discontinuity-sensitive fatigue behaviour of the M789 alloy rather than from a probabilistic fatigue-modelling standpoint. Consequently, representative discontinuity sizes and experimentally measured fatigue strengths are plotted without explicit error bars, with the KT diagram treated as a conservative bounding framework. The validation of models using the experimental results are presented in Section 4.4 of article 1.

3.14 Summary of the scientific approach

The scientific approach adopted in this thesis combines the research objectives with a structured experimental and modelling framework designed to explore the influence of discontinuities on the fatigue behaviour of LPBF M789 alloy. A series of specimen types was selected to represent different surface conditions, including as-printed specimens (FA), polished specimens (FP), pitted specimens (FC), and internally notched specimens (FIC). These specimens were characterised using non-destructive methods in order to predict the largest discontinuity size, subsequently used to predict fatigue strength. High-cycle axial fatigue tests were then performed to obtain fatigue strength data for all specimen groups, with fractography to obtain failure sizes of discontinuity. The experimental results are interpreted using discontinuity-sensitive fatigue models, including Murakami, El-Haddad, and LEFM. The fatigue behaviour was analysed within the framework of

the Kitagawa-Takahashi diagram. Together, these steps form the scientific approach that links discontinuity characterisation, fatigue performance, and fatigue strength prediction, and allows the research objectives of this thesis to be addressed systematically.

CHAPTER 4 ARTICLE 1: DISCONTINUITY-CONTROLLED FATIGUE PERFORMANCE AND MODEL ASSESSMENT FOR LPBF- M789 MARAGING STAINLESS STEEL WITH DIFFERENT SURFACE CONDITIONS

This chapter presents Article I, written towards the realisation of the research objectives and establishes the analytical basis for the subsequent chapters. The article investigates the fatigue behaviour of LPBF-processed M789 alloy in the presence of representative intrinsic and extrinsic discontinuities, with particular focus on three specimen conditions: polished (FP), as-printed (FA), and corroded (FC). The study examines how surface condition and discontinuities influence fatigue strength within the Kitagawa–Takahashi framework. The work presented in this article mainly addresses Objectives I and II, and partly Objective IV of these theses. It includes a systematic characterisation of process-induced discontinuities and corrosion pits using fatigue testing, non-destructive evaluation, and fractographic analysis, with the results interpreted using the Kitagawa–Takahashi approach. The experimental procedures described in the article are not repeated in the thesis; instead, they are referenced where necessary. The results obtained in this chapter form the basis for the analysis and discussions presented in the following chapters.

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Discontinuity-controlled fatigue performance and model assessment for LPBF-M789 maraging stainless steel with different surface conditions (K. Sanni, A. D’Andrea, M. Cova, L. Patriarca, M. Brochu, 2026).

Discontinuity-Controlled Fatigue Performance And Model Assessment For LPBF-M789 Maraging Stainless Steel With Different Surface Conditions

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Abstract

Laser Powder Bed Fusion (LPBF) M789 is a recently developed alloy for fatigue-critical tooling applications, yet its fatigue behaviour in the presence of intrinsic and extrinsic discontinuities remains largely uncharacterised. This study evaluates the discontinuity-controlled fatigue response of the alloy using polished, as-printed, and corrosion-pitted specimens, fatigue tested and complemented by non-destructive characterisation, fractography, and extreme value analysis. A Kitagawa–Takahashi diagram incorporating the experimentally measured long-crack threshold was employed to assess the applicability of LEFM, El-Haddad, and Murakami models. The experiments highlighted that failure initiated at discontinuities with contributions from the microstructure. Fatigue strength ranged from 93 to 282 MPa depending on surface condition, making the idealised peak fatigue strength unattainable. Process-induced discontinuities exhibited crack-like behaviour accurately captured by LEFM and El-Haddad models, whereas corrosion pits have a notch-controlled behaviour driven by geometry, with minimal interaction effect between neighbouring discontinuities. The geometric effects account for deviations from fatigue-strength predictions and demonstrate that discontinuity size alone is insufficient to rank severity when morphologies differ. Overall, the study establishes a discontinuity-informed basis for predicting the fatigue behaviour of LPBF M789 and supports the development of reliable design methodologies for fatigue-critical tooling components manufactured from the alloy.

4.1 Introduction

Additive manufacturing (AM) has become a mainstay technology in the manufacturing industry, with laser powder bed fusion (LPBF) spearheading its popularity. LPBF's patronage stems from its versatility in processing a wide range of materials, dimensional accuracy, and part reproducibility. Thus, industries such as aerospace, biomedical and automotive find the LPBF an attractive option for parts with an interesting strength-to-weight ratio [115].

Despite its advantages, LPBF-produced parts often contain discontinuities such as lack of fusion (LOF), non-metallic inclusions (NMI), gas pores, and surface roughness in the as-printed condition. These challenges are continually addressed through improved process control and post-processing treatments, resulting in parts with enhanced integrity. Additionally, extrinsic discontinuities such as corrosion-induced pits are being mitigated through better mechanical design and material selection. As a result, the mechanical properties of LPBF materials are becoming more consistent, with progress being made in enhancing their fatigue resistance. In the critical review of Kan et al. [116] on processed induced porosity in LPBF alloys, they identified Ti-6Al-4V, Inconel 718, AISI 316L and AlSi10Mg as the most commonly investigated LPBF alloys. However, recent research trends have shown that there is an increasing interest in LPBF tool steels like the H13 and M2 alloys [117].

In the context of fatigue, discontinuities in LPBF alloys negatively impair their resistance. This fatigue debit is attributed to stress localisation, crack initiation and propagation to failure of the part. Several authors have identified size as a key influencer of the severity of a discontinuity on fatigue strength [118-120]. However, other considerations, like the shape and location of the discontinuity, contribute to its severity. According to Nadot [39], five critical factors that influence the severity are: i) size, ii) type, iii) position, iv) morphology, and v) loading conditions. Furthermore, the rough surface of as-printed LPBF parts contributes to stress concentration, either acting singularly as troughs or in combination with other discontinuities. As such, there is a competitive quest amongst these influencing factors to drive part failure.

The typical cellular microstructures, melt pool boundaries and characteristic texturing that characterise LPBF alloy also influence their mechanical properties [121]. This influence often manifests in densely built LPBF parts with low porosity, finely polished surface, and microstructural features comparable to the discontinuity sizes. Kan et al. [116] have also identified

build quality and finishing as important in determining whether failure is driven by discontinuity or microstructure. In an earlier work, Nadot et al. [122] showed that there exists a competition for failure trigger between the microstructure and discontinuity size for identical specimen geometry, loading conditions, and discontinuity morphology. They concluded that the relative size of a discontinuity, in comparison to the characteristic microstructural dimension, was more critical to fatigue strength. These findings exemplify the role of the microstructure on fatigue failure, notwithstanding the presence of discontinuities.

In many service applications, components produced by LPBF are required to perform under cyclic loading conditions for a specified number of cycles. Typically, these requirements correspond to high-cycle fatigue regimes, i.e. 10^6 or 10^7 loading cycles. A common design tool used to assess the fatigue resistance of material containing discontinuities is the Kitagawa-Takahashi (KT) diagram proposed in 1976. The diagram depicts the relationship between the size of discontinuities and fatigue strength [123]. Originally, their work set out to investigate the applicability of fracture mechanics to short cracks and, in the course, observed the existence of a transition zone from short to long cracks. Consequently, the authors unveiled the evolution of the fatigue strength with the increasing size of discontinuities. Several models, like El-Haddad [124] and Murakami [125] (discussed briefly in Section 4.2), have been further proposed to improve the modelling accuracy of the trend of the KT diagram, and enhance conservatism. Today, researchers often leverage the KT diagram to understand the influence of discontinuities on the fatigue behaviour of LPBF alloys. Wu et al. [126] employed the KT diagram to study the anisotropic fatigue resistance of LPBF-produced AlSi10Mg alloys by shedding light on the orientation of discontinuity, relative to the build direction. In another application, Cersullo et al. [127] predicted fatigue life by incorporating local stress and size of discontinuity into the KT diagram. More recently, Bergant et al. [128] employed the KT diagram in exploring the significance of discontinuities on the behaviour of electron beam melted Ti-6Al-4V alloy. These examples, to mention but a few, underscore the versatility of the KT diagram in studying the fatigue behaviour of materials dominated by discontinuities.

Given the sensitivity of fatigue strength to discontinuities, it becomes essential to understand their influence on newly developed LPBF alloys intended for cyclic loading applications. Recently, Böhler pioneered a maraging stainless steel called M789 [129] engineered to combine the strength

of maraging steels and the corrosion resistance of precipitation hardened (PH) steels [129]. The alloy is intended for use as a tooling material for die inserts in a high cycle loading application, requiring static and cyclic strength and good corrosion resistance.

Recent research conducted on the alloy focused more on its microstructural properties and very little on its mechanical properties. For instance, Tian et al. [130] investigated the microstructure and static mechanical properties of the M789 LPBF alloy in as-printed and age-hardened conditions. They described the microstructure as having a random crystallographic texture and mechanical properties that are comparable in both the building and transverse directions. In another work [13], the same authors focused on the effect of age-hardening heat treatment on the alloy and reported the presence of plate-like and spherical Al and Ni-rich precipitates in the microstructure. As of late, Ordnung et al. [110] showed that the use of an incremental dual LPBF process can markedly enhance the fatigue performance of the M789 alloy by reducing surface roughness, lowering stress concentration and removing subsurface pores. Their fatigue findings were focused on the global fatigue performance of the alloy across different numbers of cycles and not on discontinuity influence. In a rather different approach, Laliberté-Riverin et al. [131] conducted a preliminary investigation into the alloy's discontinuities and corrosion properties, and employed theoretically estimated parameters to construct its KT diagram. Although Laliberté-Riverin et al.'s [131] work provided valuable insights on the discontinuity influence, further information is desired on the alloy's fatigue behaviour. Particularly, the influence of as-printed surfaces and extrinsic discontinuities like corrosion pits, which were not characterised in their work. Additionally, for a reliable KT diagram, Schönbauer and Mayer [132] recommend experimentally obtaining the stress intensity range threshold for long cracks, $\Delta K_{th,LC}$. The work of Laliberté-Riverin et al. [131] used a theoretical estimate of $\Delta K_{th,LC}$ of $4.42 \text{ MPa}\sqrt{m}$ for the KT diagram of the alloy, but a full evaluation of $\Delta K_{th,LC}$ is still pending. A comparison of S-N data of the M789 with alloys in the same class such as the 18Ni300 maraging steel, 17-4 PH steel and 316L stainless steel is presented in Figure 4-1. The fatigue performance of the alloy falls within the same range as other additively manufactured steels such as LPBF 18Ni300 maraging steel, LPBF 316L stainless steel, and LPBF 17-4 PH stainless steel, with differences likely influenced by discontinuity population rather than microstructure. While this trend is apparent, it triggers a need to explore the discontinuity influence associated with the more recent M789 alloy. A quick search at the time of this work on the Web of

Science returned just the two articles of Ordnung et al. [110] and Laliberté-Riverin et al. [131] on fatigue of the M789 alloy. Therefore, the fatigue behaviour of the alloy remains insufficiently understood, particularly regarding discontinuity-controlled fatigue mechanisms and their representation within the Kitagawa-Takahashi framework. This limitation hinders reliable fatigue design and qualification of M789 components intended for cyclic loading applications.

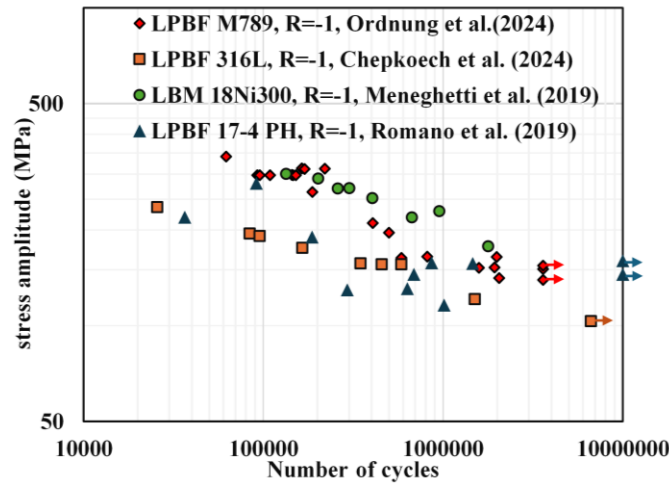


Figure 4-1. Comparison of S-N data extracted from literature of axial fatigue of LPBF M789 [110], 316L [111], 18Ni300 [36] and 17-4 PH [112].

The present study addresses the knowledge gap by investigating M789's fatigue performance in the presence of manufacturing (intrinsic) and artificially induced (extrinsic) discontinuities such as electrochemical pits, as the intended service environment predisposes the alloy to pitting corrosion. In contrast to Laliberté-Riverin et al. [131] that focused on testing only polished specimen, this work extends to as-printed and pitted specimens and constructs the Kitagawa-Takahashi diagram of the alloy using experimentally determined parameters. The predictive accuracy of the El-Haddad and Murakami models is then assessed by superimposing them on the experimental KT diagram, with the goal of identifying the discontinuity characteristics governing fatigue strength. This work provides the first experimentally validated assessment of discontinuity-controlled behaviour in LPBF M789 and is relevant to design methodologies for the use of the alloy in fatigue critical parts.

In this writing, a brief theoretical background is given in 4.2, elucidating the development of the KT diagram along with the models of El-Hadad and Murakami. Then followed by the methodology in Section 4.3, where the first part characterises the basic microstructure and mechanical properties

of tensile, microhardness and the experimental determination of the $\Delta K_{th,LC}$. The second part focuses on the production of electrochemical pits on fatigue specimens. Also included in the section are the protocols adapted for the measurement, non-destructive characterisations, prediction of critical discontinuity size in the fatigue specimens, and the fatigue testing procedure. In Section 4.4, the KT diagram is constructed using the experimentally obtained $\Delta K_{th,LC}$ and microhardness of the alloy. The models of El-Haddad and Murakami are also superimposed on the diagram. Next is Section 4.5, presenting the prediction of the fatigue strength of the specimens using the discontinuity size from non-destructive characterisation. Sections 4.6 and 4.8 respectively focus on the results and discussion, and the conclusions reached are given in Section 4.9.

4.2 Theoretical background

4.2.1 Effect of discontinuities on fatigue strength

The KT diagram describes the relationship between fatigue strength and discontinuity size by combining the discontinuity-free strength with LEFM behaviour for long cracks [70]. The diagram (Figure 4-2) unveils regimes dominated by short and long cracks [90]. The fatigue strength for discontinuity-free material is expressed by Equation 4-1.

$$\sigma_a = \sigma_{w,0} \quad \text{Equation 4-1}$$

Where $\sigma_{w,0}$ is the fatigue strength in MPa for stress ratio ‘R’ of -1 and can be estimated from the Vickers hardness (HV) using, $\sigma_{w,0} = 1.6 \text{ HV} \pm 0.1 \text{ HV}$, [71]. For components containing discontinuities, fatigue strength follows Equation 4-2 [90].

$$\sigma_a = \sigma_{FS} = \frac{\Delta K_{th,LC}}{2Y\sqrt{\pi a}} \quad \text{Equation 4-2}$$

where $\Delta K_{th,LC}$ is the stress intensity threshold in $\text{MPa}\sqrt{m}$ for long cracks, a is the crack dimension, σ_{FS} in MPa is the fatigue strength in the presence of discontinuities, and “Y” is a geometric correction.

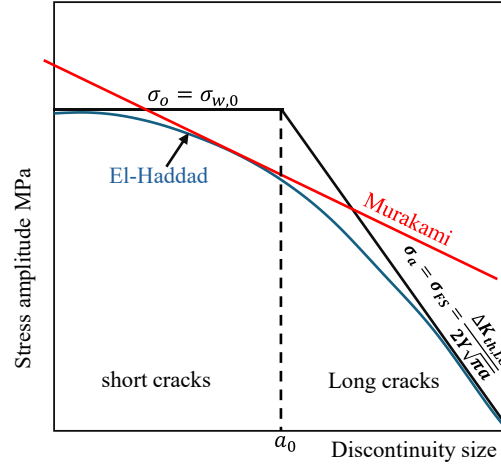


Figure 4-2. A schematic representation of the Kitagawa-Takahashi diagram showing the evolution of fatigue strength with the size of discontinuity.

4.2.2 El-Haddad's model of the KT diagram

Experimental results show a gradual transition between short and long crack behaviour. El-Haddad et al. [11] introduced a characteristic crack length (a_0) defined $a_0 = \frac{1}{\pi} \left(\frac{\Delta K_{th,LC}}{2\sigma_{w,0}} \right)^2$, to describe the transition by Equation 4-3.

$$\sigma_a = \sigma_{FS} = \frac{\Delta K_{th,LC}}{2Y\sqrt{\pi(a + a_0)}} \quad \text{Equation 4-3}$$

The critical crack length is the length at the intersection of Equations 1 and 2, expressed as $a_0 = \frac{1}{\pi} \left(\frac{\Delta K_{th,LC}}{2\sigma_{w,0}} \right)^2$.

4.2.3 Murakami model of the KT diagram

Murakami [133] proposed $\sqrt{area}$ as a representative parameter for discontinuity size and derived a fatigue strength expression for materials containing small discontinuity [133] and the associated stress intensity factor given in Equation 4-4.

$$\sigma_{FS} = \frac{k(HV + 120)}{(\sqrt{area})^{1/6}} \quad \text{Equation 4-4}$$

Where $\Delta\sigma$ is replaced with $2\sigma_{FS}$ in MPa, $k = 1.43$ and 1.56 for the surface and internal discontinuities, and $\sqrt{area}$ in μm . For stress ratios (R) other than -1 , Murakami [134] proposed correcting Equation 4-4 by the factor $\left[\frac{1-R}{2}\right]^\alpha$, where α is a material parameter given as

$$\alpha = 0.226 + HV \times 10^{-4}.$$

From the discussions above, the El-Haddad (Equation 4-3) and Murakami (Equation 4-4) models show a reliance on discontinuity size ($\sqrt{area}$) and material property of microhardness and $\Delta K_{th,LC}$. These models inherently treat discontinuities as preexisting cracks to model the fatigue strength. Nonetheless, their modelling trend differs (as seen in Figure 4-2) and can align differently with experimental results. The Murakami model leverages small discontinuities and does not explicitly cover long crack regimes. However, the El-Haddad model spans the regime of both short and long cracks, making it applicable from small to larger discontinuities. So, it is appropriate to ascertain which of the models is accurate for the newly developed LPBF-M789 alloy, which is likely to contain different discontinuity sizes before practical applications.

Terminology clarification: In this study, the term discontinuities is used uniformly to describe both intrinsic (manufacturing-related) and extrinsic (corrosion pits) features influencing fatigue behaviour. The term “discontinuity” appears only in the titles of cited references, where it reflects the authors’ original terminology.

4.3 Materials and methods

4.3.1 Raw material and as-built samples

Heat-treated M789 alloy specimens used for this research were produced by the voestalpine Additive Manufacturing Centre (vAMC), Mississauga, Ontario, Canada, using the LPBF AM process. Information on the feedstock powder and the building parameters employed during the LPBF build process for parts can be found in our previous work [131]. The chemical composition of the received alloy after part manufacturing is given in Table 4-1, where it is represented as "tested". Glow-discharge atomic emission spectroscopy was used for the measurement, while combustion analysis determined the carbon content. The results are compared to the patent specification of the alloy, where all the constituents except for Mn are in conformance. The Mn value is 0.05 wt %, which is less than the specified range of 0.50-1.50 wt%. Despite the out-of-specification value of Mn, the amount is considered to have no significant impact on this work. Though we note that Mn may be added to steel to favour the formation of MnS inclusions known to improve machinability, sometimes it is desired to be kept to a minimum to reduce the formation of inclusions [135].

Table 4-1: Chemical composition of M789 alloy (in wt%).

Elements	C	Mn	Cr	Mo	Ni	Cu	Ti	Al	Fe
Patent specification [129]	0.01-0.05	0.10-0.50	12.00-13.00	0.50-1.50	9.50-10.50	0.00-0.05	0.50-1.50	0.50-1.50	balance
Tested	<0.02	0.05	12.30	1.11	10.10	0.01	1.07	0.55	balance

Figure 4-3a and b illustrate the layout of the build plate for the samples presented in Table 2. Cubic samples (pink) for metallographic observations and microhardness tests, two tensile (in red) and fifteen fatigue samples (in yellow) were built vertically. Three rectangular bars (in purple) from the build plate in Figure 4-3b are used to machine single-edge bending (SEB) specimen, used for crack growth threshold $\Delta K_{th,LC}$ characterisation. Other samples on the plate (grey, green, blue) were produced for parallel research projects and not presented in this paper. The dimensional drawings of the tensile and fatigue specimens are shown in Figure 4-4 and Figure 4-5, respectively.

Table 4-2: Breakdown of samples used for this research, showing the dimensions, sample type, and quantity.

Sample	Dimension mm	Type	Number	Purpose
Cubic sample	17 x 17 x 17	NA Metallographic polishing	5	Microstructural characterisation and hardness test
Tensile sample	Gauge 6 x 32 x 3	Vertical	2	Tensile tests
		Horizontal	2	
Fatigue sample	Gauge 2.6 x 26 x 11.6	All vertical As-printed (FA)	5	fatigue test
		Polished (FP)	5	
		Polished specimen then corroded (FC)	5	
Rectangular bars	12.7 x 110 x 24	All vertical Polished	2	Estimation of $\Delta K_{th,LC}$ of the M789 alloy

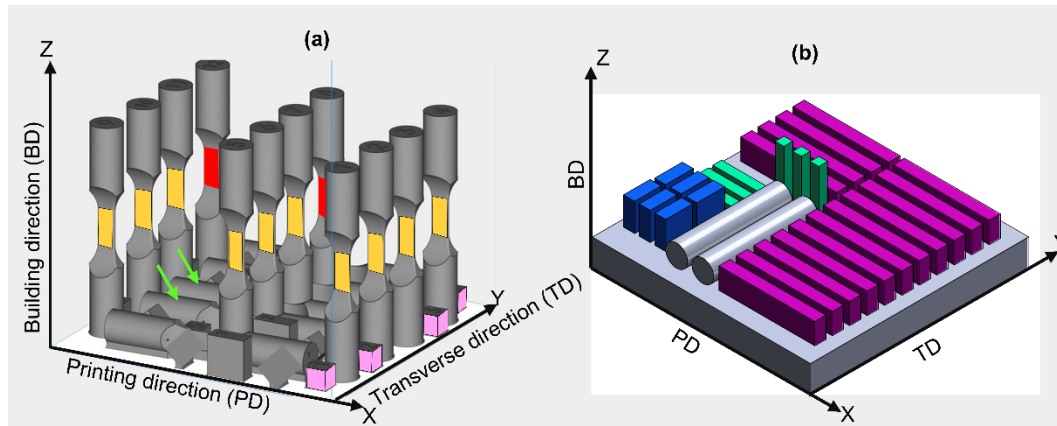


Figure 4-3. Build plate of samples (a) sample layout on build plates with reference guides for the M789 alloy, showing all specimens (b) layout of second build plate

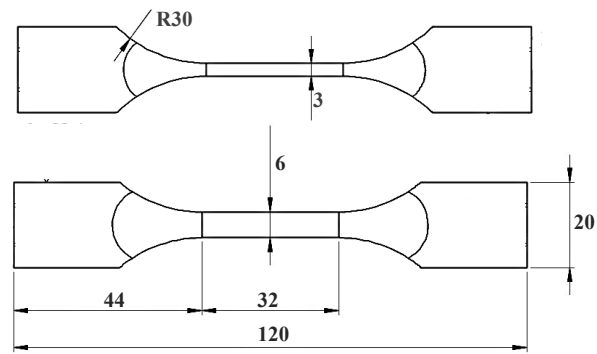


Figure 4-4. Schematic representation of a tensile specimen. All dimensions indicated are in mm.

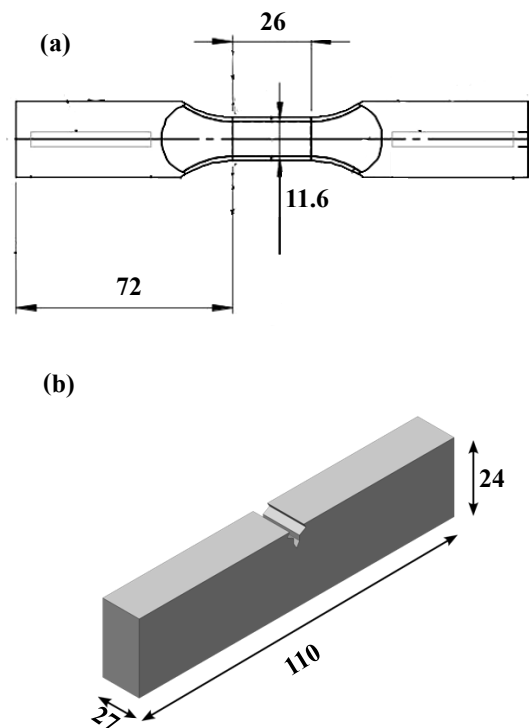


Figure 4-5. Fatigue specimen: (a) axial fatigue specimen, (b) SEB specimen for stress intensity threshold test. All dimensions are in mm.

4.3.2 Material characterisation

The results of material characterisation, excluding the fatigue behaviour, will be presented in the methodology since they are meant to provide general information about the tested material.

4.3.2.1 Microstructure

Specimens meant for microstructural characterisation were cut from the cubes using a Struers Secotom 20 machine, with cuts made in planes perpendicular (XY) and parallel (XZ) to the building direction. The extracted samples were mounted in phenolic resin and subjected to metallographic preparation. This involved sequential grinding with abrasive papers of 240, 400, 600, 800, and 1200 μm grits. Subsequently, the samples were polished employing diamond pastes with particle sizes of 6 μm , 3 μm , and 1 μm .

