HIGH STRAIN-RATE MATERIAL RESPONSE: IMPACT LOADING
OF TRADITIONALLY AND ADDITIVELY MANUFACTURED
INCONEL 718 SUPERALLOYS

by

RUSSELL ROWE
KEIVAN DAVAMI, COMMITTEE CHAIR
ANTHONY PALAZOTTO, COMMITTEE CO-CHAIR
KASRA MOMENI
SADIE BECK
ALEXEY VOLKOV
XIN WANG

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ABSTRACT

This project aimed to evaluate the mechanical properties of additively manufactured (AM) components in comparison to ones fabricated via traditional manufacturing methods. AM components were compared with traditionally manufactured (TM) components under quasi-static and high strain rate compressive loads. Additionally, high strain rate shear experiments were conducted on both AM and TM specimens and the resultant adiabatic shear bands (ASB) were compared to determine if there were any differences in shear localization susceptibility. Finally, AM nickel-based superalloy lattice structures were designed, printed, and treated with a novel heat treatment process to study how post-processing affects the specific energy absorption capabilities of these novel structures under high strain rate loading.

The research began with an exhaustive review of ASBs and the conditions required for them to form in nickel-based superalloys. A top-hat shear sample geometry was implemented to localize shear and form ASBs. Due to the extreme conditions needed for ASBs to form, high-strain rate experiments using a split Hopkinson pressure bar (SHPB) were conducted and a range of strain rates and strain were tested to determine the critical strain for the formation of the ASBs.

High strain rate shear experiments, using an SHPB, were successful in generating ASBs at shear strain rates ranging from 100,000 to 140,000 s\(^{-1}\) in the AM and TM specimens when fraction shear strain values reached 2.47 and 4.28, respectively. Transmission electron microscopy of the ASBs revealed dynamic recrystallization to occur at the center of the bands.
The effect of a novel heat treatment process on the mechanical properties of nickel-based superalloy lattice structures was carried out using quasi-static and high strain rate compressive experimentation. Three different triply periodic minimal surfaces (TPMS) designs were used: Gyroid, Primitive, and I-WP. All designs experienced an increase in yield strength, elastic modulus, toughness, and specific energy absorption following the heat treatment. However, the I-WP design received the most significant increase of 65.2% in quasi-static yield strength and 86.7% in specific energy absorption. It was concluded that the mechanical attributes of the AM lattice structures can be adjusted not only by optimizing their design and the materials they are made of but also through post-processing.
DEDICATION

I dedicate this dissertation to the unwavering support and love of my cherished family and loved ones, whose encouragement has been my guiding light throughout this journey. To my parents, Russell and Selena, whose unwavering belief in my abilities supported me to the end of this journey. Your unflinching support and encouragement have been the cornerstone of my academic pursuit, and for that, I am forever grateful. To my grandparents, Nana, Pawpaw, and Mimie, this would not have been possible without your constant belief.
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CHAPTER 1: INTRODUCTION

1.1 Motivation

Additive manufacturing of metal components is a rapidly evolving fabrication process that has made large strides since the late 1990s. Several advantages distinct to AM include the fabrication of novel materials, which are difficult to fabricate using traditional manufacturing methods, and the creation of complex geometries, such as those with an optimized strength-to-weight ratio [1]. The versatility of AM allows it to be used in the repair of TM parts as well as manufacturing or repairing at the point of need [2]. Civilian and military supply chain efficiency can also be improved using AM by eliminating the need for specialized tooling, reducing material waste, and lowering fabrication times [3].

The versatility of being able to print complex geometries using additive manufacturing is of great importance to the aerospace industry. The engine of the Vulcan 2 rocket, called the Ignus-II, was entirely built out of Inconel 718 using additive manufacturing. 95 variable area regenerative cooling channels are used to cool the combustion chamber which normally would have to be attached post-fabrication, but thanks to the versatility of 3D printing the cooling channels can be printed at the same time as the combustion chamber with no post-fabrication attachments required [4], [5]. GE is using additive manufacturing to consolidate the fabrication of parts and components with highly complex internal structures so that they can manufacture the largest high-bypass turbofan aircraft engine with 19 3D-printed fuel nozzles, which is 10% more
efficient than its predecessors[6]. With the improvements in 3D printing, Boeing expects to save $2-3 million per plane built thanks to the increased productivity [6].

Evaluating the discrepancies in mechanical properties for additively and traditionally manufactured components is essential before additive manufacturing can replace traditional manufacturing. High strain rate loading affects the mechanical properties and the microstructure of metals and alloys by thermomechanical instability. Although previous studies have thoroughly documented the dynamic impact properties and microstructural evolution of traditionally manufactured nickel-based superalloys, little work has been reported on additively manufactured Inconel 718 [7], [8], [9].

Adiabatic shear bands (ASB’s) are important failure mechanisms seen in many metals typically used in high strain rate environments including nickel alloys [10], [11], [12], titanium alloys [13], [14], [15], and steels [16]. Abrupt and/or premature failure in metals during high-rate deformation can often be caused by the formation of ASBs [17]. The conditions for the formation of ASBs are not completely understood, however, a prerequisite for their formation includes a large amount of plastic deformation in a rapid time, i.e., high strain rate applications. In metals, when there is a large amount of deformation, much of the work done is converted into thermal energy. Therefore, the material is experiencing both strain or strain rate hardening and thermal softening simultaneously. If the increase in temperature is faster than the heat diffusion away from the deformed region then the strain or strain rate hardening is overpowered by thermal softening and the material will experience a large amount of localized shear deformation, i.e., an ASB forms. Steels and titanium alloys are more prone to ASB formation in comparison to nickel alloys [17], [18].
1.2 Objectives

The critical technical barriers for this research lie in the lack of information about how ASBs form in additively manufactured Inconel 718 and the stages of shear band formation, crack initiation, propagation, and coalescence in the shear region. Though ASBs are a common occurrence in steel and titanium alloys, their formation in nickel and its alloys requires much higher strain rates. Comparing the mechanical properties and ASB formation of additively and traditionally manufactured Inconel 718 in the dynamic regime will also be conducted.

The objective of this study is to investigate the mechanical properties and effects of shear localization on additively manufactured Inconel 718 in comparison with traditionally manufactured material. This study involves 1) a literature review of ASB formation for nickel-based superalloys, 2) quasi-static and high strain rate compressive testing and dynamic shear band formation characterization of AM and TM Inconel 718, 3) quasi-static and high strain rate compressive characterization and energy absorption capabilities of AM Inconel 718 lattice structures. The outcome of this research will identify the dependence of the high strain rate response of widely used superalloys on the manufacturing technique. It will also provide the knowledge and serve as a foundation for future work in optimizing AM techniques, product design, and modeling the high strain rate behavior.

The significance of this project can be examined from several points of view. First of all, impact loading is implemented intentionally in many engineering applications. Just to name a few, key manufacturing processes such as forming, welding, cladding, and compaction of metal and ceramic powders all benefit from high strain rate loading. Therefore, a better understanding of this process will lead to heightening the knowledge necessary for the modification or prediction of the outcome of these manufacturing processes, as well as product design. In
addition to employed manufacturing techniques, catastrophic events, such as explosions, can result in high-rate shock loading conditions. In such scenarios, along with the propagation of a wave through the materials, there is another crucial aspect which is the attributed high strain rate during material deformation. The latter dictates not only the microstructure evolution characteristics and possible phase transitions of the impacted material but also influences the mechanical properties of the specimen. The study of the wave propagation and the microstructure changes in impact loadings helps with a better understanding of the response of materials to high rates of loading and, more importantly, serves as a basis for the prediction of dynamic material properties. The other significance of this work is associated with the lack of knowledge of the high strain rate material properties of additively manufactured structures. Without robust characterization of newly discovered or introduced materials, their effective application and widespread adoption will certainly be limited or at least face unprecedented challenges. This research will offer the knowledge to fill this gap. Systematic and integrated experimental investigation of the phenomenon of shear-band formation in additively manufactured materials that will be carried out for the first time in this research will lead to the prevention, containment, and control of shear instability. This will help with the prediction of its effects on the strength and failure modes of a wide range of materials. Finally, while there are several analytical and numerical models that are used to simulate the high-rate loading response of materials and successfully predict the stress-strain response under a set of prescribed loading conditions, the absence of explicit microstructure parameters remains a challenge and limits their extrapolation capability and accuracy. The microstructure analysis that will be conducted in this research and the correlation that will be drawn with the mechanical properties of the part will provide the necessary input for these models.
1.3 Background

1.3.1 Inconel 718

Superalloys are most notably characterized by their advantageous material properties at extreme temperatures making them ideal for applications in the aerospace [19], automotive [20], and power industries [21]. Inconel 718 is a nickel-chromium superalloy with high strength and corrosion resistance containing significant amounts of Fe, Co, and Mo with lesser amounts of Al and Ti [21], [22]. Specifically, Inconel 718 has been used for jet engines [19], power plants [23], and turbine wheels [24].

The primary strengthening mechanism for Inconel 718 is the precipitation of the ordered face-centered cubic $\gamma'$ (Ni$_3$(Al,Ti)) with a DO$_{22}$ structure and metastable body-centered tetragonal $\gamma''$ (Ni$_3$Nb) with a L1$_2$ structure strengthening phases [25], [26]. Post-process solution heat treatment was found to be crucial to the dissolution of the Laves phase, a deleterious phase found in the interdendritic spaces, and MC carbides thereby freeing up large amounts of Nb [27], [28]. This excess Nb is required for the precipitation of the $\gamma''$ strengthening phase which accounts for a major portion of the precipitation strengthening in this alloy, along with the precipitation of the $\gamma'$ phase. Typical heat treatment processes for Inconel 718 includes solutionizing at 1095 °C, aging for 1 hour at 955 °C to precipitate a minor amount of $\delta$ phase, aging at 720 °C for 8 hours and, finally aging at 620 °C for 8-10 hours [25], [29], [30]. Though large amounts of $\delta$-phase (Ni$_3$Nb) will cause “$\delta$-phase embrittlement” [31], [32], [33] reducing the material ductility, small amounts of needle $\delta$-phases are beneficial as they impede grain growth.

1.3.2 Additive Manufacturing of Inconel 718

Significant challenges exist with traditional fabrication of Inconel 718-based components primarily due to the high strength that the material retains at elevated temperatures as well as
difficulty in its machining [34]. High-speed machining of Inconel 718 is costly since using a cutting tool causes significant strain on the tools resulting in them wearing quickly and frequently requiring replacement [35].

Direct Metal Laser Sintering (DMLS) is a subcategory of the laser powder bed fusion additive manufacturing process where a high-power laser is used to fuse small particles of metal powders on the surface of a powder bed, layer by layer, into a mass that has the desired three-dimensional shape. The laser selectively fuses the powdered material by scanning the cross-sections (or layers) generated by a 3D modeling program from a CAD file. After each layer is built, the powder bed is lowered by one-layer thickness, typically less than 100 µm. Then a new cross-section of material is printed on top of the previous layer and the process continues until the part is fabricated [36], [37], [38]. This technique utilizes a variety of metals and alloys including stainless steels [37], [39], metal alloys [40], [41], [42], and nickel-based superalloys [36], [38] to create strong and durable parts for end-use applications and prototypes. The ability to additively manufacture Inconel 718 components with comparable material properties to TM components is of interest to civilian and military manufacturing operations [3], [34], [35], [43].

Fabrication of Inconel 718 using additive manufacturing leads to complex thermal histories for the material which can cause the precipitation of deleterious precipitates, such as Laves phases and δ phases near grain boundaries, which should be dissolved before aging and irregular grain formation [29], [44], [45]. A non-equiaxed microstructure, which indicates inhomogeneous material properties, is also seen after additive manufacturing since the heat flow during printing causes columnar dendritic grains to grow which can cross multiple print layers [29]. During additive manufacturing, the formation of the Nb-rich interdendritic Laves phase occurs in the overlapping regions due to multiple laser passes causing solute segregation of Nb
δ-phase precipitates are seen to replace the γ″ precipitates at temperatures above 750 °C since they have the same chemical composition of Ni3Nb, which can lead to a decrease in ductility [29], [47]. The heat treatment mentioned in the previous section is sufficient to dissolve the Laves phase and MC carbides into the γ matrix. The chemical composition of the Inconel 718 powder used in this investigation is shown in Table 1.

Table 1-1: Chemical composition of Inconel 718 powder: [48]

<table>
<thead>
<tr>
<th>Element</th>
<th>Ni</th>
<th>Cr</th>
<th>Fe</th>
<th>Nb+Ta</th>
<th>Mo</th>
<th>Ti</th>
<th>Al</th>
<th>Co</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt. %</td>
<td>54.05</td>
<td>18.08</td>
<td>17.69</td>
<td>5.32</td>
<td>2.93</td>
<td>0.97</td>
<td>0.45</td>
<td>0.20</td>
<td>0.03</td>
</tr>
</tbody>
</table>

Less than 0.1 wt% of Mn, Si, P, B, Cu, Ca, Mg, O, N

Due to the fabrication process of AM, the microstructure of AM Inconel 718 components is significantly different from components fabricated using TM [20], [30], [46], [49], [50], [51], [52]. Metal AM commonly builds components through the sintering of consecutive layers of material leading to non-uniform thermal gradients that cause several detrimental effects. Namely, the thermal energy from newly deposited layers propagates into the previously deposited material and creates columnar grains [51], anisotropic mechanical properties [30], [51], [53] and deleterious phase precipitation [50], [54]. Columnar grains elongated in the build direction typically form in AM samples made using sintering processes [46], [53]. This causes AM components to have lower yield strength in the build direction, approximately 10% lower for DMLS Inconel 718[52], compared to the yield strength perpendicular to the build direction [30], [51]. As-built AM Inconel 718 components have lower ultimate strength and yield strength compared to wrought Inconel 718, though higher ultimate and yield strength than cast parts [20]. Surprisingly, after proper heat treatment, AM components typically have the same or slightly higher ultimate and yield strengths than wrought components [20], [49]. Deleterious phases, such as the Laves phase, form in the interdendritic regions of AM Inconel 718 components and work
on the presence of Laves phases in wrought Inconel 718 shows that they limit component ductility [50].

