Strengthening Mechanisms in Microtruss Metals

by

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Abstract

Microtrusses are hybrid materials composed of a three-dimensional array of struts capable of efficiently transmitting an externally applied load. The strut connectivity of microtrusses enables them to behave in a stretch-dominated fashion, allowing higher specific strength and stiffness values to be reached than conventional metal foams. While much attention has been given to the optimization of microtruss architectures, little attention has been given to the strengthening mechanisms inside the materials that make up this architecture. This thesis examines strengthening mechanisms in aluminum alloy and copper alloy microtruss systems with and without a reinforcing structural coating. C11000 microtrusses were stretch-bend fabricated for the first time; varying internal truss angles were selected in order to study the accumulating effects of plastic deformation and it was found that the mechanical performance was significantly enhanced in the presence of work hardening with the peak strength increasing by a factor of three. The C11000 microtrusses could also be significantly reinforced with sleeves of electrodeposited nanocrystalline Ni-53wt%Fe. It was found that the strength increase from work hardening and electrodeposition were additive over the range of structures considered. The AA2024 system allowed the contribution of work hardening, precipitation hardening, and hard anodizing to be considered as interacting strengthening mechanisms. Because of the lower formability of AA2024 compared to C11000, several different perforation geometries in the starting sheet were considered in order to more effectively
distribute the plastic strain during stretch-bend fabrication. A T8 condition was selected over a T6 condition because it was shown that the plastic deformation induced during the final step was sufficient to enhance precipitation kinetics allowing higher strengths to be reached, while at the same time eliminating one annealing treatment. When hard anodizing treatments were conducted on O-temper and T8 temper AA2024 truss cores, the strength increase was different for different architectures, but was nearly the same for the two parent material tempers. Finally, the question of how much microtruss strengthening can be obtained for a given amount of parent metal strengthening was addressed by examining the interaction of material and geometric parameters in a model system.
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## Precursor and Microtruss Architectural Parameters

<table>
<thead>
<tr>
<th>Symbol</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>( l ) (( l_o ))</td>
<td>Strut length (initial)</td>
</tr>
<tr>
<td>( t ) (( t_o ))</td>
<td>Strut thickness (initial)</td>
</tr>
<tr>
<td>( w ) (( w_o ))</td>
<td>Strut width (initial)</td>
</tr>
<tr>
<td>( \rho_R(\rho) )</td>
<td>Relative density (absolute)</td>
</tr>
<tr>
<td>( \omega(\omega_{MAX}) )</td>
<td>Truss angle (maximum)</td>
</tr>
<tr>
<td>( \phi )</td>
<td>Perforated metal open area fraction</td>
</tr>
<tr>
<td>( I )</td>
<td>Strut moment of inertia</td>
</tr>
<tr>
<td>( A )</td>
<td>Strut cross-sectional area</td>
</tr>
<tr>
<td>( L )</td>
<td>Column length</td>
</tr>
<tr>
<td>( r )</td>
<td>Strut radius of gyration</td>
</tr>
<tr>
<td>( A_{truss} )</td>
<td>Microtruss unit cell area</td>
</tr>
<tr>
<td>( A_{strut} )</td>
<td>Strut cross-sectional area</td>
</tr>
<tr>
<td>( N )</td>
<td>Number of struts in one unit cell</td>
</tr>
<tr>
<td>( I_{width} )</td>
<td>Second moment of inertia (strut width orientation)</td>
</tr>
<tr>
<td>( I_{thickness} )</td>
<td>Second moment of inertia (strut thickness orientation)</td>
</tr>
<tr>
<td>( R )</td>
<td>Rounded-square perforation corner radius</td>
</tr>
<tr>
<td>( l_U )</td>
<td>Length of microtruss unit cell</td>
</tr>
<tr>
<td>( l_P )</td>
<td>Length of unit cell perforation</td>
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## Precursor and Parent Material Parameters

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<th>Symbol</th>
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<tbody>
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<td>Tensile stress</td>
</tr>
<tr>
<td>( \epsilon )</td>
<td>Tensile strain</td>
</tr>
<tr>
<td>( \sigma_{YS} )</td>
<td>Yield strength</td>
</tr>
<tr>
<td>( \sigma_{UTS} )</td>
<td>Ultimate tensile strength</td>
</tr>
<tr>
<td>( \epsilon_T )</td>
<td>Total failure strain</td>
</tr>
<tr>
<td>( \Delta l )</td>
<td>Change in strut length during uniaxial tension</td>
</tr>
<tr>
<td>( E )</td>
<td>Modulus of elasticity</td>
</tr>
<tr>
<td>( E_t )</td>
<td>Tangent modulus</td>
</tr>
<tr>
<td>( K, n )</td>
<td>Holloman strength coefficient, strain-hardening exponent</td>
</tr>
<tr>
<td>( \epsilon_o )</td>
<td>Ramberg-Osgood model plastic strain corresponding to ( \sigma_{YS} ) (0.002)</td>
</tr>
<tr>
<td>( N )</td>
<td>Ramberg-Osgood model strain-hardening exponent</td>
</tr>
<tr>
<td>( t_{n-NiFe} )</td>
<td>Nominal deposited thickness of nanocrystalline nickel-iron</td>
</tr>
<tr>
<td>( m_{n-NiFe} )</td>
<td>Deposited mass of nanocrystalline nickel-iron</td>
</tr>
<tr>
<td>( \rho_{n-NiFe} )</td>
<td>Density of nanocrystalline nickel-iron</td>
</tr>
<tr>
<td>( S_A )</td>
<td>Electroplated surface area</td>
</tr>
</tbody>
</table>
Microtruss Forming Parameters

- $d_M (d_{\text{MAX}})$: Perforation stretching displacement (maximum)
- $F (F_{\text{MAX}})$: Stretch forming force (maximum)
- $d$: Stretch forming displacement
- $F_B$: Bending force

Microtruss Mechanical Testing Parameters

- $\sigma_c$: Compressive stress
- $\sigma_{\text{truss}}$: Microtruss compressive strength
- $\sigma_{\text{strut}}$: Strut axial compressive strength
- $\varepsilon$: Microtruss compressive strain
- $F$: Truss compressive force
- $d$: Truss compressive displacement
- $\sigma_P$: Microtruss peak compressive strength
- $\sigma_V$: Microtruss valley compressive strength
- $E_{\text{Reload}}$: Microtruss compressive modulus (reload)
- $J_D$: Microtruss densification energy
- $\sigma_{CR}$: Strut critical buckling strength
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1 INTRODUCTION

This chapter is an introduction to cellular materials, their mechanical properties and the strengthening mechanisms that can be used to enhance their performance. Section 1.1 discusses how cellular materials can occupy different voids in material property space according to the efficiency of their internal architecture, comparing conventional cellular materials with microtruss periodic cellular materials. Section 1.2 introduces methods that have been used to strengthen cellular materials, briefly discussing several material systems and the determination of total strength. Section 1.3 covers the scope of the thesis: the assessment of strengthening mechanisms in microtruss cellular metals and their additivity.