The polished specimens were etched using a modified Fry's reagent [136], composed of 1 g CuCl_2 , 50 mL HCl , 150 mL H_2O , and 50 mL HNO_3 . Etching was performed via sequential 10-second swabbing over a total duration of 140 seconds to prevent sample burning. After each swabbing step, the samples were rinsed with water and examined under an optical microscope. Figure 4-6 presents the optical micrograph showing a random orientation of the martensitic microstructure with no visible melt pool tracks. This observation aligns with the work of Tian et al. [130], where no melt pool boundaries were visible in heat-treated M789 alloy samples. Discontinuities such as lack of fusion (LOF) and alumina non-metallic inclusions (NMI) were identified in the microstructure and are shown in Figure 4-7. The acquired sizes of discontinuity using optical microscopy from both XY and XZ planes are plotted on a frequency distribution accompanied with cumulative distribution of plot of the exponential function in the secondary axis shown in Figure 4-8. Across both planes, the measured discontinuity sizes predominantly ranged from 5 to 25 μm .

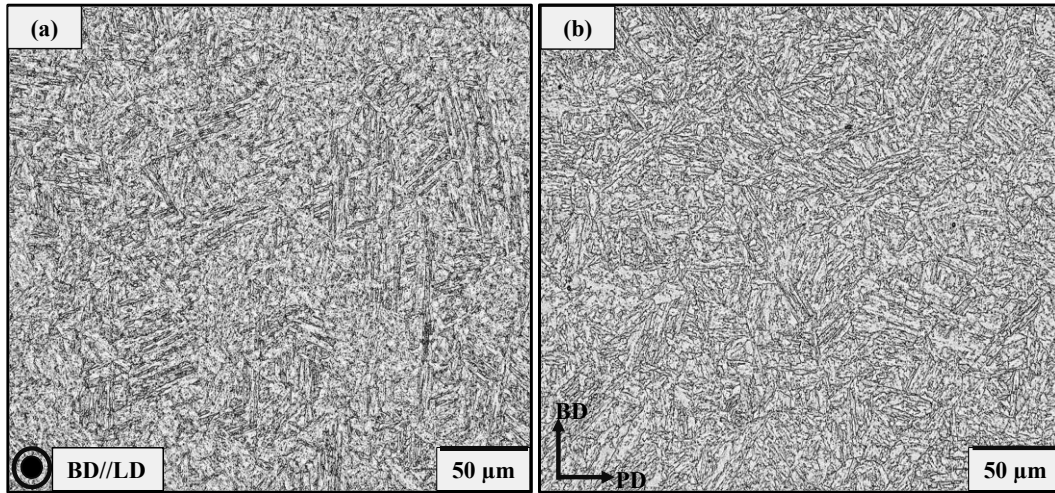


Figure 4-6. Optical microscopic images of the M789 microstructure captured with the Keyence microscope: (a) image of plane (XY) perpendicular to the BD, (b) image of plane (ZX) parallel to the BD.

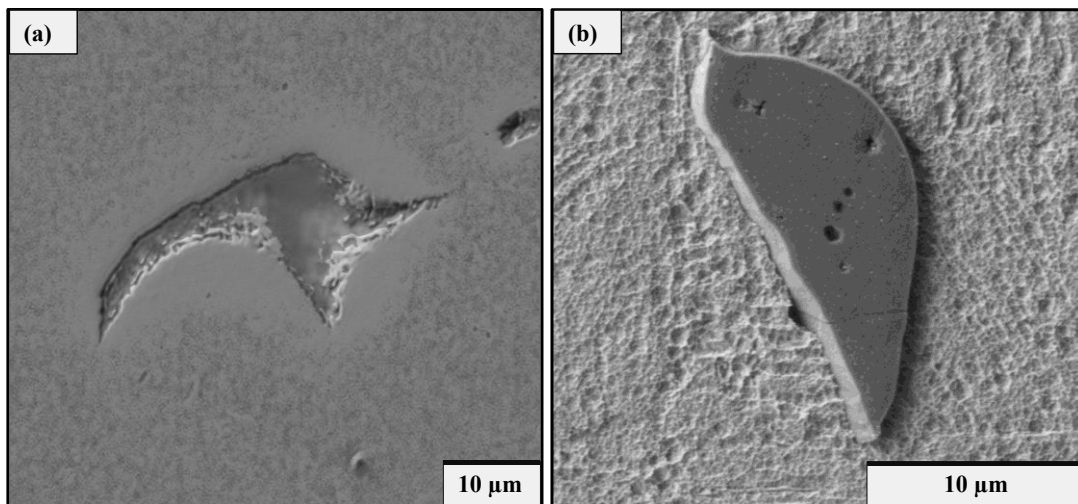


Figure 4-7. Manufacturing discontinuities observed under SEM include (a) LOF and (b) alumina oxide non-metallic inclusion.

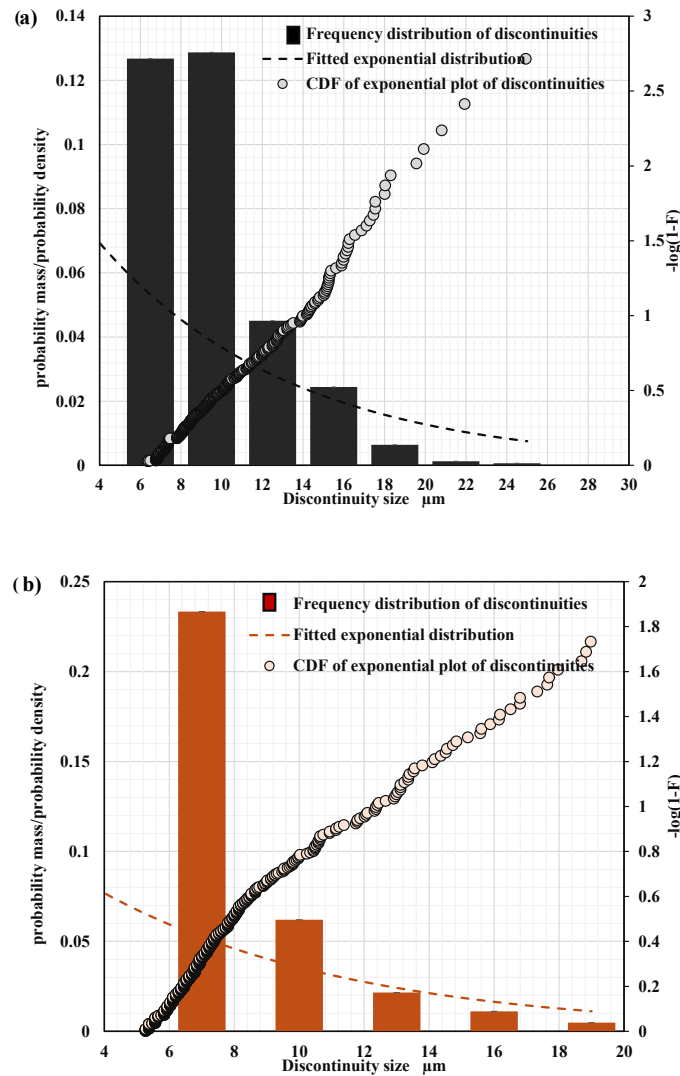


Figure 4-8. Distribution of discontinuities acquired from cubic specimens, (a) XY direction and (b) XZ direction

Revealing prior austenite grain boundaries (PAGBs) in quenched steels is often challenging, particularly in steels with low carbon content [27], and the M789 alloy proved no exception. This measurement was necessary to provide information on the size of the PAG, since it can contribute to the failure mechanism of the alloy. To address this, the oxidation method proposed by Faria et al. [28] was adapted. Metallographically prepared samples were austenitised at 750 °C in a Pyradia electric furnace for 30 minutes. The samples were then air-quenched and lightly polished to remove the oxidised surface layer. They were then etched using Villela reagent, composed of 2 g picric acid, 5 mL HCl, and 100 mL ethanol. Austenitisation induced preferential oxidation along the

austenite grain boundaries, creating topographic heterogeneities on the polished surface. Subsequent etching enhanced the optical contrast of these topographic grooves, allowing a distinction of the PAGB. The Keyence image analysis software was employed to observe and measure the parent austenite grain size (PAGS) in accordance with ASTM E112. An example of the optical image generated with the traced out PAGs is shown in Figure 4-9a, and the frequency plot of the equivalent circle diameter is shown in b. The PAG size ranged from 5 to 70 μm with an average of $22 \pm 10.5 \mu\text{m}$ and an ASTM grain size number "G" of 6.8 ± 0.47 . Notably, 90% of the grains measured were less than 50 μm .

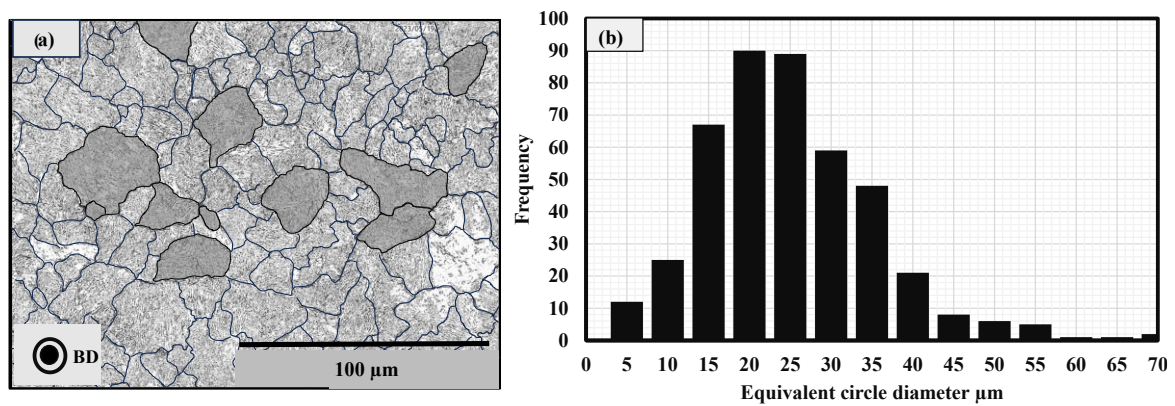


Figure 4-9. (a) Optical image of the PAGB after etching showing the grain boundaries in black, and (b) frequency distribution of the measured PAGS.

4.3.2.2 Tensile test

Four tensile tests were performed, two for each build orientation (horizontal and vertical), were performed according to ASTM E8/E8M [34] using a 100 kN servo-hydraulic machine at a nominal strain rate of 0.008 s^{-1} , with a 25.4 mm extensometer attached to the specimen. The tensile specimen had a rectangular gauge cross-section measuring $6 \times 3 \text{ mm}$ and 32 mm gauge length, as shown in Figure 4-4. The representative engineering stress-strain curves for each orientation from the test are presented in Figure 4-10, and the others are provided in Figure S.1 of the supplementary information (SI). Similar tensile behaviours are seen irrespective of build orientation. The mean ultimate tensile strength and 0.2% yield strength of vertically built specimens were 1800 MPa and 1685 MPa, respectively. Horizontally built specimens exhibited approximately 1.5% higher ultimate tensile strength and less than 0.8% higher yield strength. Additionally, the engineering

strain to failure of horizontally built specimens was approximately 0.006 mm/mm higher than that of vertically built specimens. This difference is attributed to anisotropy inherent to LPBF alloys arising from discontinuities between build layers. Sarkar et al. [137] reported that vertically built LPBF specimens contain lack-of-fusion discontinuities, oriented perpendicular to the loading axis, promoting earlier strain localisation compared with horizontally built specimens, where discontinuities are more often aligned parallel to the loading axis. The present results are consistent with those reported by Tian et al. [13] and place M789 alloy within the class of high-strength steels as categorised by Li et al. [138].

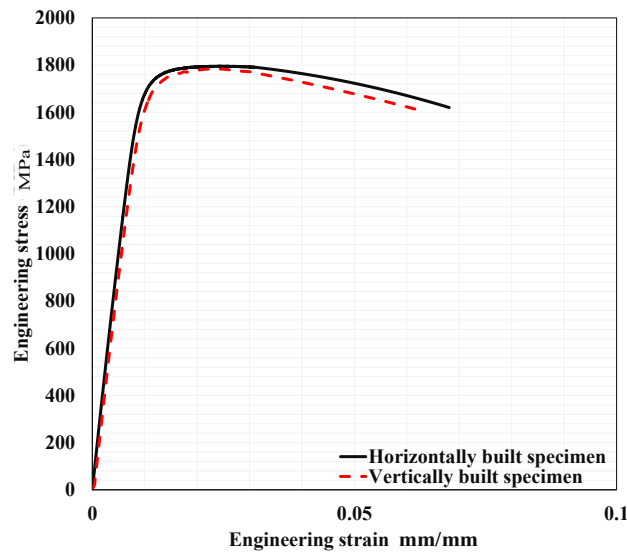


Figure 4-10. Representative engineering stress-strain plot from tensile test.

4.3.2.3 Vickers Microhardness test

Vicker's microhardness test was carried out in accordance with ASTM E384 using a BUEHLER Wilson VH1202 microhardness tester equipped with a diamond indenter. Measurements were performed on polished transverse sections of a cubic specimen. Prior to testing, the instrument's calibration was verified using a calibration reference test block. Indentations were made under a 0.5 kgf load with a dwell time of 10 s, and a total of 15 indents were placed with a minimum spacing of 2.5 times the diagonal length to avoid interaction effects. The average microhardness obtained from these measurements was 557 ± 11 HV.

4.3.3 Fatigue testing

4.3.3.1 Stress intensity threshold test

A single-edge bending (SEB) specimen with dimensions shown in Figure 4-5 was used for the crack propagation test. The fatigue testing process begins with finite element analysis (FEA) to estimate the appropriate load for pre-cracking. Using Abaqus software, the stress-strain behaviour of the material is incorporated into the model to simulate the plastic zone required to initiate a crack, as explained by Carboni et al. [139]. This precaution ensures precise control over crack formation while avoiding unnecessary material damage. Once the load is determined, pre-cracking is performed using a compression-compression method. This approach prevents the formation of a plastic wake or residual stress that could influence subsequent crack propagation, as explained by Tabernig et al. [140]. The pre-crack is carefully grown to a size of approximately 100 μm , using a hydraulic testing machine for precise load control. Then, crack propagation testing is performed following the load reduction technique outlined in ASTM E647 [141] for an R -ratio of 0.1 and a shedding rate of -0.08 mm^{-1} . To accelerate testing without compromising accuracy, a resonant testing machine operating at around 105 Hz was used. The crack growth threshold ($\Delta K_{th,LC,R=0.1}$) was then determined from linear regression on near-threshold data. The regression used data points within the linear portion of the near-threshold regime for crack growth rates below 10^{-9} m/cycle , yielding a threshold value of $4.12 \text{ MPa}\sqrt{\text{m}}$ at 10^{-10} m/cycle . This value is smaller than the $4.42 \text{ MPa}\sqrt{\text{m}}$ theoretical estimate reported by Laliberte-Riverin et al [131]. The procedure is diagrammatically summarised in Figure S.2 of SI.

4.3.3.1.1 Constant amplitude fatigue testing

The fatigue testing protocol by Maxwell and Nicholas [114] was adopted to determine the fatigue strength of the material at a specified number of cycles. In this protocol, specimens are cyclically tested at constant stress amplitude starting from an initial stress amplitude ($\sigma_{a,0}$) lower than the estimated fatigue strength of the specimen for a given number of cycles N_{life} . If the specimen survives N_{life} cycles, the stress amplitude is increased by $\Delta\sigma_a$ and tested again. Each stress amplitude is called a step. The process is repeated n times, until failure occurs before or at N_{life} . Maxwell and Nicholas [114] proposed that once failure occurs, the material fatigue strength is characterised by the stress amplitude of the load step before failure $\sigma_{a,n-1}$, the load increment

$\Delta\sigma_a$ and the number of cycles at failure N_{fail} using Equation 4-5. This methodology assumes that load steps anterior to $\sigma_{a,n-1}$ are not damaging the material.

$$\sigma_{FS} = \sigma_{w,Nlife} = \sigma_{a,n-1} + \Delta\sigma_a \left(\frac{N_{fail}}{N_{life}} \right) \quad \text{Equation 4-5}$$

where $\sigma_{w,Nlife}$ is the fatigue strength N_{life} cycles. This method has two merits: I) it is suitable for use where there is limited availability of samples, and ii) the fatigue strength at a desired number of cycles is determined for each specimen. Thompson et al. [109] conducted a study evaluating the step-loading technique for estimating fatigue strength in crankshaft-grade SAE alloys. They concluded that the method provides reliable fatigue-strength estimates and offers insight into the part's tolerance to discontinuities. Nevertheless, this fatigue testing approach can sometimes induce coxing effects. In this study, such effects are assumed to be negligible because the alloy is a high-strength material, which typically exhibits limited plasticity and minimal cyclic strengthening. This assumption is supported by the findings of Nakajima et al. [142], who reported that coxing effects diminish as material strength increases, with high-strength steels showing only minor or negligible improvements in fatigue resistance. Furthermore, preexisting intrinsic and extrinsic discontinuities are expected to dominate fatigue failure, making it unlikely that prior low-amplitude loading would significantly modify the fatigue response, especially below the endurance limit. Under these conditions, any potential deviation introduced by step loading is expected to be small (on the order of a few percent, $< 5\%$).

This test procedure was applied to 15 fatigue test specimens having the dimensions reported in Figure 4-5. Three different surface conditions were tested as reported in Table 2 and described in the next section. The fatigue tests were performed using a stress ratio of 0.1 on a servo-hydraulic machine equipped with a 100 kN load capacity. A frequency of 20 Hz, a target of $2 \cdot 10^6$ cycles at failure and a stress amplitude increment of 20 MPa were used. The choice of the number of cycles was selected to align with the high-cycle fatigue regime relevant to the intended application of the alloy, while providing reliable characterisation sufficiently beyond the low to high cycle transition region of the S-N curve. In addition, $2 \cdot 10^6$ cycles lie within the validity range of the Murakami model, defining the peak fatigue strength in the KT diagram [143]. Using a target of 10^7 cycles would significantly increase testing time without substantially altering the results of this study.

As with other fatigue testing methods, prior knowledge of the alloy's fatigue strength is required to ascertain the starting stress amplitude in the step loading method. Thus, referencing the work of Laliberte et al. [131], a start stress amplitude of 180 MPa was chosen for the FP specimens and 100 MPa for the FA specimens. As for the FC specimens, 60 MPa was chosen as the starting stress based on the largest observed pit $\sqrt{area}_{max}$ during nondestructive characterisation (discussed in Section 4.3.5.3).

4.3.3.2 Selection of starting stress amplitude for the step loading protocol

A starting stress amplitude of 180 MPa was selected for the FP specimens based on the minimum fatigue strength reported for polished conditions in Laliberté-Riverin et al. [131]. As for the as-printed specimens (FA), no literature data were available at the time of testing, therefore, a starting amplitude of 100 MPa was used, guided by their predicted minimum fatigue strength given in Table 4-3 of Section 4.5. For the pitted FC specimens, a starting stress amplitude of 60 MPa was chosen based on the predicted minimum fatigue strength associated with the largest measured pit size $\sqrt{area}_{max}$ also reported in Table 4-3. Collectively, these starting values were selected to ensure that all specimens began testing at stress levels below their expected fatigue strength and avoid premature failure.

4.3.4 Fractography protocol

After fatigue testing, a fractographic analysis of the failure surface for each specimen was conducted to identify and measure the discontinuity responsible for crack initiation. The fractured surface was separated from the specimen gauge section using the Struers secotom-20 machine for easy manipulation during the microscopic examinations. Each half of the fractured surface was sonicated in acetone with Branson 3800 equipment for 15 minutes to clean the surface of any organic material and impurities. The methodology used to measure the $\sqrt{area}$ of the discontinuities is presented in Figure 4-11, showing 4 typical cases.

Case 1: tracing contours around the absolute shape of the discontinuity (LOF, NMI) found at the failure origin of polished or pitted specimens and computing the $\sqrt{area}$.

Case 2: tracing smooth contours to enclose the shape of the discontinuity (LOF, NMI) and its associated ligaments (a_1 and a_2) between it and the free surface. This applies to both polished and

pitted specimens when the discontinuity is not open to the surface. The $\sqrt{area}$ is computed using Equation 4-6.

$$\sqrt{area} = \sqrt{area + a_1 + a_2} \quad \text{Equation 4-6}$$

Case 3: failures from closely situated discontinuities are considered as a single discontinuity if the distance of separation is less than the size of the smallest discontinuity [125]. Applicable to polished and pitted specimens. The entire area in the red dashed line is used for the computation of the $\sqrt{area}$. This applies to the pitted specimen.

Case 4: failures from troughs for as-printed specimens measured by determining the depth of the trough found at the origin and treating it as a shallow, slender crack using the equation $\sqrt{area} = c\sqrt{10}$, where c is the depth.

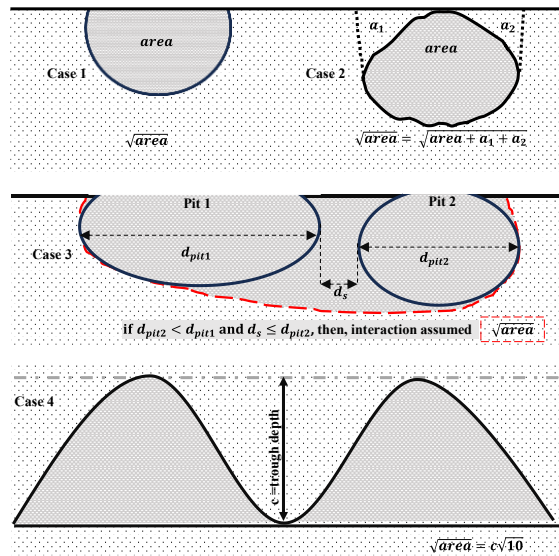


Figure 4-11. Schematic description of the measurement of the discontinuity at the failure origin. Cases 1 and 2, surface and subsurface discontinuity at fracture origin. Case 3 closely situated discontinuities at the fracture origin. Case 4, failure from troughs of as-printed surfaces.

4.3.5 Surface preparation and non-destructive characterisation (NDC)

4.3.5.1 Polished specimens (FP)

A hand rotating tool was used to polish the 4 surfaces in the gage length and fillet areas of 10 fatigue specimens. Special attention was given to rounding the edges, as crack initiation from these features can influence fatigue test results. Besides the tool used to polish, the polishing protocol described in Section 4.3.2.1 was adopted. Five of these polished specimens were tested as-polished (FP), and the other five were submitted to the corrosion protocol described in Section 4.3.5.3.

For the FP specimens, it is expected that the fatigue crack will initiate from natural discontinuities (internal or opened to the surface) such as LOF and NMI. Based on this assumption, the largest discontinuities observed during metallographic observation are critical. Ten regions, from the metallographically prepared cubic surface, each having an area of 20 mm^2 were inspected, and the $\sqrt{\text{area}}$ of the largest discontinuity in each region was recorded. An image of the acquisition is given in Figure B- 4 of SI. The quantification was performed automatically under an optical microscope at a magnification of X200 using colour thresholding in-built image analysis software. These measurements will later be used in extreme value analysis (in Section 4.5) for the prediction of the largest discontinuity expected to cause failure in the fatigue specimen.

4.3.5.2 As-printed specimens (FA)

For the FA specimens, no surface preparation was performed; they were used as manufactured. However, fatigue cracks are expected to initiate from surface roughness, either alone or in combination with microstructural discontinuities such as lack of fusion (LOF) or non-metallic inclusions (NMI). Therefore, surface roughness was characterised using an optical microscope with 3D reconstruction at $200\times$ magnification, a $1.5 \text{ }\mu\text{m}$ height step, and auto-stitching to generate images of $1.4 \text{ mm} \times 1.1 \text{ mm}$ per tile. To cover the entire gauge length, ten-line profiles, parallel to the loading direction (BD/Z axis), as shown in Figure 4-12 (for FA01), were extracted from each as-printed fatigue specimen. In accordance with ASTM F3624[144], a cutoff length of 2.5 mm was applied, and the maximum valley depth ($R_{v,max}$) was determined over five sampling lengths (SL#) for each line. By this protocol, ten $R_{v,max}$ is acquired for each FA specimen. The acquired $R_{v,max}$ values were transformed to $\sqrt{\text{area}}_{max}$ with the assumption of case 4 described in 4.3.4 where $C = R_{v,max}$.

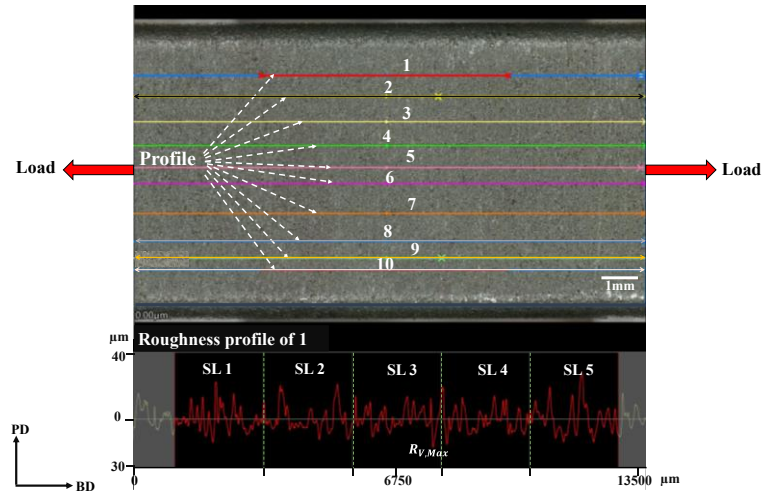


Figure 4-12. Image of the acquisition of the roughness profile for specimen FA01.

4.3.5.3 Corroded specimens (FC)

Pits were introduced onto FP specimens using an in-house-designed electrochemical cell. A 1 cm² area on the specimen's gauge length was isolated for pitting using Gamry masking tape, after which the remaining region was coated using a 3M electrical insulating sealer and left to dry for 15 minutes. Figure4-13a shows a coated fatigue specimen.