1.3.3 High Strain Rate Properties of Inconel 718

Material strain rate dependence affects properties such as strength, ductility, failure mechanisms, and shear band formation. In general, a significant increase in yield strength is seen when increasing the strain rate from the quasi-static regime to the dynamic regime (Figure 1-1) [8], [55], [56], [57]. Another difference is that close inspection of lower strain rate stress-strain curves reveals that the quasi-static, isothermal loading exhibits a steady increase in the load-carrying capacity of the material after yielding, however at high strain rates where the process is adiabatic, a sudden drop in stress after yielding is seen followed by work hardening, see Figure 1-2 [12], [58], [59]. This difference is attributed to the formation of ASBs due to the high rate of heat generation during dynamic testing which does not occur during quasi-static testing because of the low rate of heat generation allowing adequate time for the heat to dissipate evenly. However, at high strain rates heat is rapidly generated and does not have time to dissipate away from highly deformed regions resulting in material softening and thus the formation of a shear band, which will be discussed in more detail later [17].
Several investigations into the high strain rate behavior of Inconel 718 have been conducted using a SHPB, some of which show the strain rate dependence of several material properties [8], [12], [21], [53], [60]. The tensile yield strength of as-cast Inconel 718 was reported to be 100 MPa higher in the dynamic regime compared to the quasi-static regime [60]. Demange et al. observed that the yield strength in TM Inconel 718 approximately doubles under dynamic strain rates compared to quasi-static loading in compression [12]. Another dynamic strain rate investigation by Wang et al. revealed that as the strain rate increases from 5000 to 11000 s\(^{-1}\), the yield strength gradually increases[21]. As the temperature increases from room temperature to 800 °C, the strength gradually decreases at all strain rates. However, the increase in strength due to higher strain rates is reduced at higher testing temperatures [21]. High strain rate testing at elevated temperatures also reveals that the yield strength and work hardening coefficient of wrought Inconel 718 is conversely related to the strain rate though inversely related to the temperature [8], [21]. This was attributed to the increase in dislocation density due

![Figure 1-1: Dependence of Flow Stress on strain rate for Inconel 690 [108]](image1)

![Figure 1-2: Comparison of flow stress in quasi-static and dynamic regime for Inconel 718 [12]](image2)
to a higher rate of deformation and a dissolution of dislocations when tested at high temperatures [8]. Most of the previously mentioned works were on traditionally manufactured Inconel 718, however a recent study by Kouraytem et al. [61] on additively manufactured Inconel 718 reported that the yield strength doubled as a result of the precipitation of strengthening phases during aging heat treatment. Experiments by Forni et al., Figure 1-3, showed that the tensile yield strength of as-built AM Inconel 718 is 554 MPa in the quasi-static regime and increases to 669 MPa and 703 MPa at strain rates of 200 and 800 s⁻¹ respectively, and that the strain hardening coefficient is higher than that of as-cast TM Inconel 718 [60]. Reducing the porosity of AM Inconel 718 does increase the dynamic compressive and tensile flow stresses, however as the porosity decreases by 95% only a 10% increase in yield stress is achieved suggesting that other microstructural features, such as the precipitation of strengthening phases, are the primary strengthening mechanism [62]. Kouraytem et al. also noted that samples cut from the x-y plane were deformed isotopically and had an equiaxed microstructure whereas the specimens cut from along the x-z and y-z plane deformed elliptically which was attributed to their columnar grain structure [53].
1.3.4 High Strain Rate Adiabatic Shear Band of Inconel 718

Under high strain rates, steels and titanium alloys commonly used in military applications are especially susceptible to adiabatic shear banding (ASB)[17], [18]. ASBs have typically been viewed as forming without warning and leading to abrupt instability and catastrophic failure. However recent research into the formation of ASB in titanium alloys shows that under shear stress, the microstructure gradually evolves from small to large areas of dynamically recrystallized regions that locally soften the material allowing cracks to form and propagate through the region more easily [11]. The occurrence of ASB indicates the presence of mechanical instability in the material which functions as nucleation points. A typical nucleation point, such as a void or other defects intrinsic to the AM process, will intensify the stress resulting in more heat being generated from plastic deformation which causes more thermal softening locally than that experienced by the surrounding material [59]. Interestingly, numerous shear bands can form parallel to each other if multiple nucleation points are adjoining one another [63], [64].

Figure 1-3: Tensile stress-strain data for as-built Inconel 718 [60]
Two types of ASBs can form, deformation bands are regions that only experience intense plastic shear and transformation bands are regions that experience intense plastic shear and phase transformation [18], [65], [66]. Deformation bands typically form in materials with a low density such as Al and Ti however reports have shown them to also form in high-density steels and uranium alloys [66], [67], [68]. Under high shear strains, the locally deformed material will experience a large increase in temperature within a short time not allowing adequate heat dissipation thus the material experiences a local softening which forms a shear band, however, the material does not experience any phase transformation. Transformation bands are primarily seen to form with steel and titanium alloys. Early work by Zener and Holloman in 1944 on steels showed that temperatures within the shear band could exceed 1000 °C in micro- or milli-seconds and then be immediately quenched by the surrounding material, this would cause ferrite to be transformed into FCC austenite which can then form martensite after quenching [69]. Work by Derep in 1987 on chromium armor steel indicates that temperatures in the shear band could even reach just below the melting point [70].

High strain rate shear experiments conducted on Inconel 718 by Song et al., Figure 1-4, [71] and Johansson et al. [72] have shown that ASBs form at strain rates around 80,000 s⁻¹ and that the critical shear strain required for ASB formation is approximately 4.5 for aged TM Inconel 718 and between 2.4 to 3.2 for solution heat-treated TM Inconel 718. While ASBs are not cracks in themselves, they are commonly seen as precursors to localized ductile or brittle failure [17]. Severe strain within the shear region can cause an increase in dislocation density, work hardening, and an increase in hardness of 50-100 HV [10]. This increase in hardness in the vicinity of ASBs has been noted for many materials including stainless steels [73], low alloy steels [74], titanium alloys [75], and HEAs [76] with the increase in hardness being higher at
larger amounts of deformation. Though the material experiences an increase in hardness within the shear region, the high rate of material deformation causes a large increase in temperature within a short amount of time. A significant rise in temperature causes thermal softening in the shear region. If this softening surpasses the increase in hardness resulting from work hardening, it triggers the formation of an Adiabatic Shear Band (ASB) within the material [17].

![Image of ASB formation and crack propagation for Inconel 718 at high strain rates](image)

**Figure 1-4:** ASB formation followed by crack propagation for Inconel 718 at high strain rates [72]

1.3.5 Lattice Structures

The fabrication of intricate structures with precise architectures using additive manufacturing (AM) has great potential in impact/blast attenuator applications due to their high capacity for absorbing kinetic energy [77], [78], [79], [80], [81], [82], [83], [84]. Building lattice structures using AM allows the ability to manipulate specific attributes such as stiffness, strength-to-weight ratio, and density, to suit a wide range of engineering applications including thermal insulation, structural aircraft parts, and vehicle components [85], [86], [87]. Many experimental studies on the mechanical properties of AM structures have been conducted. Some
have been comparative studies investigating which design has the highest strength-to-weight ratio [88] or greatest energy absorption [89], [90], whereas other studies focused on optimizing printing conditions to get the lowest porosity or best consistency between layers [91], [92]. There have even been investigations on the high-strain-rate mechanical properties of 3D printed lattices [93], [94], [95], [96], [97], [98], [99], however, these studies have not yet investigated how heat-treated AM metal lattices behave under high-strain-rate conditions [53], [100], [101]. As mentioned previously, Inconel 718 has exceptional mechanical properties at low [102] and high [103] operational temperatures, high corrosion resistance [104], and good creep resistance [105] therefore being able to 3D print lattice structures out of this superalloy would open doors to a wide variety of potential applications.

While many studies are focused on the mechanical properties of metallic lattice structures, one aspect that must be considered is the post-fabrication treatments done on the structures to alleviate detrimental artifacts from the printing process. The microstructures of AM structures typically have columnar grain due to directional heat flow during the cooling of printed parts as well as a higher void-to-volume ratio than TM components. Hot isostatic pressing (HIP) is one common method used to reduce porosity in AM structures [106]. Another effective method for lowering the porosity in AM IN718 fully dense structures is a post-process heat treatment, typically one that relieves stress and solutionizes the material, however, the effects these treatments have on the mechanical properties of lattice structures still need more study [101], [106]. An investigation that correlates the post-processing of AM lattice structures is essential to optimize their design and performance, ensuring their structural integrity in real-world scenarios.
1.4 Dissertation Overview

This dissertation encompasses several scientific works pertinent to the topic, either already published or currently in the process of review for publication. Chapter 1 establishes the groundwork for this research, elucidating the motivation behind it, summarizing the findings of studies featured in subsequent chapters, providing context for the materials under analysis, and establishing the viability of additive manufacturing as a possible replacement for traditional manufacturing in select scenarios.

Chapter 2, published in Metals, offers a concise review of adiabatic shear band formation in nickel and nickel-based superalloys. A discussion of multiple experimental methods for inducing ASB formation is provided concerning each method's advantages, limitations, and requirements. Additionally, the reason nickel and nickel-based superalloys have a higher resistance to ASB formation than other metals was presented based on reports in the current literature.

Chapter 3, published in Materials Characterization, presents a comparative mechanical study for additively and traditionally manufactured Inconel 718. Experiments included quasi-static and high strain rate compression, high strain rate shear testing with specimens designed for localizing of ASB formation, and micro indentation hardness testing. A brief microstructural study of the ASBs is available in which dimensions of the bands and recrystallized grain diameters were measured using SEM and TEM. The primary result of this thorough study combines the mechanical and microstructural data to elucidate which fabrication process creates material that is more vulnerable to ASB formation.

Chapter 4, submitted to The International Journal of Advanced Manufacturing Technology, presents a comprehensive comparison of three popular tripy-periodic-minimal-surface lattice structures (I-WP, Primitive, and Gyroid) built out of Inconel 718. A novel heat
treatment for Inconel 718 was used that extended the time held at the solution treatment
temperature from 1 to 4 hours. Experiments include quasi-static and high-strain rate
compression. This data was used to determine each design's yield strength, elastic modulus,
ductility, toughness, and specific energy absorption. The primary result of this comparative study
was identifying how this heat treatment affected each design, particularly which design had the
largest increase in energy absorption after heat treatment.

Chapter 5, details a summary of the significance and contributions of each work as well as
suggestions as to how future endeavors can improve and streamline this technology to be better
suited for wider applications and material.
1.5 References


CHAPTER 2: ADIABATIC SHEAR BANDING IN NICKEL AND NICKEL-BASED SUPERALLOYS: A REVIEW

This chapter was published in the journal *Metals*.

**Abstract**

This review paper discusses the formation and propagation of adiabatic shear bands in nickel-based superalloys. The formation of adiabatic shear bands (ASBs) is a unique dynamic phenomenon that typically precedes catastrophic, unpredicted failure in many metals under impact or ballistic loading. ASBs are thin regions that undergo substantial plastic shear strain and material softening due to the thermo-mechanical instability induced by the competitive work hardening and thermal softening processes. Dynamic recrystallization of the material’s microstructure in the shear region can occur and encourages shear localization and the formation of ASBs. Phase transformations are also often seen in ASBs of ferrous metals due to the elevated temperatures reached in the narrow shear region. ASBs ultimately lead to the local degradation of material properties within a narrow band wherein micro-voids can more easily nucleate and grow compared to the surrounding material. As the micro-voids grow, they will eventually coalesce leading to crack formation and eventual fracture. For elevated temperature applications, such as in the aerospace industry, nickel-based superalloys are used due to their high strength. Understanding the formation conditions of ASBs in nickel-based superalloys is also beneficial in extending the life of machining tools. The main goal of the review is to identify the formation
mechanisms of ASBs, the microstructural evolutions associated with ASBs in nickel-based alloys, and their consequent effect on material properties. Under a shear strain rate of 80,000 s$^{-1}$, the critical shear strain at which an ASB forms is between 2.2 and 3.2 for aged Inconel 718 and 4.5 for solution-treated Inconel 718. Shear band widths are reported to range between 2 and 65 microns for nickel-based superalloys. The shear bands widths are narrower in samples that are aged compared to samples in the annealed or solution treated condition.

2.1 Introduction

ASBs form in many metals at high strain rates and large plastic deformation due to the localization of plastic flow into a concentrated region. This localization occurs mainly at high strain rates once a critical shear strain value is reached and can produce large amounts of heat as approximately 90% of the work of deformation is converted into thermal energy \([1]\). At high rates of deformation, this heat may not have sufficient time to dissipate away from the shear region leading to thermal softening. Since the region is plastically deformed, it also undergoes work hardening and thus this region of adiabatic shear is under a plastic instability phenomenon. Predicting at what stage an ASB will form is difficult, but a general rule is that an ASB can form when the strength loss due to thermal softening exceeds the strength gained from work or strain rate hardening \([2,3]\). This naturally means that materials with low strain/strain rate hardening coefficients and/or low thermal conductivity have an increased risk of ASB formation. ASBs commonly occur in aluminum, titanium, uranium alloys, and steels during ballistic impact \([4,5]\), however, nickel-based alloys have a higher resistance to localized shear band formation \([2]\). The initiation of shear bands can be furthered by the presence of microstructural defects or voids as these artifacts can further localize stresses in the shear region \([6]\). Understanding the formation mechanisms for ASBs can aid in the prediction of material behavior under high-speed
deformation specific when loaded under torsion or compression, during dynamic impact, high-speed machining, and explosive fragmentation.

ASBs have been classified by Rogers [7] into two distinct categories: “transformation bands” and “deformation bands”. “Transformation bands” are bands in which there is a crystallographic phase change or change in the microstructural orientation as a result of the generated heat, plastic deformation, and/or rapid cooling. “Deformation bands” experience no phase change and are purely plastically deformed bands. ASBs occur in various materials differently and are sensitive to multiple mechanical variables such as cutting tool/workpiece geometry, rate of deformation, preheating, and temperature as well as material properties such as strain-rate sensitivity, the temperature dependence of flow stress, strain-hardening rate, thermal conductivity, specific heat, and phase transformation kinetics [8,9,10,11]. Depending on the metal used and the type of deformation, temperatures in the ASBs could be several hundred degrees higher than the surrounding material, with cooling rates on the order of $10^7$ K/s [12,13].

ASBs form intermittently throughout the shear region and eventually coalesce to form a singular band [14]. Material that has been additively manufactured (AM) [15] commonly contains imperfections, such as microvoids, which create stress concentrations and allow ASBs to form at lower strains compared to traditionally manufactured (TM) material [16]. In TM materials, this occurs in a straight line, primarily along the plane of maximum shear stress. However, in AM materials, microvoids may be irregularly distributed in the shear region and therefore cause ASBs to form in an irregular path. At a cutting speed of 2 m/s, Inconel 625 that was manufactured through casting did not show evidence of ASB formation (Figure 2-1a), however, AM Inconel 625 built using selective laser melting did show distinct ASBs (Figure 2-1b red arrow) under the same cutting conditions. As the cutting speed increased to 4 m/s, ASBs
began to arise in the cast material, forming along a straight path in contrast to the shear bands formed in the AM samples which took an irregular curved path. Bhavsar et al. stated that the irregular path the ASB takes supported their hypothesis that the shear band formation in the AM samples occurs near microvoids inherent in the selective laser melting (SLM) process [16].

![ASB formation comparison](image)

**Figure 2-1.** Under the same post-processing cutting operation, (a) no visible ASB can be seen in cast IN 625, (b) however distinct ASB formation is observed in SLM IN 625 showing that AM material is more susceptible to ASB formation [16]. Reprinted from [16] with permission from Elsevier.

ASB formation leads to a degradation of mechanical properties by weakening the material in the shear band thus creating a pathway for cracks to preferentially travel along [14]. Identification of a shear band optically in an etched material is relatively simple because the boundary between the band and the surrounding material is very distinct and can be distinguished from other types of shear failure in Nickel-based superalloys by the presence of recrystallized grains in the shear band. One can estimate the width of a shear band analytically using the half-width model proposed by Dodd and Bai [17], where \( \delta \) is the predicted half-width of the shear band, \( k \) is the material's thermal conductivity, \( T \) is half of the melting temperature, \( \tau \) is the shear stress and \( \dot{\gamma} \) is the shear strain rate:

\[
\delta = \left( \frac{kT}{\tau \dot{\gamma}} \right)^{\frac{1}{2}}
\]  

(1)
Understanding what factors increase the likelihood of ASB formation in metals, such as heat treatments [18,19] or elevated temperature [10], is important as this information can be used to optimize cutting parameters for high-speed machining [20,21,22]. In this literature review, the following will be discussed for Nickel-based superalloys: shear band formation and propagation kinetics, and how the microstructures around the ASBs evolve when shear localization occurs.