1.1 FILLING VOIDS IN MATERIAL PROPERTY SPACE

Hybrid material design presents unique opportunities to take advantage of material combinations and configurations in order to optimize each material’s attributes for an overall more efficient structure given a set of design requirements. Hybrids can be broadly classified into four groups (Figure 1.1) based on their configuration: composite, sandwich, lattice and segment [1]. Stochastic metal foams and periodic cellular metals are two classes of hybrid materials that offer weight advantages in terms of improved strength-to-weight and stiffness-to-weight ratios at low density compared to traditional structural materials [2]. Conventional cellular materials can be described as metal foams. Their structure in terms of pore connectivity, open (Figure 1.2) or closed cell, and ligament size and shape will depend on the particular processing method. Although foams offer low relative density (density of the cellular solid divided by the density of the
parent material), an inherent drawback is their structural irregularity and thus unpredictability in mechanical performance.

Figure 1.1. Four families of configurations of hybrid materials: composites, sandwiches, lattices and segmented structures. The functionalities offered by each family are listed on the right (Figure from [1]).

Failure modes in this sandwich/core configuration are through face yield, indentation and core shear [3]. Foams are often used as the core in sandwich structures adhesively bonded to outer face panels. Foam can represent a challenge because a minimum number of cells are required in order to reflect a statistical distribution of
random imperfections [4]. In addition, loading causes a deformation gradient whereby the cells in closest proximity to the point of loading suffer the most damage [5,6]. Furthermore, cells on the outer edges of the sample generally contribute less to the overall mechanical performance due to boundary-layer and constraint effects [3,7-10].

Figure 1.2. Example of an open-cell foam showing the topology of an irregular array of struts (Figure from [2]).

Periodic cellular materials are characterized by repeating units of supporting ligaments that provide the structure’s connectivity. They can serve as the core in sandwich structures with outer face sheets in addition to being stackable with either solid or perforated face sheets offering the capability of multidimensional flow, Figure 1.3. For a given relative density, microtrusses will have a higher strength than their foam counterparts offering potentially increased functionality [2].
As shown in Figure 1.4, both of these classes of materials are capable of filling gaps in the low density region of materials property space, making them an attractive option for structural engineering applications when weight-savings is an issue [2,3]. In addition to their good mechanical properties at low density, cellular materials can also exhibit multifunctionality in terms of thermal management and energy absorption [1,2,12] and can offer energy savings that can lead to environmentally-friendly material systems [13].
Figure 1.4. Materials property map showing Young’s modulus, \( E \), and relative Young’s modulus, \( \bar{E} \), versus density where \( \rho \) is the density for the metals and polymers and \( \bar{\rho} \) is the density of the cellular material. Stretch-dominated lattices (e.g. microtruss materials) show enhanced mechanical stiffness compared to bending-dominated lattices (Figure from [2]).

The overall deformation mechanism of a cellular architecture can be considered in terms of Maxwell’s stability criterion [2], shown in two-dimensions in Figure 1.5. Failure in cellular materials will be bending-dominated or stretch-dominated depending on the connectivity of their internal structure (Figure 1.5). Ideally, the stretch-dominated structures are more stable on a per weight basis. For the two-dimensional case:

\[
M = b - 2j + 3
\]

(1.1)

where \( b \) is the number of struts and \( j \) is the number of joints. In three-dimensions:

\[
M = b - 2j + 6.
\]

(1.2)

If \( M < 0 \) for a pin-jointed frame, the joints are able to move freely and the frame has at least one degree of freedom when loaded. The stiffness and/or strength of the material are
not a factor since the joints will control the collapse of the structure. In the case of fixed joints, the frame becomes bending-dominated as the struts bend near the nodes creating plastic hinges. If $M = 0$, then the structure is considered stretch-dominated. Instead of collapsing directly, a stretch-dominated and pin-jointed structure will internally resolve an external load such that there is a balance between tensile and compressive forces in the struts. As shown in Figure 1.5, stretch-dominated microtruss architectures satisfying Maxwell’s Stability Criterion can have significantly greater structural efficiency than the bending-dominated architectures of conventional metal foams [2]. As shown in Figure 1.4 and 1.6a, for a given density, stretch-dominated lattices have higher stiffness values and have improved performance compared to bending-dominated lattices given their slopes of 1 and 2 respectively.

Figure 1.6 focuses on the material property map region showing cellular materials to contrast bending-dominated lattices with stretch-dominated lattices. In both cases, stretch-dominated lattices demonstrate better relative stiffness (Figure 1.6a) and relative
strength (Figure 1.6b) as a function of relative density. An ideal bending-dominated and stretch-dominated material originating from the point (1,1) will have moduli that scale as $(\tilde{\rho}/\rho_S)^2$ (slope = 2) and $(\tilde{\rho}/\rho_3)$ (slope = 1) respectively, Figure 1.6a [1]. Honeycombs loaded parallel to the hexagons exhibit this ‘ideal’ behaviour. Pyramidal lattices, a subset of microtrusses, show stretch-dominated properties but lie a factor of 3 below the idealized line [1]. The heterogeneous nature of foams translates into weak structural regions limiting both stiffness and strength of the cellular material. Figure 1.6b shows the relative strength versus the relative density of both bending-dominated and stretch-dominated lattices. Similar to the case of the relative modulus versus relative density shown in Figure 1.5a, an ideal bending-dominated and stretch-dominated material originating from the point (1,1) will have strengths that scale as $(\tilde{\rho}/\rho_S)^{3/2}$ (slope = 1.5) and $(\tilde{\rho}/\rho_3)$ (slope = 1) respectively [1]. In the case of strength, honeycombs compressed parallel to the hexagon axis tend to buckle easily due to their thin cell walls while the strength of metallic foams is limited by their heterogeneous structure [1]. In contrast, both pyramidal and Kagome lattices can approach ideal stretch-dominated strength behaviour.
Figure 1.6. Materials property map showing (a) relative Young’s Modulus versus relative density and (b) relative strength versus relative density (where relative density is defined as density of the cellular material, $\tilde{\rho}$, divided by density of the parent material, $\rho_s$). Stretch-dominated lattices (e.g. microtruss materials) show enhanced mechanical stiffness compared to bending-dominated lattices (Figures from [1]).
Microtruss structures are typically used as the core component in sandwich panels. Three modes of failure are possible in this assembly: face sheet wrinkling and yielding, core failure and core-face sheet shear [14]. The focus of this thesis is on microtruss cores, which typically fail by tensile or compressive yielding and by buckling [15], as shown in Figure 1.7 [16]. Since the buckling strength of a slender column in compression will typically be substantially lower than the yield strength of the material, it is the critical buckling stress of the composite strut that will greatly influence the overall mechanical properties, underlying the importance of studying the strength of microtruss core members (note that in this study the buckling resistance of the microtruss struts will be measured directly in uniaxial compression).