Subsequently, the specimen was affixed to the base of the electrochemical cell, sealed with a silicone coating (Figure4-13b) and allowed to dry for 48 hours. When dried thoroughly, the Gamry tape was carefully removed to expose the 1 cm² region. Next, the electrochemical cell was assembled and filled with the 3.5% NaCl solution to a level sufficient to submerge the exposed surface (Figure4-13c). The resulting electrochemical setup consisted of the fatigue specimen (exposed surface in the electrolyte) acting as the working electrode (WE), a platinum wire as the counter electrode (CE), and an Ag/AgCl as the reference electrode (RE). Then, with a computer-controlled Biologic SP-150 potentiostat, a cyclic potentiodynamic polarisation (CPP) technique was employed to induce pitting, using a scan rate of 0.167 mV/s within the range of -1 V to 1 V relative to the open circuit potential. The scan was allowed to complete automatically, and the resulting CPP plot is presented in Figure B- 3 of Appendix. The process was then repeated on another 1 cm² area of the specimen's surface. Five polished specimens were pitted using this technique.

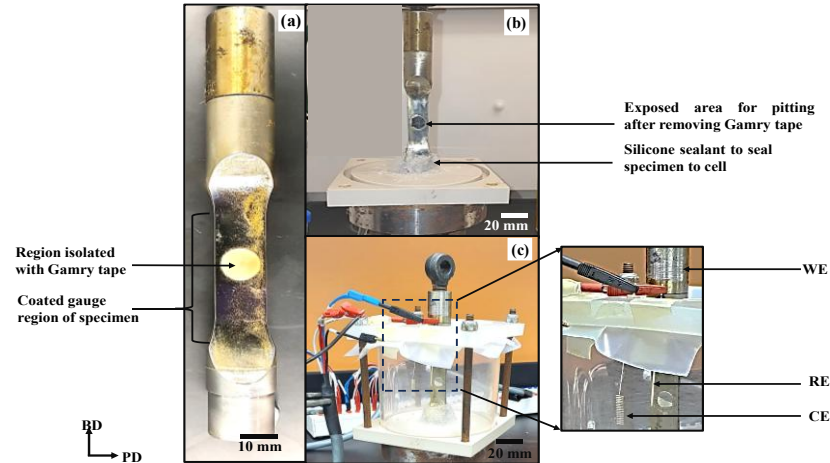


Figure4-13. (a) Coated specimen showing the isolated area (b) Specimen mounted on the electrochemical cell and sealed at the base using silicone (c) Coupled electrochemical cell showing the WE, RE and CE.

The corrosion pits produced are anticipated to be the critical feature from which the fatigue crack will initiate. Therefore, the largest pit on each specimen was characterised using an optical microscope with 3D reconstruction capability to acquire the $\sqrt{area}_{max}$. An example of the two-step procedure is given in Figure 4-14. Firstly, an inspection of the surface with the microscope is carried out to identify the largest pit, and a 3D reconstructed image of the pit is acquired. Figure 4-14a shows the 3D image of the largest pit acquired with a depth composition with a step of 1.5 μm . Height colouration mapping helped to identify the deepest point (dark blue) of the pit, as labelled in the image. Then, a section line A-A cutting through the deepest point and perpendicular to the loading axis or BD of the specimen is used to help define the $\sqrt{area}$ critical to fatigue failure of the specimen. Figure 4-14b shows a 2D representation with the sectioned line A-A (in black dashed line), the $\sqrt{area} = 421 \mu\text{m}$ and pit depth is 224 μm . The process was repeated for all the corroded samples (FC01-FC05) and the $\sqrt{area}_{max}$ for each was acquired and reported in Table 4-3.

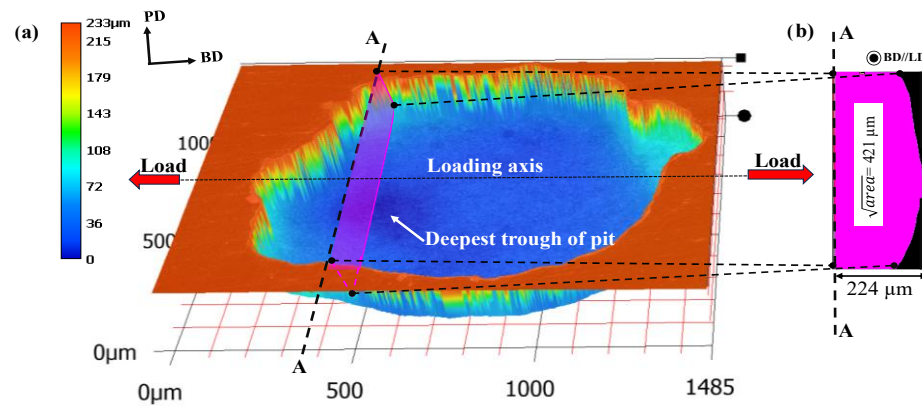


Figure 4-14. Characterisation of severe pits using the Keyence microscope for specimen FC03. (a) 3D colour mapping image of the largest pit (b) 2D image of the characterised.

Representative images of the specimens after undergoing the surface preparations are shown in Figure 4-15. The $\sqrt{area}_{max}$ of the discontinuity sizes acquired from the metallographic, as-printed (for specimen FA01) and the corroded specimens are shown in Figure 4-16. A Gumbel's probability density function (PDF) is fitted to the $\sqrt{area}_{max}$. From the diagram, the $\sqrt{area}_{max}$ of internal discontinuities in metallographic specimens ranged from 8 to 27 μm, which is less than the 47 to 62 μm of surface discontinuities estimated for the as-printed (FA01) specimen and the 182 to 409 μm for the corroded specimen (FC). Thus, with the size difference and surface condition, it can be anticipated that pits will most probably be the critical discontinuity across all specimens. These results demonstrate that testing the three different surface conditions allows studying a wider size range of discontinuities.

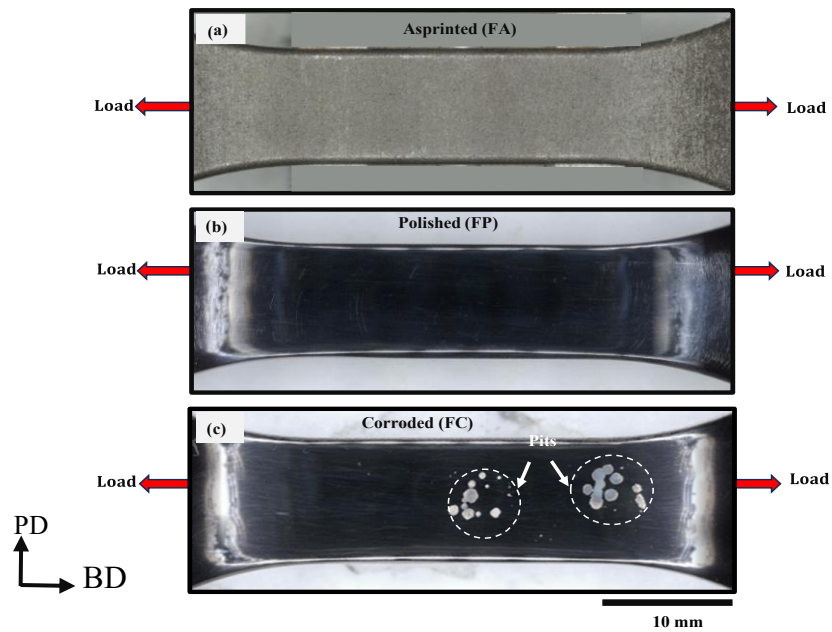


Figure 4-15. Optical image of the Fatigue specimens after preparation, (a) as-printed specimens (FA), (b) polished specimens (FP), (c) corroded specimens (FC) with pits as indicated by the white arrows.

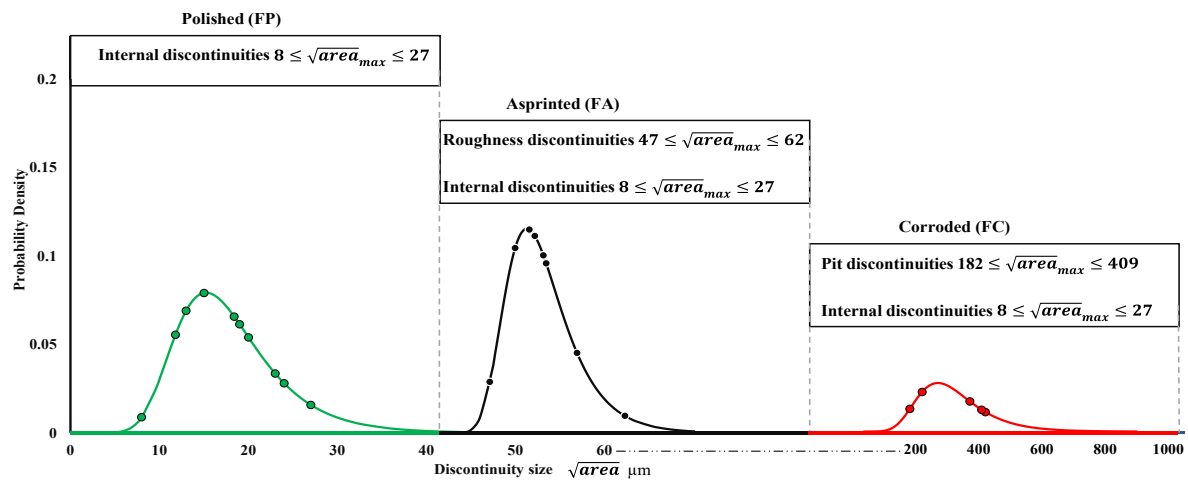


Figure 4-16. PDF Gumbel fit discontinuity size comparison for as-printed, polished and corroded surface conditions.

4.4 Generic KT diagram of M789 alloy

To construct the KT diagram of the M789 alloy for a stress ratio of 0.1 and $2 \cdot 10^6$ cycles, we require the fatigue strength of the alloy in the discontinuity-free or pristine state and the varying stress intensity threshold of the alloy $\Delta K_{th,LC}$. The values are respectively input to the models given by Equation 4-1 and Equation 4-2, after which the plot shown in Figure 4-2 is generated for the alloy. The author has chosen to draw the KT diagram for a Y value of 0.65, which is associated with failure from the surface as estimated by Murakami [72]. This approach has been adopted because all the specimens failed from the surface, as will be seen in Section 4.6, where the results are presented.

4.4.1 Fatigue strength of discontinuity-free or pristine M789 alloy

Laliberte-Riverin et al. [131] proposed a KT diagram for M789 using theoretical estimates of the $\Delta K_{th,LC}$ at $R = 0.1$. Here, we establish the diagram using experimentally obtained material properties of hardness and the stress intensity threshold range at $R = 0.1$ ($\Delta K_{th,LC,R=0.1}$).

Firstly, the hardness of the M789 alloy determined in Section 4.3.2.3 is used to estimate the discontinuity-free fatigue strength “ $\sigma_{w,0}$ ” which is calculated as 880 MPa using the relation $\sigma_{w,0} \cong 1.6 HV \pm 0.1 HV$ for $R = -1$. Thereafter, Goodman’s [145] equation is used to estimate the fatigue strength, where S_u is the tensile strength and $R=0.1$ as;

$$\sigma_{w,0.1} = \frac{1}{\frac{1}{\sigma_{w,0}} + \frac{1+R}{S_u(1-R)}} \quad \text{Equation 4-7}$$

The computed value of $\sigma_{w,0.1}$ was 550 MPa and is intrinsically related to the microstructure of the material in the absence of discontinuities. This value of $\sigma_{w,0.1}$ represent the horizontal line of the KT diagram.

4.4.2 The construction of the KT diagram of the M789 alloy

The $\Delta K_{th,LC,R=0.1}$ and $\sigma_{w,0.1}$ are employed to construct the KT diagram of the M789 alloy shown in Figure 4-17. Also for comparison, the prediction models proposed by El-Haddad (Equation 4-3) and Murakami (Equation 4-4) are superimposed on the diagram. The value of the El-Haddad

parameter $\left(a_0 = \frac{1}{\pi} \left(\frac{\Delta K_{th,LC}}{2Y\sigma_{W,0}} \right)^2\right)$ is 10.58 μm which is less than the 15.1 μm for the intersection point of LEFM and the Murakami model.

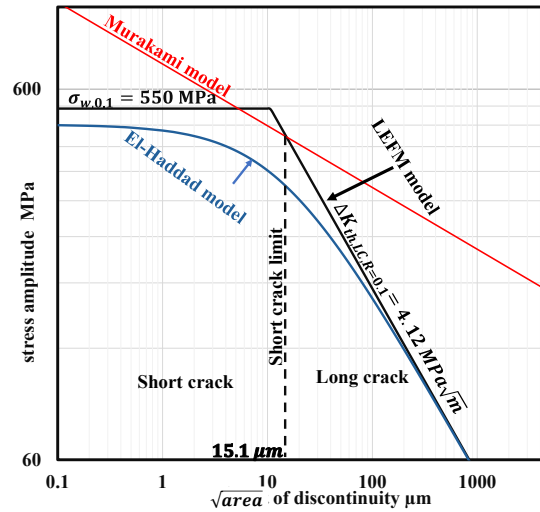


Figure 4-17. The KT diagram of the M789 alloy was developed based on experimentally obtained hardness and $\Delta K_{th,LC}$ of the M789 alloy.

4.5 Fatigue strength prediction

To predict the fatigue strength, the largest discontinuity expected for each specimen condition was first estimated. The estimated discontinuity size was then entered into the El-Haddad Equation 4-3 to obtain the predicted fatigue strength. The following subsection describes how the largest discontinuity was determined for each specimen condition.

4.5.1 FP specimens

EVS were applied to the measured $\sqrt{area}_{max}$ values of internal discontinuities obtained from the metallographic characterisation in Section 4.3.5.1 to predict the largest expected discontinuity in FP specimens. The value obtained was then used to estimate the fatigue strength of the FP specimen. The inspection area $A_0 = 20 \text{ mm}^2$ of the metallographic specimen is multiplied by the average size of the discontinuities to yield the inspection volume V_0 and the return period “ T ” is computed using Equation 4-8.

$$T = \frac{V_{ref}}{V_0} \quad \text{Equation 4-8}$$

Where V_{ref} is the fatigue specimen's gauge volume. The maximum likelihood method, as described in ASTM E2283[146], is used to estimate the parameters of the EVS, and Equation 4-9 computes the expected largest discontinuity in the gauge volume of FP specimens.

$$\sqrt{area}_{max} = my + c \quad \text{Equation 4-9}$$

$$y = -\ln\left(-\ln\left(\frac{(T-1)}{T}\right)\right)$$

y is the reduced variate, m and c are the EVS distribution parameters, $\sqrt{area}_{max}$ is the expected largest discontinuity in the FP specimen, and T is the return period. A graphical illustration of the prediction of the EVS prediction is shown in Figure 4-18. Only one prediction is performed for FP specimens using the metallographic data. The scheme of the EVS has been detailed in our previous publication [131].

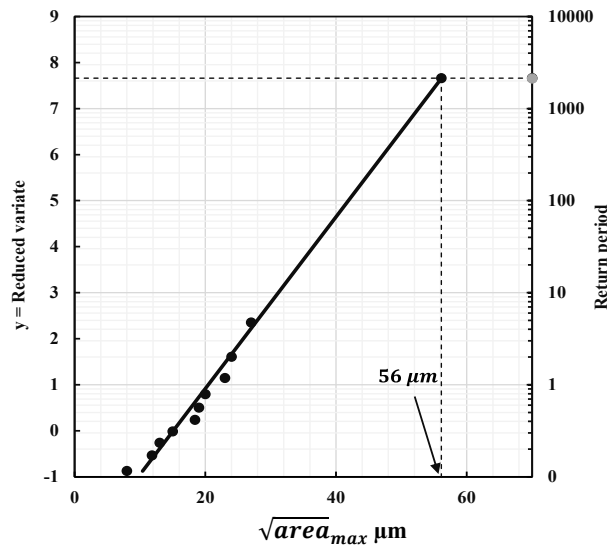


Figure 4-18. $\sqrt{area}_{max}$ values of metallographic discontinuities on cubic surfaces and the prediction of the largest $\sqrt{area}_{max}$ in FP specimen.

4.5.2 FA specimens

Similarly, for the FA specimens, EVS was applied to the measured $R_{v,max}$ values from the roughness characterisation in Section 4.3.5.2. The gauge section of the specimen served as the reference area “ A_{ref} ”, while each profile multiplied by the average $R_{v,max}$ served as the inspection area A_0 used in the computation of T . The $R_{v,max}$ obtained from the prediction is subsequently used to estimate the $\sqrt{area}_{max}$ per Equation 4-10.

$$\sqrt{area} = R_{v,max} \sqrt{10} \quad \text{Equation 4-10}$$

A graphical illustration of the prediction for FA01, chosen as a representative, is shown in Figure 4-19, and the remaining four are presented in the SI.

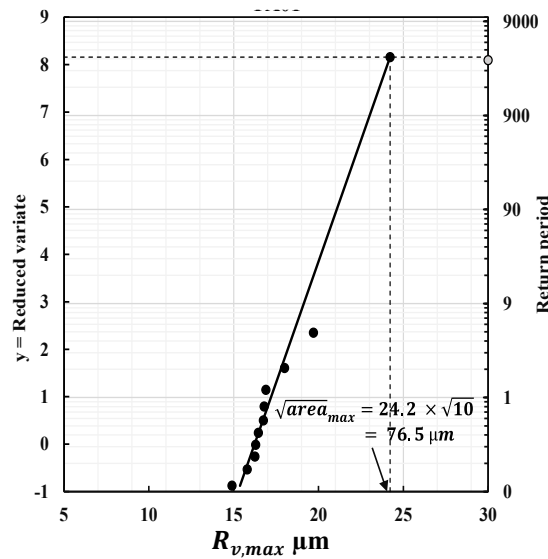


Figure 4-19. Plot of EVS of the $R_{v,max}$ values for the prediction of the expected largest $R_{v,max}$ for specimen FA01. $R_{v,max}$ is converted to $\sqrt{area}$.

4.5.3 FC specimens

In the case of the FC specimens, the computed $\sqrt{area}_{max}$ value from the characterisation of the largest pit on the corroded surface in Section 4.3.5.3 was directly used to estimate the fatigue strength of each FC specimen.

Using the results of EVS predictions for FP and FA specimens and the characterised largest pits in FC specimens, the fatigue strengths were predicted using the El-Haddad model in Equation 4-3, and the results are reported in Table 4-3.

Table 4-3: Summary of the results obtained from the NDC prediction of the fatigue strength (in amplitude) for FP, FA and FC specimens.

Specimen ID	Predicted $\sqrt{Area}_{max}$ μm	Predicted fatigue strength MPa	Comment
FP01	56	219	Discontinuity measured on a metallographic specimen.
FP02			
FP03			
FP04			
FP05			
FA01	77	192	$\sqrt{area}$ predicted using $R_{v,max}$ of roughness values for each fatigue specimen. See Section 4.3.5.2.
FA02	126	153	
FA03	122	155	
FA04	162	136	
FA05	96	173	
FC01	182	129	$\sqrt{area}$ measured from the largest pit on the pitted specimen.
FC02	372	91	
FC03	421	86	
FC04	221	117	
FC05	309	100	

4.6 Results

4.6.1 Results of the fatigue test

The fatigue test results obtained using the step loading protocol are presented in Figure 4-20 and reported in Table 4-4. The results show that the specimens experienced different numbers of steps before failure, and this is attributed to the inherent fatigue scatter caused by variation in discontinuity size, geometry and distribution across the specimens. The results are rounded to the nearest whole number and ordered in increasing stress amplitude at failure for each specimen condition. All crack initiation sites observed are located at the surface (S) or subsurface (SS), irrespective of the surface condition. The last failure stress level for each specimen is reported in the $\sigma_{a,n-1}$ column and the number of cycles sustained at failure is reported in column N_{fail} . The computed fatigue strength using Equation 4-5 is reported in the 4th column. Representative images of the fractographic observation are presented in Figure 4-22, and the discontinuity sizes are reported in the column of $\sqrt{area}$.

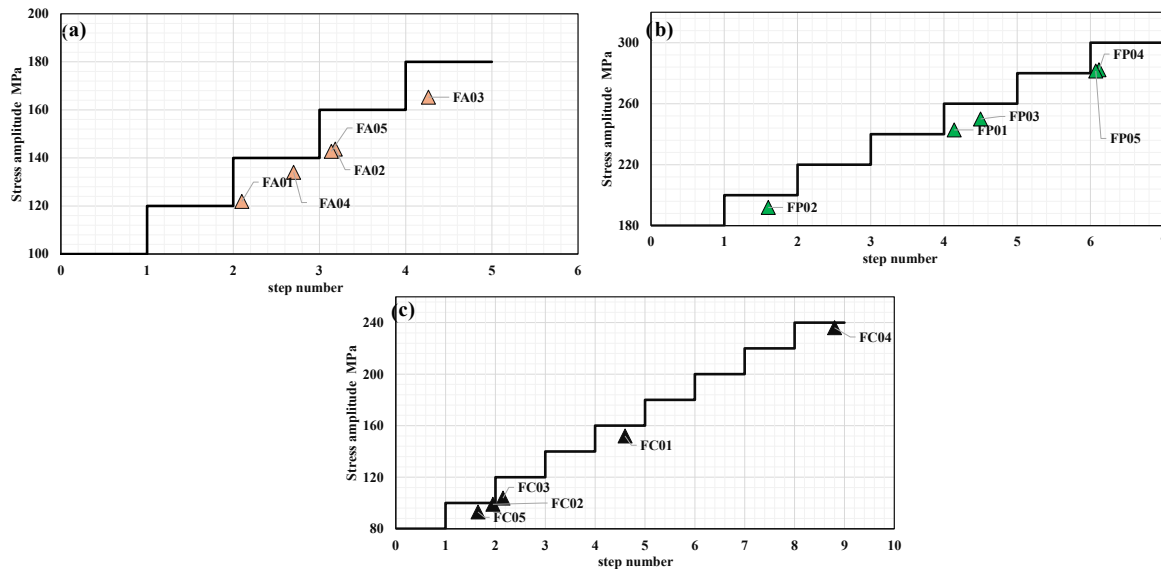


Figure 4-20. Steploading protocol applied to the fatigue specimens showing the survived steps before failure. (a) as-printed-FA specimens, (b) polished-FP specimens, and (c) corroded-FC specimens.

Table 4-4. Fatigue test results obtained for all specimens showing the computed fatigue strength, size type and location of discontinuities at the failure origin.

ID	$\sigma_{a,n-1}$ MPa	N_{fail}	$\sigma_{FS} = \sigma_{w,Nlife}$ MPa	$\sqrt{area}$ μm	Type discontinuity	of Location
FP02	180	$1.2 \cdot 10^6$	192	59	LOF	SS
FP01	240	$2.0 \cdot 10^5$	242	46	LOF	SS
FP03	240	$1.0 \cdot 10^6$	250	51	NMI	SS
FP05	280	$1.0 \cdot 10^5$	281	44	NMI	SS
FP04	280	$2.0 \cdot 10^5$	282	42	LOF	SS
FA01	120	$2.0 \cdot 10^5$	122	117	Trough	S
FA04	120	$1.4 \cdot 10^6$	134	95	Trough	S
FA05	140	$3.0 \cdot 10^5$	143	170	Trough	S
FA02	140	$4.0 \cdot 10^5$	144	158	Trough	S
FA03	160	$5.0 \cdot 10^5$	165	202	Trough	S
FC05	80	$1.3 \cdot 10^6$	93	260	Pit	S
FC02	80	$1.9 \cdot 10^6$	99	361	Pit	S
FC03	100	$3.0 \cdot 10^5$	103	643	Pit (combined)	S
				465	Pit(isolated)	
FC01	140	$1.2 \cdot 10^6$	152	123	LOF	SS
FC04	220	$1.6 \cdot 10^6$	236	226	Pit	S

4.6.2 Fatigue performance of the M789 alloy

The fatigue performance of the M789 at $2 \cdot 10^6$ cycles with respect to the studied specimen groups (FA, FP and FC) is illustrated using the chart in Figure 4-21. The coloured background bands represent different performance regimes relative to the peak fatigue strength (PFS) of 550 MPa determined in Section 4.4.2. The diagram also includes the fatigue grade classification proposed by Murakami [147], which provides a common reference framework for interpreting fatigue performance in AM materials. None of the specimens was able to approach the peak fatigue strength of 550 MPa, as they are all below the PFS. The FP specimens were the highest in performance (192 to 282 MPa), reaching the 30-70% band of the PFS, hence spanning fatigue grades 2 and 3. In contrast, the respective fatigue strength of FA (122 to 165 MPa) and FC (103 to 226 MPa) specimens falls within the lower range of roughly 10–30 % of the PFS. Although we note that one FC specimen was in the 30 – 50 % band but it shall be discussed in Section 4.8.1.4. Overall, the chart highlights the clear differences in fatigue performance among the three surface conditions of the alloy. The next section will focus on the observed discontinuities at the failure origin.

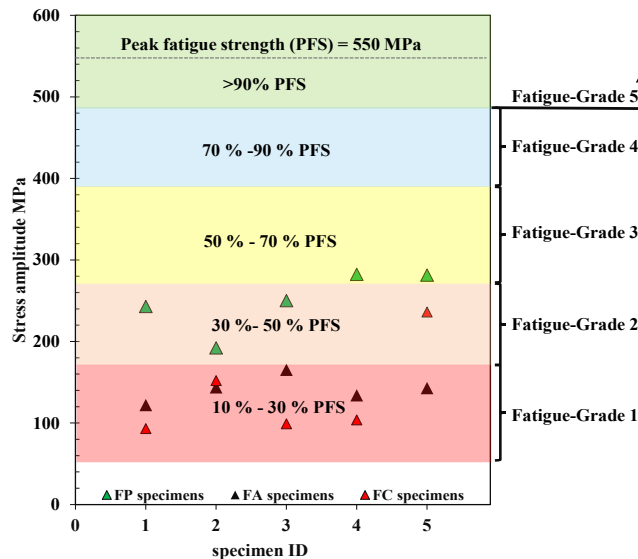


Figure 4-21. Illustration of the fatigue performance of the group of specimens relative to the peak fatigue strength of the M789 alloy.