2.2 Testing Methodology

Numerous testing methods have been developed to characterize the formation of ASBs and further understand the effect of shear localization in a material. The most common are torsional testing [23,24,25,26] (Figure 2-2a), forced shear compression testing [25,27,28,29] (Figure 2-2b,c), and high-speed cutting with a quick stop device [30] (Figure 2-2d). The reasoning behind using these testing methods is that the geometries and loading conditions encourage shear deformation in a localized region. The compact forced simple shear (CFSS) sample geometry is the most recent design for testing materials under simple shear [28,31]. This design is an improvement upon the previously used top hat geometry which had significant issues such as rotation of the shear surface during testing and radial expansion of the brim causing a multi-axial stress state [28]. The CFSS sample design allows a narrow plane to be under simple shear loads, shown by the arrows in Figure 2-2b, as the two ends are pushed together with no compressive loading in the shear section. One significant advantage that the top-hat geometry does have over the CFSS is that stopper rings may be used with the top-hat samples to limit their deformation which allows for characterizing the formation process of ASBs more accurately [32].
High-speed cutting experiments are primarily used to determine the optimal cutting speed to prevent serrated chip formation, which is merely repeated ASB formation, as their formation increases the wear on cutting tools [30,33]. Cutting experiments are typically done using a quick-stop device to disengage the cutting tool and “freeze” the cutting state for analysis. The repeated formation of ASBs causes micro-cracks on the cutting tool increasing the wear rate and lowering the tool’s lifetime. Factors such as workpiece feed rate and the cutting angle can affect the stresses on the workpiece however cutting speed has the most obvious influence on shear band formation [8,9].

During a compressive test on a top hat sample, Figure 2-2c, the tip of the sample will be pushed into the hollow brim section of the sample creating a shear region shown in red. During this test the radial expansion of the brim is limited and thus most of the strain measurement is
from shearing. The shear strain achieved can be limited by using a stopper ring made of stronger material, preferably the same material as the incident/transmitted bars. The compressive and torsional Hopkinson bars are shown in Figure 2-3a,b, respectively. Assuming that the specimen is in equilibrium, the stress $\tau_s$, strain $\gamma_s$, and strain rate $\dot{\gamma}_s(t)$ can be evaluated using the following equations listed for a top-hat sample [32,34].

\[
\tau_s(t) = \frac{E_{bar}A_{bar}}{A_{specimen}} \epsilon_T(t) \cos(\theta) \quad (2)
\]

\[
\gamma_s(t) = -2 \frac{C_{bar}}{L_s} \int_0^t \epsilon_R(t) \, dt \quad (3)
\]

\[
\dot{\gamma}_s(t) = -2 \frac{C_{bar}}{L_s} (\epsilon_R(t)) \quad (4)
\]
Figure 2-3. (a) Compressive and (b) torsional split Hopkinson pressure bar system.

Here, $A_{\text{bar}}$ and $A_{\text{specimen}}$ are the cross-sectional area of the bar and the specimen, respectively; $E_{\text{bar}}$ is the elastic modulus of the incident and transmission bar material; $C_{\text{bar}}$ is the wave speed of the bars; $L_s$ is the length of the shear region; $\varepsilon_{R(t)}$ is the strain of the reflected wave in the incident bar at time ($t$), and $\varepsilon_{T(t)}$ is the strain of the transmitted wave in the transmitted bar at time ($t$).

The overall shear stress $\tau_s$ and shear strain $\gamma_s$ from tests with the torsional bars can be determined using the following equations [35,36]:

34
\[ \tau_s = \frac{2T}{\pi D_s^2 t_s} \]  

\[ \gamma_s(t) = \left( \frac{2c_s D_s}{L_s D} \right) \int_0^t \dot{\gamma}_R(t) \, dt \]  

\[ \dot{\gamma}_s(t) = \left( \frac{2c_s D_s}{L_s D} \right) \gamma_R(t) \]  

In the previous equation, \( T \) is applied torque during the test; \( D_s \) and \( D \) are the mean diameter of the torsional sample and diameter of the bar, respectively; \( t_s \) is the thickness of the test section; \( \dot{\gamma}_R \) and \( \gamma_R \) is the shear strain rate and shear strain, respectively, from the reflected wave; \( L_s \) is the width of the gauge section; \( c_s \) is the torsional wave velocity in the bar. During torsional testing, the test specimen is twisted around the central axis of the torsional bars. The torque-generating mechanism is commonly a loading arm or rotary actuator that rapidly twists the incident bar while the transmitted bar is fixed.

2.3 Shear Band Formation and Propagation Kinetics

The shear region along with a fully formed ASB in a top-hat shear sample is shown in Figure 2-4a. Previous work showed that dynamic recovery [37], dynamic recrystallization [38], phase transformation [39], melting, and amorphization [40] can occur during the formation and propagation of an ASB [41]. Hines and Vecchio [42] proposed that progressive lattice rotations, termed PRiSM (progressive subgrain misorientation), occur during shear localization due to strain-induced recrystallization. Prior to the formation of an ASB, a larger area referred to as the shear affected zone (SAZ), or sometimes the transition layer, undergoes localized plastic shear deformation [43]. In this region, it is possible for grains to become elongated in the direction of the shear load or for grains to rotate and form high-angle grain boundaries. Large microstructural variations are observed from the outside of the SAZ to the core in which an ASB may form. The
outermost sections of the SAZ have well-defined subgrains with a gradual transition to elongated subgrains near the core of the SAZ. If an ASB had formed, the subgrains at the core of the SAZ would partition into smaller cells with highly misoriented grains due to dynamic recrystallization, though in these experiments no recrystallization was seen to occur. Distinguishing the perimeter of the SAZ and the base material is often difficult because the grains still resemble the undeformed size and shape. Electron backscatter diffraction (EBSD) scans (Figure 2-4b,c) can be used to get a qualitative idea of the shear deformation to identify the SAZ through the average intragranular misorientation (AIM), which is an average of the misorientation between each data point and its nearest neighbor within a grain [43]. An AIM value of 1° is often used to determine the width of a SAZ as shown in Figure 2-4c.

![Figure 2-4](image)

**Figure 2-4.** (a) Shear region including the shear band for a top-hat Inconel 718 sample (b) IPF of an ASB region and (c) observation of SAZ using the AIM values [43]. The scale bar for (a) is 150 microns and 100 microns for (b,c). Reprinted from [5,43] with permission from Elsevier.
Landau et al., studied possible indicators of ASB formation by observing the microstructure surrounding ASBs [14]. They reported that ASBs do not form from a single initiation point but intermittently within the shear region, shown in Figure 2-5a. TEM lamellas lifted out from regions adjacent to cracks within the ASB, marked ‘A’ in Figure 2-5a, showed dynamically recrystallized grains. Figure 2-5b. TEM images from lamellas lifted from regions between crack tips that formed in the ASB, marked ‘B’ in Figure 2-5a, showed heavily deformed large grains with intermittent islands of dynamically recrystallized grains, Figure 2-5c. The final TEM lamellas were taken 5 microns to the side of cracks within the ASB region, marked ‘C’ in Figure 2-5a. The TEM images taken from the region slightly outside the ASB still showed heavily deformed large grains but very sparse dynamically recrystallized grains, Figure 2-5d. Rittel et al. reported similar findings that dynamic recrystallization is a precursor to ASB formation [44]. Landau et al. experiments show that ASBs form in the shear region after sparse islands of dynamically recrystallized grains gradually evolve and coalesce into a single band of recrystallized grains [14].
ASBs are often seen in high-strength materials, such as steels, titanium, and aluminum alloys, susceptible to ASB formation under high strain rate loading conditions. Nickel is more resistant to shear banding and requires a much higher rate of deformation compared to other materials used in aero-engines, such as Titanium alloys, due to their high yield strengths even at elevated temperatures, low thermal softening coefficients, and low thermal conductivity, all of which affect the formation of ASBs [11]. It was observed by Hokka et al. that Inconel 625 required significantly more strain during cutting operations for ASBs to form compared to Ti-6246 [11]. The reason a higher strain is required for ASBs to form for Inconel 625 is that the thermal conductivity for Inconel 625 is twice as high as Ti 6246, therefore heat is dissipated more quickly for Inconel 625, and that Inconel 625 has a higher strain hardening exponent. This allows Inconel 625 to sustain more strain before the effects of thermal softening surpass those of strain hardening. Shear band formation during the cutting of Ti6Al4V and Inconel 718 was compared at a cutting speed of 10 m/s by Cai et al. [45]. They found that the spacing between the shear bands was approximately 80 microns for Ti64 and approximately 92 microns for Inconel.
Inconel 718 (Figure 2-6). Similar to the situation in the research by Hokka et al., Inconel 718 has almost double the thermal conductivity and nearly a 3 times higher strain hardening exponent, and thus the material was able to withstand more deformation before plastic instability could lead to the formation of a new ASB.

![Figure 2-6](image100x412_to_512x610.png)

**Figure 2-6.** The shear band spacing during the high-speed cutting of (a) Ti64 is smaller than that seen in (b) Inconel 718 [45]. Both scale bars are 50 microns. Reprinted from [45] with permission from Elsevier.

Aging heat treatments are generally beneficial for precipitation-hardened material by increasing the yield strength through the growth of strengthening phases. However, doing so also decreases the strain required for an ASB to form due to a reduction in ductility. For Inconel 718, after aging heat treatments, when the $\gamma'$ and $\gamma''$ strengthening precipitates reach a critical size, i.e., their radii become larger than $\sim$10 nm, a change in their shearing mechanism leads to a lower strain hardening coefficient [46]. Research by Song et al. [18] conducted dynamic shear tests using a top hat sample geometry and stopper rings of various thicknesses on solution treated (Figure 2-7a) and aged (Figure 2-7b) Inconel 718. While the aged material did show a higher shear yield strength than the solution-treated material, the critical strain for an ASB to form was lower for the aged material. Under the same testing conditions, the aged material formed ASBs
between strains of 2.4 and 3.2 while the solution-treated samples required a strain of 4.5 for an ASB to form. A summary table of dynamic shear experiments on nickel-based superalloys is shown in Table 2-1.

Figure 2-7. Shear stress vs. strain graphs for (a) solution treated and (b) aged samples [18]. Reprinted from [18] with permission from Elsevier.
### Table 2-1. Summary of dynamic shear testing under various heat treatments and testing strain rates [5, 18, 47, 48, 49].

<table>
<thead>
<tr>
<th>Year</th>
<th>Author</th>
<th>Material</th>
<th>Tested strain rate ((s^{-1}))</th>
<th>Heat treatment</th>
<th>Width of ASBs (microns)</th>
</tr>
</thead>
<tbody>
<tr>
<td>2003</td>
<td>Clos</td>
<td>Steel Inconel 718</td>
<td>Shear: (10^6-10^7)</td>
<td>-</td>
<td>4-20</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Annealed</td>
<td>2-4</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Precipitation hardened</td>
<td>65.8</td>
</tr>
<tr>
<td>2009</td>
<td>Demange</td>
<td>Inconel 718</td>
<td>Shear: (5\times10^4)</td>
<td></td>
<td>33.6</td>
</tr>
<tr>
<td>2016</td>
<td>Johansson</td>
<td>Inconel 718</td>
<td>Global: approximately 1500</td>
<td>Precipitation hardened</td>
<td>7</td>
</tr>
<tr>
<td>2018</td>
<td>Song</td>
<td>Inconel 718</td>
<td>Shear: (8\times10^4)</td>
<td>Solution treated Aged</td>
<td>10-13</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Precipitation hardened</td>
<td>10</td>
</tr>
<tr>
<td>2020</td>
<td>Colliander</td>
<td>Inconel 718</td>
<td>-</td>
<td></td>
<td>7</td>
</tr>
</tbody>
</table>

| 2.4 Microstructures in an ASB |

In addition to dynamic recrystallization, the microstructure within the ASB and SAZ experience significant changes in terms of precipitate morphology. In particular, the dissolution of strengthening phases in the ASB region has been observed, though the mechanism behind their dissolution requires further research [49]. Atom probe tomography (APT) specimens were taken from undeformed and deformed Inconel 718 material. An optical image of an ASB in the deformed sample is shown in Figure 2-8a. This material shows clusters of aluminum, titanium, and niobium which indicates the presence of \(\gamma'\) and \(\gamma''\) strengthening phases, Figure 2-8b,d. However, APT specimens from an identified ASB in deformed cylindrical top hat samples showed no clustering which is evidence of an absence of strengthening phases, Figure 2-8c. Estimations of the temperatures in the ASB obtained by numerical integration of the thermal balance equation (assuming adiabatic conditions and neglecting thermo-elastic effects) were between 864 °C and 1118 °C. These values are in the range of solvus temperatures for the \(\gamma'\) and
\( \gamma'' \) strengthening phases, 850 °C and 900 °C respectively. The surrounding material does not increase in temperature; thus, it effectively quenches the ASB and limits the time the ASB is above the solvus temperature of \( \gamma' \) to less than one \( \mu \)s, which prevents the nucleation of precipitates.

![Image](image.png)

**Figure 2-8.** (a) Image of the ASB from which APT samples were lifted out from. Chemical distribution of precipitate elements from (b) undeformed bulk material, (c) ASB region, and (d) iso-surface reconstructions of precipitates from the undeformed bulk material. A 1 mm scale bar is used for (a) and (b–d) scale bars are 20 nm. Reprinted from [49] with permission from Taylor and Francis.

Inconel 718 has been shown to have a higher resistance to ASB formation compared to low-carbon steel [47]. The microstructure in Inconel 718 was observed to be more stable after shear localization compared to low-carbon steel by Clos et al. [47]. In their study, microstructural observations of flat top-hat dynamic shear samples of low-carbon steel showed a nearly linear growth of shear band thickness from 4 microns to 20 microns after shear localization. Similar Inconel 718 samples showed a constant band width, approximately 2–4 microns, but no increase in size after shear localization. The critical strain values for the formation of ASBs were 2.2 for the steel and 1.7 for Inconel 718. Following the onset of unstable shear localization, the stress
was reduced, and the remaining deformation was confined to a narrow zone. Temperature measurements using InSb-detectors and InGaAs-detectors on the flat top hat samples measured temperatures of 500 °C after initial shear localization and 800 °C in the post-localization phase of the experiments with a heating rate on the order of \( 10^7 \) K/s.

Johansson et al. [48] compared the formation process of ASB between Inconel 718 sheared at a high strain rate in a top hat sample geometry and under a high-speed cutting operation. The top hat samples and the samples mechanically cut under high speeds showed a similar appearance on the microstructural level with both having visible shear bands approximately 4–5 microns wide. The Inconel samples showed no evidence of a transition zone between the ASB and the base material. Grains near ASB were observed to be in the process of subdivision by shearing. Evidence of dynamic recrystallization was not observed using EBSD due to low confidence indexes in the ASB region (Figure 2-9b), however, band contrasts maps taken from a transmission electron microscopy (TEM) foil removed from the internal section of the ASB outlined in red (Figure 2-9a) showed ultra-fine equiaxed grains, Figure 2-9c, with a size between 50 and 300 nm. This further reinforced that the region is an ASB because it has undergone recrystallization and is not simply a SAZ.
Severe plastic deformation in the shear region can lead to work hardening by an increase in the dislocation density, Figure 2-10a. Since the deformation is localized primarily along the shear plane, the surrounding microstructure is mostly unaffected, but sometimes shows slight grain elongation and rotation along the direction of the shear load. Using this information, ASBs can be identified using EBSD as the grains will be slightly misoriented. If the material in the ASB has undergone dynamic recrystallization, the grains will be very small and require a very low step size to accurately measure the recrystallized grain’s size. Identification of the approximate size of the ASB is possible using EBSD even though the step size is not small.
enough to record the size of the recrystallized grains because the ASB will consist of a band of points with a very low confidence index [18] (Figure 2-10b).

**Figure 2-10.** (a) Band contrast and (b) IPF maps of the ASB and surrounding material [18]. Both scale bars are 200 microns. Reprinted from [18] with permission from Elsevier.