Figure 1.7. Kagome structure made by investment casting undergoing 3-point bending showing strut buckling (Figure from [16]).

1.2 STRENGTHENING MECHANISMS IN CELLULAR MATERIALS

The major focus of previous microtruss studies covered the architectural aspect of these materials with less attention paid to the effect of the internal structure (i.e. microstructural effects). In terms of material design, the structure’s mechanical performance is a combination of both its architecture and its microstructure. Figure 1.8 illustrates the hierarchy of internal structures that control the overall material performance. In the case of the present study, microstructural effects are seen indirectly,
through the effect that the various thermomechanical treatments have on the overall mechanical properties of the microtruss cores. Longer term objectives for this area of research, beyond the scope of the present thesis, will include in-depth microstructural characterization to directly link these twin aspects of structure illustrated below.

![Figure 1.8](image)

**Figure 1.8.** Schematic diagram illustrating the hierarchy of (a) architecture and (b) microstructure (Figure from [17]), which control the overall mechanical properties.

As far as the overall mechanical properties of the cellular hybrid are concerned, microstructural design is just as important as the architecture. For conventional metal foams, strength can be obtained by using insoluble reinforcing particles in the melt [18] and second phase precipitation [19]. Precipitation hardening has also been widely used in microtruss cellular metals [16,20]. In addition to these mechanisms, deformation-formed microtruss cellular metals can also be strengthened by fabrication-induced work hardening [21]. Conventional metal foams and microtrusses can also be reinforced by structural coatings, such as the electrodeposition of a high strength nanocrystalline sleeve. The coating creates an interconnected network of nanocrystalline tubes which reinforce the overall structure [22-25]. For aluminum alloys, even greater specific
strength increases can be achieved by anodizing the parent metal to create a hard external ceramic shell [25]. A key question concerns the relative significance of these different strengthening mechanisms, and the ways in which they might contribute to the overall strength if more than one was present.

For several material systems, simple rules have been identified for expressing the contribution of more than one structural element to the overall material strength. For example, the yield strength of a low-carbon steel system, can be described as [26]:

\[
\sigma_{ys} = \sigma_i + k_1 \sqrt{c} + k_2 d^{-\frac{1}{2}} + \alpha G b \sqrt{\rho}
\]

where \(\sigma_i\) is the Peierls stress, \(k_1 \sqrt{c}\) is the strength from solid solution strengthening (e.g. where \(c\) represents the martensite solute atom concentration), \(k_2 d^{-\frac{1}{2}}\) describes grain boundary hardening (e.g. from the Hall-Petch relation for grain size \(d\)) and \(\alpha G b \sqrt{\rho}\) is the strength from strain hardening (e.g. work hardened shear modulus \(G\), Burgers vector \(b\) and dislocation density \(\rho\)). This relationship essentially states that to a first approximation each microstructural component acts independently towards increasing the overall strength.

Another example is composites. In this case, an upper and a lower bound are established and the volume fraction of each component is taken into account. The upper bound applies to the case where fibres are used as the reinforcement in the matrix [1]:

\[
(\sigma_f)_u = f(\sigma_f)_r + (1 - f)(\sigma_f)_m
\]

where \((\sigma_f)_r\) is the strength of the reinforcement, \((\sigma_f)_m\) is the strength of the matrix and \(f\) is the volume fraction of the reinforcement. The lower bound describes the case of strong reinforcing particles in a continuous ductile matrix. The strength of the particulate-
reinforced composite system is simply the yield strength of the matrix with the minimal addition of the plastic constraint of the particles [1]:

\[
(\sigma_f)_t = (\sigma_f)_m \left[ 1 + \frac{1}{16} \left( \frac{f^{1/2}}{1 - f^{3/2}} \right) \right].
\] (1.5)

With the capability to determine strength in material systems where multiple strengthening mechanisms are at play, the underlying question is now to determine what controls the additivity of strengthening mechanisms in microtruss systems. The underlying complexity in addressing this question for microtruss cellular materials is the role of the cellular architecture.

1.3 SCOPE OF THESIS

The objective of this thesis is to examine the buckling resistance of individual struts in new types of composite microtruss cores. The interaction and additivity of strengthening mechanisms is also examined in order to develop a general framework for predicting strength additivity in hybrid microtruss systems. Strengthening mechanisms were selected according to the parent metal of the microtruss. For copper-based microtrusses, the strengthening mechanisms studied were: fabrication induced work hardening and the application of structural nanocrystalline NiFe coatings. The copper system allowed the cumulative effects of work hardening to be investigated due to its large capacity for plastic deformation as seen via the Holloman exponent of \( n = 0.44 \) for annealed copper [27]. While copper alloy microtrusses are desirable for their combined potential of structural and thermal management applications; aluminum alloy microtrusses are desirable for lightweight structural applications. The aluminum system allowed the combination of work hardening and precipitation hardening to be
investigated, further reinforced with an anodized coating which allowed strengthening in a fracture dominant system to be investigated.

The following section, Chapter 2, provides a background to previous studies of cellular copper and aluminum conducted to date. Chapter 3 details the experimental methods used in the forming and mechanical testing of periodic cellular microtrusses. Chapter 4 investigates the potential of work hardening and nanocrystalline electrodeposition as strengthening mechanisms in copper microtrusses. Chapter 5 studies strengthening mechanisms in aluminum alloy microtruss cores reinforced with a ceramic coating, further examining the effect of perforation shape. Chapter 6 examines the question of how architecture determines the additivity of strengthening mechanisms in microtruss materials. Finally, Chapter 7 provides a set of conclusions and recommendations for future work.