4.7 Fractography

Representative fractographic images of the discontinuities observed at the origin of FP and FA specimens are shown in Figure 4-22.

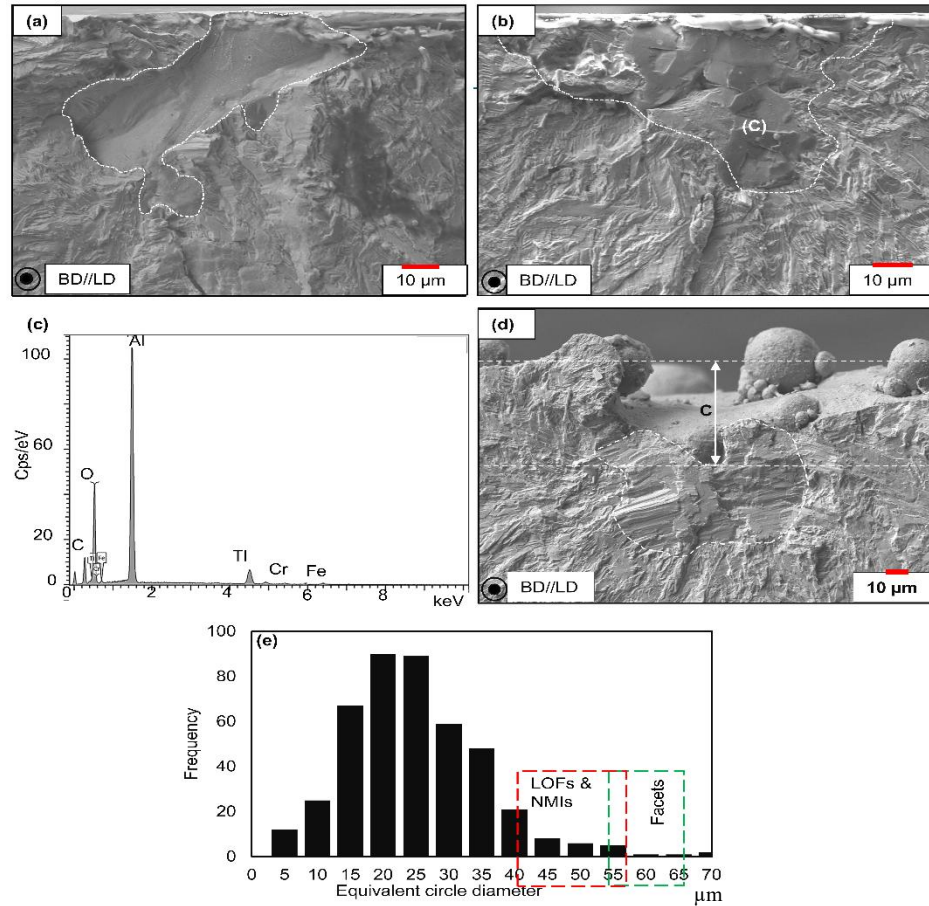


Figure 4-22. Fractography showing (a) LOF at the failure origin of FP01, (b) NMI at the failure origin of FP02, (c) EDS results of the NMI in b, (d) a failure from a trough from an as-printed surface for FA05, (e) size overlap of LOFs, NMI and facets on the PAGS. Other facet images are shown in Figure B- 6 of Appendix B.

The origins were characterised by the presence of either LOF (Figure 4-22a) or NMI (Figure 4-22b) situated at the surface, highlighted with the red dashed lines. NMIs accounted for the failure of 2/5 of the FP specimens with $\sqrt{\text{area}}$ of 44 to 51 μm. They exhibited irregular shape and have an alumina-based stoichiometry, as seen in the EDS result in Figure 4-22c. The remaining failures were triggered by LOFs in the FP specimen, with a size range of 42 to 59 μm.

On the other hand, as-printed specimens failed from the surface asperities (troughs) as seen in Figure 4-22d. The computed $\sqrt{area}$ using Equation 4-10 with ‘c’ is the trough depth for the ranged from 95 to 202 μm . The white arrowed line in Figure 4-22d measures the depth (c) of the trough taken from the surface. The fatigue strength ranged from 122 to 165 MPa. Some as-printed specimens had crystallographic facets at the failure origin with $\sqrt{area}$ facet size of 62-69 μm . Evidence of a facet-like feature (marked in white, broken dashed line) is seen beneath the trough in Figure 4-22d. Other fractographic images showing facets are presented in Figure B- 6 of Appendix B. In Figure 4-22e, we can see that the size of LOF, NMI and crystallographic facets are comparable in size to parent austenite grain sizes of 40-67 μm .

To unveil the underlying features of the LOF and provide an idea of its geometry, focused ion beam (FIB) milling was used to mill out $15 \times 10 \times 5 \mu\text{m}^3$ material from the failure origin of one of the LOFs at the fractographic origin illustrated in Figure 4-23. The LOF (Figure 4-23a) marked in red dashed marks was milled on the X-X' section shown in Figure 4-23b and c, while Figure 4-23d is a magnified image of the milled region. The LOF exhibits a flat and irregular shape, while the FIB-milled region reveals the presence of interconnected subsurface LOF networks. These underlying micro-LOFs can contribute to the severity of the discontinuity, causing failure.

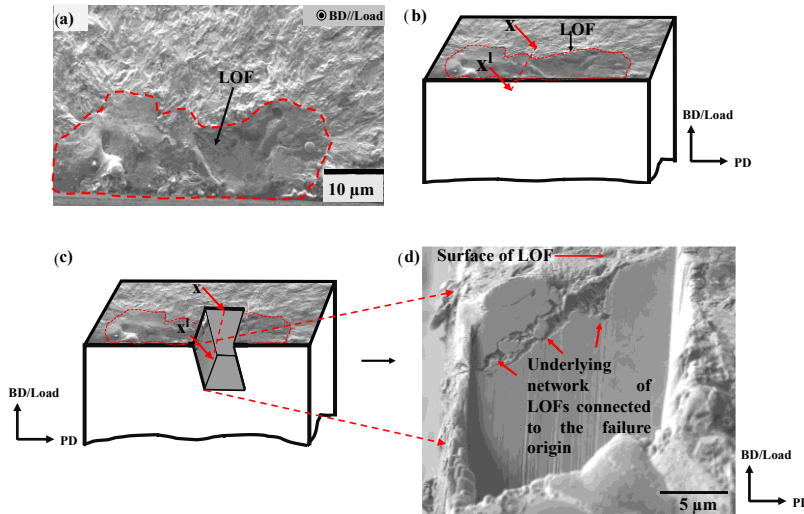


Figure 4-23. Image of sectioned LOF (a) fractographic image of LOF (b) schematic illustration of the identified section to be milled (c) representation of the milled LOF and (d) region of milled LOF showing the underlying LOF network.

The pitted specimens (FC) mostly failed from the pit, except for FC01, which failed from an LOF. Their fatigue strength ranged from 103 to 226 MPa, while the computed $\sqrt{Area}$ of the pit at the failure site was in the range of 123 to 643 μm . The representative fractography shown in Figure 4-24 for specimen FC03 reveals the existence of two pits (named Pit 1 and Pit 2) close to each other at the failure origin. The pits have an area of 376 μm and 465 μm , respectively. Murakami [72] proposes to calculate an overall area covering the pits and the connecting ligament when the distance (100 μm) between discontinuities is smaller than the size of the smallest discontinuity (376 μm). The application of Murakami's method would lead to a critical discontinuity size of $\sqrt{Area} = 643 \mu\text{m}$.

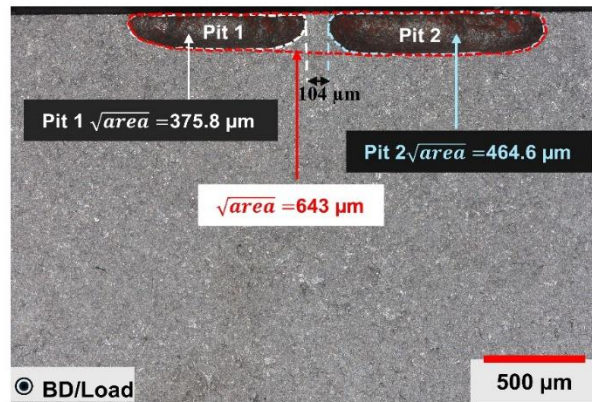


Figure 4-24. Failure from two closely situated pits at the failure origin of pitted specimen FC03.

FC01 is an exception among the pit specimens as it failed from an LOF having $\sqrt{area}$ of 123 μm , despite the presence of an electrochemical pit of $\sqrt{area}$ 182 μm . Figure 4-25a presents an optical image of the largest pit identified before fatigue testing. Figure 4-25c and d show the specimen after failure, with the pit remaining unaffected and the fractographic origin marked by a lack of fusion (LOF) encircled in red dashed lines, respectively.

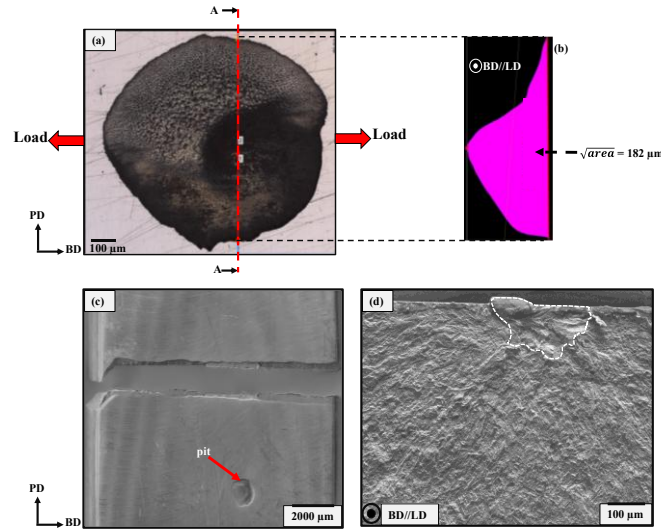


Figure 4-25. Specimen FC01 with (a) the most severe pit on the surface, (b) section A-A of the pit, (c) after fracture, optical image of the specimen with failure away from the pit, (d) LOF discontinuity observed at the failure origin instead of the specimen.

4.7.1 Presentation of the fatigue results on the KT diagram

The fatigue test data and the computed $\sqrt{Area}$ of discontinuities at the fracture origin of all specimens are presented on the KT diagram in Figure 4-26.

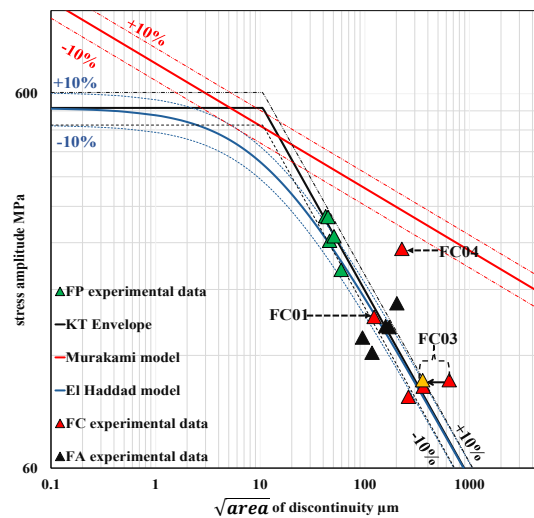


Figure 4-26. KT diagram of the M789 alloy with the superimposition of experimental data of all specimens.

Except for specimen FC04, the fatigue strength is well predicted by LEFM applicable to long cracks. The observed discontinuities range from 42 μm to 643 μm . For FC03, which failed due to two pits at the origin, its fatigue strength is represented using the combined (ligament-inclusive) $\sqrt{Area}$ of both pits (red data point), as recommended by Murakami [72] and the $\sqrt{Area}$ of the larger pit (brown data point). The use of the larger pits $\sqrt{Area}$ (464.3 μm) yields a more precise prediction using LEFM, whereas using the Murakami's approach of combined $\sqrt{Area}$ (643 μm) underestimated the fatigue strength. Notably, FC04 exhibits a fatigue strength higher than predicted by LEFM but lower than Murakami's estimate. Possible reasons for this behaviour are discussed in Section 4.8.1.6.

4.7.2 Comparison between NDC predictions and fracture data

The comparison between the NDC characterisation (see Table 4-3) for discontinuity size and fatigue strength is illustrated in Figure 4-27a and b, respectively. The dashed 1:1 line represents perfect prediction accuracy, data above it indicates overprediction, and those below indicate underprediction.

In Figure 4-27a, for size comparison, overprediction corresponds to conservative estimates, while underprediction represents non-conservative ones. The results reveal a conservative trend for polished (FP) specimens, but non-conservative predictions for as-printed (FA) and pitted (FC). The average differences between predicted and experimental discontinuity sizes were $7.6 \pm 6.8 \mu\text{m}$ (FP, conservative), $-31.9 \pm 59.0 \mu\text{m}$ (FA, non-conservative), and $-32.3 \pm 107.9 \mu\text{m}$ (FC, non-conservative). The negative sign denotes underprediction. The FC03 outlier skewed the FC's mean because of the assumed pit interaction (red data point in Figure 4-27a). When this assumption is removed and only the largest pit is considered (brown data point in Figure 4-27a), the prediction becomes conservative with a mean difference of $3.3 \pm 32.9 \mu\text{m}$. Interestingly, the pit size for FC04 was accurately predicted despite appearing as an outlier in the KT diagram (see Figure 4-26), suggesting that the fatigue resistance model lacked precision in estimating fatigue strength.

For fatigue strength (Figure 4-27b), points above the dashed line indicate unsafe predictions where predicted strengths exceed experimental results, while points below represent conservative predictions. Using the NDC and LEFM models together yielded safe predictions for eight out of fifteen specimens. Fatigue strength was generally underpredicted for FP and FC specimens and

overpredicted for FA specimens. The average differences were -30.0 ± 36.8 MPa (FP, conservative), 20.0 ± 31.4 MPa (FA, non-conservative), and -32 ± 49.9 MPa (FC, conservative). This pattern aligns with the inverse relationship between fatigue strength and discontinuity size. Notably, although the discontinuity size of FC03 was underpredicted (due to neglecting proximity effects), its predicted fatigue strength (86 MPa) remained conservative compared to the experimental value (103 MPa). Similarly, for FC04, the experimental fatigue strength (236 MPa) was significantly higher than the predicted value (117 MPa), a point further discussed later.

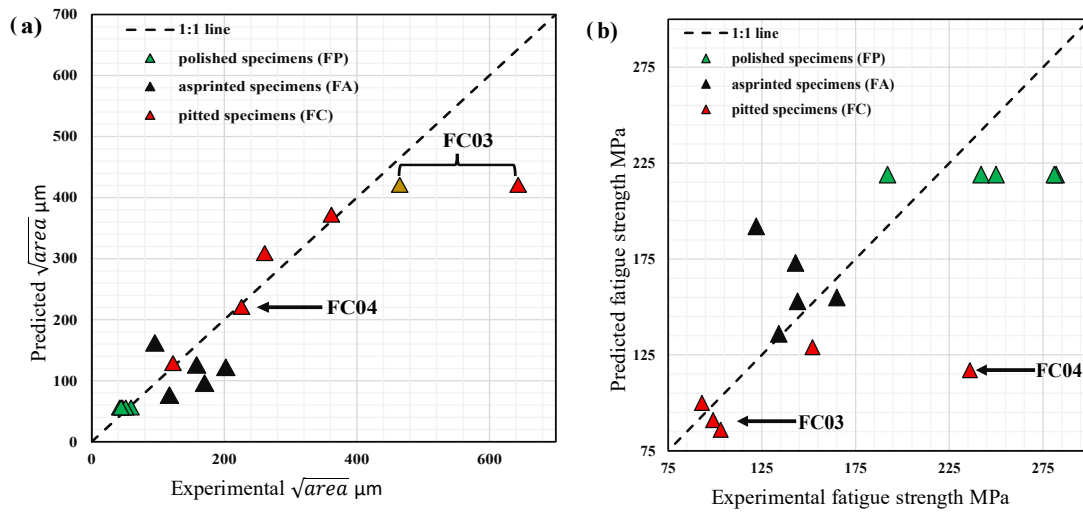


Figure 4-27. Plot for comparison between predicted and experimental data: (a) discontinuity size and (b) fatigue strength.

4.8 Discussion of results

The fatigue test results unveil the scatter in fatigue strength for the LPBF-M789 alloy at 2.10^6 cycles across the different fatigue grades. The higher performance shown by the results (in Figure 4-21) of the FP specimens reveals the benefit of the polishing on the fatigue performance. The FP specimens are of fatigue grade 2 to 3 zone, indicating a more than 48.7 % reduction from the PFS of 550 MPa. In contrast, those of the as-printed and pitted surface conditions are predominantly in the fatigue grade 1 classification, where the fatigue strength is between 10 to 30 % of the peak fatigue strength. It appears that even in the polished condition, the fatigue strength is unable to come close to the PFS. However, the limited number of tested specimens and the specimen geometry constrain the range of discontinuity sizes observed at failure origins, which may influence

the measured fatigue strength distribution. The result here, when compared to those in a recent work by Ordnung et al. [110] on the same M789 alloy show similar trend despite using an incremental dual laser LPBF technology to improve the surface finish quality of the alloy. These findings reveal the limit of the peak fatigue performance of the M789 alloy in the polished condition at $2 \cdot 10^6$ cycles. The next sections focus on discussing the results, failure trigger discontinuities, failure mechanisms, and result comparison to the NDC predictions.

4.8.1 The discontinuities at the failure origins

The KT diagram presented in Figure 4-26 reveals the evolution of the fatigue strength with the size of the discontinuities found at the failure origin for the tested specimens. Notably, because of the smooth surface finish, the polished specimens exhibited higher fatigue strength, ranging from 192 to 282 MPa, compared to that of the other specimen conditions, which ranged from 93 to 236 MPa. Complimentarily, polished specimens had the smallest discontinuities (42 to 59 μm), followed by the as-printed (95 to 202 μm) and pitted (123 to 643 μm) specimens. These discontinuities are not necessarily the largest within the specimens (further explanation in 4.8.1.6), as one cannot rule out the presence of larger discontinuities capable of causing failure.

4.8.1.1 Failure mechanism of polished specimens (FP)

Naturally occurring discontinuities, particularly LOFs and NMI, were the main influencers of the fatigue strength of polished specimens of the alloy. The discontinuities showed no convergence towards any unique shape as they all exhibited different geometrical features, hence are deemed irregular. Their size range of 42 to 59 μm places them in the long-crack region of the KT diagram (Figure 4-17), where the transition size between the Murakami model and LEFM is 15 μm . This positioning suggests a crack-like failure mechanism, with negligible cycles devoted to crack nucleation. The irregularity, especially with the alumina-based NMI, known for its brittle characteristics, enhanced the chances of intra-cracking and decohesion from the alloy's matrix, causing crack propagation.

4.8.1.2 Failure mechanism of as-printed specimens (FA)

As regards the as-printed specimens, their surface heterogeneity was critical to the fatigue strength. The range of the fatigue strength (122 to 165 MPa) for the as-printed specimens is smaller compared to that of the polished specimens (192 to 282 MPa) because of the rough surface. The

presence of surface valleys and troughs, which are typical of LPBF alloys, often intensifies strain localisation under cyclic loading. This localised deformation, coupled in the highly stressed volume region, promotes a crystallographic failure mechanism evidenced by the presence of facets (see Figure 4-22d of FA05) at the failure origin.

4.8.1.3 The short to long crack behaviour and the effect of microstructure

The behaviour of the alloy in the short crack region was not observed as the transition size of 15.1 μm (see Figure 4-17) is smaller than the discontinuities measured at the failure origins. This made it challenging to validate the El-Haddad and Murakami models in the short crack regime, where microstructural influence is pronounced. One possible reason for this is the large size of the specimen's gauge volume, which biased failure towards larger discontinuities. A similar observation was reported in the work of Hu et al. [148] on LPBF-built Ti-6Al-4V. They did not find intrinsic discontinuity sizes below 30-72 μm at the failure origin despite extensive testing. To trigger failure from smaller discontinuities, or the microstructure, would require a reduction of the specimen geometry or post-process operation like hot isostatic pressing (HIP). Masuo et al. [149], in their work on Ti-6Al-4V alloy, have reported an improvement in fatigue strength of Ti-6Al-4V alloy produced using electron beam melting. The improvement was to an extent of the fatigue grade 5 (peak fatigue strength), and it was attributed to the elimination and reduction in the size of discontinuities in the alloy realised through HIP. Another approach proposed by Pessard et al. [99] is to reduce the size of the highly stressed volume, which consequently reduces the likelihood of triggering larger size discontinuities. Accordingly, future work will focus on validating the El-Haddad and Murakami models in the short crack regime by modifying the highly stressed volume or using smaller sized M789 specimens to promote crack initiation from microstructurally small discontinuities. In contrast, the El-Haddad and LEFM models were validated by the measured discontinuities at the failure origin and the associated fatigue strength, thereby unveiling their long crack-like behaviour.

Regarding the microstructure, the presence of crystallographic facets at the failure origin of some FA specimens hints at the influence of the hierarchical martensitic structure of the M789 alloy. Martensite typically exhibits a hierarchical structure in decreasing size order: PAG > packets > blocks > laths. Any of these substructure units can participate in the failure mechanism of the alloy and leave faceted features on the fracture surface [150]. In the case of martensitic stainless steels

like the M789, cleavage facets are often associated with fracture along the (100) family of planes, which represent low atomic packing density and, consequently, the preferred cleavage path [151, 152]. Facets are also linked to large preferentially orientated subsurface grains, which are weak microstructural links due to ease of slips and crack initiation. Kobayashi et al. [153] observed in their work that the size of the facet feature seen on the fracture surface of 718 alloy was consistent with large subsurface grains. As unveiled in Figure 4-22, the facet sizes (62-69 μm), LOFs and NMI (42-59 μm) are comparable with the large PAGS (40-70 μm). Having such comparable sizes can breed a competition for crack initiation and propagation between the discontinuities and the microstructure. Although our results showed that cracks initiated from the discontinuities larger than the 15.1 μm theoretical long crack transition size, the microstructure influenced their subsequent propagation.

4.8.1.4 Failure mechanism of pitted specimens (FC)

The data of the electrochemically pitted specimens were also positioned in the long crack region of the KT diagram; important observations on the results of specimens FC01, FC03 and FC04 are discussed in the subsequent sections.

4.8.1.5 Competition between pits and LOF discontinuities

The competing effect between the pits and LOF discontinuity was seen in the result of specimen FC01. Despite the size of the pit with $\sqrt{\text{area}}$ of 182 μm on the specimen, the fatigue failure emerged from an LOF of a smaller size of 123 μm . The irregularity of the LOF likely enhanced its severity, making it more detrimental to the fatigue strength than the relatively smooth and regular pit (Figure 4-25). The size difference between the two discontinuities (59 μm) was insufficient for the pit to control the fatigue strength of the specimen. This suggests the pit size in FC01 is not large enough to dominate the LOF discontinuity as the primary cause of failure. The observation is further supported by the prediction results (see Table 4-3), which show that the pit in FC01 is the smallest critical pit among all the FC specimens and was the only case where failure emanated from an internal LOF rather than a pit. It is also noteworthy that the size of the LOF discontinuity (123 μm) exceeds the estimated largest size of 56 μm from the EVS analysis of the metallographic specimen performed in section 4.5. This observation highlights the inherent stochastic nature of

discontinuities in the M789 alloy, which can account for deviations from analytical predictions and consequently alter the expected fatigue behaviour.

4.8.1.6 The interaction and shape effect of pits at the failure origin

The assumption of the interaction effect, consistent with Murakami's position on fatigue strength, is particularly relevant for specimen FC03. This is because the smallest discontinuity ($375\text{ }\mu\text{m}$) exceeded the separation distance ($104\text{ }\mu\text{m}$), as shown in Figure 4-24. Aman et al. [154] validated this interaction criterion when evaluating the influence of interacting small discontinuities on the fatigue limits of pure iron and bearing steel. This assumption of the pit interaction explains why FC03 appears as an outlier on the KT diagram plot (Figure 4-26). Conversely, when interaction is not assumed, and only the largest pit is considered as the failure origin, the data align well with the KT diagram in Figure 4-26, suggesting that failure may have been governed by the largest discontinuity in isolation. The noninteraction assumption may be valid since M789, being a high-strength steel ($\text{HV} > 400$), exhibits low ductility and intuitively signifies a rapidly decaying stress field. Nonetheless, regardless of the assumption, the KT diagram trend remains conservative compared to the experimental results (Figure 4-26).

For the shape, its influence is apparent on the fatigue strength of specimen FC04 (236 MPa, Figure 4-26), which is about twice the value (116 MPa) predicted by the KT diagram. This deviation is likely due to the geometry of the pit. Depending on its shape, a pit can behave either like a crack or a notch [150]. In the crack-like case, the fatigue strength is governed by the stress intensity factor (K) and a stress field exponent of 0.5, making it suitable for modelling with LEFM. Conversely, a notch-like geometry produces a lower stress field exponent (< 0.5), where fatigue strength is mainly influenced by stress concentration rather than K , making LEFM predictions conservative but less accurate (see Figure 4-26). Fractographic analysis of specimen FC04 (Figure B- 8 of Appendix) and finite element analysis (Figure B- 9, Appendix) returned a stress concentration factor of 2.06 and a stress field exponent of 0.16. These findings suggest that the pit geometry likely induced a notch-like behaviour, resulting in the higher fatigue strength observed in FC04. However, further targeted studies are needed to fully characterise the geometry-induced notch-like response of the alloy and to expatiate the influence of discontinuity proximity and interaction on the fatigue strength. These aspects will be explored in future work.

4.8.1.7 Disparity of NDC prediction from experimental data

In general, the NDC prediction and experimental data unveiled varying behaviour across the specimens. For the polished specimens (FP), EVS yielded conservative fatigue strength estimates, aligning well with predictions. This is not surprising, as many literature works have successfully applied it to internal discontinuities (LOFs and NMIs) to predict fatigue strength.