An image of the microstructure in and around an ASB is shown in Figure 2-11a. Measurements of the hardness values in the SAZ and ASB regions can be used as qualitative estimations of their width. It was reported previously that the hardness values in the SAZ increase due to an increase in dislocation density [5, 18, 43, 50, 51, 52]. Inconel 718 specifically contains both FCC Ni which facilitates slipping and BCC Fe which favors the formation of mechanical twins [5]. The microhardness measurements taken from the SAZ region in Inconel 718 which was dynamically deformed under shear loads are shown in Figure 2-11b with a maximum value at the center of the ASB. The indents of the hardness measurements are shown in Figure 2-11a and the dashed red line traces the path of the ASB. Hardness values are reported
as averages of each column with the maximum and minimum values from each column shown by the error bars. Since material in the ASB region is recrystallized and exhibits a smaller average grain, the material hardens due to the Hall-Petch relationship which describes the inverse relationship between yield strength and grain size.

**Figure 2-11.** (a) OM image showing the SAZ region of the shear band and (b) micro-hardness measurements perpendicular to the ASB. The scale bar in (a) is 30 microns.
2.5 Conclusions

ASBs are rare and unpredictable dynamic failure mechanisms that occur at very high strain rates. ASBs form because the heat generated by rapid plastic deformation does not have sufficient time to dissipate to the surrounding material and thus thermally softens the material within the shear band, leading to localized shear failure. In order for an ASB to form both a critical strain and critical strain rate must be reached, though given that this is a probabilistic process it is not guaranteed to occur even if the conditions are met. The critical strain value for aged Inconel 718 has been recorded to be around 2.2 and 3.2 while the solution-treated material required strain values of 4.5 in samples tested at a shear strain rate of $8 \times 10^4$ s$^{-1}$. The higher resistance to shear localization in the solution-treated sample was attributed to the presence of $\gamma'$ and $\gamma''$ phases which pin grain boundaries thereby assisting with work hardening. Furthermore, AM Inconel 718 has a lower critical strain value compared to cast material, which has been attributed to the voids and defects inherent in the AM process. Shear band widths vary widely for Nickel-based superalloys ranging between 2 microns and 65 microns. This large variation is due to the effect heat treatments have on ASB formation. The band width for Inconel 718 is 65.8 microns after annealing and 33.6 after aging heat treatments. Unlike other materials that commonly experience ASBs, no new phases have been recorded to form in the shear bands of Inconel. Few investigations into the formation of ASBs in AM materials have been conducted on nickel-based superalloys and is limited to the high-speed cutting of AM Inconel 625. Future work on this topic could include comparing the shear localization behavior of AM and TM materials using the same sample geometry, such as the top hat, under the same shear strain rates. This will provide a better understanding of the ASB formation mechanism in AM materials and help with developing more comprehensive and versatile models. In conclusion, ASBs are a
probabilistic failure phenomenon that occurs under a high strain rate in Nickel-based superalloys, and understanding their formation mechanisms in nickel-based superalloys is key to the study of high-strain damage mechanisms in aerospace-grade materials.

Notable findings from the works referenced in this paper include the following:

1. ASB bandwidths vary between 2 microns and 65.8 microns for Nickel-based superalloys.

2. Aging heat treatments on nickel-based superalloys decrease the strain required for an ASB to form from 4.5 to between 2.2 and 3.2 and nearly halves the band widths of the ASB.

3. No new phases precipitate during ASB formation in Inconel 718 however $\gamma'$ and $\gamma''$ strengthening phases are reported to dissolve.
2.6 References


CHAPTER 3: A COMPARISON OF HIGH STRAIN RATE RESPONSE AND ADIABATIC SHEAR BAND FORMATION IN ADDITIVELY AND TRADITIONALLY MANUFACTURED INCONEL 718

This chapter was published in the journal *Materials Characterization*.

**Abstract**

3D printing of Inconel 718 has been increasingly commercialized, as such research comparing the mechanical properties of additively manufactured (AM) and traditionally manufactured (TM) components in both quasi-static and high rate regimes should be conducted. Herein, the high strain rate material response, as well as the formation of adiabatic shear bands (ASBs) in AM and TM Inconel 718 are studied using a split Hopkinson pressure bar (SHPB) system. A top-hat cylindrical specimen geometry was used to convert the compressive load to shear and conduct high rate shear tests in the SHPB. The quasi-static and high rate stress-strain data reported herein indicates that TM components have at least a 70 MPa higher compressive yield strength. Interestingly, the AM specimens were more susceptible to ASB formation than their TM counterparts. The critical shear strain at which an ASB forms was higher for the TM specimens compared to the AM specimens. This was attributed to the TM material having a higher critical stress for dynamic recrystallization initiation, which is required to ASB formation. The microstructures of the shear zone in the top-hat specimens were examined using an optical microscope (OM), scanning electron microscope (SEM), and transmission electron microscope (TEM). Dynamic recrystallization was observed at the center of the shear affected zone (SAZ) using TEM from material lifted out from the ASB. Due to the plastic deformation in the shear region, the material strain hardens and therefore has a higher hardness than the bulk material. At
the center of the shear region, a peak in nano-hardness of 585 HV was recorded on the ASB. The outcome of this research will lead to the prevention, containment, and control of shear instability in AM structures.

3.1 Introduction

Aircraft and passenger safety is of utmost importance to aero-engine manufacturers, which makes containing the failed engine components, such as turbine fan blades, an essential design consideration. Containment systems design typically consist of a thin metal ring made from steel [1], titanium alloys [2], [3], or aluminum alloys [4], though other systems with composite rings, layers of woven fabrics, and/or honeycomb structures also exist [4]. During high-speed ballistic impact, such as when a fan blade fails at top rotational speeds and impacts the containment shell, plastic deformation can be localized at the edge of where the projectile impacts leading to shear plugging or bending [5]. Shear plugging is a result of adiabatic shear bands (ASBs) forming at the edges of where the projectile impacts, which weakens the material leading to premature failure in the localized region. Therefore, choosing a material that has a high resistance to ASB formation, such as nickel-based superalloys, would better ensure the safety of both aircraft and passenger by containing the blade fragments.

In recent years, additively manufactured (AM) components have seen increased use in civilian and military aircraft [6]–[9] due to the versatility and capability of printing complex geometries that was previously deemed impossible or impractical by traditional manufacturing (TM) methods. Laser powder bed fusion (LPBF) is an advanced additive manufacturing process that creates metal parts directly from 3D CAD data. This technique can be used to print a variety of metals and alloys including stainless steels [10], [11], aluminum alloys [12], [13], titanium alloys [14], and nickel-based superalloys [15], [16] to create strong and durable parts for
prototypes and end-use applications. Evaluating the discrepancies in mechanical properties for AM and TM components would further increase the acceptance of 3D printing critical components. Additionally, replacing TM methods with AM can reduce manufacturing costs by eliminating the need for special tooling, reducing material waste, and lowering the fabrication time, which all lead to an increase in the supply chain efficiency for civilian and military applications [17].

The response of TM nickel-based superalloys and their sensitivity to temperature, strain rate, and stress states have been thoroughly reported in the literature by Zhou and Baker [18], Kuo et al. [19], Smith et al. [20], Thomas et al. [21] and many others. Quasi-static tensile and compressive testing along with high rate testing have shown how different stress states affect the yield strength and plastic flow behavior of nickel-based superalloys [19], [20], [22]. Kouraytem et al. [23] investigated the microstructure, print orientation, and the effects of heat treatment on the high rate yield strength of AM Inconel 718 specimens. They reported that the high rate yield strength of AM Inconel 718 specimens doubled due to the precipitation heat treatment. It was noted that specimens oriented in the x-y plane were deformed isotropically and had an equiaxed microstructure whereas the specimens oriented along the x-z and y-z planes were deformed elliptically and this was attributed to their columnar grain structure. Lee et al. [24] and Wang et al. [25] performed impact tests at an elevated temperature, ranging from room temperature to 800 °C, with a split Hopkinson pressure bar to characterize the high strain rate and temperature sensitivity of Inconel 718. They both observed that the yield strength increased with strain rate, up to 11,000 s⁻¹, but decreased with increasing temperature. It was also documented that an increase in strain rate increased the dislocation density, thus, resulting in a greater strain-hardening effect.
ASBs are important failure mechanisms seen in various metals including steels [26], titanium alloys [27]–[29], and nickel-alloys [5], [30], [31], under impact and ballistic loading. Abrupt and/or premature failure in metals during high-rate deformation can often be caused by the formation of ASBs [32]. While ASBs are not cracks themselves, they are commonly seen as precursors to localized ductile or brittle failure. The conditions for the formation of ASBs are not completely understood, however, a prerequisite for their formation includes a large amount of plastic work within a short duration of time, i.e., high strain rate applications. In metals, when there is a large amount of deformation, much of the work done is converted into thermal energy. Therefore, the material experiences both strain or strain rate hardening and thermal softening simultaneously. If the increase in temperature is faster than the heat diffusion away from the deformed region, then the strain or strain rate hardening is overpowered by thermal softening and the material experiences a large amount of localized shear deformation, i.e., an ASB forms.

Steels and titanium alloys are more prone to ASB formation in comparison to nickel alloys [33]. Compared to other materials used in aero-engines, such as titanium alloys, nickel based alloys require a higher rate of deformation for an ASB to form. Hokka et al.[34] noted that Inconel 625 could handle substantially higher strains during high-speed cutting operations before an ASB would form compared to Ti-6264 at the same cutting speed. Higher strains are required for an ASB to form in Inconel 625 because it has a higher heat transfer coefficient than Ti-6264. When deformed at a strain rate of 80,000 s$^{-1}$, the critical strain at which an ASB could form for Inconel 718 is 4.5 for solution-treated material and ranges between 2.4 and 3.2 for the aged material [35]. Song et al.[36] attributed the higher ASB resistance in the solution-treated Inconel specimen to the presence of δ-phase precipitates in the solution-treated specimens, which pin grain boundaries leading to significant strain hardening.
Due to the high temperatures that form during ASB formation, some materials experience phase transformations, such as the formation of martensite in Ti64 [37] or δ-ferrite in chromium armor steel [38]. Interestingly, no new phases have been reported to precipitate during ASB formation in Inconel 718, although dissolution of the face-centered cubic $\gamma'$ ($\text{Ni}_3(\text{Al,Ti})$) and body-centered tetragonal $\gamma''$ ($\text{Ni}_3\text{Nb}$) strengthening in the ASB region has been observed by Colliander et al. via atom probe tomography [39]. Using electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM), Johansson et al. measured the widths of shear bands that form in wrought Inconel 718 to be approximately 4-5 $\mu$m, in which substantial recrystallization has occurred resulting in grains that are tens of nanometers in diameter [40].

Although previous studies have thoroughly documented the high rate impact properties and microstructural evolution of TM nickel-based superalloys, little work has been reported on AM Inconel 718 [24], [40], [41]. There are many factors that affect the formation of ASBs including the strain rate exponent, strain rate sensitivity, hardness, microstructure, and presence of imperfections such as precipitates and inclusions [42], [43]. While there are many reports studying these parameters for TM specimens, there is not much information for their AM counterparts. Due to the anisotropy exhibited in AM components, this information cannot be implemented to accurately predict the formation of ASBs in AM Inconel 718 components. If additively manufactured components are going to be used in aircraft engines, the susceptibility of AM components to shear band formation and response at high strain rates must be thoroughly investigated.

The present investigation consists of high strain rate compressive and shear testing using a split Hopkinson pressure bar (SHPB). Microstructural observations of ASB formation are conducted on the shear specimens to identify what shear strain is required for an ASB to form in
both AM and TM Inconel 718 specimens. The widths of the shear bands and changes in the microstructure such as recrystallization or grain refinement have been compared between the AM and TM specimens. The results of the presented work here are expected to improve the scientific and industrial communities knowledge concerning the high rate properties of Inconel 718 as well as the differences in behavior between AM and TM components. The goal of this work is to guide future developments for aero-engine designs as well as many other critical components.

3.2 Experimental Section

The TM Inconel 718 specimens were obtained from Huntington Alloys Corp. (West Virginia, USA) and the AM specimens were provided by the Air Force Institute of Technology (AFIT, Ohio, USA). The TM specimens were fabricated via a hot rolling manufacturing process while the AM specimens were printed using an LPBF process, where a 400 W continuous-wave Ytterbium fiber laser was used to deposit layers 20 µm in thickness. A pre-alloyed Inconel 718 alloy in fine powder form with globular particles with diameters in the range of 25-40 µm was used for fabrication. The chemical composition of the Inconel 718 powder, determined by spectrometry, is shown in Table 3-1. To minimize the edge effect in the impact test results, disk-shaped specimens with a one-to-one diameter-to-length ratio are required. Cylindrical specimens with a diameter and thickness of 5 mm were cut from the TM IN718 using a wire electric discharge machining (EDM) technique and the AM IN718 specimens printed to be with a diameter of 10 mm and a thickness of 10 mm.

Table 3-1: Chemical composition of Inconel 718 powder. [44]

<table>
<thead>
<tr>
<th>Element</th>
<th>Ni</th>
<th>Cr</th>
<th>Fe</th>
<th>Nb+Ta</th>
<th>Mo</th>
<th>Ti</th>
<th>Al</th>
<th>Co</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt. %</td>
<td>54.05</td>
<td>18.08</td>
<td>17.69</td>
<td>5.32</td>
<td>2.93</td>
<td>0.97</td>
<td>0.45</td>
<td>0.20</td>
<td>0.03</td>
</tr>
</tbody>
</table>

Less than 0.1 wt% of Mn, Si, S, P, B, Cu, Ca, Mg, O, N
Specimens were separated into 4 different categories and heat treated using a KSL-1100x high-temperature muffle furnace. The first specimen category is as built AM, which did not receive any heat treatment after printing. The second category is solution-treated TM which received a solid solution heat treatment at 1095°C (1 hour, air-cooled) to dissolve carbides and Laves phases back into the matrix. The third and fourth categories are aged AM and TM, which received the same solution-treatment as the solution treated TM specimen but were then aged at 955°C (1 hour, air-cooled), to precipitate δ phases, then secondly at 720°C (8 hours, furnace-cooled with a rate of 57 K/h to 620°C), and finally at 620°C (8 hours, air-cooled) to precipitate γ' and γ'' phases. The solution treatment stage is crucial for the aged specimens because it dissolves the niobium-rich Laves phase which allows a larger abundance of the niobium-based γ'' strengthening phase to precipitate.

Table 3-2: Specimen heat treatment conditions

<table>
<thead>
<tr>
<th>#</th>
<th>Specimen Name</th>
<th>Post Process Heat Treatment</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>As-Built AM</td>
<td>None</td>
</tr>
<tr>
<td>2</td>
<td>Solution Treated TM</td>
<td>1095°C for 1 hour followed by air cooling</td>
</tr>
<tr>
<td>3</td>
<td>Aged AM</td>
<td>1095°C for 1 hour followed by air cooling, 955°C for 1 hour followed by air cooling, 720 °C for 8 hours followed by cooling in the furnace to 620 °C then held for 8 hours followed by furnace cooling</td>
</tr>
<tr>
<td>4</td>
<td>Aged TM</td>
<td>1095°C for 1 hour followed by air cooling, 955°C for 1 hour followed by air cooling, 720 °C for 8 hours followed by cooling in the furnace to 620 °C then held for 8 hours followed by furnace cooling</td>
</tr>
</tbody>
</table>

A Split-Hopkinson Pressure Bar (REL Inc.) was used for the high-strain rate compression and shear characterization. SHPB testing is one of the most popular and reliable methods to study the high rate behavior of materials at strain rates from 200 to 10⁴ s⁻¹. Figure 3-1 shows a schematic of the setup. The setup consisted of three striker bars with lengths of 44.5 cm, 29.21 cm, and 15.24 cm, a 182.25 cm incident bar, and a 182.25 cm transmission bar. The lengths of the striker bars correspond respectively to the desired testing strain rates. The striker bar, incident
bar, and transmission bar were made of 350C maraging tool steel with identical diameters of 38 mm. The momentum trap located on the incident bar is made of 350C maraging tool steel and is located 125 cm from the specimen. The incident bar has a strain gauge located 90.17 cm away from the specimen and is placed parallel to the length of the bar. The same strain gauge setup is used for the transmission bar. The specimens were polished by hand on 400-grit SiC grinds pads to maximize the contact with the SHPB incident and transmitted bars. To reduce the frictional impact on the specimen multi-purpose synthetic grease with syncolon was applied to the end of the specimen in contact with the transmission and incident bars. Compression specimens were tested at strain rates of 1000, 1700, and 2200 s\(^{-1}\).