1.4 REFERENCES


2 LITERATURE REVIEW

The following chapter reviews the fabrication methods and strengthening mechanisms in cellular metal systems. Perhaps the largest single area of research has been the development of aluminum foams, which is now a relatively mature and well-established technological area [1-6]; nine process-routes are available to make metal foams, five of which are established in industry [1].

Several examples of metal foams are depicted in Figure 2.1 [1,7]. Figure 2.1a shows a melt/gas injection technique where gas is bubbled through a viscous melt into cooling liquid metal. Figure 2.1b shows a melt/gas forming particle method where foaming agent particles are added into a liquid metal and subsequently heated to cause the particles to decompose and release gas (e.g. H\textsubscript{2}). Figure 2.1c shows a semi-solid/gas forming particle method, similar to the melt/gas forming particle method described above, with foaming agent particles added into a solidified powder matrix instead. Figure 2.1d shows the melt cast/foam template mould and is based on investment casting using a polymer foam template. Figure 2.1e uses the vapour deposition/foam template method where polymer foam templates are used with chemical vapour deposition. Figure 2.1f shows foam made by gas atomized hollowed spheres. In this method, gas is used to produce hollow spheres which are hardened and joined.

This literature review will focus on three areas: first, the synthesis of copper-based stochastic foams, second, the fabrication of microtruss materials, and third, the strengthening mechanisms that have been used to reinforce both foams and microtruss materials.
2.1 SYNTHESIS OF COPPER FOAMS

Stochastic foam can be produced with open or closed connectivity. The type of processing method will dictate the type of metal foam that is available to be made with a range of relative density, cell size and cell type accordingly. The ideal foam should have a high degree of uniformity in terms of strut ligament and void dimensions, be free of impurities and show good density distribution [8]. Structural applications for copper foam include energy absorption where porosity, specific strength, ductility in compression and cost are important considerations. These terms often dictate a closed-cell structure that is composed of fused hollow metal spheres [9]. When used as an electrode, the foam must have sufficient mechanical strength to avoid physically disintegrating over time [10]. Functional applications for copper foam include thermal insulation, heat dissipation, catalyst support and sound absorption; these applications require an open-cell to allow for continuous fluid flow [9,11].
One issue that has often been noted is the introduction of strength-limiting defects into the cellular architecture during fabrication. For example, in the electrodeposition technique, dendrite formation is common [12-14]. In the traditional sintering methods, residual material may remain between the fused metal particles thereby weakening the joints and there is a limitation in the control of thickness and porosity of the final foam product [8,9,11,15]. In each case, the fabrication-induced deformation decreases the mechanical strength of the copper foam.

Copper foam can be broadly classified into 3 methods of fabrication: sintering, electrodeposition and solidification; these methods are discussed below.

The sintering method to make copper foam involves coating a template with a metal slurry, heating to decompose the template followed by sintering at high temperature. Zhang et al. [16] and Xie et al. [11] produced copper foam by coating a polyurethane template with the metal slurry and removing the template using an appropriate heat treatment thereby creating a solid network of copper struts. The copper is oxidized when heated in air and reduced during sintering. This is a multi-stage process that can exhibit inconsistencies in the final product; shrinkage, polymer loss and copper oxidation take place concurrently during the decomposition of polyurethane resulting in inorganic residues between metal particles leading to a decrease in mechanical strength. Figure 2.2a shows a sintered open-cell copper foam structure with porosity ranging from 85-91%. Figure 2.2b is a close-up of the foam showing porosity resulting from sintering [11].
Another sintering method involves the sintering of a compact metal powder, also known as Lost Carbonate Sintering (LCS) [8]. This approach is advantageous for producing a foam with controlled cell shape, cell size and porosity distribution [8]. The final cell size of foam depends on the starting metal powder particle size. Its limitations include residual carbonate particles from the sintering process that weaken the foam ligaments.

Figure 2.3 presents the microstructure of copper foam of 35% porosity synthesized by using poly(methyl) methacrylate (PMMA) beads as space holders [16]. This sintering synthesis method requires mixing polymer beads that are greater in size than the metal powder, followed by sintering to evaporate the polymer beads leaving behind voids. Macro-pores, ~200-500 µm, shown in Figure 2.3a, are left over from PMMA burnout and prevent irregularly-shaped pores from forming. Although the purpose of the space holders is to keep the size and shape of the voids uniform, volume shrinkage is observed after sintering. This structure is considered closed-cell based on macro-pores; however, the macro-pores are often filled with micro-pores, ranging from
10-20 μm, which are considered defects. The micro-pores formed due to high friction between the metal powder particles and are found on cell edges and faces (Figure 2.3b). Binder residue also hinders fusion between adjacent metal particles, and affects the strength of the overall structure.

Figure 2.3. Microstructure of copper foam of 35% porosity synthesized by using PMMA space holder technique, (a) large spherical pores are created using PMMA starting powder, (b) micro-voids between the powders (Figure from [16]).

The second broad category of synthesis method for creating copper foams is electrodeposition [12-14], which involves the reduction of metal ions from an aqueous solution. Pore size and the thickness of the wall structures are dictated by adjusting the deposition conditions [12-14]. Shin et al. [12] used electrodeposition in conjunction with hydrogen evolution to make a porous metal that had pore size increasing with increasing distance from the substrate. The structure of this metal foam is more suitable for applications that include electrodes in fuel cells, batteries and sensors where fast
transport of electroactive species is important. Producing copper foam by
electrodeposition has resulted in the formation of dendrites and powder-like crystallites
as observed by Nikolic et al. [13], which decreases the mechanical strength of the foam
by interrupting the uniformity of the deposit. Aly et al. [10] observed the fracture of open
cell copper foam under tension and noted that non-uniform strut thickness or struts
containing voids weakened the foam’s mechanical properties.

Finally, Nakajima et al. [17] produced a copper foam by unidirectional
solidification of liquid copper under applied hydrogen pressure. It was shown that pore
size and pore growth depended on melting temperature and the resulting pore size and
porosity distribution were non-uniform due to the difficulty in controlling the gas
pressure and freezing direction. Figure 2.4 shows copper foam of 32.6% porosity (Figure
2.4a) and 44.6% porosity (Figure 2.4b). It was shown that most hydrogen in the molten
copper cannot dissolve in the solidified copper because of the low solubility in the solid
state; hydrogen exits at the solid-liquid interface and forms elongated pores which align
parallel to the solidification direction.
Figure 2.4. Optical microscopy images of transverse (above) and longitudinal cross-section (below) of porous copper: (a) 32.6% porosity; (b) 44.7% porosity (Figure from [17]).