In contrast, the predictions of fatigue strength for the FA specimens were nonconservative, as the predicted sizes of failure discontinuity were smaller than what was observed on the fracture surfaces. One possible explanation for the variations is the complexity of the as-printed surface condition of the FA specimen. Features like surface profile, solidification cracks, overhanging powders and other discontinuities not considered can enhance the severity of the troughs and even trigger crystallographic failure mechanisms. Also, the contributing microstructural influence, as seen with the emergence of facets at the failure origin, cannot be understated. Hence, relying solely on trough depth for fatigue prediction in as-printed conditions may oversimplify the actual damage-driving mechanisms. However, it is worth mentioning that the use of EVS for as-printed surfaces yields a mixed bag of success stories in the literature. Nikfar et al. [60] recently applied the EVS to estimate the deepest valley of an as-printed Ti-6Al-4V alloy and reported mean errors of 5 -15 %. They, however, noted the limited reliability of the approach that may be attributed to quality control. The reader is also referred to the review of Beretta on 25 years of EVS [62].

For the pitted specimens, using the largest surface discontinuity as the basis for fatigue predictions resulted in conservative estimates. However, deviations observed in the outlier data can be attributed to pit interactions or the absence thereof, as illustrated by specimen FC03. Additionally, the tendency of pits to exhibit notch-like behaviour contributes to the disparity with the predicted values. These outliers highlight the need for caution when relying solely on pit discontinuity size to estimate fatigue strength, particularly when the full geometry of the pit cannot be accessed during non-destructive characterisation due to its complex morphology.

4.9 Conclusions

This study examined the fatigue behaviour of recently developed LPBF-manufactured M789 alloy under polished, as-printed, and corroded surface conditions using a KT framework at 2.10^6 cycles.

Based on the experimental, fractographic, and modelling analyses, the following conclusions are drawn:

- I. Fatigue behaviour of LPBF M789 is governed by both manufacturing (intrinsic) and extrinsic discontinuities, with surface condition influencing performance levels corresponding to 50 % of peak fatigue strength for polished specimens and 10-30 % for as-printed and corroded specimens.
- II. Fatigue strength of the alloy is well predicted by LEFM and El-Haddad models, as manufacturing discontinuities have propagation behaviour comparable to long cracks, making the Murakami model unsuitable. In contrast, extrinsic discontinuities, such as corrosion pits, adopt a notch behaviour.
- III. Non-destructive characterisation and EVS provide conservative predictions of discontinuity size and the fatigue strength of the alloy for polished and pitted conditions, but not for the as-printed.
- IV. The alloy has limited sensitivity to interactions between neighbouring discontinuities, as fatigue strength predictions based on the largest discontinuity are 59% more accurate than predictions using combined discontinuity sizes.
- V. Microstructural features contribute to the alloy's failure mechanisms in the long-crack region, as crystallographic facet sizes at crack initiation sites are comparable to prior austenite grain sizes.

CHAPTER 5 PRELIMINARY RESULTS OF FIC SPECIMENS

In this chapter, the fatigue results of the FIC specimens, which were not included in the Article of Chapter 4, are presented. The graphical results illustrating the progression of loading steps leading to failure for the specimens are shown in Figure 5-1. In addition, the results are tabulated in Table 5-1 with the associated specimen IDs, step numbers, failure cycles, and final computed fatigue strengths. $\sqrt{area}$ and failure location. The protocol of fatigue testing has already been discussed in Chapter 4 (Article 1). The results show the FIC specimens having a fatigue stress range of 75 to 89 MPa, corresponding to an average of 80 ± 5.9 MPa (mean $\pm$ standard deviation). The small standard deviation indicates low scatter, suggesting that all specimens failed likely due to a similar failure mechanism. Additionally, for the FIC specimens, 3/5 of the specimens failed from the channel, while the rest failed outside the channel. The fractographic image is given in the next subsection.

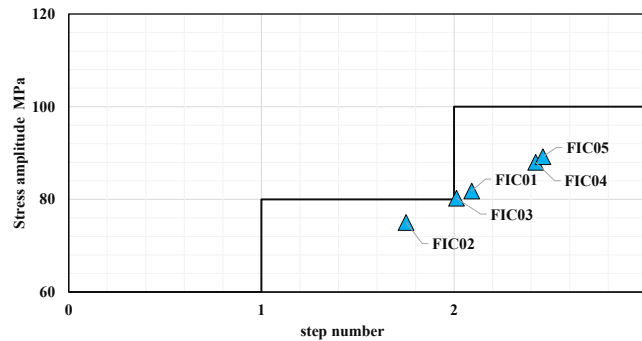


Figure 5-1. Step loading plot of the FIC specimens.

Table 5-1. Table of fatigue test results for the FIC specimens.

Specimen ID	$\sigma_{a,n-1}$ MPa	N_{fail}	σ_{FS} $= \sigma_{w,Nlife}$ MPa	$\sqrt{area}$ μm	Type of discontinuity	Location
FIC02	60	1.5E6	75	2865	Trough	Channel
FIC03	80	2.5E4	80	NA	Trough	Out channel
FIC01	80	1.8E5	81	NA	Trough	Out Channel
FIC04	80	8.4E5	88	2865	Trough	Channel
FIC05	80	2.0E5	89	2865	Trough	Channel

5.1 Fractography of FIC specimens

A representative fractographic image of specimen FIC04 is given in Figure 5-2a. The $\sqrt{area}$ of the exposed channel (traced in red dashed line) was comparable across the specimens. The failure origin is in the region labelled 'side 1' and outlined in black. The channel surface is as-printed and accounts for the observed rough features seen at the origin in the SEM fractographic image in Figure 5-2b. On the region labelled 'side 2', ratchet marks highlighted with red lines are visible. These features suggest elevated stress at the region (site 2) acting on the remaining ligament after crack initiation from side 1.

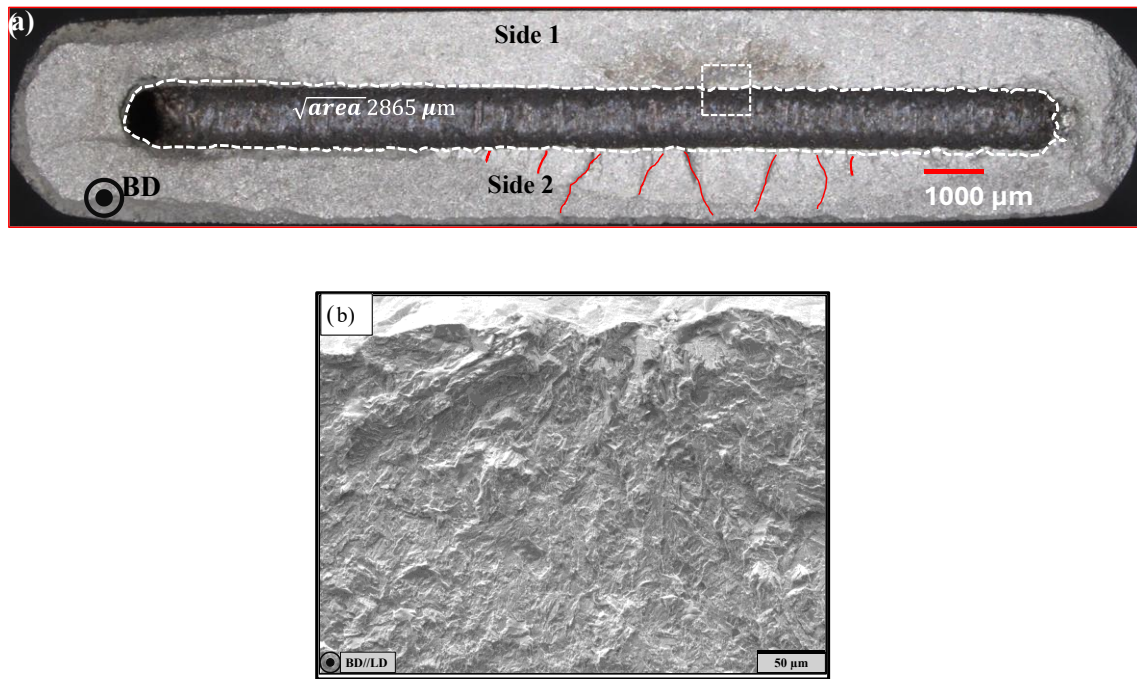


Figure 5-2. (a) Fractographic image of FIC04 specimen with the channel as the failure origin. (b) SEM image of the failure origin of FIC04.

5.2 KT diagram of all specimens

The fatigue results for the FIC specimens are plotted on the KT diagram of the M789 in Figure 5-3. For perspective, the diagram also includes other specimen types (FA, FP and FC) drawn from article 1 in Section 4.7.1 of Chapter 4.

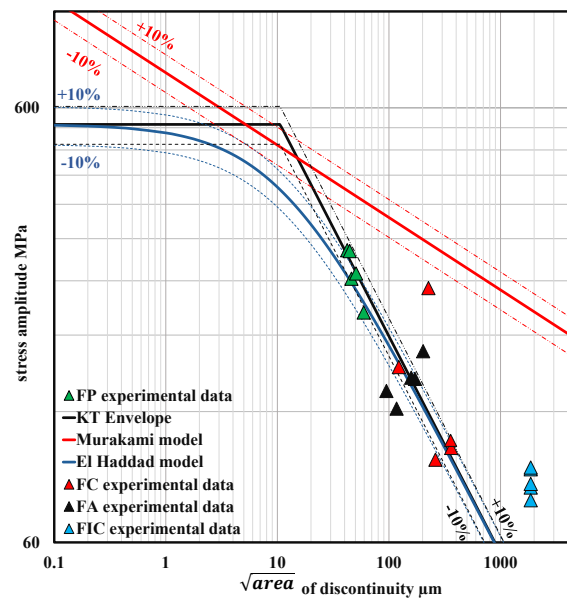


Figure 5-3. The KT diagram of the M789 alloy shows all the results of fatigue strength and discontinuity size for all the tested specimens.

Expectedly, the FIC specimens displayed lower fatigue strength compared to the other specimen types. This is due to the large $\sqrt{area}$ of 2862 μm relative to discontinuities of the other specimens. Also, the fatigue strength of the FIC significantly exceeds the value anticipated from the KT diagram, making it emerge as an outlier just like specimen FC04. These outliers are conservative with respect to the KT diagram but require further understanding of such behaviour. For this reason, the next section explores the use of local stress to present the results of FIC and FC04 on the KT diagram.

CHAPTER 6 LOCAL STRESS PRESENTATION OF THE KT DIAGRAM

In this section, a local stress approach is used as an alternative method to present the results of the FIC and FC04 specimens instead of the commonly used nominal stress approach. This will help in the verification of its suitability in treating the outlier data.

6.1 The local stress approach

In this approach, the results of FC04 and FIC specimens are treated as notches since their outlier data are attributed to the notch effect on the fatigue strength. The geometry of the discontinuities is assumed to have influenced their behaviour to cause a notch effect on their fatigue strength [155]. For the FC04 specimen, the pit geometry is influencing the notch behaviour, while for the FIC specimen, the internal channels' geometry. Thus, to effectively capture the impact of these extrinsic discontinuities on the alloy, a local stress approach is adopted to present the results, particularly for these specimens. The steps used to accomplish this are as follows:

Step 1. Estimation of the stress concentration of the pit on FC04 (see Section A.4 of the Appendix) and for internal channel (see Section 3.9)

Step 2. Estimation of the stress concentration of roughness asperity at notch root (see Section A.4.1 of the Appendix).

Step 3. Estimation of the local stress at the pit of FC04 and the channel of FIC using Equation 6-1

$$\sigma_{f,local} = k_t \times \sigma_{f,nominal} \quad \text{Equation 6-1}$$

Step 4. Plotting of the local stress estimated from step 3, for both the FIC specimens and the pit on FC04.

The results obtained for local stress for the FIC and FC04 pit specimens are reported in Table 6-1. The table shows the stress amplitude at failure, the estimated $\sqrt{area}$ for the failure origin of the channel, and the stress field equation from FE analysis.

Table 6-1. Result on the estimation of the local stress from the equivalent position of the roughness trough from the notch root of the channel.

ID	stress amplitude MPa	depth μm	$\sqrt{area} = c \times \sqrt{10}$ μm	stress exponent γ	field $k_{t,R}$	$\sigma_{f,local}$, MPa
FIC05	89	41	130		1.6	142
FIC04	88	53	168		1.5	132
FIC02	75	86	272	-0.256	1.4	105
FC04	236	NA	226	-0.16	$k_t=2$	472

Figure 6-1 shows the KT diagram presented using the local stress results for the FICs and specimen FC04- pit. The results are plotted on a size of $0.1 \mu\text{m}$, which is arbitrarily chosen as the size to represent the microstructure. The outlier data are not present anymore, and this allows all the results to be described accurately with the KT diagram.

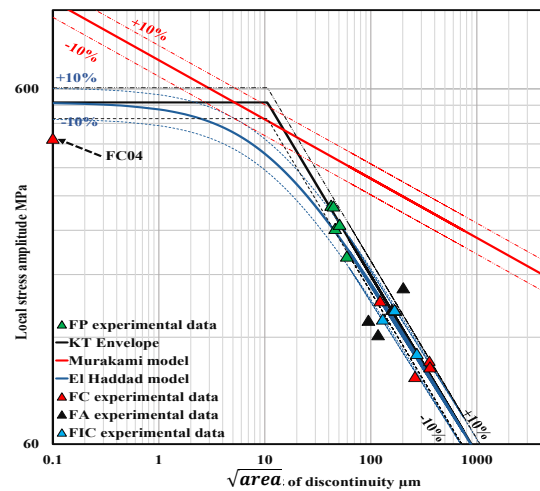


Figure 6-1. KT diagram of the specimens represented using the local stress.

6.2 Discussion on the local stress approach

The use of the local stress approach helped to realign the outlier data with the KT diagram. The use of the stress concentration of the FC04 specimen to compute the local stress led to results of 462 MPa, which is closest to the discontinuity-free fatigue strength of the alloy compared to the results from other specimens. This result suggests that fatigue initiation was controlled by the microstructure, since no significant discontinuities were present at the notch root. Similar studies by Benedetti et al.[156] have shown that fatigue predictions become more accurate when local stress or equivalent stress approaches are used instead of nominal stress values, because they better represent the real conditions at the crack initiation site.

Furthermore, the 462 MPa obtained falls in the upper limit of fatigue performance (70-90 % of the PFS) of the alloy that may be achievable when discontinuities are minimised or absent, for example, through post-processing treatments such as hot isostatic pressing. The result can be attributed to the microstructure, as there are no discontinuities at the notch root. This is also indicative of the possible region of fatigue strength that is attainable when post-processing operations like hot isostatic pressing are performed on the alloy. Under such conditions, discontinuities in the short crack region could dominate the fatigue of the alloy.

For the FIC specimens, the results obtained using the local stress approach fall within the long-crack region of the KT diagram and are consistent with the fatigue strength of the as-printed

specimens. This indicates that fatigue failure was mainly controlled by the roughness asperities along the internal channels, rather than by the nominal stress concentration of the channel geometry.

Overall, the local stress approach brings all the experimental results into the crack-controlled region of the KT diagram and removes the apparent outliers observed when local stresses were used. This produces a more realistic and conservative KT diagram for the M789 alloy because it reflects the actual stress conditions at the fatigue initiation sites. As a result, the local-stress-based KT diagram provides a more accurate representation of fatigue behaviour for the alloy across the different surface or specimen types.

CHAPTER 7 ANALYSIS OF OUTLIER EXTRINSIC DATA USING THE ATZORI-LAZZARIN DIAGRAM

In this section, the Atzori–Lazzarin diagram is constructed for the M789 alloy using the geometric characteristics of extrinsic discontinuities. The focus is on specimen data that appeared as outliers in the KT diagram, particularly specimen FC04 and the FIC specimens. Hence, it is hinged on the treatment of FC04 and FIC specimens as notches. The analysis aims to determine whether the geometric features of these discontinuities can explain the deviations observed in the fatigue behaviour.

The analytical step involved in the construction is as follows

- i. Retrieval of stress field exponents ‘ γ ’ from stress field plots for FC04 (Figure A- 8) and FIC (See Figure 3-7) specimens.
- ii. Estimation of β parameter dependent on notch opening angle using Equation 7-1 proposed by Atzori et al. [90]

$$\beta = \frac{\sqrt{2}}{2^\gamma} \pi \left(\left(\frac{1}{2} \right)^{-\gamma} \right) \quad \text{Equation 7-1}$$

- iii. Estimation of notch stress intensity threshold $\Delta K_{I,th}^V$ using Equation 7-2

$$\Delta K_{I,th}^V = \beta \cdot 2\sigma_{w,0.1}^{(1-2\gamma)} \cdot \Delta K_{th,LC}^{2\gamma} \quad \text{Equation 7-2}$$

Where $2\sigma_{w,0.1}$ is the fatigue strength in stress amplitude for the discontinuity-free state at $R = 0.1$ and $\Delta K_{th,LC}^{2\gamma}$ is the stress intensity threshold range for long cracks for the LPBF M789 alloy

- iv. Representation of crack size using the effective crack/notch dimension a_{eff} . The estimation of the Y -factor is given in Appendix A.
- v. Construction of the Atzori-Lazzarin diagram using Equation 7-3

$$\Delta \sigma_{th} = \frac{\Delta K_{I,th}^V}{2\sqrt{\pi}(a_{eff} + a_o^V)^\gamma} \quad \text{Equation 7-3}$$

A summary of all the parameters used to plot the ALD for the pit of FC01, FIC specimens, is shown in Table 7-1. The stress concentration k_t and stress field exponent from the FEA analysis in sections extracted from Table 6-1 for the FIC and FC04 specimens, respectively, are shown in the table, as well as the corresponding stress range intensity threshold.

Table 7-1. Summary of parameter values used to construct the Atzori Lazzarin diagram.

	k_t	γ	β	$\Delta K_{I,th}^V$ MPa m^γ	a_0 μm	a^* μm	Comments
FC01-pit	2	0.16	1.87	343.81	19.48	730.29	Shape factor used 1.12
FIC	2.89	0.26	1.57	98.58	8.65	77.21	Shape factor used 1.05

Note

$$\Delta K_{I,th}^V = \beta \cdot \Delta \sigma_0^{(1-2\gamma)} \cdot \Delta K_{th,LC}^{2\gamma}$$

$$\beta = \frac{\sqrt{2}}{2^\gamma} \pi^{\left(\left(\frac{1}{2}\right)-\gamma\right)}$$

$$a^* = a_0 \cdot k_t^{1/\gamma}$$

The pit of FC01 is treated as a semi-elliptical pit, and a shape factor of 1.12 was used.

The channel is treated as a through all hole with a shape factor of 1.05

Using the information from the table, the fatigue strength evolution for the internal channels and the pit of FC04 is constructed based on the framework of the Atzori Lazzarin diagram. The plot is made using the fatigue strength versus the effective length of the discontinuity defined by Equation 7-4.

$$a_{eff} = a \cdot Y^{1/\gamma} \begin{cases} a = \sqrt{area} \text{ for cracks} \\ a = depth \text{ for notches} \end{cases} \quad \text{Equation 7-4}$$

Where a is the discontinuity size, Y is the shape factor, and γ is the stress field exponent retrieved from finite element analysis.

The conversion of the discontinuity sizes using Equation 7-4 is reported in Table 7-2. Only the FC04 and the FIC specimens were used for other shape factors due to the anticipated notch influence on their fatigue strength. Also reported in the table are the fatigue strength (FS) and the

singularity exponent. All discontinuities across the different specimen conditions that are assigned the typical crack exponent 0.5, except for the outliers of FC04-pit and the FIC specimens.

Table 7-2. Conversion of crack sizes to effective crack size using the Y factors adopted for the cracks and notches (FC04-pit and FIC specimens).

ID	FS MPa	Crack size μm	Y value	classification	Singularity exponent	$a_{eff}\mu\text{m}$
FP02	192	59	0.65	crack	0.50	24.93
FP01	242	46	0.65	crack	0.50	19.44
FP03	250	51	0.65	crack	0.50	21.55
FP05	281	44	0.65	crack	0.50	18.59
FP04	282	42	0.65	crack	0.50	17.75
FA01	122	117	0.65	crack	0.50	49.43
FA04	134	95	0.65	crack	0.50	40.14
FA05	143	170	0.65	crack	0.50	71.83
FA02	144	158	0.65	crack	0.50	66.76
FA03	65	202	0.65	crack	0.50	85.35
FC05	93	260	0.65	crack	0.50	109.85
FC02	99	361	0.65	crack	0.50	152.52
FC03	103	465	0.65	crack	0.50	196.46
FC01	152	123	0.65	crack	0.50	51.97
FC04	236	45	1.12	notch	0.16	91.37
FIC02	75	375	1.05	notch	0.26	453.73
FIC04	88	375	1.05	notch	0.26	453.73
FIC05	89	375	1.05	notch	0.26	453.73

In Figure 7-1, the ALD diagram is drawn using the effective length of the crack, defined as a_{eff} and the fatigue strength. Note that the peak fatigue strength used remains 550 MPa, while the sloped region for the FIC specimens and the FC04-pit is defined by the respective $\Delta K_{I,th}^V$ given in Table 7-1.

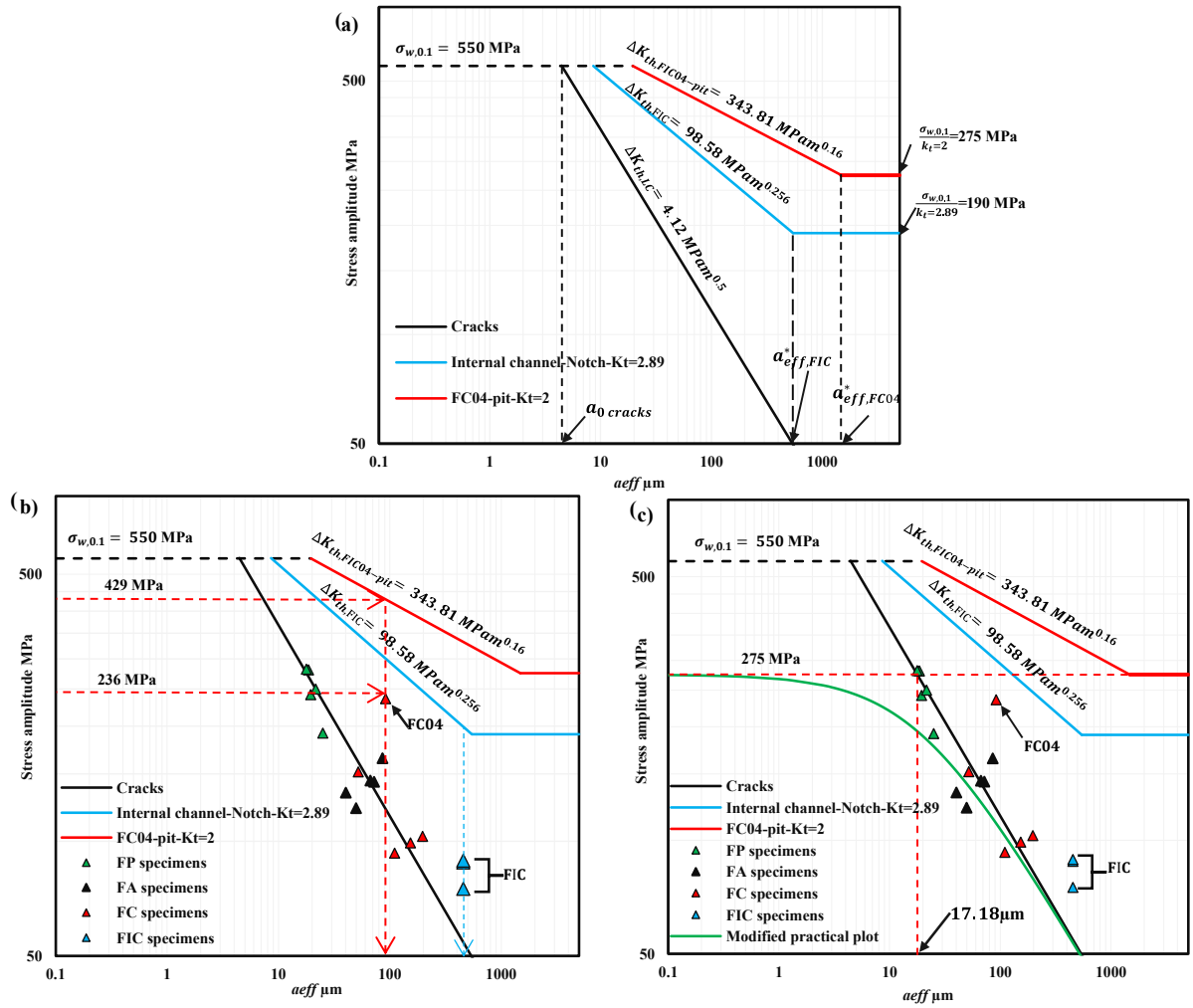


Figure 7-1. Kitagawa–Takahashi diagrams showing: (a) the constructed ALD diagram for FC04-pit specimens with $k_t = 2$ and FIC specimens with $k_t = 2.89$, (b) experimental fatigue results superimposed on the diagram; and (c) the proposed practical KT diagram for LPBF M789 alloy.

7.1 Discussion on the ALD for extrinsic discontinuities

In the ALD of Figure 7-1b and c, the full-notch fatigue strength for FC04 is 275 MPa ($550/k_t$ with $k_t = 2$). The effective pit size of FC04 is in the stress-field controlled region, therefore, extrapolation of the ALD trend predicts a fatigue strength of approximately 429 MPa. However, the experimentally measured fatigue strength was 236 MPa. This discrepancy indicates that the fatigue response was not governed solely by the pit geometry. Fractographic observations revealed

a crystallographic facet measuring approximately 52 μm , which is comparable to the size of the prior austenite grain size. The crack initiation from the notch was likely limited by a microstructure-controlled weak link and not by the “ideal notch” behaviour assumed in the ALD trend. Simm et al. [157] demonstrated that in maraging steels, the PAG size may act as the dominant microstructural weak link and control fatigue strength. For the FC04, the pit likely raised the local stress and created a highly stressed volume around the notch root. Within this highly stressed volume, the probability of encountering an unfavourable microstructural region (large PAGB, packet boundary, or other weak feature) increases and causes crack initiation at a stress level lower than predicted by the ALD. In other words, the pit primarily acted as a stress concentrator while the microstructure set the real fatigue strength. So, the specimen failed at a stress level similar to that of intrinsic discontinuities rather than approaching the stress-field prediction of the ALD.