Figure 3-1. A schematic of the SHPB system used for high strain rate experiments.

For high strain rate testing, the striker bar is loaded into the launch chamber where the chamber is then pressurized based on the desired strain rate, between 68 and 550 kPa. After pressurizing the chamber, the striker bar is propelled down the chamber, at speeds between 5 and 35 m/s, until it impacts the incident bar. At the impacted end of the incident bar, a stress wave,
referred to as the incident wave, is created, which propagates through the bar from the left to the right toward the specimen. If the momentum trap is enabled, the wave will split into two waves on its way toward the specimen called the incident wave and the trapped wave. Upon reaching the specimen the incident wave splits into two smaller waves. One of the smaller waves, the transmitted wave, travels through the specimen and into the transmitted bar, straining the specimen and causing plastic deformation. The other smaller wave, called the reflected wave, is reflected from the specimen and travels back down the incident bar. The strain gages installed on the bars measure the strains due to the incident, transmitted, and reflected waves. Assuming deformation in the specimen is uniform, the stress and strain can be calculated from the amplitudes of the incident, transmitted, and reflected waves. The trapped wave will then pull the incident bar away from the specimen to avoid a second loading.

The voltage data collected by the strain gages on the incident and transmitted bar are converted to individual strain measurements using Equations 1 and 2.

\[
\varepsilon = \frac{\text{Voltage Measurement} \times \text{Gauge Resistance}}{\text{Calibrated Voltage} \times \text{Gauge Factor} \times (\text{Gauge Resistance} + \text{Calibration Resistance})} \tag{1}
\]

For the strain gauges used on the SHPB, the gauge resistance is 120Ω, the voltage during calibration was 2 V, the gauge factor is 2, and the resistance of the resistor used for calibrating the tool was 59,940 Ω. The formula for converting the voltage data into strain data with all constants applied is shown in Equation 2.

\[
\varepsilon_{I\ or\ T} = V(\text{incident or transmitted}) \times 0.0005 \tag{2}
\]

Assuming that the specimen is in equilibrium, the stress \(\sigma_s\), strain \(\varepsilon_s\), and strain-rate \(\dot{\varepsilon}_s\) were evaluated using Equations 3-5 listed below [45].

\[
\sigma_s(t) = \frac{E_{\text{bar}}A_{\text{bar}}}{A_{\text{specimen}}} \varepsilon_T(t) \tag{3}
\]
Here, $A_{\text{bar}}$ and $A_{\text{specimen}}$ are the cross-sectional area of the bar and the specimen, respectively; $E_{\text{bar}}$ is the elastic modulus of the 350C maraging steel bars (200 GPa); $C_{\text{bar}}$ is the wave speed of the bars (4970 m/s); $L_s$ is the thickness of the specimen; $\varepsilon_R(t)$ is the strain of the reflected wave in incident bar at time ($t$), and $\varepsilon_T(t)$ is the strain of the transmitted wave in the transmitted bar at time ($t$). Quasi-static and high rate yield strength values were determined using the 0.2% offset method.

High-strain rate shear experiments were also conducted using the SHPB with a modified top-hat specimen geometry proposed by Johansson et al. [46]. The top-hat is depicted in Figure 3-2a and the dimensions are shown in Figure 3-2b. TM specimens were cut from a commercially available annealed Inconel 718 rod. AM specimens were cut from the previously mentioned LPBF specimens from AFIT. During a compressive test on a top hat specimen, the tip of the specimen is pushed into the hollow brim section of the specimen creating a shear region shown by the red lines in Figure 3-2b. During this test, the radial expansion of the brim is limited and thus promoted shear failure. The shear strain achieved can be limited by using a stopper ring, Figure 3-2c, made of a stronger material, preferably the same material as the incident/transmitted bars. The desired target final shear strain values were separated into a low (0-3), medium (3-6), and high (above 6) categories. A list of the shear specimens is shown in Table 3-3.
Table 3-3: Shear Specimen Naming Nomenclature

<table>
<thead>
<tr>
<th>Specimen Name</th>
<th>Final Shear Strain</th>
<th>Stopper ring thickness</th>
<th>Shear displacement</th>
</tr>
</thead>
<tbody>
<tr>
<td>AM Shear 1</td>
<td>2.47 (Low)</td>
<td>1.79</td>
<td>0.58</td>
</tr>
<tr>
<td>AM Shear 2</td>
<td>5.37 (Medium)</td>
<td>1.17</td>
<td>0.94</td>
</tr>
<tr>
<td>AM Shear 3</td>
<td>5.12 (Medium)</td>
<td>No stopper ring</td>
<td>-</td>
</tr>
<tr>
<td>TM Shear 1</td>
<td>4.28 (Medium)</td>
<td>1.17</td>
<td>0.92</td>
</tr>
<tr>
<td>TM Shear 2</td>
<td>6.3 (High)</td>
<td>0.34</td>
<td>1.23</td>
</tr>
<tr>
<td>TM Shear 3</td>
<td>6.87 (High)</td>
<td>No stopper ring</td>
<td>-</td>
</tr>
</tbody>
</table>

The Hopkinson bars collect voltage data through strain gauges attached to the incident and transmission bars. This voltage data can be converted into relevant stress-strain data using equations specific to the experimental method used. Assuming that the specimen is in equilibrium, the shear strain rate $\dot{\gamma}_s(t)$ can be evaluated using Equation 6 for a top-hat specimen [45], [47].

$$\dot{\gamma}_s(t) = -2 \frac{C_{\text{bar}}}{L_s} \left( \varepsilon_R(t) \right)$$  \hspace{1cm} (6)

The dimensions needed for Equation 6 are the same as in Equations 3-5, except $L_s$, which is the width of the shear region that can be calculated by taking the difference in diameters between the tip and the hole at the back of the specimen and dividing by 2. The SHPB system can propel the 15.24 cm striker bar up to 35 m/s allowing shear testing strain rates to reach 140,000 s$^{-1}$. The final shear strain experienced by the specimen is calculated using Equation 7.

Where $\Delta H_T$ is the change in total height of the shear test specimen, $\Delta H_S$ is the change in the height of the section with the bigger diameter, $D_T$ is the diameter of the tip, and $D_H$ is the diameter of the hole after the shear SHPB experiments. An illustration of the dimension is shown in Figure 3-2d.

$$\gamma_s(t) = \frac{\Delta H_T - \Delta H_S}{D_T - D_H}$$ \hspace{1cm} (7)
Following high strain rate shear testing, the top-hat specimens were cut in half using electric discharge machining (EDM) to prevent thermal effects or plastic deformation destroying the shear bands in the shear region. The cross-section specimens were then mounted and ground using SiC pads and then polished using 3 and 1 μm diamond suspension. OM images were taken using a Keyence Microscope. After OM images were taken, SEM images were taken using a Thermo-Fischer Apreo SEM which also has an electron backscatter detector (EBSD) that was used to prepare inverse pole figures (IPF) and observe the grain structure in the shear region.
Transmission electron microscopy (TEM) was conducted using a FEI Tecnai F20 TEM to collect a higher resolution image of the microstructure in the shear region.

3.3 Results and Discussion

3.3.1. Comparison of Material Responses of AM and TM Specimens Under High Rate Loading

The behavior of AM and TM Inconel 718 under quasi-static and high rate compressive loads has been compared in Figure 3-3. The engineering stress-strain data was determined using Equations 3-5 and then converted into true stress and true strain using Equations 8 and 9:

\[
\sigma_{\text{true}} = \sigma_{\text{engineering}} \times (1 + \varepsilon_{\text{engineering}}) \\
\varepsilon_{\text{true}} = \ln (1 + \varepsilon_{\text{engineering}})
\]

Figure 3-3a showed the true stress-strain response for the as-built AM and solution treated TM specimens. The as-built AM specimens had a yield strength of 703 MPa whereas the solution-treated TM had a yield strength of 773 MPa. AM and TM specimens with the same geometry were tested under strain rates of 1000, 1700, and 2200 s\(^{-1}\), and their true stress-strain results are shown in Figures 3-3b, c, and d, respectively. The AM specimens exhibited a yield strength ranging from 824 to 878 MPa in the as-built condition and between 1218 to 1380 MPa after aging heat treatment. Likewise, the TM specimens had yield strengths ranging from 957 to 1132 MPa in the solution-treated condition and after aging heat treatment the strength increased to between 1463 and 1576 MPa. The increase in yield strength after aging heat treatments is attributed to the precipitation of the \(\gamma'\) and \(\gamma''\) strengthening phases. Recently, we showed that double aging Inconel 718 at 720 °C and 620 °C for 8 hours results in high density nanoscale \(\gamma'\) and \(\gamma''\) strengthening phase precipitation [48].
The yield strength for Inconel 718 is strain rate dependent as shown in Figure 3-3e, which shows estimated yield strength values calculated from the high strain rate tests (≥1000 s⁻¹) compared to the quasi-static tests (<0.1 s⁻¹). The yield strength of the AM materials was lower than the TM material in both the quasi-static and high rate regimes. Similar results can be seen in the work done by Forni et al. [49] in which the authors tested the quasi-static and high rate tensile response of as-cast (TM) and as-built (AM) Inconel 718 and reported a higher yield strength for the as-cast material. The AM specimens tested here and in the work by Forni et al. [49] were both loaded in the build direction. Previous research by Kouraytem et al. [50] tested LPBF Inconel 718 under strain rates between 2200-2600 s⁻¹ which exhibited a 10% lower yield strength in the build direction as opposed to material loaded perpendicular to the build direction.

In this paper, the difference in yield strength between the AM and TM material is larger under high rate loads compared to quasi-static loads, ~200 MPa compared to ~70 MPa. One reason for the lower yield of the AM specimens in comparison with the TM specimens may be the numerous defects which are introduced during the AM process [31]. LPBF Inconel 718 specimens have a high void-volume ratio. These voids may act as points for pre-mature failure of the AM specimens resulting in lower yield strength. A void volume ratio as high as 0.5 % was measured in Inconel 718 specimens printed using an LPBF process by Bahrami et al. [51]. Bahrami et al. [51] showed that aging heat treatment at 1065.5 °C for 1.5 hours reduces void volume by 29.7 % in LPBF specimens.

For the LPBF specimens in the high rate regime, the various stages of heat treatment cause the material to behave differently as the strain rate increased. The as-built AM specimens only experienced a 50 MPa increase in yield strength when the strain rate increased from 1000 to 1700 s⁻¹, comparing Figure 3-3a-b, while the solution-treated TM specimen experienced a 175
MPa increase over the same interval. In agreement with the conclusions by Forni et al. [49], the TM material in this study before aging heat treatment has a higher strain-rate sensitivity compared to the AM material. This relation between the AM and TM material does not hold true after aging heat treatment. The aged AM specimens showed an increase of 162 MPa between strain rates of 1000 and 1700 s⁻¹ but the aged TM specimens only showed a 113 MPa increase. Additionally, by comparing Figure 3-3a with Figure 3-3b, the change in yield strength from quasi-static to the high rate regimes was observed to be 184 MPa for the TM specimens and 121 MPa for the AM specimens.
Figure 3-3. Compression true stress-strain results from specimens tested under strain rates of (a) 0.002 s\(^{-1}\), (b) 1000 s\(^{-1}\), (c) 1700 s\(^{-1}\), and (d) 2200 s\(^{-1}\) and (e) a graph comparing the yield strength values at each strain rate.
3.3.2 Comparison of material properties under high rate shear loads

Multiple high rate shear tests were conducted at a velocity of approximately 35 m/s, which generated shear strain rates between 100,000 and 140,000 s\(^{-1}\), but with varied amount of plastic deformation determined by the thickness of the stopper rings. Contact with the stopper rings during the tests can be observed in the voltage data from a representative high rate shear experiment displayed in Figure 3-4a. Load displacement graphs are displayed for the TM and AM specimens in Figure 3-4b and c, respectively. It is known that there are limitations with the Hopkinson pressure bar for high strain rate measurements [52]. It needs to be pointed out that Equations 3-5 can only be used to calculate the simple cylindrical specimens high rate stress and strain. Due to the non-uniform plasticity that occurs during the shear experiments, new equations need to be derived to account for the uneven displacement across the specimen. For this reason, the load vs displacement graphs were used to accurately show the high rate response of the top-hat shear specimens. Each specimen experienced different levels of displacement, which is useful for tracking the development of ASBs for the TM and AM specimens. Additionally, specimens TM Shear 3 and AM Shear 3 were tested without a stopper ring to show the load-displacement curves without interference. By comparing specimens TM Shear 3 and AM Shear 3 with the other shear experiments, a drastic increase in recorded load after yielding is observed in the shear specimens upon contacting the stopper ring.
Figure 3-4. (a) Voltage data for a high rate shear experiment, Load vs Displacement data for the (b) annealed TM specimens and (c) as-built AM specimens. Specimens TM Shear 3 and AM Shear 3 were tested without a stopper ring.

3.3.3 Microstructure Evaluation in the Shear Region

The shear region for specimens that reached medium and high shear strain regimes is shown in Figure 3-5 for the TM Inconel 718 top-hat shear specimens. Intermittent cracking can be seen in the shear region of the TM top-hat specimens as shown in Figure 3-5a and EBSD scans are conducted between these cracks. As the amount of allowed shear strain by the stopper ring increases during testing for different specimens, these cracks gradually coalesce until complete failure occurs and the entire tip portion of the shear specimen breaks away from the base cylinder. Figure 3-5b is from TM Shear 1 which experienced a medium amount of shear strain and a partially formed shear band was observed. TM shear 2 experienced a final shear strain in the high regime (above 6) and developed a fully formed shear band. The ASB was observed in the IPF map of the shear region as shown in Figure 3-5c along with the shear
affected zone (SAZ). It can be seen in Figure 3-5c that grains near the ASB region were plastically deformed. A high magnification EBSD scan of the region in the red box in Figure 3-5c is shown in Figure 3-5d where the width of the shear band was approximated as 10 μm. These results are in good agreement with previous studies on shear band formation in TM Inconel 718. Johansson et al. [46] reported 5-10 μm wide shear bands in TM Inconel 718 tested under a shear strain rate of 20,000 s⁻¹ and Song et al. [30] reported a 10-13 μm wide for solution heat treated and 10 μm for aged TM Inconel 718 under a strain rate of 80,000 s⁻¹.