2.2 MICROTRUSS FABRICATION

The fabrication of microtruss materials falls into four broad categories: investment casting, honeycomb structures, textile lay-up and deformation of metal precursors (some examples are depicted in Figure 2.5). Reviews have been conducted recently by Wadley [7] and Sypeck [18]. This section summarizes examples of Al-based and Cu-based microtrusses made to-date.
Figure 2.5. Microtrusses can be honeycomb, prismatic or lattice truss. Shapes for honeycomb include hexagonal, square and triangular. Shapes for prismatic include triangular, diamond and navtruss. Lattice truss shapes (which can be based on solid or hollow trusses) range from tetrahedral, pyramidal, three-dimensional Kagome, diamond textile, diamond collinear and square collinear (Figure from [7]).

Investment casting has been used to produce both aluminum [19] and copper alloy sandwich cores [20]. As shown in Figure 2.6, a polymer template is produced by rapid prototyping before being coated with a ceramic slurry and hardened. Finally, molten metal is poured into the shell with the polymer burning out to create a single piece (i.e. outer panels and microtruss core).

Figure 2.7a shows a tetragonal lattice structure made from Cu-2% Be alloy, chosen for its high fluidity, made by rapid prototyping followed by investment casting [21]. Another example of investment casting is shown in Figure 2.7b which shows a 3-dimensional Kagome sandwich panel made from Cu-1.8 % Be [22].
Given the small internal strut cross-sections, these architectures are often subject to casting defects which are detrimental to the mechanical performance [21]. Chiras et al. [21] achieved a relative density of 2% however disadvantages include casting defects such as porosity in the region of the node and face sheet (example shown in Figure 2.8).
Microtrusses can also be made using a textile approach: weaving or stacking. Weaving involves laminating woven meshes but it is difficult to achieve a relative density of less than 10% [18]. Stacking involves the organized layering of hollow tubes or solid wires in a periodic fashion seen in Figure 2.9 in a square or diamond orientation with a relative density as low as 3% [23].

Figure 2.9 shows a metal textile lay-up where metal wires or rods can be stacked in an alternating pattern resulting in a woven sandwich core. Tian et al. [24,25] produced copper microtruss woven mesh using wires, presented in Figure 2.10, made from copper alloy C11000 using transient liquid phase bonding, then brazing plane weave copper meshes. The wire mesh offers high surface area but with less porosity ranging from ~68-82% [25] compared to what is achievable with conventional copper foam that can have porosity as high as 97% [11].
Figure 2.9. Fabrication of PCMs using textile lay up using both solid and hollow trusses in square and diamond orientation (Figure from [23]).
Finally, microtrusses have been produced by deforming perforated sheet metal precursors using the bending brake and perforation-stretching methods. Figure 2.11 shows the bending brake approach using a mated V-die set. A pre-perforated flat metal sheet is passed through the die at regular intervals to form the sheet. The result is a regular periodic bend/crimp in the metal sheet. The drawbacks to this approach are that the v-shape (angle) must be decided beforehand, a die must be made specifically for the task, there is wastage from the perforated material and each row of the architecture must be deformed one step at a time.
In the stretch forming approach, a method pioneered by Sypeck [18], a mated pin die is used, Figure 2.12. Again, a pre-perforated metal sheet is placed between the mated dies and a force is applied causing a vertical displacement of the nodes. The metal is stretched. This method is attractive because it is simple and quick in terms of industrial widespread scaled-up production. Drawbacks include the wastage from the perforated material and the use of multiple presses depending on the formability of the metal precursor and the desired amount of deformation. In addition, Sypeck stated that the stretch forming approach was limited to materials with high formability, ruling out aluminum alloys [18]. However, it has been shown that technologically important alloys including AA3003 [27-30], AA6061 [31], high Mn steel [32], austenitic stainless steel
and electrolytic tough pitch copper are all capable of being formed into microtrusses using the stretch bend approach.

Figure 2.12. Periodic cellular metal formed by placing a flat perforated metal sheet into a press and using a stretch-bend method (Figure from [18]).

2.3 STRENGTHENING MECHANISMS

Comparatively less attention has been given to the application of strengthening mechanisms in cellular metals. This section first examines mechanisms that have been applied to stochastic foams and then examines mechanisms that have been applied to microtruss cellular metals.

2.3.1 Strengthening Mechanisms in Foam

The bending-dominated and lightweight nature of foams makes them the attractive material choices for energy-absorbing applications [1]. Attempts to improve
their mechanical properties have been made using both microstructural and coating strategies.

Foam strength has been increased by the addition of dispersions and particle reinforcements into the melt. Insoluble reinforcing particles such as SiC [e.g. 33,34], Y₂O₃ [35], Al₂O₃/SiC (e.g.[36]), reinforcing fibers [e.g. 37] and second-phase precipitation from the solid state (precipitation hardening in heat treatable AA6061 aluminum alloy [38]) have been used to increase the mechanical properties of stochastic foams. Figure 2.13 shows two examples of aluminum foam strengthened using SiC and Si precipitates [33].

![Figure 2.13. Scanning electron micrographs showing the microstructures of (a, b) Alcan and (c, d) Alporas aluminum alloy foam ligaments strengthened using Si precipitates and SiC particle inclusions (Figure from [33]).](image-url)
In the study performed by Margevicius et al. [39], aluminum AA6101 alloy foam was compressed (from an initial relative density of 9.8%) using two methods: cold isostatic pressing and rolling, at 25°C and 200°C (Figure 2.14). Cold isostatic pressing increased the relative density of the foam from its starting relative density of 9.8% to 20%. Multiple passes of rolling increased the relative density of the foam from 9.8% to 18-19% at 25°C, and from 9.8% to 23% at 200°C. The bend tests of the foams showed similar bending pathways with applied load; however, the reduced performance of the sample heated to 200°C (compared to samples of the same relative density but processed at room temperature) was indicative of recovery [39].