For the FIC specimens (see Figure 7-1b and c), the ALD full-notch prediction corresponds to approximately 190 MPa ($550/k_t$ with $k_t = 2.89$), but the measured fatigue strength is only 75-90 MPa, which lies closer to the crack-controlled region of the KT diagram than to the notch-controlled ALD prediction. This suggests that the internal channel did not adopt a full notch behaviour despite having an effective size of $\sim 453 \mu\text{m}$, which is near the transition size $a_{eff,FIC}^* \sim 546 \mu\text{m}$. A likely explanation is that the roughness of the walls of the internal channel influences a crack-like behaviour of the channel, which contains roughness valleys, troughs, and partially fused features. In addition, as discussed in the literature review (see Figure 2-18), beyond a critical stress concentration factor, denoted k_t^* , crack-like behaviour predominates. The stress concentration factor of the channel, $k_t = 2.89$, may be beyond the k_t^* of the alloy, meaning a crack-like condition dominates. Typical values of k_t^* for high-strength steel materials is 1.5 to 2.5 depending on the geometry of the notch. This can be seen in the work of Smith and Miller [86], where the transition from notch-controlled to crack-like behaviour occurs relatively at low stress concentration factors ($k_t < 2.5$) was observed. In this work, we have not characterised the k_t^* for the alloy, but it is inferred that the $k_t = 2.89$ of the channel is too high for it to adopt a notch behaviour.

Based on the foregoing observations, a practical Kitagawa–Takahashi diagram was proposed (green line in Figure 7-1c). In this diagram, the upper fatigue strength limit is taken as 275 MPa, corresponding to the full notch behaviour defined by $k_t = 2$. This value is close to the maximum

fatigue strength measured in polished specimens (281 MPa) and corresponds to an effective discontinuity size of about 17 μm , which is used as the intrinsic discontinuity size a_0 in the El-Haddad equation. This intrinsic size is also close to the effective size of 22 μm of the smallest facet observed at failure origin. Furthermore, the unvalidated idealised peak fatigue strength of 550 MPa obtained using the empirical relationship, 1.6 (HV), may have been an overestimation. As such, the experimentally supported value of 275 MPa represents a more realistic upper fatigue-strength limit governed by intrinsic discontinuities and prior austenite grain size. The proposed diagram is, therefore, intentionally conservative and focuses on the region where competition between different discontinuities is most likely. This approach is reasonable because the El-Haddad and Murakami models could not be fully validated in the short-crack region due to the limited experimental data. The results also suggest that discontinuities with $k_t < 2$ are likely governed by intrinsic discontinuities or microstructural features, making it difficult to reach the fatigue strength prediction of ALD.

Key conclusions

- Large parent austenite grains within the highly stressed volume of pits limit the attainment of the notch fatigue strength predicted by ALD.
- Roughness of internal channel and the associated stress concentration of the internal channels influence the adoption of a near crack-like behaviour.

CHAPTER 8 GENERAL DISCUSSIONS

This thesis investigated the fatigue behaviour of the LPBF-processed Böhler M789 maraging stainless steel, with specific emphasis on the role of intrinsic and extrinsic discontinuities and their integration into Kitagawa–Takahashi (KT)-based fatigue design. The work was motivated by the intended industrial application of M789 as a die-insert material operating under high cyclic loading with cooling liquids flowing through internal channels. The study addressed four main objectives: (i) understanding the influence of intrinsic and extrinsic discontinuities, (ii) assessing fatigue strength prediction models, (iii) identifying factors controlling discontinuity severity beyond size, and (iv) clarifying fatigue failure mechanisms and microstructural contributions.

Objective I. Unveil the influence of intrinsic (LOFs, NMI, Roughness asperities, pores) and extrinsically induced discontinuities (electrochemical pits and internally designed channels) on the fatigue strength of the M789 alloy.

In fulfilment of objective I of this study, the influence of extrinsic and intrinsic discontinuities on the alloy's fatigue behaviour was unveiled using the chosen surface conditions of the specimens: polished (FP), as-printed (FA), pitted (FC), and specimens with internal channels (FIC). This was particularly achieved in Article 1 of Chapter 4. Therein, we observed that the Fatigue performance of the alloy is governed by discontinuities to such an extent that none approached the discontinuity-free fatigue strength of 550 MPa. This was particularly interesting to see the extent to which the discontinuities influence and degrade the alloy's fatigue strength. Even the highest fatigue strength measured in an FP specimen remained approximately 48% below 550 MPa, while for the extrinsic discontinuities, it is approximately 10–30% of 550 MPa. The Maxwell and Nicholas [158] step loading protocol was useful in estimating the fatigue strength of individual specimens. Post-processing operations like hot isostatic pressing could hypothetically allow the alloy to achieve 80 % of the 550 MPa, although this was not explored in the present work. Overall, the findings confirm that the discontinuities set the fatigue limit of LPBF M789.

Nevertheless, the fulfilment of this objective should be interpreted in light of the limited number of samples tested for each condition in this study. Although the accelerated testing method used was meant to mitigate the use of limited specimen but, it may also subject the alloy to coxing during the test at various stress levels. This coxing effect was assumed to be insignificant in this work due to the limited plasticity of the alloy, which is characteristic of high-strength steels.

However, a dedicated investigation would be required to quantify the extent of such effects on the results obtained.

Objective II. Ascertaining which of the fatigue strength prediction models of El-Haddad, LEFM, and Murakami is most accurate for the prediction of the fatigue strength of the M789.

The evaluation of fatigue-strength prediction models presented in Article 1 (Chapter 4) showed that the El-Haddad and LEFM approaches provided the most consistent agreement with the experimental results. Fractographic analysis revealed that the sizes of all discontinuities at the failure origin were within the long-crack regime ($> 15 \mu\text{m}$) of the Kitagawa–Takahashi (KT) diagram. This indicates that none of the failure-initiating discontinuities fell within the short-crack region of the alloy. Consequently, models that depend on the long-crack threshold, such as the El-Haddad and LEFM approaches, showed better agreement with the experimental results than the Murakami model. This does not imply that the Murakami model is unsuitable; rather, it was not validated within the range of discontinuity sizes observed in this study. As shown in Article 1, the Murakami model is primarily applicable to discontinuity sizes smaller than $15 \mu\text{m}$. A similar observation was reported by Andreau et al. [92], who applied the KT framework to model fatigue behaviour in 316L stainless steel and found that their results also lay outside the short-crack regime. Evaluating the accuracy of fatigue models in the short-crack region would require failure to be initiated from smaller discontinuities ($< 15 \mu\text{m}$). This could be achieved by testing smaller specimens with a reduced highly stressed volume, thereby increasing the likelihood of smaller discontinuities acting as failure origins. Therefore, the fulfilment of Objective II is limited by the discontinuity sizes observed at the failure origins of the tested specimens.

Furthermore, the nondestructive characterisation, along with EVS carried out in Article 1 in Chapter 4, provided additional predictive capability along with the LEFM model, although accuracy varied with surface condition. Conservative predictions were obtained for FP specimens, whereas FA specimens exhibited non-conservative predictions, likely due to the complexity of roughness-induced stress concentrations not fully captured by the roughness parameter. For FC and FIC specimens, EVS-based predictions were generally conservative, although several outliers highlighted the influence of discontinuity shape and associated stress-concentration effects. On the whole, the EVS prediction should be viewed from the perspective of the size alone as the defining factor of discontinuity severity and other influences like microstructural contribution or even the

morphology or discontinuity interaction are not considered during prediction. In addition, the FA surface condition was characterised using representative roughness parameters and discontinuity measurements. However, the full 3D complexity of surface roughness morphology and local asperity interactions was not completely captured within the modelling framework

Objective III. Assessing the critical factors controlling the severity of discontinuities in the M789 other than the size through the exploration of the fractography and the Kitagawa Takahashi diagram.

The fulfilment of this objective is supported by the analyses presented in Article 1 (Chapter 4) and Chapter 6. Although discontinuity size emerged as the primary driver of fatigue strength, several observations indicate that discontinuity geometry also contributes to overall severity.

The results obtained from pitted (FC) and internally channelled (FIC) specimens were particularly important in revealing this behaviour. It was observed that competition between discontinuities for crack initiation can result in a smaller but sharper intrinsic discontinuity dominating over larger but smoother extrinsic discontinuities, such as pits. In such cases, large discontinuities may behave according to notch theory due to the associated stress concentration, rather than following crack-based approaches. This behaviour results in higher fatigue strength values (observed as outliers) compared to predictions from linear elastic fracture mechanics (LEFM)-based KT models.

Similarly, small irregular discontinuities can dominate fatigue behaviour despite the presence of larger discontinuities on the surface. This becomes particularly critical when fatigue design decisions are based on large extrinsic discontinuities that are expected to control failure due to their size. Under such conditions, caution must be exercised, as failure may instead originate from smaller intrinsic discontinuities.

In one case, the geometry of an outlier discontinuity produced a stress concentration that contributed to its deviation from LEFM-based predictions. To further investigate the influence of geometry and address these outliers, a local stress approach was applied in Chapter 6 to reconstruct the KT diagram, thereby accounting for stress concentration effects in the FC and FIC specimens. The improved alignment with the KT diagram supports the significance of discontinuity geometry. Overall, these observations highlight the importance of discontinuity shape in determining whether a discontinuity behaves in a notch-like or crack-like manner. However, further work involving

controlled discontinuity geometries and sizes is required to fully quantify their influence before definitive conclusions can be drawn.

Interaction effects between closely spaced discontinuities were also examined. Even when Murakami's geometric criteria for interaction were satisfied, fatigue failure was governed by the dominant discontinuity, with limited evidence of interaction. This behaviour may be attributed to the high strength and limited plasticity of the M789 alloy, which could suppress interaction effects. Nonetheless, these observations warrant further targeted investigation to better understand the conditions under which interaction becomes significant.

Objective IV. Identifying the fatigue failure mechanisms of the alloy and evaluating the influence and potential contribution of its microstructural characteristics.

This objective was addressed in Article 1 (Chapter 4), where microstructural features were observed in the FA and FC specimens in the form of crystallographic facets at the fatigue crack initiation site. The dimensions of these facets were comparable to the prior austenite grain (PAG) size, suggesting that the PAGs may act as alternative microstructural weak links for crack initiation. The initiation of cracks from crystallographic facets with dimensions comparable to the PAG size has been widely reported in additively manufactured steels [159].

In the FC specimens, the stress concentration generated at the root of a critical pit may activate these microstructural weak links, leading to crack initiation manifested as crystallographic facets at the failure origin. This mechanism likely contributed to the inability to attain the anticipated fatigue strength of 426 MPa predicted from the constructed ALD for the FC04 pit in Chapter 7. When geometrical discontinuities and microstructural features exist at comparable length scales, multiple crack-initiation pathways may compete. Under such conditions, the relative size of the discontinuity with respect to the governing microstructural feature becomes a critical factor in fatigue-strength evaluation.

This observation is consistent with the findings of Nadot et al. [122], who demonstrated that the ratio between grain size and surface hemispherical discontinuity size plays an important role in fatigue-strength behaviour. When failure originates from large prior austenite grains situated near discontinuities, the resulting fatigue strength may fall below predicted values, particularly in cases where the discontinuity follows notch-dominated behaviour, such as pits with well-defined

geometry. Therefore, identifying the governing microstructural weak link, whether prior austenite grain size, block size, or lath size, is essential when assessing newly developed alloys.

For the LPBF M789 alloy, the results suggest that microstructural features, although not dominant, may still contribute to fatigue crack initiation even in the presence of discontinuities within the long-crack regime. Nevertheless, the present study did not explicitly quantify the relative contributions of prior austenite grain size, packet size, block size, and lath morphology to fatigue crack initiation behaviour.

Implications of findings for KT-Based fatigue design

From a practical design standpoint, the nominal stress approach used in the KT diagram approach is a conservative framework for M789 in the long crack region, as all experimental data lie above the predicted bounds. This reflects the intrinsic conservatism of treating all discontinuities as crack-like, the most severe assumption for a given size. However, outlier behaviour observed for large, smooth extrinsic discontinuities demonstrates that such conservatism may lead to overly pessimistic or even misleading life predictions when notch-like behaviour prevails.

In this regard, the local stress-based KT approach offers a more realistic representation by incorporating stress concentration effects and local failure mechanisms. Although this approach requires the further step of estimating the stress concentration, it effectively reconciles outlier data and ensures that fatigue life predictions remain anchored to realistic initiation and propagation mechanisms using the long crack threshold. For safety-critical applications, this crack-based approach, informed by local stress analysis, emerges as the more reliable.

Alternatively, we proposed a KT diagram for the alloy based on an effective transition size of $17\text{ }\mu\text{m}$ determined at the intersection of the fatigue strength at full notch behaviour ($k_t = 2$) in the ALD and the LEFM prediction line. The proposed diagram effectively focuses on the discontinuity-dominant regions of the alloy and is conservative even in the short crack region.

CHAPTER 9 CONCLUSIONS, CONTRIBUTIONS AND RECOMMENDATIONS

This thesis investigates the fatigue behaviour of M789, a recently developed maraging steel produced via the Laser Powder Bed Fusion (LPBF) process. The study focuses on understanding how discontinuities associated with different surface conditions, as-printed (FA), polished (FP), pitted (FC) and polished with internal channels (FIC), influence fatigue performance and failure mechanisms. The motivation for this work was to investigate the fatigue behaviour of the alloy in the presence of discontinuities that are likely to occur during service. This need arises from the intended industrial application of the alloy in die inserts, where components are subjected to cyclic loading conditions. Since fatigue failure in LPBF alloys is often controlled by discontinuities, understanding their influence on fatigue performance is essential for the safe and reliable deployment of the alloy in practical applications. Based on the experimental investigations and findings obtained throughout this research, the following conclusions are drawn:

I. The fatigue behaviour of LPBF M789 alloy is governed by discontinuities.

Intrinsic discontinuities in polished specimens ($\sqrt{area} = 42\text{--}59 \mu\text{m}$) allow the highest practical fatigue strengths (192–282 MPa), corresponding to about 52% of the discontinuity-free fatigue strength of 550 MPa, whereas those in as-printed specimens ($\sqrt{area} = 95\text{--}202 \mu\text{m}$) reduces fatigue strength to 122–165 MPa. For the larger extrinsic discontinuities, such as pits ($\sqrt{area} = 226\text{--}471 \mu\text{m}$) and internal channels ($\sqrt{area} = 2865 \mu\text{m}$), they further reduce fatigue strength to 93–236 MPa and 75–89 MPa, respectively, limiting fatigue strength to 10–30% of the ideal fatigue strength.

II. The El-Haddad and LEFM models accurately predict the fatigue strength of the alloy in the long crack regime, whereas the Murakami model overestimates it. All experimentally observed failure-origin discontinuities ($\sqrt{area} 42\text{--}2865 \mu\text{m}$) are consistent within the long-crack regime, above the estimated short-crack limit of approximately 15 μm . In contrast, the Murakami model is less suitable for the tested specimens, since its predictive strength lies in the short-crack regime, which was not reached in any of the performed tests.

III. **NDC and EVS are reliable in predicting the fatigue strength of the alloy except for the as-printed condition.** The largest surface-measured discontinuity in pitted and channelled specimens was sufficient to yield conservative results.

IV. **The geometry and discontinuity proximity may contribute to the severity of fatigue strength.**

A small-sized LOF discontinuity of 123 μm that follows LEFM theory caused fatigue failure despite the presence of a larger-sized extrinsic pit discontinuity of 182 μm . The increasing size of a discontinuity enhances the tendency to adopt well-defined geometries that influence a higher fatigue strength than that predicted by LEFM. Results obtained from FEA of stress concentration of 2.89 and 2 for FIC and FC04 specimens, respectively, suggest an influence on fatigue strength.

V. **The microstructure of the alloy contributes to its fatigue failure mechanism.**

Crystallographic facet size range ($\sqrt{\text{area}} = 52 - 67 \mu\text{m}$) found at the failure origin are comparable to the large parent austenite grain size of the alloy. These can serve as weak links and particularly influence fatigue strength where the discontinuity is of comparable size or when situated at the root of a blunt discontinuity like flat shallow pits.

VI. **The local stress approach within the KT framework is conservative across all surface conditions.** Particularly when discontinuities such as corrosion pits and the internal channels give outlier data on the nominal stress-based KT diagram. In these cases, the local-stress approach supported by FEA provides a reliable basis for safe-life decisions, because it repositions notch-controlled failures within the KT framework defined by cracks.

VII. **A fatigue strength of 275 MPa is a safe operating stress across all the tested specimen conditions, in contrast to the idealised 550 MPa of the alloy.**

9.1 Key contributions of this work

Based on the findings of this study, the following contributions towards the understanding of the fatigue behaviour of the M789 alloy have been established:

- A systematic evaluation of the influence of intrinsic and extrinsic discontinuities under different surface conditions on the fatigue behaviour of M789 alloy, providing insight into its discontinuity tolerance.

- Clarification of the role of the alloy's microstructure in its fatigue failure mechanism, showing that microstructure is a contributing factor in the alloy's failure, even when discontinuities are present.
- Identification of factors beyond the size of discontinuity that contribute to the severity and cause a deviation from the predicted fatigue strength of the M789 alloy.
- Demonstration of the usefulness of EVS, combined with non-destructive surface characterisation, in predicting the fatigue strength of the alloy for different surface conditions.
- Development of a KT diagram tailored for the M789 alloy, leveraging the average facet sizes observed at failure origin, the smallest observed discontinuity size at failure origin and the crack size coincident with the fatigue strength under full notch behaviour at $k_t = 2$ to define a new a_0 within the El-Haddad framework. The resulting diagram relieves the use of the unrealisable ideal fatigue strength and focuses on the region where discontinuities predominate.

9.2 Industrial relevance of this work

The insights generated in this study contribute to the scientific understanding of the fatigue behaviour of the recently developed M789 alloy. By revealing the governing fatigue-failure mechanisms and the alloy's limitations despite its attractive static properties, this work kickstarts research work and partly fills the knowledge gap in the limited literature available on the alloy. The mechanistic interpretation of the results supported by fractography, Kitagawa-Takahashi analysis, and discontinuity-sensitive modelling offers a foundational understanding of the alloy's discontinuity tolerance. In addition, it identifies future research topics on the alloy that are relevant to the theme of this work to expand the influence of the discontinuities in the alloy on its fatigue strength.

For our industrial partners, the findings offer input for the safe and reliable deployment of M789 in fatigue-critical components. The validated models can be employed within the studied scope of this work and their in-house design criterion for service applications. Furthermore, it supports informed decision-making in the selection of an appropriate fatigue-design philosophy, whether crack or notch-based theories should be adopted, depending on the surface condition. For instance, the specimens with channels (FIC), which are meant to replicate the internal channels, emerged as

outliers from the KT diagram developed based on long cracks. The use of the long crack theories, although conservative, may result in overprediction of fatigue life. The notch theory would likely be suitable as it allows for the use of the notch stress intensity threshold that becomes part of the input criteria for design against fatigue and the accurate prediction of fatigue life. This will help reduce the uncertainty in life cycle prediction for components subjected to fatigue loading. In addition, the demonstrated accuracy of the KT diagram and the evaluated discontinuity-sensitive models provide a basis for incorporating discontinuity effects into design calculations. This is particularly relevant for industries considering M789 for high-cycle loading conditions in estimating the safe operating loading conditions.

9.3 Recommendations for future work

The following research avenues emerged from the findings of this work and are proposed as future works towards improved understanding of the fatigue behaviour of the LBF M789 alloy.

Short-crack regime characterisation

Further work should investigate the transition between short- and long-crack behaviour by employing smaller specimen geometries and controlled discontinuity sizes. Capturing this regime is essential for validating the predictive accuracy of discontinuity-sensitive fatigue models, particularly the Murakami approach, which is most applicable when cracks are comparable to microstructural dimensions.

Behaviour of HIPed M789 alloy

Hot Isostatic Pressing (HIP) is known to reduce or close internal discontinuities in LPBF components. Evaluating the fatigue behaviour of HIPed M789 would clarify the extent to which discontinuity mitigation improves fatigue strength, discontinuity tolerance, and the applicability of LEFM-based design approaches. This comparison would be valuable for industries considering post-processing routes for performance optimisation.

Geometrically sensitive modelling approaches

Future studies should incorporate tailored discontinuities such as notches with controlled radii and varying stress-concentration factors to quantify the notch sensitivity of M789. This will enable the

development of geometry-dependent fatigue-design rules and improve the accuracy of performance predictions for components with unavoidable stress raisers.

Determination of the k_t^* stress-concentration factor for M789 alloy

A systematic notch-fatigue investigation across a range of stress-concentration factors and discontinuity sizes is needed to determine the critical k_t^* for M789. Establishing this parameter will support the integration of the alloy into discontinuity-sensitive design frameworks and enhance the reliability of fatigue-strength predictions for components with geometric discontinuities.

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APPENDIX A EXPERIMENTS AND RESULTS

A.1 Estimation of residual stress in the M789 alloy

The residual stresses in the M789 were measured on the cubic sample using a μ -X360 FULL 2D portable X-ray residual-stress analyser (PULSTEC, Tokyo, Japan; v3.1.0) [28]. The instrument uses the $\cos \alpha$ method: with one fixed X-ray angle, a two-dimensional detector captures the full Debye–Scherrer ring in a single exposure, and the software converts the ring’s 0–360° strain pattern into the in-plane normal and shear stresses. This single-shot approach avoids repeated tilts and helps keep measurements fast and consistent. Key settings were Cr-K α radiation, 30 kV, and 1.0 mA tube power, a 35° incident angle, a 1.0 mm collimator, and 30 s exposure per point (Table A-1).

To sample the surface evenly, two perpendicular lines were drawn through the centre of the measured face (X and Y), and three readings were taken along each line. Because the geometry is fixed, the sample was rotated by 90° to switch from X to Y. The central position is common to both lines, so the face contributes five unique locations in total (See Figure A-1). Phase-specific measurements were taken for both ferrite (BCC) and austenite (FCC).

The measured stresses were low across the board. Average values were about 74 MPa (ferrite-X), 60 MPa (ferrite-Y), 90 MPa (austenite-X), and 81 MPa (austenite-Y), with individual readings ranging from roughly 36 to 123 MPa (Figure A- 2). Compared with the alloy’s yield strength (≈ 1600 MPa), these are small, typically $\leq 6\%$ of σ_y , and even the highest local value is $< 8\%$. These results of residual stresses are unlikely to significantly alter the mechanical behaviour. Their low level is consistent with the post-build solution-anneal, which is expected to relieve most process-induced stresses. In the text of Suresh, he mentioned that residual stresses would typically be impactful when they are $> 30\%$ of the yield strength of the alloy. In this case, 30% of the σ_y is 480 MPa which is higher than the measured results.

Table A- 1. The μ -X360 FULL 2D conditions used for the measurement of the residual stress in the M789 alloy

Measuring device	μ -X360 FULL 2D Portable X-ray Residual Stress Analyser		
X-ray tube target	Cr-K α	Diffraction surface	Fe(α)
X-ray tube voltage	30 kV	X-ray tube electricity	1.0 mA
Measuring method	Single incident angle method	X-ray incident angle	35.0 deg
Exposure time	30 sec	Collimator spot size	1.0 mm

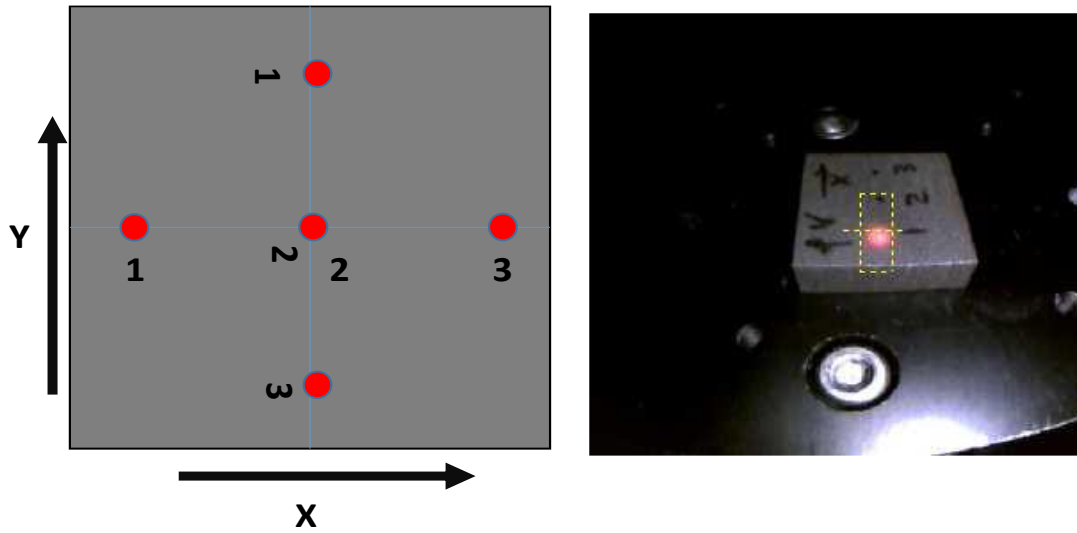


Figure A-1.(a) The specimen marked with the BD (Y) and the PD (X), and (b) the specimen in the μ -X360 chamber.