Figure 3-5. IPF maps of (a) cracks in the shear region, (b) a partially formed shear band (shear strain of 4.28), (c) a fully formed shear band (shear strain of 6.3), and (d) high magnification image of the red box in panel (c) of the fully formed shear band in the annealed TM specimens.
The AM Inconel 718 specimens were much more susceptible to shear band formation. Figure 3-6 shows the shear region of AM specimens that reached low (0-3) and medium (3-6) shear strain regimes. Intermittent cracking is shown in the shear region of the AM top hat specimens similar to what was observed in the TM specimens. The IPF map shown in Figure 3-6b was taken from AM Shear 1 that experienced a low amount of shear strain. Figures 3-6c was taken from AM Shear 2 that experienced a medium amount of shear strain. Figure 3-6d shows the IPF map of AM Shear 2 with a lower step size and higher magnification for an increased resolution for estimating the band width. The AM material formed shear bands in the low regime whereas TM material formed shear bands in the medium regime. Interestingly, the width of the shear band in the AM specimen was only 2 μm wider than the width of the shear bands in the TM specimens.
Figure 3-6. (a) cracks in the shear region, (b) partially formed shear band (shear strain of 2.47), (c) fully formed shear band (shear strain of 5.37), (d) high magnification image of the fully formed shear band indicated by the red box in (c) in the as-built AM specimens.

Table 3-4 shows the low, medium, and high shear strain regimes for both the AM and TM top-hat specimens. In the table, a check mark indicates a fully formed shear band was observed using EBSD, an X indicates that a fully formed shear band was not observed using EBSD and a “—” indicates that the microstructures was not studied. Shear bands were observed at the lowest shear strain in the AM top-hat specimens reported in this work. A partially formed shear band was observed in the medium shear strain regime for the TM specimens. A fully formed shear band did not form until the high shear strain regime. It is assumed that shear bands did not form in the low
shear strain regime for the TM specimens since shear banding does not start until the medium strain regime.

Table 3-4: Shear Strain Regime Where Fully Formed Shear Bands Were Observed

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Low (0-3 shear strain)</th>
<th>Medium (3-6 shear strain)</th>
<th>High (6+ shear strain)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AM Top-Hat</td>
<td>✓</td>
<td>✓</td>
<td>—</td>
</tr>
<tr>
<td>TM Top-Hat</td>
<td>—</td>
<td>X</td>
<td>✓</td>
</tr>
</tbody>
</table>

The microstructure within the shear band was not able to be resolved using SEM, therefore a TEM foil was prepared perpendicular to the shear bands in multiple AM and TM specimens. The location where the TEM lamella was lifted out of the TM specimen is shown in Figure 3-7a-b. The TM specimens that experienced final shear strains in the high regime (above 6) formed ASBs as indicated by the dynamically recrystallized grains observed via TEM. Figure 3-7c shows the thinned TEM lamella and Figure 3-7d shows the crack along with the microstructure on both sides of the ASB. The grains shown in Figure 3-7 e-f have diameters below 300 nm, which is substantially smaller than the grain diameters of the material outside the shear region that can be seen in Figure 3-5b and c. These smaller grains are a result of dynamic recrystallization from adiabatic heating when the shear region rapidly deforms. Dynamic recrystallization near an ASB has been investigated by Landau et al. [31] who observed dynamic recrystallization in pure α-titanium. Landau et al. [31] reported that before ASBs form in the shear region, the microstructure first forms sparse islands of dynamically recrystallized grains that coalesce to form massive dynamically recrystallized regions as more heat is generated in the shear region. Nesterenko et al. [53] similarly observed that shear bands initiate in favorably oriented grains and start as discontinuous segments, however these segments gradually coalesce into a single continuous band. This may cause crack formation in the shear region to form in
non-continuous segments which gradually coalesce together which was observed in Figure 3-7b and 8b and indicated with red arrows. Landau [31] also proposed that the dynamically recrystallized grains in the middle of the SAZ cause local softening allowing the growth of shear bands. This was supported by their observation that ASBs form in regions where a high density of dynamic recrystallization was observed however no ASBs were seen in regions that had little to no dynamically recrystallized grains. The grain structure in Figure 3-7e and f shows that the dynamic recrystallization is uniform for both sides of the crack at the center of the ASB.
Figure 3-7. (a) The location of the region on from the top-hat specimen, (b) an OM of the lift-out region, (c) the thinned TEM lamella showing the crack that forms at the center of the ASB (d) a TEM image of the center of the ASB, (e) a zoomed-in TEM image on the left side of the ASB, (f) a zoomed-in TEM image on the right side of the ASB region. All images come from a TEM foil lifted out of a TM Shear 2.

The location where the TEM foil was lifted out can be seen in Figure 3-8a-b for the AM specimen. The TM specimens that experienced final shear strains in the medium regime (3-6)
formed ASBs as indicated by the dynamically recrystallized grains observed via TEM. The IPF map from the corresponding shear region was previously presented in Figure 3-6c-d. Figure 3-8c shows the thinned TEM lamella and Figure 3-8d shows the crack along with the microstructure on both sides of the ASB. A close-up image of the nanoscale grains on the left side of the ASB is shown in Figure 3-8d. The grains in the vicinity of the crack in Figure 3-8d are approximately 50-200 nm in size, which is substantially smaller than the grains observed in Figure 3-6. Figure 3-8f is a zoomed-in image of the right side of the ASB and the grains are still on the nanoscale but range between 200 and 500 nm, which is much larger than the grains on the left side. In the compression section of this paper, the AM material was shown to be weaker than the TM material, therefore the critical stress for the initiation of dynamic recrystallization is lower for the AM material.

Analysis of the grain structure near ASBs have been conducted by multiple groups [5], [54], [55]. However, studies are typically limited to SEM analysis of the microstructure in the shear region and rarely extend to a TEM analysis. The different grain size on the left and right side of the ASB has not been reported previously for top hat shear specimens, however, the grain size on both sides of the ASB in a different compact force shear (CFS) geometry reported previously was the same size [31], [56]. The grain size following recrystallization has been the subject of previous research conducted by Derby et al. based on multiple works [57]. It was found that the grain size is directly related to the deformation stress in the region undergoing recrystallization. Derby et al. found that the normalized grain size following recrystallization was smaller when the deformation stress was greater [57]. Due to the geometry of the top hat shear specimens, there could be a higher stress on one side of the shear band since the geometry on both sides of the shear region are not symmetric. This issue may be avoided by the CFS
geometry since both sides of the shear region are symmetric for that design. Further research is needed to determine the cause of the differing grain sizes.

Figure 3-8. (a) The location of the region on from the top-hat specimen, (b) an OM of the lift-out region, (c) the thinned TEM lamella showing the crack that forms at the center of the ASB (d) a
TEM image of the center of the ASB, (e) a zoomed-in TEM image on the left side of the ASB, (f) a zoomed-in TEM image on the right side of the ASB region. All images come from a TEM foil lifted out of an AM Shear 2.

Figure 3-7b and 3-8b show that the ASBs and cracks do not propagate intergranularly but instead travel through the grains. Higher magnification images are shown in Figure S1 and S2 in the Supplementary Information section. The grains are outline by dotted white lines and the cracks traveling through the grains are indicated by the blue arrows. Average grain sizes in the baseline material from the top hat specimens were 36.9 µm for the AM specimens and 13.9 µm for the TM specimens. While the AM specimens have more than double the average grain size of the TM specimens, based on the images presented herein the difference is not significant enough to alter the ASB formation route or cracking path.

3.3.4 Nanomechanical Characterization in the Shear Region

Nano-hardness measurements from the SAZ, shown in Figure 3-9, indicate the hardness increases as one approaches the ASB. A peak in hardness can be observed on the ASB, which is in agreement with the literature [35], [58]. This increase in hardness is attributed to work hardening of the material in the SAZ where a high degree of plastic deformation occurs. In Figure 3-9a, the red arrow points toward the ASB. The red data point in Figure 3-8b corresponds to the indents that were obtained on the ASB in Figure 3-9a. The hardness data shown in the graphs is an average of the 3 indentations in each column with the error bars showing the standard deviation of each column. There is a slight increase in the hardness measurements as they approach the ASB with a sudden spike up to 585 HV in Column 6, which lies on the ASB. Nanoindentation hardness data from the center of the ASB is of particular interest because very few studies have reported nano-hardness data from within the ASB. In a previous study by Gwalani et al. [59] the authors reported inconsistent results concerning nano-hardness data from
measurements taken within the ASB, where both an increase and a decrease in the nano-hardness were observed in their specimens.

Figure 3-9. (a) OM image showing the ASB and the nano-indents in the area, (b) average hardness measurements from the 3 indents in each column.
3.4 Conclusion:

In this report, the high strain rate compressive properties, high strain rate shear properties, ASB formation and propagation mechanisms, as well as the effect of shear localization in AM and TM Inconel 718 are discussed. The yield strength of the as-built AM specimens was determined to be 773 MPa in the quasi-static regime and between 824-878 MPa in the high rate regime. The yield strength of the AM specimens after aging heat treatment was determined to be between 1218-1380 MPa in the high rate regime. The yield strength of the TM specimens after aging heat treatment was between 1463-1576 MPa in the high rate regime. Both the AM and TM materials experienced an increase in high rate yield strength of at least 400 MPa after aging heat treatments, which can be attributed to the precipitation of $\gamma'$ and $\gamma''$ strengthening phases. The yield strength of the as-built AM material was lower than the solution-treated TM specimens in the high rate and quasi-static regimes. Likewise, after aging both the AM and TM materials, the TM material exhibited a higher quasi-static and high rate yield strength. The critical shear strain required for ASB formation was lower in the AM material. ASBs form in the low shear strain regime (0-3) for the AM specimens whereas the TM specimens required shear deformation in the high regime (above 6) for an ASB to fully form. The material in the shear region undergoes strain hardening due to the localized plastic deformation and therefore has a higher hardness than the bulk material. At the center of the shear region, a peak in hardness was recorded on the ASB. The higher susceptibility to ASB formation seen in the AM specimens is attributed to the lower critical stress for dynamic recrystallization initiation compared to the TM specimens.
3.5 References


[41] A. Iturbe et al., “Mechanical characterization and modelling of Inconel 718 material behavior for machining process assessment To cite this version: HAL Id: hal-02283291,” 2019.


CHAPTER 4: EFFECTS OF A MODIFIED HEAT TREATMENT ON THE QUASI-STATIC AND DYNAMIC BEHAVIOR OF ADDITIVELY MANUFACTURED LATTICE STRUCTURES

This chapter is under review in the journal *The International Journal of Advanced Manufacturing Technology*.

**Abstract:**

The flexibility of additive manufacturing techniques that produce parts from powders layer-by-layer directly from a digital model, enabled the fabrication of complex lightweight lattice structures with precisely engineered mechanical properties. Herein, an investigation of the quasi-static and dynamic behavior of additively manufactured (AM) triply periodic minimal surface (TPMS) lattice structures before and after a novel post-process heat treatment step is conducted. The specimens were fabricated out of Inconel 718, a nickel-chromium-based superalloy, using a selective laser melting technique with three different topologies, namely, Gyroid, Primitive, and I-WP. The quasi-static tests were conducted at a strain rate of 0.002 s\(^{-1}\) and dynamic experiments were conducted using a split Hopkinson pressure bar at three different strain rates, 600 s\(^{-1}\), 800 s\(^{-1}\), and 1000 s\(^{-1}\). It was shown that while the strain rate does not significantly affect the mechanical responses of the lattice structures, the heat treatment step dramatically changes their behavior. Results demonstrated that after the heat treatment, the yield strength of the I-WP specimens increased by 65.2% under a quasi-static load. Also, flow stress after yielding in the dynamic tests was shown to increase around 9.6% for I-WP specimens and up to 12.8% for Gyroid specimens. The specific energy absorption values were 10.5, 19.1, and
10.7 for I-WP, Gyroid, and Primitive, respectively, before the heat treatment, and changed to 19.6, 19.8, and 15.4 after the heat treatment. The results confirm that by precisely designing the architecture of a lattice structure and implementing a modified heat treatment process, it is possible to optimize the weight, strength, and energy absorption capability of this type of metamaterial.

4.1 Introduction

Metal 3D printing is already influencing a broad range of industries by allowing the production of complex metallic components such as heat exchangers and cellular lattice structures [1]–[4]. This relatively novel technology allows the fabrication of various parts at the point of need and in an efficient and cost-effective manner with minimum waste material compared to commonly used traditional manufacturing methods such as casting, sheet metal working, and forging. Additively manufactured (AM) cellular structures with intricate geometries and typically precise architectures have the potential to be implemented in critical applications such as impact/blast attenuators due to their capacity to absorb kinetic energy when compressed to large strains [5]–[12]. By precisely engineering the architecture of lattice structures and changing the spatial configuration of their constituent unit cell or their relative densities it is possible to adjust their responses to impact loading in blast and impact protection systems [13]. It is also feasible to manipulate other attributes such as stiffness, strength-to-weight ratio, and density, to achieve desirable macro-scale material properties for a wide variety of other engineering applications including thermal insulation, structural aircraft, and vehicle components [14]–[16]. Although studies to date have extensively reported the high-strain-rate mechanical properties of 3D printed lattices [17]–[23], the body of knowledge pertaining to the behavior of heat-treated AM metal lattice structures under high strain-rate dynamic loads is somewhat
lacking [24]–[26]. Investigating the response of these structures under dynamic loading conditions and how post-processing can affect their behavior, is necessary in order to optimize their design and ensuring their structural integrity.

Laser powder bed fusion (LPBF) is a commonly used AM technique, and it facilitates the fabrication of lattice structures [27]–[30]. LPBF consists of a laser that selectively melts \(\sim 20\)–\(150\) \(\mu\text{m}\) metal particles and creates the part from a digital model layer by layer. While this technique has many advantages, the fabricated parts might have voids, porosities, and other defects. Also, due to the heat gradient, the grain growth in the AM specimens is directional [31], causing anisotropic mechanical properties [32], [33]. Furthermore, the rapid melting and solidification process in LPBF might lead to tensile residual stresses, trapped gas bubbles, and a heterogeneous microstructure [34], [35]. The tensile residual stresses encourage surface-related failure mechanisms [36], [37]. Also, the defects act as an ideal location for the stress concentration and reduce the fatigue life in AM parts [38]–[43] in addition to causing undesirable surface roughness. Although significant technological advancements and extensive research into process parameters and optimization have addressed many of these concerns to some degree, post-processing steps like heat treatment [44]–[48] and laser peening [49]–[52] remain integral components of additive manufacturing.

3D printing of lattice structures from a superalloy like nickel-based superalloys is highly of interest not only because it offers various benefits such as maintaining remarkable mechanical properties at high [53] and low [54] operational temperatures, high strength against corrosion [55], and creep strength [56], but also since it opens doors to unprecedented potential applications. Nickel-based superalloy's high temperature stability is attributed to a combination of solid solution strengthening or precipitation strengthening. Inconel 718 (IN718) is a
precipitation-hardened nickel-chromium alloy, ideal for extreme conditions, that has been widely implemented in various critical industries such as aerospace, energy, marine, and aircraft engine industries [57]–[59]. The stable mechanical response at high-temperature is attributed to the precipitation of \( \gamma' \) (\( \text{Ni}_3(\text{Al,Ti}) \)) and \( \gamma'' \) (\( \text{Ni}_3\text{Nb} \)) intermetallic phases within the \( \gamma \) matrix [60]–[63]. The fabrication of AM IN718 parts, controlling the microstructure, composition, and porosity have been reported numerously in various reports [64]–[66]. Previous studies by Zhang et al. [67] have investigated IN718 manufacturing defects, cracking, and residual stresses, that particularly occur during the fabrication using the LPBF technique. Zhang et al. [68] attributed these to the low thermal conductivity and high coefficient of thermal expansion inherent to nickel-based superalloys. Post-processing heating and aging treatments have been suggested as an effective way to optimize the mechanical properties of AM IN718, minimize its porosity [69], significantly reduce its micro-structural and mechanical anisotropies, and induce the precipitation of \( \gamma' \) and \( \gamma'' \) phases [38]. Several reports suggested various types of heat treatments in AM IN718 at different ranges of time and temperature [26], [38], [70].