![Figure 2.14: Room temperature bend tests of foams compacted by rolling and cold isostatic pressing (CIP) at 25°C and 200°C (Figure from [39]).](image)

Finally, nanocrystalline coatings have also been used to strengthen aluminum foams in order to create a hybrid cellular structure. Nanocrystalline materials are advantageous in terms of mechanical properties due to the Hall-Petch strengthening effect which becomes significant at grain sizes below 100 nm, large volume fraction of
grain boundaries and triple lines act as barriers to impede dislocation motion [40,41]. When grain size is reduced to the nanometer scale (5-100nm), significant increases in yield strength, tensile strength and hardness can be obtained [e.g. 41,42].

Electrodeposition is a non-line-of-sight technique, and thus it is well-suited to coat the irregular 3-dimensional internal structure of foam, creating a continuous sleeve of nanocrystalline material on the cellular foam structure. Figure 2.15 shows the cross-section of an aluminum foam electrodeposited with nanocrystalline nickel [43] where a 5-fold increase in peak strength was observed after electrodeposition. Electrodeposition, however, also introduced a weight penalty to the structure, increasing the density by a comparable factor [43]. The results of this study were similar to those seen by Boonyongmaneerat et al. [44] who also observed a significant strength increase after electrodepositing a nanocrystalline Ni-W alloy on aluminum foam.

Figure 2.15. Scanning electron micrograph showing an aluminum foam coated with nanocrystalline nickel (Figure from [44]).
2.3.2 Strengthening Mechanisms in Microtrusses

Microtruss structures are typically used as the core component between face sheets, forming a sandwich panel. The struts of microtruss cores generally fail by tensile or compressive yielding and by buckling [45]. Since the buckling strength of a slender column in compression will typically be substantially lower than the yield strength of the material, it is the critical buckling stress of the composite strut that will greatly influence the overall mechanical properties. This is typically studied by measuring the compressive strength of the structure.

In microtrusses, strengthening has been achieved by precipitation hardening in a cast Cu-2%Be alloy [21] and in a wrought AA6061 alloy [26]. Kooistra et al. [26] studied the mechanical performance of tetrahedral truss aluminum 6061 alloy cores (Figure 2.16), in the annealed and age hardened condition with relative densities, $\rho_{R}$, ranging from 2–8.3%. At the highest relative density of 8.3%, the peak strength of an age hardened core (Figure 2.16b) can be increased by ~133% compared to its annealed counterpart (Figure 2.16a). Compressive stress-strain testing showed that the mechanical compressive performance is affected by two factors: the relative density of the core and the metallurgical state of the core alloy [26].
Figure 2.16. The nominal compressive stress-strain curves for five relative densities of (a) annealed and (b) age-hardened tetrahedral truss aluminum AA6061 alloy cores (Figure from [26]).

Two fabrication methods can induce work hardening in microtrusses: stretch-bending and bending-brake. In the stretch-bend approach, a perforated sheet is mechanically deformed by stretching alternating nodes out of plane, where plastic strain is accumulated during fabrication [30,46,47]. While aluminum alloy microtrusses have
shown to be challenging to stretch-bend due to early pin punch-through failure [18], Bele et al. [30] showed that work hardening introduced during fabrication can double the compressive peak strength in non-precipitation hardenable microtrusses of AA3003. The strength of the microtruss was preserved despite the localized heat-affected zones from the brazing of sandwich panel to microtruss core [30]. The range of architectures is limited by the maximum displacement depth of the press before sheet failure [18,27]. Starting sheets are typically softened using a pre-annealing treatment prior to forming in order to increase formability [27,46]. The effect of work-hardening and post-fabrication annealing is shown in Figure 2.17 for AA3003 aluminum alloy and SS304 stainless steel alloy [28].

In the bending-brake approach, the deformed microstructure is present before fabrication [48] from the perforated or expanded starting sheets, which are subsequently formed in a bending press and formed into an accordion-style structure. Plastic deformation is concentrated in the hinge region (see Figure 2.18a) where the common failure mode is transverse cracking at the convex surface of the hinge/bend, related to a minimum value of bending radius-to-sheet thickness [49,50]. By partially annealing the work hardened starting sheets, it was possible to fabricate microtruss materials having a significant degree of work hardening in the final architecture (Figure 2.18b).
Figure 2.17. Microtruss mid-strut microstructures of (a, b) work hardened AA3003 aluminum alloy; (b) AA3003 aluminum alloy post-fabrication annealing; (c) work hardened SS304 stainless steel alloy and (d) SS304 stainless steel alloy post-fabrication annealing (Figure from [28]).
Figure 2.18. (a) Microhardness map from finite element analysis of an aluminum alloy microtruss core hinge region post-annealing and (b) deformation banding shown by the superimposed arrows within grains in regions of greatest strain (Figure from [48]).

The addition of a nanocrystalline coating on a microtruss core significantly increases its mechanical properties, resulting in the hybrid structure failing during uniaxial compression by a combination of inelastic buckling collapse and nanocrystalline coating fracture mechanisms [31,51]. An electrodeposited nickel coating has been shown to create a continuous sleeve on a microtruss with either a metal core (e.g. aluminum core with nanocrystalline NiFe coating [31]) or a polymer core with nanocrystalline Ni coating [52] whereby the coating is optimally positioned away from the neutral bending
axis of the microtruss struts, creating a large second moment of area in the electrodeposited sleeve, see example in Figure 2.19 [31,51].

Finally, anodizing can also be used to reinforce aluminum alloy microtrusses. Bele et al. [53] showed that a ceramic coating using aluminum oxide of AA3003 aluminum alloy can increase compressive strength by over 140% without the weight penalty associated with the use of nanocrystalline coating. Further, the compressive strength and failure mechanism is dependent on coating thickness: at low thicknesses, compressive strength is controlled by the group buckling of the internal struts, while at higher thicknesses, the compressive strength is controlled by coating fracture and local deformation in the hinge region of the strut [53].
2.4 SUMMARY

In summary, strength-limiting defects are often introduced into the cellular architecture of foams during fabrication, decreasing their mechanical properties. Stretch-bending can be used to fabricate low relative density architectures that are comparatively defect-free. Limitations with this method may include the need for multiple deformation cycles to achieve the desired architecture for comparatively lower formability of metal precursors.

Unlike the effect of architecture, strengthening mechanisms in both foam and microtrusses has been given comparatively little attention. In microtrusses, both precipitation hardening and work hardening have been successfully used as strengthening mechanisms, as have structural coatings. In spite of these earlier studies, a strategy for the systematic co-selection and design of materials and geometry is underdeveloped. The remaining chapters begin to address these issues and focus on how material modification (before, during and after fabrication) can affect the buckling resistance of metal-based microtrusses.