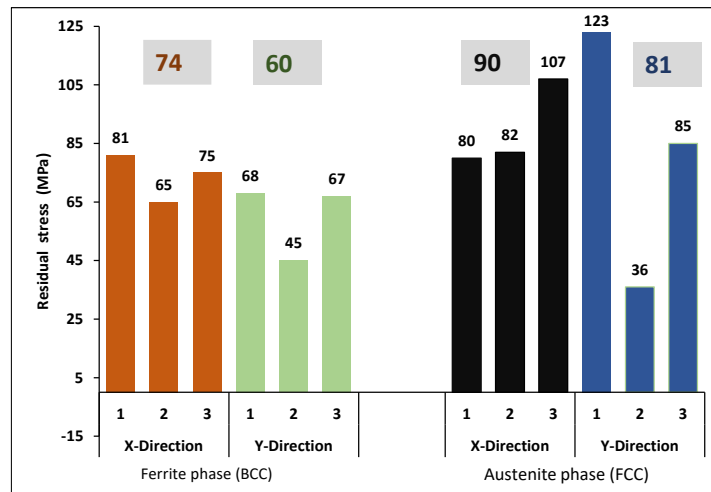


Figure A- 2. Residual stress measurement of the M789 alloy in both the ferrite and austenite phases.

A.2 Pit production and nondestructive characterisation

An in-house-designed electrochemical cell was used to induce pitting on the fatigue specimens. A 3.5 wt% NaCl solution served as the electrolyte. A saturated calomel electrode (SCE; $E = + 241$ mV vs SHE at 25°C) was employed as the reference electrode (RE), and a platinum electrode served as the counter electrode (CE). Pitting was performed on specimens designated **FC#**.

Step 1: Surface preparation:

After polishing as per Section 4.3.5.1, the FC specimens were subjected to a brief chemical cleaning step using 5% hydrochloric acid (HCl) for ten seconds. The polishing may leave behind residual oxide films, embedded abrasive particles, and organic contaminants from the polishing suspension. These surface residues can interfere with corrosion measurements by altering the initial electrochemical response and reducing reproducibility between tests.

A.2.1 Step 2: Isolation of the pitting area

A 1 cm^2 area on the specimen surface was isolated using Gamry tape to define the region for electrochemical pitting. The remaining exposed surfaces were then coated with a 3M electrical insulating sealer and allowed to dry for 15 min (see Figure4-13a). Subsequently, the specimen was affixed to the base of the electrochemical cell, the interface was sealed with a silicone coating, and the assembly was left to cure for 48 h (see Figure4-13b).

A.2.2 Step 3: The pitting process

Once the silicone had fully cured, the Gamry tape was removed to expose the 1 cm^2 test area for the pitting experiment, and the cell was fully coupled as shown in Figure4-13c. The electrochemical cell is filled with the 3.5% NaCl solution to a level sufficient to cover the exposed surface of the specimen and the CE, and the RE coupled to it. Finally, the whole setup is connected to a potentiostat controlled by an EC-Lab v11.43 (Bio-Logic), which helped to apply the scanning voltage used for the pitting. The resulting setup is a case where only a 1 cm^2 area of the alloy is exposed to the NaCl solution for corrosion. A schematic sketch of the entire setup is shown in Figure A- 3.

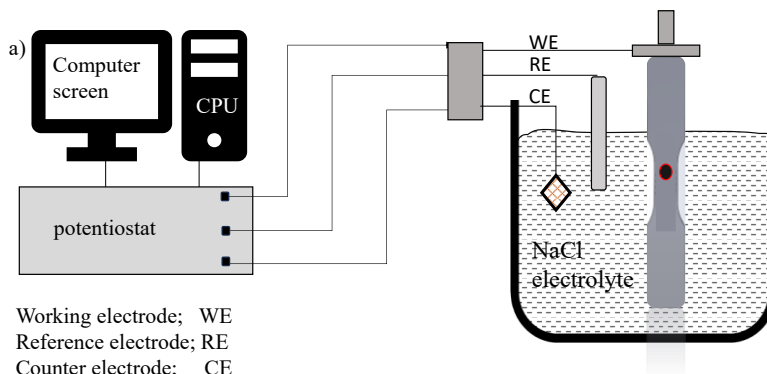


Figure A- 3. Schematic representation of the electrochemical set-up connected to the potentiostat.

The measurements acquired during the electrochemical process are described as follows:

A.2.3 Measurement of open circuit potential (OCP)

The electrochemical experiment commenced by running the open-circuit potential E_{ocv} for 2 hrs to ensure the alloy's free potential was established in the NaCl electrolyte before applying external polarisation. This provided information on the alloy's susceptibility to corrosion in the tested environment while also ensuring a stable electrochemical state is attained before any imposition of potential. An example of the recorded evolution of the potential with time is shown in Figure A- 4. The potential is seen to converge towards -0.08 V, signifying that a near steady state free potential had been reached in the NaCl solution.

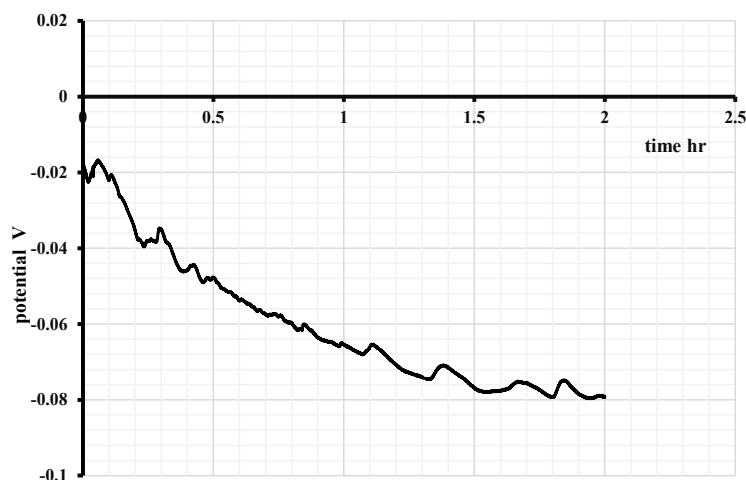


Figure A- 4. Open circuit potential data obtained for one of the specimens.

A.2.4 Cyclic potentiodynamic polarisation (CPP) of the M789

Then, the next step was the cyclic potentiodynamic polarisation technique used to apply potential to the setup. This technique, while being able to produce pits on the specimen, also provided information on the pitting potential ' E_{pit} ' of the alloy. The potential application was applied to sweep from 0.30V below the stabilised open-circuit potential and progressing to 2.0 V (vs. the reference electrode) at a rate of 0.167mVs^{-1} . The scheme used during the CPP process is shown in Figure A- 5, and the corresponding experimental data obtained are presented in Figure A- 6.

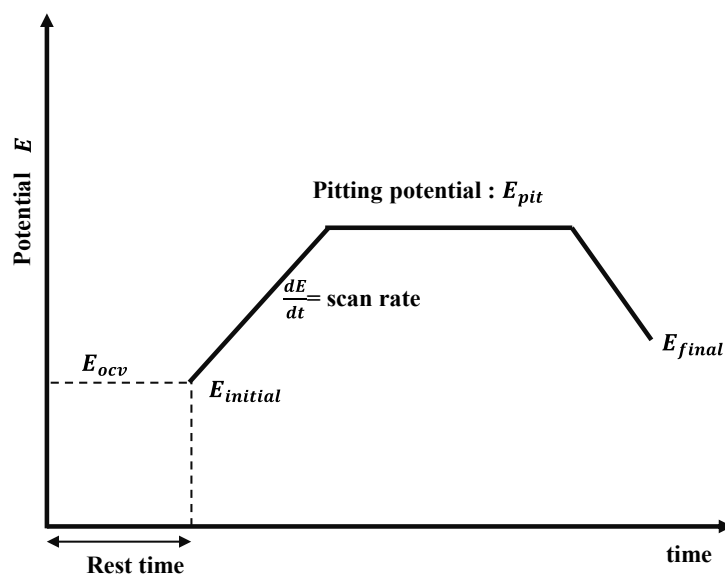


Figure A- 5. CPP scheme of potential scan for electrochemical pitting.

The illustration shows the scan starts from the open circuit potential and scans to a maximum of an imposed potential limit, after which it is held there for some time, during which the pitting occurs, and finally begins to descend. The scan limits were chosen to scan past the pitting potential (E_{pit}) sufficiently. The current limit was set to 10 mA to avoid the destruction of the specimen. It can be seen that the E_{ocv} is 0.08 V, and the E_{pit} is 1 ± 0.11 volts. These results are consistent with earlier corrosion studies performed by Laliberte Riverin et al.[131]. The pitting process was conducted on each side of the specimen, adopting the described protocol. Following the conclusion of the pitting process, the pits are characterised by taking note of the most severe pit with the largest $\sqrt{area}$ perpendicularly projected in the direction of testing.

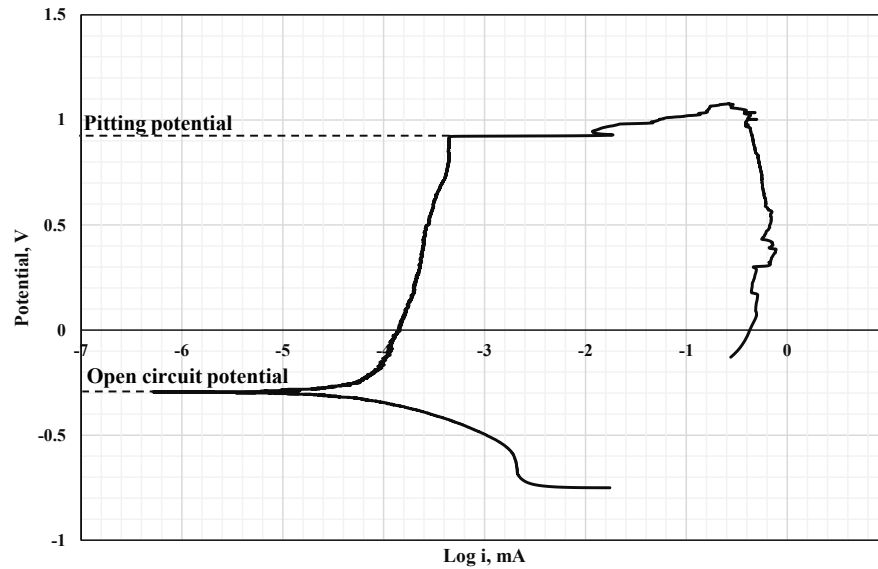


Figure A- 6. Result of the CPP performed on the M789 for different specimens.

However, a second electrochemical technique was employed to allow the generation of a variety of pit geometries. The insights from the CPP technique, particularly the E_{pit} was used as a guide in implementing the potentiation pitting process. Not that this procedure was not for comparison of technique, but rather just for the sole purpose of altering the variety of pits.

A.3 Fatigue results of all specimens

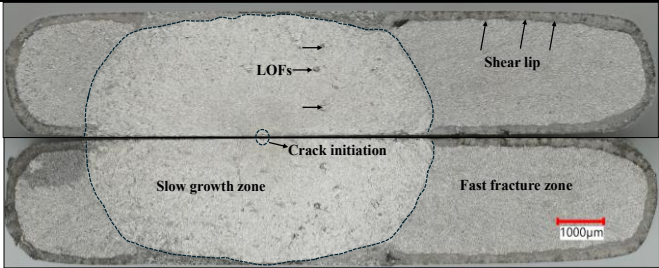
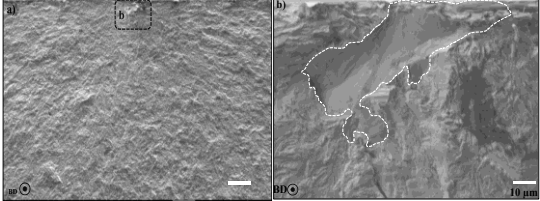
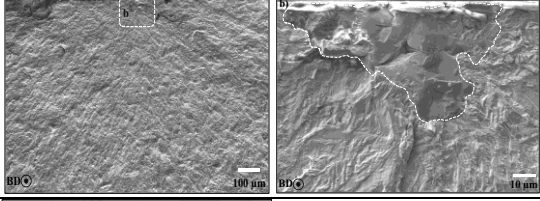
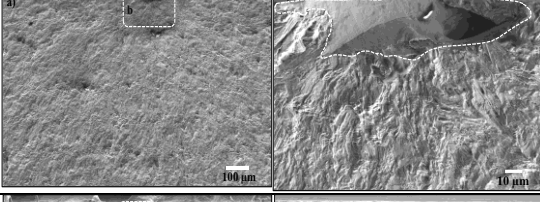
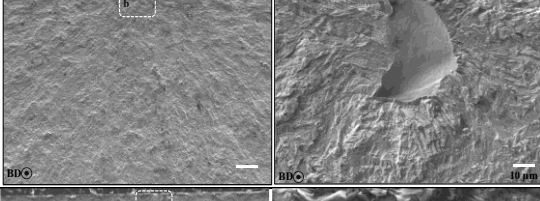
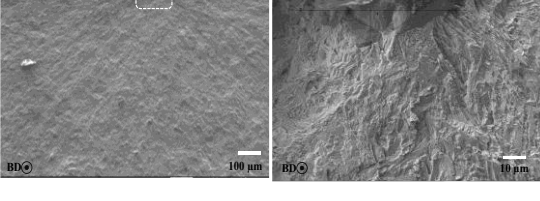
In this section, the results of the fatigue test for all the tested specimens are shown in Table A- 2 and reflect the stress amplitude of the last step before failure ($\sigma_{a,n-1}$), the number of cycles sustained at the last step (N_{fail}), the computed fatigue strength ($\sigma_{w,Nlife}$), the $\sqrt{area}$ of the discontinuity at the failure origin, the type and location, while Table A- 3-6 shows all fractographic results of specimens. The results for FA, FP and FC specimens are used for Article 1 of Chapter 4, while that of the FIC specimens is discussed in Chapter 5.

Table A- 2. Fatigue test results obtained for all specimens showing the computed fatigue strength, size type and location of discontinuities at the failure origin

ID	$\sigma_{a,n-1}$ MPa	N_{fail}	σ_{FS} $= \sigma_{w,Nlife}$ MPa	$\sqrt{area}$ μm	Type of discontinuity	Location
FP02	180	1.2E6	192	59	LOF	SS
FP01	240	2.0E5	242	46	LOF	SS
FP03	240	1.0E6	250	51	NMI	SS
FP05	280	1.0E5	281	44	NMI	SS
FP04	280	2.0E5	282	42	LOF	SS
FA01	120	2.0E5	122	117	Trough	S
FA04	120	1.4E6	134	95	Trough	S
FA05	140	3.0E5	143	170	Trough	S
FA02	140	4.0E5	144	158	Trough	S
FA03	160	5.0E5	165	202	Trough	S
FC05	80	1.3E6	93	260	Pit	S
FC02	80	1.9E6	99	361	Pit	S
FC03	100	3.0E5	103	643	Pit (combined)	S
				465	Pit(isolated)	
FC01	140	1.2E6	152	123	LOF	SS
FC04	220	1.6E6	236	226	Pit	S
FIC02	60	1.5E6	75	1840	Trough	Channel
FIC03	80	2.5E4	80	NA	Trough	Out channel
FIC01	80	1.8E5	81	NA	Trough	Out Channel
FIC04	80	8.4E5	88	2865	Trough	Channel
FIC05	80	2.0E5	89	2865	Trough	Channel

A.3.1 Fractography of polished specimens

Table A- 3. Fractographic images of FP specimens showing the general failure surface and the respective failure origin of each specimen.

Specimen ID	Fractography	$\sqrt{area} \mu m$
		
FP01		46
FP02		51
FP03		59
FP04		42
FP05		44

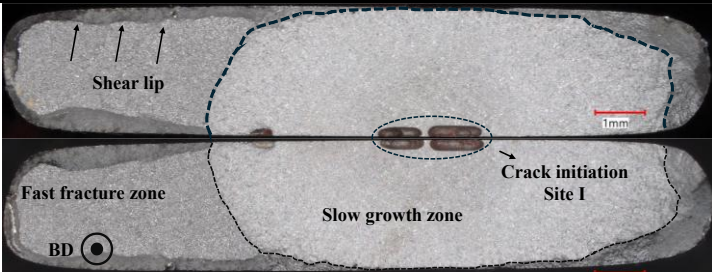
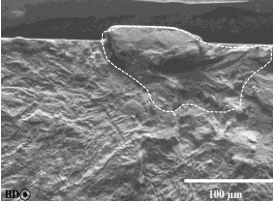
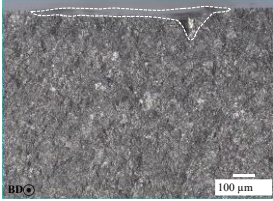
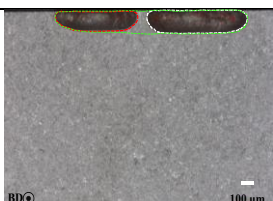
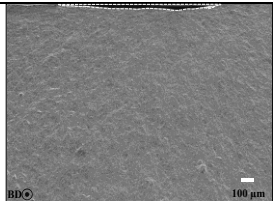
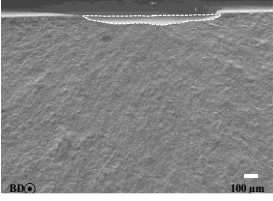
A.3.2 Fractography of as-printed specimens

Table A- 4. Fractographic images of FA specimens showing the general failure surface and the respective failure origin of each specimen.

Specimen ID	Fractography	$\sqrt{area}$ μm
FA01		117
FA02		158
FA03		202
FA04		95
FA05		170

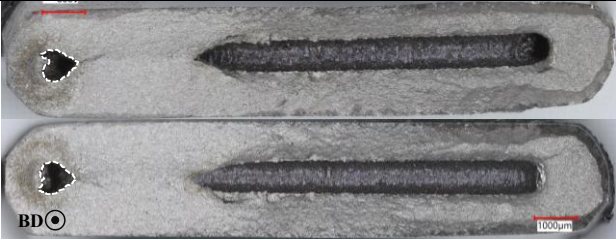
A.3.3 Fractography of corroded (FC) specimens

Table A- 5. Fractographic images of FC specimens showing the general failure surface and the respective failure origin of each specimen.

Specimen ID	Fractography	$\sqrt{area}$ μm
		
FC01		123
FC02		361
FC03		643 465
FC04		226
FC05		260

A.3.4 Fractography of specimens with channels (FC) specimens

Table A- 6. Fractographic images of FIC specimens showing the general failure surface of each specimen.

Specimen ID	Fractography	$\sqrt{area}$ μm
FIC01		NA
FIC02		2865
FIC03		NA
FIC04		2865
FIC02		1840

A.4 Estimation of stress concentration of pit on FC04-pit

The fractographic image for the pit of FC04 is shown in Figure A- 7 .A depression can be seen at the failure origin below which a crystallographic facet is present. This was triggered by the high stress concentration associated with the location and can be the critical part of the pit that drove failure. The influence of the stress concentration was considered as the reason for its emergence as an outlier in the data on the KT diagram. To this end, it was appropriate to estimate the associated stress concentration using FEA analysis.

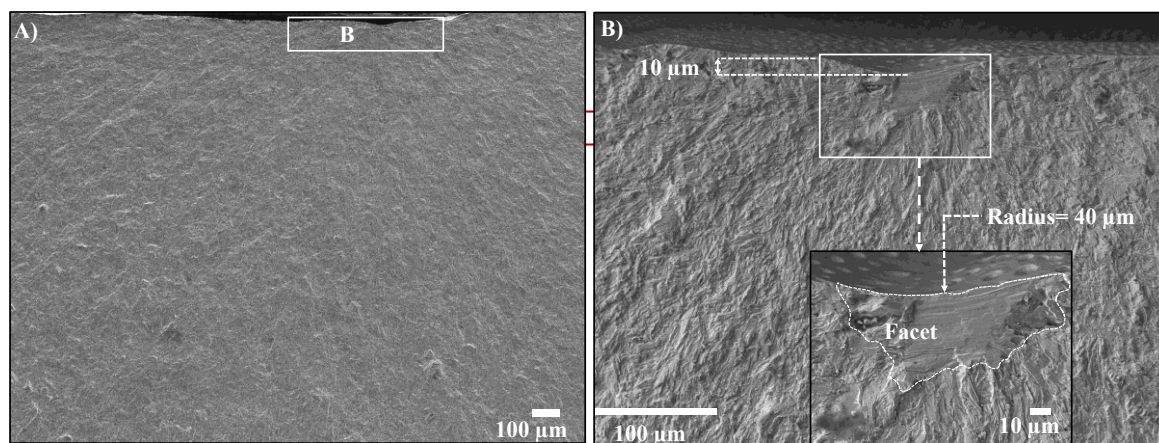


Figure A- 7. SEM image of pit FC04 at failure origin, (a) Failure origin from the pit, (b) Higher magnification image of the failure origin showing the pit dimension.

The finite element analysis (FEA) was carried out using Abaqus CAE software. The material properties used for the simulation are Young's modulus of 200 GPa and a Poisson ratio (ν) = 0.3. A half-symmetry of the specimen's thickness was used for the simulation to reduce the number of elements and the computation time of the simulation. The model was simulated, ensuring a rigid condition while a load of 1 MPa was applied at the opposite end in the Z-direction. A tetrahedral element shape with an interpolation order of C3D10 was employed in the model. The total number of elements in the model was 111496, and the number of nodes was 166383 with a global size of 3 mm and a fine mesh of 3 μ m. A convergence study was performed to ensure that the results are consistently optimised, considering time and computing requirements, so that the final results are unaffected.

The simulation results of the pit are presented in Figure A- 8, and the stress field is shown in Figure A- 9. The stress concentration is situated at the stress raiser in the pit. The stress concentration “ k_t ” obtained is 2.06 with a stress field exponent of 0.16.

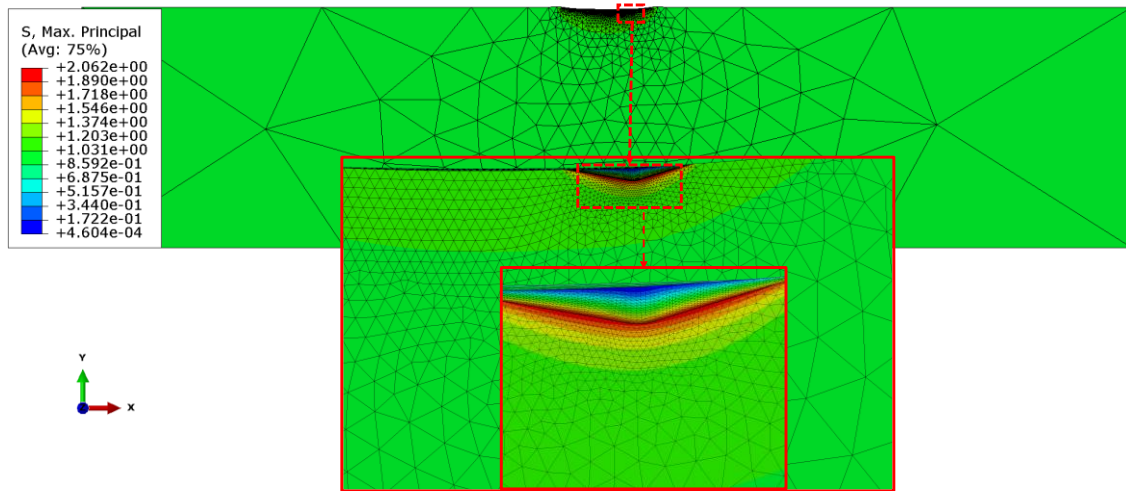


Figure A- 8 FEA result of the simulation of specimen FC04 showing the maximum stress concentration location.

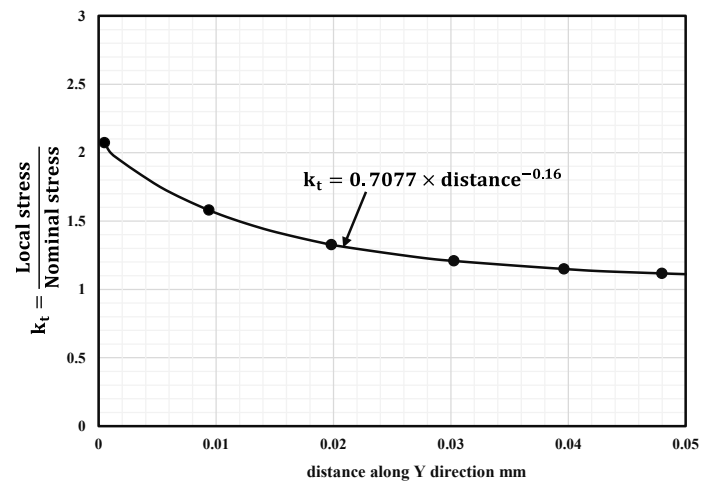


Figure A- 9. Plot of the stress field gradient associated with the maximum stress concentration.

A.4.1 Estimation of stress concentration of internal channel of FIC specimens

The FEA simulation result for the FIC specimen was performed in Section 3.9, where a k_t of 2.89 was obtained as seen in Figure 3-7.

A.5 Local stress estimation for FC04 and FIC specimen

The local stress $\sigma_{f,local}$, at the root of the pit and the channel can be estimated using the relation given in Equation 6-1.

FC04

For the FC04 specimen, the returned local stress was 462 MPa and is associated with a failure arising from the microstructure.

FIC specimens

To estimate the local stress associated with the roughness asperity, we adopt the approach of Pessard et al. [99] where they estimated the local stress at a discontinuity at the root of a notch using the projected value of stress concentration from the notch stress field. In the case the FIC specimen, the channel wall lining is in the as-printed condition, and the failure origin is characterised by a roughness trough (see Figure A- 10). The roughness asperity is treated as a discontinuity in the vicinity of the channel root. The $\sqrt{area}$ of the asperity is obtained using Equation A- 1.

$$\sqrt{area} = Asperity\ depth \times \sqrt{10} \quad \text{Equation A- 1}$$

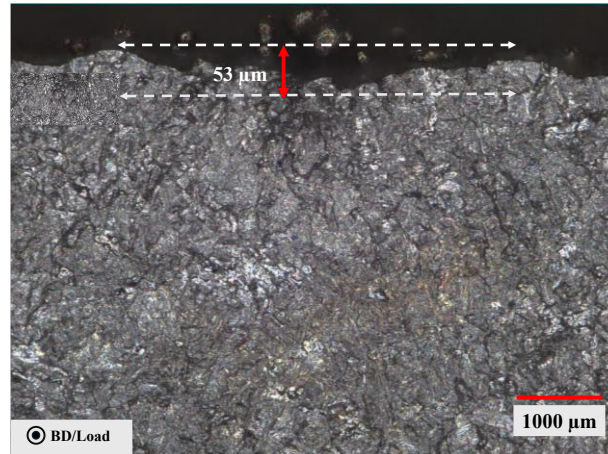


Figure A- 10. Measurement of the roughness trough at the failure origin of the internal channel for specimen FIC05.