There is a large number of reports on the mechanical behavior of AM lattice structures [17]–[23], but only a limited number of reports have focused on the effects of heat treatment on the mechanical response of the lattice structure under dynamic loads [24]–[26]. Kouraytem et al [26] studied the quasi-static and dynamic-loading behavior of simple cylindrical AM IN718 specimens before and after direct-age hardening heat-treatment. The heat-treatment was conducted at 720 °C for 24 h to precipitate the strengthening phases (\( \gamma' \) and \( \gamma'' \)). In the end, the specimens were air-cooled to room temperature. However, no significant changes in the microstructure were reported and the authors attributed this to the direct age heat treatment temperature that was below 1010 °C. The quasi-static tests were performed at a strain rate of \( 10^{-2} \)
s\(^{-1}\) and the dynamic tests were conducted at a strain rate of 10\(^3\) s\(^{-1}\). The heat treatment dramatically increased the yield strength in the quasi-static regime and led to an increase from \(~750 \pm 50\) MPa for the as-built specimens to 1400 \(\pm 100\) MPa for the heat-treated ones. The specimens showed a higher yield strength in the dynamic regime as well, and the heat treatment almost doubled the yield strength from \(~900\) MPa to approximately 1600 MPa. The authors concluded that both the process parameters and heat treatment significantly affect the response of the material under quasi-static and dynamic loads.

Hazeli et al. [24] studied the microstructure-topology relationship effects on the behavior of AM lattice structures under high strain rate loads. Four different topologies, namely Octet Truss, Rhombic Dodecahedron, Diamond, and Dode-Medium were tested with a relative density of 30\%, 20\%, 20\%, and 15\%, respectively. Three different post-processing steps were studied: a stress relief step at 1065 °C for 1.5 h, a hot isostatic pressing (HIP) step at 1163 °C for 4 h at 101.7 MPa, and a solution aging heat treatment that was conducted according to [71]. Interestingly, it was shown that the heat treatment steps were able to omit porosities in node regions considerably more than in the struts. The solution aging method significantly enhanced the yielding strength of the lattice structure, and this was attributed to the formation of the strengthening phases during the heat treatment. While depending on the type of heat treatment the lattice structures showed significantly different behaviors, it was proven that the topology of the structures played a more crucial role in how the material responds under different loading conditions. Also, the solution aging step led to a significant increase in the flow stress in the diamond topology compared to other designs, indicating again the effects of the topology. The HIP and solution aging heat treatments were shown to have a negligible effect on the mechanical strength of lattice structures with a unit cell size smaller than 2 mm. Babamiri [13] developed a
topology-microstructure-based optimization technique for the design of lattice structures. Quasi-static and dynamic behavior of AM IN718 Octet truss (OT) and Rhombic dodecahedron (RD) lattice structures with a 30% relative density before any heat treatment and after solution heat treatment and aging (STA) were compared. The heat treatment process included a homogenization step at 1065 °C for 1.5 h followed by argon purge cooling. Subsequently, the specimens were aged at 760 °C for 10 h, followed by furnace cooling to 650 °C. In the last stage, STA specimens were held at this temperature for a total precipitation time of 20 h and cooled by argon purging. The novel topologies were tested under quasi-static and dynamic loads and showed higher yield strength, enhanced flow stress, and improved energy absorption capacity compared to their traditionally fabricated counterparts. While the STA heat treatment increased the yield strength of OT and RD topologies by 42% and 56%, respectively, a 50% and 17% drop in their flow stresses were reported. This drop was attributed to the damage localization in the constituent components, in particular, the struts.

While there is a large number of studies on the mechanical behavior of lattice structures under quasi-static and dynamic loading, a comprehensive study on the post-processing (e.g., heat treatment)-performance (e.g., dynamic behavior) correlation in AM lattice structures is still missing. Such an investigation is essential in order to optimize their design and performance, ensuring their structural integrity in real-world scenarios. Also, from a scientific point of view, it lays the groundwork for studying a variety of physical, materials science, and mechanical engineering concepts. Herein, the results of such a study for three different types of AM lattice structures before and after a novel heat treatment process are presented.
4.2 Experimental Method:

4.2.1. Design:

The designs of lattice structures are determined by their unit cell, and those unit cells are either made of struts or surfaces. Strut-based unit cells are composed of prismatic components connected at nodes, while surface-based unit cells are characterized as continuous surfaces that connect a set of points, and these surfaces are then defined by a given function[72]. In this research, Inconel 718 triply periodic minimal surfaces (TPMS) were designed using computer-aided design (CAD) methods and fabricated utilizing LPBF. The TPMS were produced in three different topologies, and therefore they are defined by three distinct functions. The three different designs, (1) Schoen I-WP, (2) Primitive (Schwarz P), and (3) Gyroid (Schwarz D), are represented by trigonometric functions, as shown in Table 4-1 and described further in [73].

Table 4-1: Topologies and associated trigonometric function.

<table>
<thead>
<tr>
<th>Topology</th>
<th>Trigonometric function</th>
</tr>
</thead>
<tbody>
<tr>
<td>I-WP</td>
<td>(2(\cos mx \cos my + \cos mx \cos mz + \cos my \cos mz) - \cos 2mx \cos 2my \cos 2mz = 0)</td>
</tr>
<tr>
<td>Primitive</td>
<td>(\cos mx + \cos my + \cos mz = 0)</td>
</tr>
<tr>
<td>Gyroid</td>
<td>(\sin mx \sin my \sin mz + \sin mx \cos my \cos mz + \cos mx \sin my \cos mz + \cos mx \cos my \sin mz = 0)</td>
</tr>
</tbody>
</table>

4.2.2 Fabrication:

The IN718 powder was produced by Powder Alloy Corporation (Loveland, OH, USA), and its chemical composition is shown in Table 4-2.
Table 4-2: Chemical Composition of the IN718 Powder.

<table>
<thead>
<tr>
<th>Alloy Element</th>
<th>Ni</th>
<th>Cr</th>
<th>Cb+Ta</th>
<th>Mo</th>
<th>Ti</th>
<th>Al</th>
<th>Co</th>
<th>Mn</th>
<th>Si</th>
<th>Cu</th>
<th>C</th>
<th>P</th>
<th>S</th>
<th>B</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weight (%) min</td>
<td>50</td>
<td>20.5</td>
<td>4.75</td>
<td>2.8</td>
<td>0.65</td>
<td>0.2</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>balance</td>
</tr>
<tr>
<td>Weight (%) max</td>
<td>55</td>
<td>23</td>
<td>5.5</td>
<td>3.3</td>
<td>1.15</td>
<td>0.8</td>
<td>1</td>
<td>0.35</td>
<td>0.3</td>
<td>0.8</td>
<td>0.015</td>
<td>0.015</td>
<td>0.006</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

For the AM processing, a Concept Laser M2 SLM machine equipped with a 400 W continuous-wave Ytterbium fiber laser was utilized to produce the specimens, and a high-purity nitrogen shield gas was used to inert the build chamber. Also, the thin-surface nature of the TPMS lattice designs used in this research project required near-surface parameters of 120 W laser power, 280 mm/s scan speed, 50 μm spot size, and a 90 μm contour offset. Furthermore, the layer thickness was set to 40 μm, based on the IN718 powder’s largest diameter (D90 with a diameter of 36.72 μm). The topologies defined by the trigonometric functions in Table 4-1 are better described in Table 4-3. As observed in Table 4-3, for each of the three specimen designs (I-WP, Primitive, and Gyroid), the individual cells were composed of topologies enclosed within cubes measuring 2.5 mm on each side. The specimens representing all topologies took the form of cylinders with an approximate diameter of 10 mm and a height of approximately 10 mm. Relative density was calculated by using equation 1 below where $\rho^*$ is the density of the cellular structure and $\rho_s$ is the density of the base material [13], [74].

$$\rho_{rel} = \frac{\rho^*}{\rho_s}$$

(1)
Table 4-3: Unit cell as well as specimen’s description of the structures studied in this research segregated by topology: (1) Schoen I-WP, (2) Primitive (Schwarz P), and (3) Gyroid (Schwarz D).

<table>
<thead>
<tr>
<th>CELLS</th>
<th>3D IMAGES</th>
<th>SPECIMENS</th>
<th>RELATIVE DENSITY (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>(1) I-WP</td>
<td><img src="image1" alt="3D Image" /></td>
<td><img src="image2" alt="Specimen Image" /></td>
<td>25.73</td>
</tr>
<tr>
<td>(2) PRIMITIVE</td>
<td><img src="image3" alt="3D Image" /></td>
<td><img src="image4" alt="Specimen Image" /></td>
<td>26.26</td>
</tr>
<tr>
<td>(3) GYROID</td>
<td><img src="image5" alt="3D Image" /></td>
<td><img src="image6" alt="Specimen Image" /></td>
<td>21.34</td>
</tr>
</tbody>
</table>

*BD = Build Direction.*
4.2.3. Heat treatment:

Three different TPMS topologies were utilized, and divided into two groups, one without heat treatment, and the other group underwent a novel heat treatment step. The modified heat treatment (MSHT) consisted of subjecting the specimens to [38]:

- A homogenization heat treatment, at a temperature of 1100 °C for 4 hours: to convert the dendritic formation into an equiaxed microstructure.
- A double-aging treatment through exposing the specimens to a temperature of 720 °C for 8 hours, followed by 8 hours at 620 °C, and finally, air-cooling to room temperature to promote the formation of the γ’ and the γ” phases, and strengthen the material.

The microstructure of the specimens with and without the heat treatment were analyzed in the XZ direction (build direction, BD) via optical microscopy, and their mechanical responses were characterized by uniaxial compression tests, under quasi-static and dynamic loading conditions.

4.2.4. Mechanical Testing:

The quasi-static tests were performed using a dual-column Instron 5581 universal testing load frame with a +/- 50 KN load cell until approximately a 0.8 strain was achieved. The load cell recorded the force exerted as the frame compressed at a rate of 0.05 mm/s, and the applied force then was divided by the initial cross-sectional area of the specimens, and the engineering stresses were determined. Strain values were found by dividing the extension of the platen by the original height of the specimen.

The dynamic tests were conducted using a split-Hopkinson pressure bar (SHPB), at three different strain rates of 600 s⁻¹, 800 s⁻¹, and 1000 s⁻¹ (Figure 4-1). The SHPB system consisted of incident and transmission bars with a length of 182.25 cm; multiple striker bars with different
lengths to create the desired strain rate (Table 4-4); a shock absorber; and a gas gun to launch the striker bar (Figure 4-1(a)). All bars had a diameter of 38 mm and were made from maraging tool steel. Linear strain gauges are attached to the incident and transmission bars, approximately 90.17 cm from the specimen-bar interface (Figure 4-1(a)). Figure 4-1(b) shows a representative specimen between the incident and transmission bars.

**Table 4-4:** Bar lengths and exerted pressures during the Split-Hopkinson Pressure Bar (SHPB) tests.

<table>
<thead>
<tr>
<th></th>
<th>Strain Rate 600 s(^{-1})</th>
<th>Strain Rate 800 s(^{-1})</th>
<th>Strain Rate 1000 s(^{-1})</th>
</tr>
</thead>
<tbody>
<tr>
<td>Bar length (mm)</td>
<td>445</td>
<td>292</td>
<td>292</td>
</tr>
<tr>
<td>Exerted Pressure (KPa)</td>
<td>131</td>
<td>158</td>
<td>228</td>
</tr>
</tbody>
</table>
Figure 4-1: (a) Split-Hopkinson Pressure Bar (SHPB) compression system, (b) a representative specimen before impact between the incident bar and transmission bar.
The stress, strain, and strain rate values during the SHPB experiments were determined by Equations 2-4.

\[
\sigma_s(t) = \frac{E_{\text{bar}}A_{\text{bar}}}{A_{\text{specimen}}} \varepsilon_T(t) \tag{2}
\]

\[
\varepsilon_s(t) = 2 \frac{c_{\text{bar}}}{L_s} \int_0^t \varepsilon_R(t) \, dt \tag{3}
\]

\[
\dot{\varepsilon}_s(t) = \frac{c_{\text{bar}}}{L_s} (-2 \varepsilon_R(t)) \tag{4}
\]

Where \(A_{\text{bar}}, E_{\text{bar}},\) and \(c_{\text{bar}}\) are the cross-section area, elastic modulus, and wave speed of the maraging steel bars; \(A_{\text{specimen}}\) and \(L_s\) are the cross-sectional area and thickness of the specimen; \(\varepsilon_T(t)\) and \(\varepsilon_R(t)\) are the strain of the transmitted and reflected waves in the transmission and incident bar at the time \(t\), respectively.

Figure 4-2 depicts voltage signals obtained from a typical octet truss topology specimen. Despite the relatively significant magnitude of the reflected wave signal, a distinct contrast remains evident between the incident and reflected signals. It's essential to highlight that the transmitted signal's magnitude, at 500 mV, alleviates any concerns regarding signal-to-noise resolution.
**Figure 4-2.** Voltage signals for the incident, transmitted, and reflected waves, obtained from a typical I-WP specimen.

### 4.3 Results and Discussion

**Quasi-static:**

Engineering stress vs. engineering strain graphs for the as-built specimens are shown in blue lines in Figure 4-3. The I-WP specimen (a) shows a densification process that begins almost immediately after the initial yielding point with a small value of strain before densification starts. Densification occurs when every unit cell collapses and contacts the surrounding cells. Alternatively, the Primitive (b) and Gyroid (a) specimens seem to deform more before densifying leading to greater densification strain values. The densification phase of lattice structures is usually caused by the struts contacting together after the buckling, yielding, or fracturing of the material [75]. The early densification phase for the I-WP specimens is
attributed to the structure’s higher relative density, 38.73%. The Primitive and Gyroid specimens have lower relative densities, 26.26%, and 21.34% respectively, allowing more plastic deformation before layer/strut contact initiates densification. The Primitive and the Gyroid specimens demonstrate a similar densification phase likely due to their similar relative density values. Despite the Gyroid specimens having a slightly lower density in comparison to the Primitive topology, the Gyroids outperform the Primitive specimens in several parameters, including yield strength, 55.1 MPa compared to 50.6 MPa, and toughness, 45 MJ/m$^3$ compared to 30.9 MJ/m$^3$, respectively. This superior performance of Gyroid lattice structures has also been noticed by other authors [76]. The quasi-static tests demonstrated that the I-WP topology had the highest yield strength of 98.4 MPa. However, the I-WP specimens had a significantly lower ductility value at 5.9%, compared to the Primitive specimens with a ductility value of 51.6%, or the Gyroid specimens with a ductility of 58.3%.

Engineering stress vs. engineering strain graphs for the aged specimens are shown in red lines in Figure 4-3. The aging heat treatment increases the elastic modulus of I-WP specimens to 4290.9 MPa, which is an increase of 100.4% and greater than the other topologies in this study. The Gyroid specimens only experience a 32.5% increase, up to 3196.5 MPa, to the elastic modulus after aging. Primitive specimens experience the smallest elastic modulus increase of 26.8%, up to 2468.5 MPa, after aging. All the specimens subjected to quasi-static compressive tests showed an increase in yield strength after aging heat treatment. The I-WP specimens have the highest increase of 65.2% from 98.4 MPa to 162.6 MPa. The Primitive specimens had the next highest increase of 63.8% from 50.6 MPa to 82.9 MPa. The Gyroid had the lowest increase of 54.8% from 55.1 MPa to 85.3 MPa.
Figure 4-3: Engineering Stress (MPa) Versus Engineering Strain (mm/mm) Among (a) I-WP, (b) Primitive, and (c) Gyroid specimens. The curves are separated by heat treatment and all quasi-static tests were performed under the same strain rate (0.002 s\(^{-1}\)).