2.5 REFERENCES


3 EXPERIMENTAL METHODS AND MATERIALS

Pyramidal microtruss cores were formed using copper (C11000) and aluminum (AA2024) as the starting parent materials. C11000 (99.9 wt% Cu, 0.04 wt% O) is an electrolytic tough pitch copper alloy widely used in industry because of its mechanical and electrical properties and its excellent formability [1]. The high formability of C11000 allowed the accumulating effects of work hardening during stretch-bend fabrication to be studied in detail. AA2024 (93.5 wt% Al, 4.4 wt% Cu, 1.5 wt% Mg, 0.6 wt% Mn), is an aluminum alloy typically used in industry for applications such as aircraft structures, rivets, hardware, and truck wheels [2]. AA2024 allowed the interactions of fabrication induced work hardening and precipitation hardening within microtruss cellular metals to be studied. This chapter details the forming, heat treatment, structural coating techniques (electrodeposited nanocrystalline NiFe for C11000 and hard anodized Al₂O₃ for AA2024), mechanical testing (tensile, confined compression of microtruss cores, microhardness), and characterization (scanning electron microscopy, optical stereoscopy) of the two material systems.

3.1 MICROTRUSS FABRICATION

3.1.1 C11000 Pyramidal Cores

Pyramidal copper microtruss cores were produced for the first time using stretch-bend fabrication following a process first described by Sypeck and Wadley [3]. Samples were produced from square-punched electrolytic tough pitch copper (C11000) sheet, purchased from Woven Metal Products, Inc. (Alvin, TX). The 90.82 mm² square perforations were punched on a 2-dimensional square lattice of unit cell sizes 12.4 mm x 12.4 mm (Figure 3.1a), creating an open area fraction of $\phi = 0.56$ where open area is
defined by the area fraction that has been punched out of the starting sheet. Strut geometry is shown in Figure 3.1a where \( w_0 \) is the initial width, \( l_P \) is the length of perforation and \( l_U \) is the length of microtruss unit cell.

![Diagram of perforated sheet and microtruss unit cell](image)

**Figure 3.1.** Schematic diagram of a pyramidal microtruss core (a) Perforated starting sheet of C11000 copper alloy where \( w_0 \) is the initial width, \( l_P \) is the length of perforation and \( l_U \) is the length of microtruss unit cell and (b) strut parameters of a microtruss showing the perforation stretching displacement \( h \) and strut geometry showing thickness, length and width corresponding to \( t \), \( l \) and \( w \) respectively.

Pyramidal microtruss cores (Figure 3.1b) were fabricated from the perforated sheet by placing the coupons in a specially designed mechanical press (Figure 3.2a) such that alternating nodes of the precursor sheet were deformed above and below the starting plane in order to give a three-dimensional microtruss architecture. A constant displacement rate of 1 mm/min was applied using a Shimadzu AG-50KNI screw-driven compression platform. The pins of the perforation-stretching die had a 3.2 mm diameter and a fillet radius of 1 mm at the point of sheet metal contact, designed to just fit inside of the 3.2 mm x 3.2 mm nodes.
Figure 3.2. (a) Mechanical press used to stretch form perforated metal sheets into pyramidal microtrusses. prior to forming, (b) Schematic diagram of a pyramidal microtruss unit prior to and (c) after stretch forming.

The effect of pre-annealing on C11000 microtruss fabrication was investigated by annealing perforated sheets in a 500°C nitrate salt bath for times ranging from 30 s to 24 h. Non-standard tensile coupons, Figure 3.3, were cut from perforated starting sheets and were used to measure the tensile properties of as-received and heat treated C11000. Tensile testing was conducted using a Shimadzu AG-50KNI screw-driven load frame at a crosshead displacement rate of 1 mm/min and loaded until fracture. At first, individual struts were tested in tension (Figure 3.3a) however the relatively short grip region resulted in slippage, seen as displacement at nearly constant load (Figure 3.4). Increasing the grip area by a factor of ~4.5 (Figure 3.3b) allowed individual struts to be tested without slippage.
Prior to fabrication, the C11000 perforated sheets were recrystallization annealed at 500°C for 30 min in a nitrate salt bath [1]. To increase the range of accessible cellular architectures, a two-step forming process was also introduced for a subset of the microtruss cores. Samples were stretched to 80% of their maximum forming limit after the initial recrystallization anneal and then annealed again for 30 min at 500°C, before stretch forming for a second time.
3.1.2 AA2024 Pyramidal Cores

Because of the lower formability of AA2024 compared to C11000, the effect of perforation geometry was also studied in order to enhance the amount of work hardening that could be induced during fabrication and also to increase the range of accessible architecture space. Rounded corners have been shown to eliminate the stress concentrations that lead to failure under the pin by re-distributing the plastic strain to the mid-strut region [4].

Three starting sheet geometries, designated 0.6\(w_o\), 0.7\(w_o\) and 0.9\(w_o\), were used in this thesis and are shown in Figure 3.5 where \(R\) is the radius of curvature, \(w_o\) is the initial width (3.2 mm), \(l_p\) is the length of perforation and \(l_U\) is the length of microtruss unit cell.

Figure 3.4. Stress-strain curve showing the effect of grip area when tensile testing non-standard C11000 coupons.
Dimensions for $0.6 w_o$, $0.7 w_o$ and $0.9 w_o$ are summarized in Table 3.1. The number designation of “0.6”, “0.7” and “0.9” refer to the fractional width at mid-strut of the rounded-edge perforations compared to the conventional square-edge geometry. Sheets were obtained from The Conard Corporation (Glastonbury, CT) who used photochemical etching to perforate the starting sheets to the specified measurements. Photochemical etching is a multi-stage process which involves the placement of a photoresist stencil on a cleaned metal followed by etching and subsequently photoresist stripping [5]. The use of photochemical etching allows the fabrication of high resolution and complex geometrical shapes in a flat sheet of metal [5], without introducing mechanical or thermal stresses into the workpiece [6].