Thus, the stress distribution plot for the channel is used to quantify the stress concentration factor $k_{t,R}$ for a position equivalent to the $\sqrt{\text{area}}$ of the trough at the failure origin. This estimation is illustrated graphically in Figure A- 11, where the result of each FIC specimen is identified on the plot. Thus, associated $k_{t,R}$ for specimen FIC02, FIC04 and FIC05 are 1.35, 1.54, and 1.63 respectively.

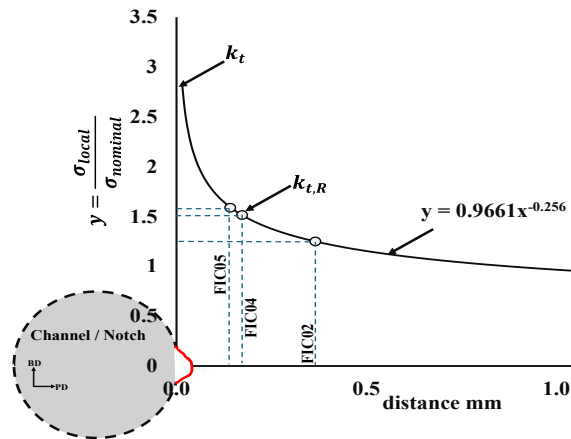


Figure A- 11. Stress distribution in depth for the internal channel, showing the estimation of the $k_{t,R}$ for the FIC specimens.

A.6 Atzori- Lazzarin diagram

The experimental results are also plotted using the Atzori-Lazzarin diagram. To realise this, we use the results of only the specimens that were off the KT diagram trend, i.e. FC04 and FIC specimens. Firstly, the stress field exponent is retrieved from the FEA analysis previously performed for the internal channels (See 3.9) and the FC04 pit (See Figure A- 8). The retrieved value is subsequently input into Equation 7-1 proposed by Atzori et al. [90] to compute β which is a parameter dependent on notch opening angle. Thereafter β is inputted into Equation 7-1 to compute the notch stress intensity threshold $\Delta K_{I,th}^Y$ using Equation 2-21. Note that β parameter in the equation is estimated using the Equation 7-1 proposed by Atzori et al. [90].

A.7 Calculation of Y-factor of internal channel (FIC) and FC01 pit

Y-factor for internal channel (FIC)

The internal channel is considered to be a through all hole along the width of 11.6 mm with its diameter of 0.75 mm centrally located in the 2.6 mm thickness of the specimen, and the tensile load is applied axially (see Figure A- 12). The shape factor, Y , is calculated as 1.05 using Equation A- 2 proposed by Tada et al.[160].

**APPENDIX B APPENDIX OF ARTICLE 1: DISCONTINUITY-
CONTROLLED FATIGUE PERFORMANCE AND MODEL
ASSESSMENT FOR LPBF-M789 MARAGING STAINLESS STEEL
WITH DIFFERENT SURFACE CONDITIONS**

B.1 Tensile plot of horizontal and vertical specimens

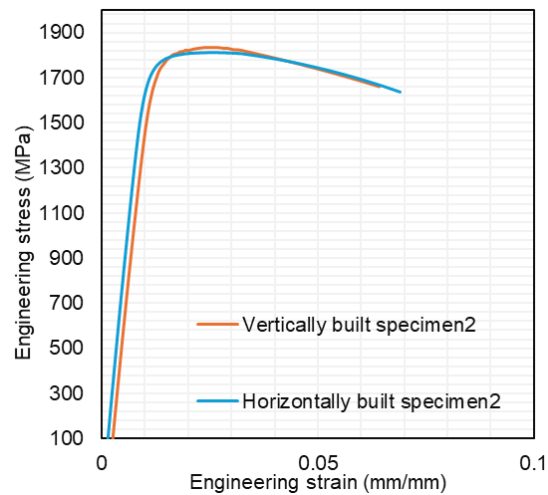


Figure B- 1. Engineering stress strain plot of LPBF M789 specimens for horizontal and vertical orientation

B.2 Determination of fatigue crack growth of the M789 alloy

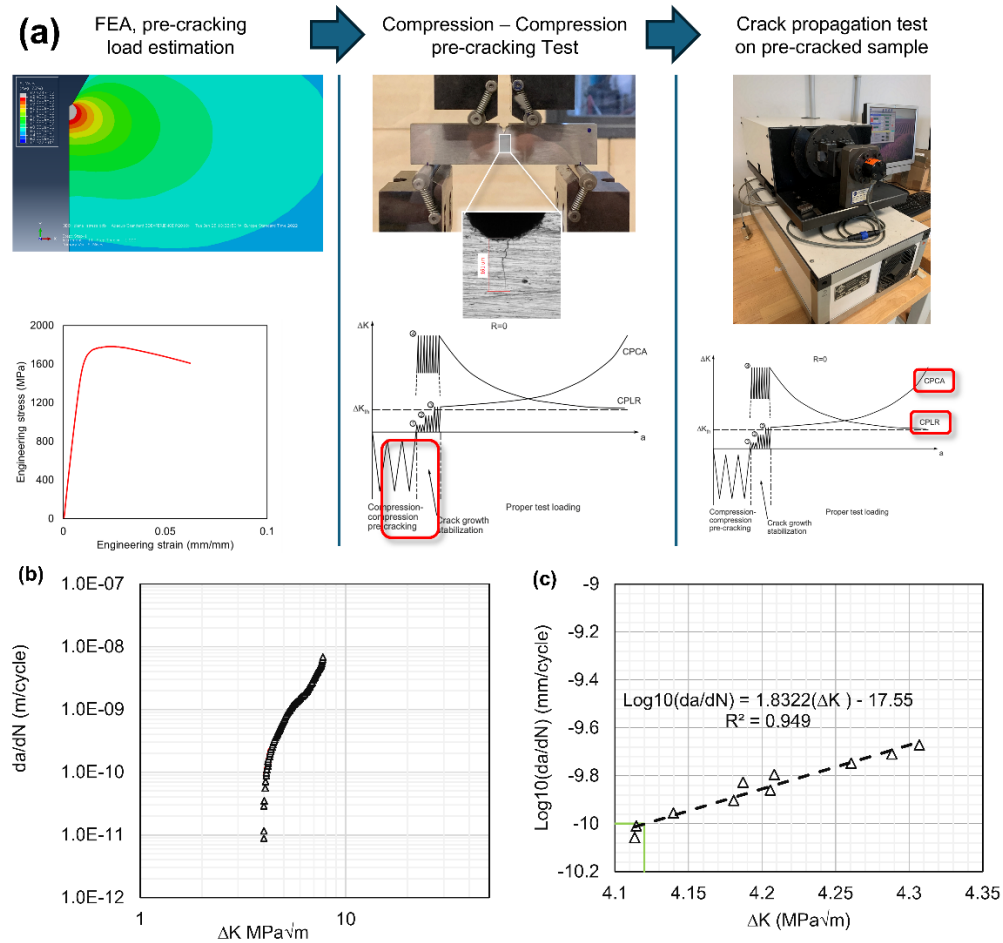


Figure B- 2. Steps used of the fatigue crack growth test to estimate the varying long crack threshold.

B.3 Cyclic potentiodynamic polarisation (CPP) curve of the M789 alloy

The result of the CPP plot for the M789 is shown in Figure B- 3. The plot shows the pitting potential of the alloy to be 0.9 V, while the open-circuit potential of the alloy in the 3.5% NaCl solution is -0.25 V.

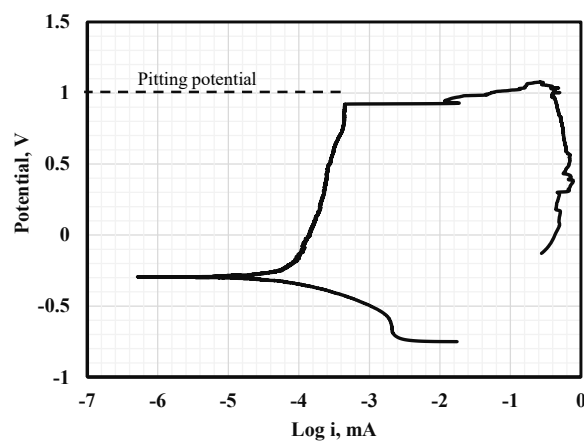


Figure B- 3. CPP plot of the M789 fatigue specimen obtained from electrochemical pitting.

B.4 Measurement of the largest discontinuity in cubic specimen

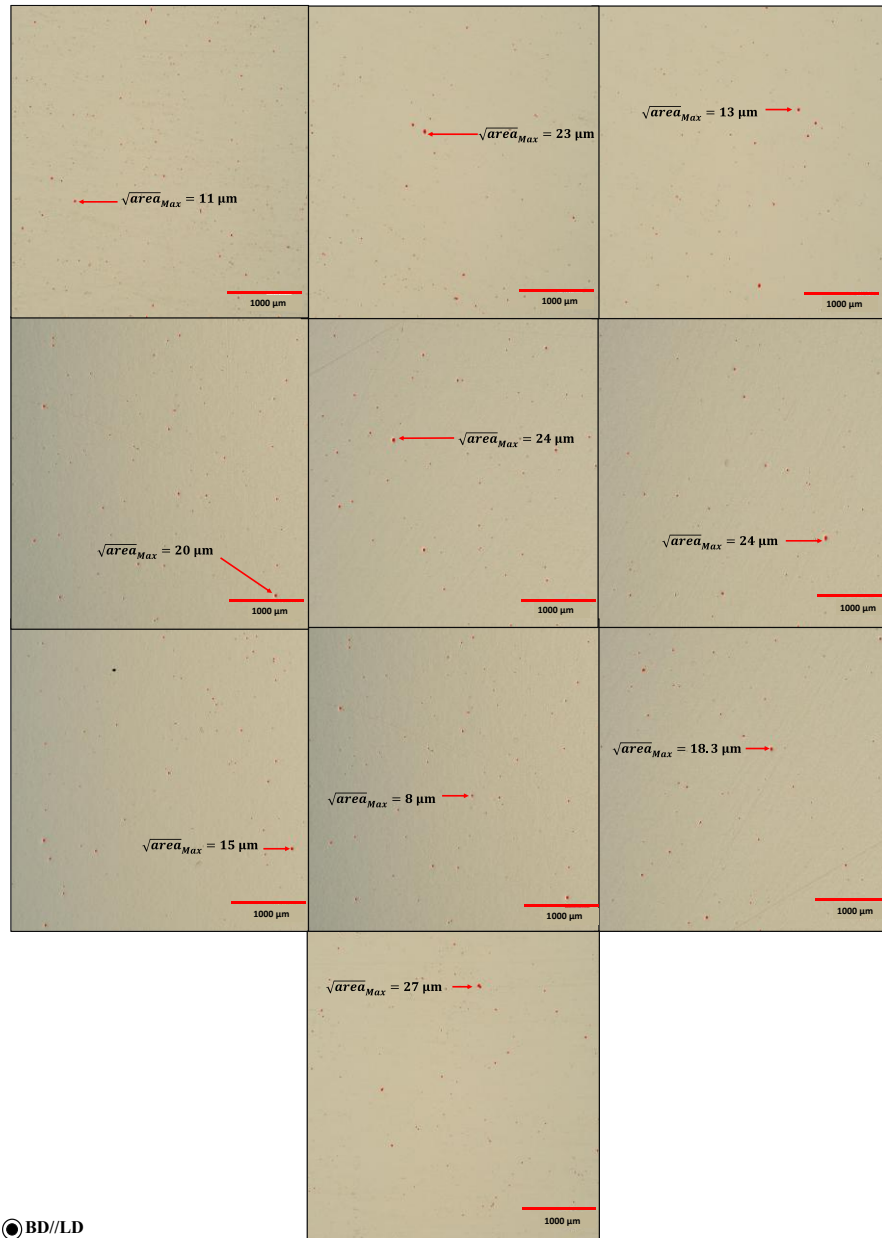


Figure B- 4. Images of the inspected areas showing the identification of the largest $\sqrt{area_{max}}$ discontinuity.

B.5 Statistics of extreme value plots for polished and as-printed specimens

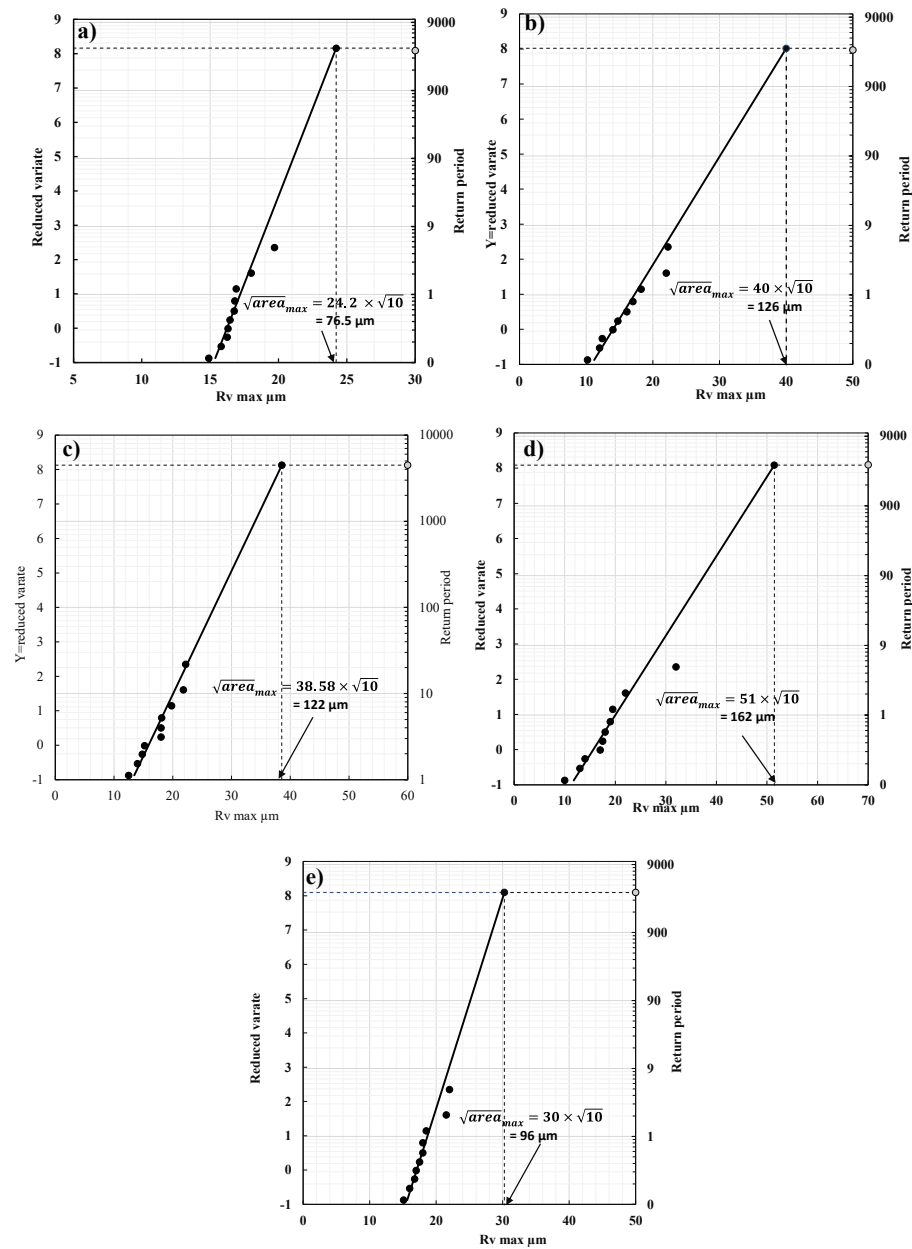


Figure B- 5. EVS predictions of the largest discontinuity in as-printed fatigue specimens sequentially shown. (a) FA01, (b) FA02, (c) FA03, (d) FA04, (e) FA05.

B.6 Fracture surface of an as-printed specimen with a facet

The fractographic image of specimen FA01 is given in Figure B- 6 a), while the image in b) shows a close-up image with evidence of a facet encircled in a black line. The size of the facet is 62 μm .

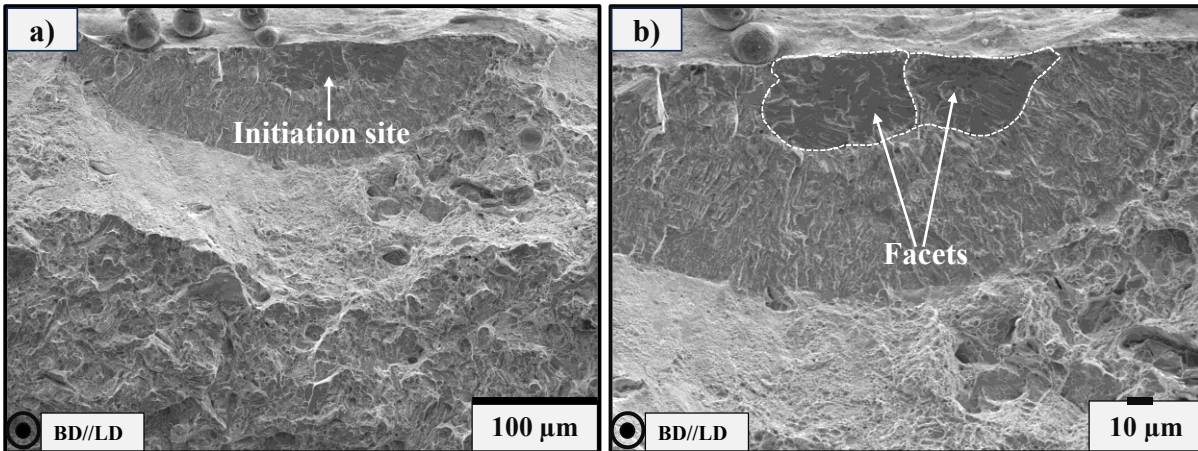


Figure B- 6. Failure origin of specimen FA01: (a) crack initiation site, (b) close-up image showing crystallographic facet.

B.6.1 Fractography of FC04

The pitted surface of specimen FC04 is shown in Figure B- 7. Prior to fatigue testing of the specimen, the pit labelled '1' was predicted to be the critical discontinuity to cause failure along the plane marked in red because its $\sqrt{\text{area}}$ of 221 μm and its depth of 57 μm is larger than the 162.53 μm and 38.5 μm of the pit labelled '2' respectively. A value of 236 MPa was obtained from the fatigue test with failure emerging from pit 2 as expected. An optical and SEM image of the fracture surface are presented in Figure B- 7b. Figure B- 7c shows the presence of a crystallographic facet at the failure origin. A close-up SEM image is given in Figure B- 8 in where the facet is clearly seen and its $\sqrt{\text{area}}$ measured to be 52 μm which is comparable to the large parent austenite grains of the alloy.

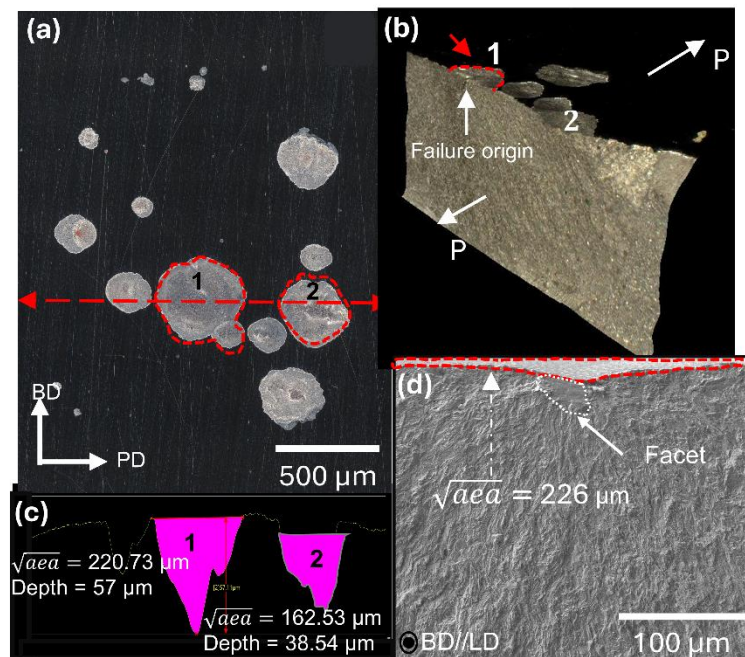


Figure B- 7. FC04 pits show (a) the prediction of the critical pit before the test, (b) a side view of the pit from the failure surface, (c) the characterisation of the pit before the test and (d) an SEM image of the failure origin from the pit.

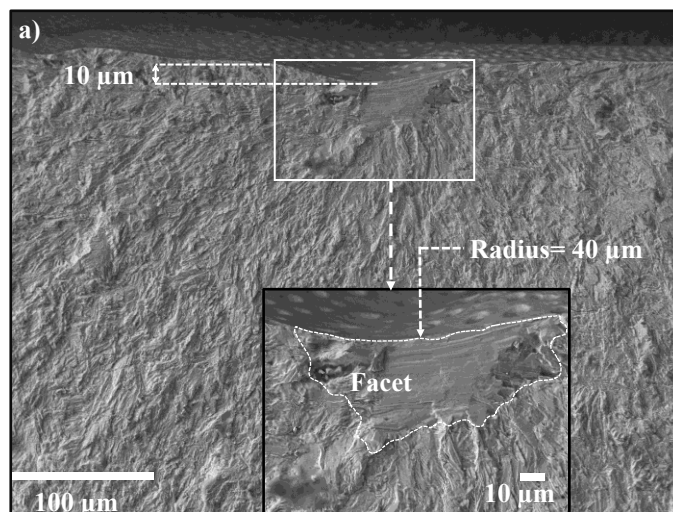


Figure B- 8. SEM image of FC04 pit at higher magnification

B.6.2 FEA simulation of stress concentration of pit FC04

The complex and irregular geometry of the pits makes predetermining the stress concentration k_t challenging before testing using FEA. To simplify this, FEA analysis was only performed after fatigue testing for specimens that showed behaviour that indicated the possible influence of stress concentration k_t on the fatigue strength. This was the case for specimen FC04, which had its fatigue strength from testing to be 236 MPa, which is higher than the LEFM prediction of 1176 MPa. The pit shape from the fractography in Figure B- 8 is adapted for the simulation, leveraging the depth and radius.

A convergence study was first performed, after which a C3D10 element type was chosen with a global mesh size of 7000 μm , 1 μm spacing around the pit and 9 μm away from it. The results of the simulation are presented in Figure B- 9 and the stress field distribution in Figure B- 10. The stress concentration is situated at the stress raiser in the pit. The maximum k_t obtained is 2.06 with a stress field exponent of 0.16. Applying notch theory, the fatigue strength is expected to be the peak fatigue strength of 550 MPa divided by 2.06, which yields 264 MPa. This value is closer to the 236 MPa obtained from the fatigue testing and is more accurate than the 117 MPa obtained from the LEFM prediction.

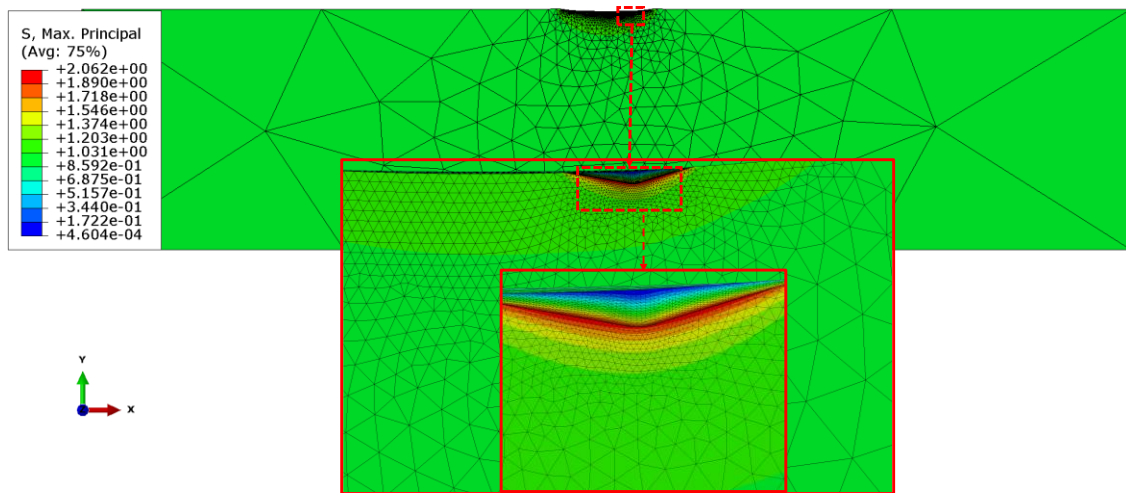


Figure B- 9. FEA result of the simulation of specimen FC04 showing the maximum stress concentration location.

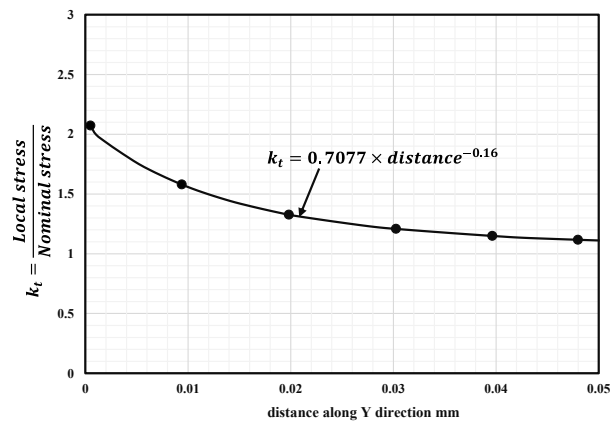


Figure B- 10. Plot of the stress field gradient associated with the maximum stress concentration.