Figure 4-4 shows the ductility, toughness, and elastic modulus values for the as-built and heat-treated specimens. The numerical values of material properties for the various topologies are shown in Table 4-5. The elastic modulus was calculated based on the slope of the stress-strain...
curve before the first yielding point. Toughness was calculated using equation 5 below where $\varepsilon_f$ is the strain upon failure and $\sigma$ is stress:

$$ T = \int_0^{\varepsilon_f} \sigma \, d\varepsilon $$

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**Figure 4-4:** (a) Ductility, (b) toughness, and (c) elastic modulus for only quasi-static data.

**Dynamic:**

Dynamic engineering stress vs. engineering strain graphs for the as-built and heat-treated specimens are shown in Figure 4-5. Uniaxial compression tests under dynamic loading
conditions were performed at strain rates of 600 s\(^{-1}\) (black), 800 s\(^{-1}\) (red), and 1000 s\(^{-1}\) (blue), shown in Figure 4-5 with squares, triangles, and diamond markers, respectively. Similar to the quasi-static tests, the I-WP specimens had much larger yield strength, than the Primitive or Gyroid specimens. The as-built I-WP specimens have higher yield strengths under quasi-static loading, 98.4 MPa, compared to dynamic strain rates at 600 s\(^{-1}\) and 1000 s\(^{-1}\), 77.1-85.3 MPa, respectively, shown in Table 4-5 and Figure 4-6. This is interesting since Inconel 718 has a higher yield strength at higher strain rates in fully dense specimens\[77\]. The as-built Primitive and Gyroid specimens behave differently and have higher yield strengths at the highest dynamic strain rate, 1000 s\(^{-1}\). However, at the lowest dynamic strain rate, 600 s\(^{-1}\), the Primitive and Gyroid specimens have lower yield strengths, 46.5 and 50.9 MPa, respectively, than under quasi-static loading conditions, 50.6 and 55.1 MPa, respectively, as shown in Table 4-5 and Figure 4-6.
Figure 4-5: Dynamic engineering stress (MPa) versus engineering strain (mm/mm) among (a-b) I-WP, (c-d) Primitive, and (e-f) Gyroid specimens.

The yield strength of the heat-treated specimens exhibited higher values compared to the as-built specimens (Figure 4-6, Table 4-5, and Figure 4-5 b, d, and f). Following aging heat treatments, all specimens experienced at least a 35% increase in yield strength under dynamic loading conditions. The largest increase in yield strength was in the I-WP specimens tested at a strain rate of 600 s\(^{-1}\) where the yield strength nearly doubled from 77.1 MPa to 154.1 MPa. The strain rate sensitivity of the heat-treated lattice structures can be seen in Figure 4-5 b, d, and f.
The heat-treated I-WP specimens behave like the as-built structures and have lower yield strengths at higher strain rates. The heat-treated Primitive and Gyroid specimens have similar yield strengths to those tested under quasi-static strain rates; they both showed lower yield strengths at the dynamic strain rate, 600 s\(^{-1}\), with values similar to the as-built specimens. The highest yield strengths at the highest dynamic strain rate, 1000 s\(^{-1}\), for the Primitive and Gyroid specimens. Typically, materials yield strength increases as the strain rate increases, which is observed in all the as-built topologies. However, the yield strength of the aged I-WP specimens lowered as the strain rate increased.

Table 4-5: Material properties of the Lattice Structures

<table>
<thead>
<tr>
<th></th>
<th>Yield Strength (MPa)</th>
<th>Elastic Modulus (MPa)</th>
<th>Ductility (mm/mm)</th>
<th>Toughness (MJ/m(^2))</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>(0.005 s(^{-1}))</td>
<td>(600 s(^{-1}))</td>
<td>(800 s(^{-1}))</td>
<td>(1000 s(^{-1}))</td>
</tr>
<tr>
<td>Primitive (as-built)</td>
<td>50.6</td>
<td>46.5</td>
<td>53.6</td>
<td>54.50</td>
</tr>
<tr>
<td>Primitive (heat-treated)</td>
<td>82.9</td>
<td>71.8</td>
<td>72.7</td>
<td>81.9</td>
</tr>
<tr>
<td>Gyroid (as-build)</td>
<td>55.1</td>
<td>50.9</td>
<td>57.1</td>
<td>58.4</td>
</tr>
<tr>
<td>Gyroid (heat-treated)</td>
<td>85.3</td>
<td>76.1</td>
<td>79.2</td>
<td>87.4</td>
</tr>
<tr>
<td>I-WP (as-built)</td>
<td>98.4</td>
<td>77.1</td>
<td>83.1</td>
<td>85.3</td>
</tr>
<tr>
<td>I-WP (heat-treated)</td>
<td>162.6</td>
<td>154.1</td>
<td>150.1</td>
<td>124.18</td>
</tr>
</tbody>
</table>

![Yield strength for quasi-static and dynamic experiments](image)

**Figure 4-6:** Yield strength for quasi-static and dynamic experiments
The effect of heat treatment on specific energy absorption:

The specific energy absorption (SEA) of the lattice structures under the quasi-static loading was calculated before and after the heat treatment process. The SEA value is a good indicator of the potential of the lattice structure for energy absorption and expresses the energy-absorbing capability per unit mass. It has been widely used as an accurate means of evaluating the structure’s energy absorption capacity, particularly, for lattice structures [13], [21]. The SEA can be calculated using the equation below [78]:

\[
SEA = \frac{EA}{m} = \frac{\int_0^\delta Fd\delta}{m}
\]  

(6)

In this equation, \( F \) represents the compressive force, \( m \) signifies the mass, and \( EA \) corresponds to the total energy absorbed by the cellular structure, calculated as the area beneath the load-displacement curve up to the point of densification.

Figure 4-7 summarizes the results of the SEA calculation. The results illustrate that before the heat treatment process is conducted, under quasi-static loading, the Gyroid structure shows the highest performance in specific energy absorption, 19.1 mJ/g, significantly higher than the other two structures, namely, the I-WP with an SEA value of 10.5 mJ/g, and Primitive structures with an SEA value of 10.7 mJ/g (Figure 4-7). Interestingly, after the heat treatment process is performed, under the quasi-static loading conditions, the I-WP structure shows an SEA value of 19.6 mJ/g, almost equal to that of the Gyroid specimens, 19.8 mJ/g, in terms of specific energy absorption. As can be seen here, structures exposed to the heat treatment process demonstrates higher SEA values. This can be attributed to the heat treatment effects on material properties, and the formation of strengthening precipitates that affect the process of dissipating the energy from external loading by plastic deformation or fracture, through enhancing the strength of the materials [38]. As a result, there is a significant elevation in the plateau stress level, resulting in a larger area beneath the
stress-strain curve in the heat-treated specimens compared to their untreated counterparts. Specifically, the I-WP structure stands out by achieving the highest SEA value increase from 10.5 mJ/g before heat treatment to 19.6 mJ/g after the heat treatment. Furthermore, the SEA values for the Gyroid and Primitive structures are reported as 19.1 J/g and 10.7 J/g, respectively, before the heat treatment to 19.8 J/g and 15.4 J/g, respectively, after the heat treatment, as illustrated in Figure 4-7. The increase in the SEA value due to the postprocessing in all three types of specimens emphasizes that the heat treatment process is a means for manipulating the SEA value through microstructure modification and consequently, mechanical property optimization.

![SEA values of the lattices before and after heat treatment](image)

**Figure 4-7.** Quasi-static SEA values of the lattices before and after heat treatment.

### 4.4 Conclusion

In this research, the quasi-static and dynamic behaviors of three different types of additively manufactured lattice structures—Gyroid, Primitive, and I-WP—made of a nickel-based superalloy, Inconel 718, were comprehensively studied. The quasi-static loading was conducted under a compression load, and a Hopkinson pressure bar was used for the dynamic behavior characterization. The behaviors of the structures before and after a recently developed heat
treatment process were compared. The heat treatment process constituted of a homogenization heat treatment step followed by a double-aging treatment. The results clearly showed that the strain rate did not have a significant impact on the Gyroid or Primitive lattice structures' mechanical responses, and the heat treatment process dramatically changed their responses of all structures. After heat treatment, I-WP specimens showed a remarkable 65.2% increase in yield strength under quasi-static compression loads. In dynamic regime, flow stress post-yielding rose ~9.6% for I-WP and approximately up to 12.8% for Gyroid specimens. Specific energy absorption values, an indicator of the potential of the lattice structure for energy absorption, were initially 10.5, 19.1, and 10.7 for I-WP, Gyroid, and Primitive, and after the heat treatment increased to 19.6, 19.8, and 15.4. The Gyroid topology showed superior mechanical properties, including higher yield strength, elastic modulus, ductility, and toughness in comparison with the Primitive topology, despite its lower relative density. Surprisingly, the I-WP topology, despite having the highest relative density, showed a 65.2% increase in yield strength after the heat treatment under quasi-static loads. These improvements were attributed to the formation of the strengthening precipitates during the heat treatment process. In the dynamic regime, the Gyroid specimens experienced an initial drop in yield strength at a testing strain rate of 600 s⁻¹, followed by a rise to 57.1 MPa under a strain rate of 800 s⁻¹, surpassing their quasi-static strength. In contrast, I-WP specimens demonstrated between an 8.5 to 38.4 MPa lower yield strength at higher strain rates compared to quasi-static conditions.
4.5 References


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CHAPTER 5: SUMMARY, SCIENTIFIC CONTRIBUTIONS, AND FUTURE WORK

This chapter contains the main conclusions from the previous chapters. Additionally, suggestions for future work or research are outlined.

5.1 Summary

The work presented herein was conducted to build a foundation for a better understanding of how additively manufactured nickel-based superalloys behave at high strain rates. This includes how such structures differ in mechanical behavior from traditionally manufactured counterparts, and if adiabatic shear band formation in 3D-printed nickel-based superalloys varies from traditionally manufactured material. Finally, ways to improve the strength and specific energy absorption capabilities of additively manufacturing lattice structures made of nickel-based superalloys are investigated.

5.2 Scientific Contribution

The scientific community has extensively investigated the occurrence of adiabatic shear bands (ASBs) in titanium alloys, steels, and other traditional metallic materials. Surprisingly, there is a notable lack of studies exploring their formation in additive manufacturing (AM) nickel-based superalloys. This dissertation contains the first scientific study of adiabatic shear band formation in additively manufactured Inconel 718 created using SHPB shear testing. It also compares the mechanical properties of AM and TM Inconel 718 under high strain rate compressive and shear loading.
There is also a void in the literature on the behavior of triply periodic minimal surface (TPMS) IN718 lattice structure designs following post-process heat treatments. A novel heat treatment was developed and used to increase mechanical properties such as yield strength, elastic modulus, and specific energy absorption of three distinct TPMS structures. The mechanical response of three different designs, Gyroid, Primitive, and I-WP, was measured under quasi-static and high strain rates before and after the novel heat treatment.

This dissertation conveys in its entirety a concise review of the formation of ASB in nickel and nickel-based superalloys, how ASB formation differs in AM and TM Inconel 718, and how heat treatment affects the mechanical response of nickel-based superalloy lattice structures. These contributions will hopefully prove beneficial to the application of AM as a fabrication method for fully dense and lattice structures. Furthermore, this research helps to fill the void of knowledge in the literature around adiabatic shear bands and their formation in additively manufactured structures.

5.2.1 High Strain Rate Experimentation for Mechanical Characterization and ASB

Formation Analysis

Compressive mechanical characterization of AM and TM Inconel 718 before and after aging heat treatment was conducted. A comparison of TM and AM yield strengths revealed the TM specimens had higher strengths in both the quasi-static, 773 MPa compared to 703 MPa, and high strain rate regimes, 957-1132 MPa compared to 824-878 MPa, before any aging heat treatment. Following aging heat treatments, AM and TM specimens experienced at least a 400 MPa increase in yield strength which can be attributed to the precipitation of γ’ and γ” strengthening phases. Even after aging heat treatments, the TM material still had a higher yield strength.
Additionally, the high susceptibility to ASB formation in AM Inconel 718 compared to TM material was attributed to the lower critical stress for dynamic recrystallization initiation for the AM material. This dissertation uncovered ASB formation differences in AM and TM Inconel 718 using SEM and TEM. SEM was used to measure the widths of ASBs to be approximately 10 μm for the TM material and 12 μm for the AM material. ASBs were observed via SEM and EBSD, to start forming at a critical shear strain value of 4.28 for the TM material and 2.47 for the AM material. This critical strain is geometry dependent but can be used to determine if the AM specimens are more susceptible to ASB formation than the TM specimens. TEM was able to show dynamic recrystallization on either side of the cracks that formed following ASB formation.

5.2.2 Exploration of a Novel Heat Treatment on the Mechanical Behavior of AM Lattice Structures

To the best of the author’s knowledge, this was the first study of this modified solution heat treatment with aging for Inconel 718 being used to treat metallic lattice structures. The novel heat treatment used a higher than average solutionizing temperature, 1100°C compared to the typical 980-1060°C, and a longer time at said solutionizing temperature, 4 hours compared to the typical 1 hour, to allow the complete dissolution of deleterious Laves phases and conversion of the dendritic microstructure to equiaxed afterward. After the solution treatment, the material was air-cooled. Afterward, specimens were then aged at 720 °C for 8 hours, then allowed to cool in the furnace to 620 °C and held for another 8 hours, finally specimens were allowed to air-cool to room temperature.

Three different TPMS designs were compared in this study, Gyroid, Primitive, and I-WP. Each was evaluated based on its yield strength, elastic modulus, ductility, and toughness. Results from mechanical compressive experiments at low and high strain rates showed no significant strain
rate dependence for any of the designs. Of the three designs observed herein, the I-WP most benefited from post-process heat treatments. The I-WP design received a 65.2% increase in yield strength in the quasi-static regime and an 86.7% increase in specific energy absorption.

5.3 Future Work and Recommendations

Though the research presented herein provided a thorough understanding of ASB formation kinetic in nickel and nickel-based superalloy this work still could be expanded upon. Some recommendations for future studies include the following:

1. The present study ran experiments on AM samples that were only loaded in the build direction. Since AM structures are anisotropic, a future study that measures properties perpendicular to the build direction and microstructural analysis of ASB that form is advised.

2. Performing Transmission Kikuchi diffraction (TKD) of the recrystallized microstructure in the ASBs would reveal any preferential orientation of the grains.

3. High-temperature experiments would more accurately simulate the service conditions for Inconel 718 components. Inconel 718 is commonly used in aerospace engines and is therefore exposed to temperatures as high as 2,000 °C. Therefore, running high strain rate shear experiments at high temperatures and investigating ASB formation would more precisely replicate the environment of an aerospace engine.

Additionally, this work developed a novel heat treatment process for use with Inconel 718 lattice structures that can increase their energy absorption capabilities. This work still could be expanded upon with the following suggestions:

1. The effects of the heat treatment on the microstructure of the lattice structures were not studied comprehensively. Therefore, microstructural analysis using advanced microscopy
techniques is advised to draw a correlation between the mechanical response and the microstructure.

2. An experimental study for each TPMS design where specimens are printed with a wide range of densities. This way, the specific energy absorption capability for each TPMS design is known with relation to a given density.

3. A further extension of the previous point would be to run a simulation study of the three different TPMS designs to determine the best design and density to achieve the highest specific energy absorption value.

4. The use of a higher solutionizing temperature with different cooling steps, such as water quenching compared to the air-cooling used herein, may lead to a variety of microstructural changes that could affect the mechanical properties of the lattices.