<table>
<thead>
<tr>
<th>Strut width, $w_o$</th>
<th>Initial sheet thickness, $t_o$ (mm)</th>
<th>Radius of curvature, $R$ (mm)</th>
<th>Length of perforation, $l_P$ (mm)</th>
<th>Length of microtruss unit cell, $l_U$ (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.6</td>
<td>0.79</td>
<td>2.22</td>
<td>10.78</td>
<td>12.70</td>
</tr>
<tr>
<td>0.7</td>
<td>0.79</td>
<td>1.67</td>
<td>10.46</td>
<td>12.70</td>
</tr>
<tr>
<td>0.9</td>
<td>0.79</td>
<td>0.58</td>
<td>9.82</td>
<td>12.70</td>
</tr>
</tbody>
</table>

In addition to the effect of perforation geometry, the effect of initial material state was also investigated. Perforated sheets were stretch-formed until failure from both the solutionized and O-temper state. A solutionizing temperature of 493°C [1] was used for 30 min while the O-temper involved a solutionized treatment of 493°C for 30 min + 413°C for 3 h [1]. Artificial aging was conducted at 191°C for times ranging from 3 to 12 hrs [1]. While both O-temper and solutionized materials contain large insoluble $\text{Al}_7\text{Cu}_2\text{Fe}$ precipitates, the O-temper material contains a coarser distribution of $\text{CuAl}_2$ and $\text{CuMgAl}_2$.
precipitates [7]. A multi-step forming procedure, similar to that developed for the C11000 microtrusses but with up to 4 annealing/forming steps, was developed to fabricate the AA2024 microtruss cores.

3.2 STRUCTURAL COATINGS

Structural coatings were also applied to the microtruss cores in order to increase their mechanical performance. Nanocrystalline NiFe coatings were electrodeposited on the copper microtruss cores by Integran Technologies Inc. (Toronto, CA), while a hard
anodized ceramic coating was introduced on the aluminum microtruss cores by CP Tech, a division of Technimeca International Corporation (Montréal, CA).

### 3.2.1 Electrodeposition of Nanocrystalline NiFe on Copper Microtrusses

The microtruss copper cores reinforced by nanocrystalline electrodeposition had a final height of 10.1 ± 0.21 mm and a truss angle of 45 ± 1.2°, corresponding to a relative density of 3.2 ± 0.2%. Electrodeposition was conducted after [8,9]: a modified Watt’s nickel bath containing iron chloride (source of iron), sodium citrate (complexing agent) and saccharin (stress reliever, grain refinement) with a consumable Ni anode at a pH of 2.5 and using pulsed deposition in the range of 2.5 to 10 ms and pulse-off times in the range of 10 to 100 ms. The nominal thickness for the electrodeposited nickel-iron was calculated assuming a uniform distribution of mass over the microtruss core using

$$\frac{m_{n-NiFe}}{S_A \rho_{NiFe}} = \frac{m_{n-NiFe}}{S_A \rho_{NiFe}}$$

where $m_{n-NiFe}$ is the electrodeposited mass, $S_A$ is the electroplated surface area, and $\rho_{NiFe}$ is the density of nickel-iron sleeve. Sleeve thickness (ranging from ~18 µm to 120 µm) was controlled by varying the length of the electrodeposition time. Alloy composition was determined to be an average of 52.9 wt% Fe-42.1 wt% Ni along the strut length using energy dispersive X-ray spectroscopy (EDS). A backscattered SEM image of the cross-section of the electrodeposited NiFe coating ($t_{n-NiFe} = 120$ µm) on a C11000 alloy microtruss core is shown in Figure 3.6.
Figure 3.6. Representative SEM micrograph of a C11000 microtruss strut with an electrodeposited coating of nanocrystalline NiFe ($t_{n-NiFe} = 120 \mu m$).

The nanocrystalline structure of the NiFe sleeve was confirmed through X-ray diffraction (XRD) using Co-Kα X-rays ($\lambda = 0.179$ nm) on reference coupons that were electroplated using the same deposition conditions as the copper microtruss (Figure 3.7). The Scherrer formula was used to estimate the grain size of the crystals, $t$, and to confirm the nanocrystalline scale of the grains:

$$t = \frac{0.9\lambda}{B \cos \theta_B}$$

where $\lambda$ is the X-ray wavelength, $B$ (measured in radians) is the instrument corrected peak full width at half maximum intensity and $\theta_B$ is the angle at $B$ [10]. The grain size was determined to be 15 nm, which is typical of pulse current electrodeposited NiFe [8].
3.2.2 Hard Anodized Coating on Aluminum Microtrusses

Hard anodizing of AA2024 is typically conducted by immersing the component in an electrolyte of 10-15wt% sulphuric acid, maintained at an operating temperature ranging from 0°C to 10°C and passing an electric current through the electrolytic solution [11]. AA2024 microtrusses with 0.6w₀ perforation geometry were formed to a truss height of 10.8 ± 0.12 mm and a truss angle of 45 ± 1.1°, corresponding to a relative density of 3.6%. AA2024 microtrusses were also formed with 0.9w₀ perforation geometry to a truss height of 9.82 ± 0.1 mm and a truss angle of 45 ± 1.0°, corresponding to a relative density of 5.3%. The samples of the present study were anodized at CP Tech in Montreal, Quebec with a coating thickness of 40 ± 1.3 µm (the maximum achievable
coating thickness for AA2024 [2]) measured using optical microscopy and based on the average of 10 measurements taken from the coated top and bottom edges. A hard anodized ceramic coating was produced with no sealant.

Representative SEM micrographs taken in backscattered mode of a composite AA2024 microtruss strut, with a perforation geometry of 0.9\(w_o\) and a 40 \(\mu\)m thick hard anodized \(\text{Al}_2\text{O}_3\) coating are shown in Figure 3.8. Figure 3.8a shows a uniform coating on the top and bottom surfaces but uneven/highly non-uniform side surfaces and pores visible within the coating while Figure 3.8b shows a close-up of the side region having uneven and thin to no coating adherence as well as pores near the surface. The difference in surface morphology and anodizing-ability is related to the different final processing steps for each surface: the well-anodized surfaces were in the as-rolled condition, while the other was in the as-chemically etched condition. Photochemical etching uses a strong alkaline or acid etchant to dissolve the exposed metal; grain boundaries and individual grain surfaces are both subject to the chemical attack which can lead to an uneven surface [12]. Typical etchants for AA2024-T3 are \(\text{NaOH} + \text{Na}_3\text{S}\) and \(\text{NaOH} + \text{triethanolamine}\) [12]. The type of etchant selected will impact both etching rate and appearance [13]. For example, pure Al reacts very slowly compared to an aluminum containing copper (e.g. AA2024) that demonstrates an increased rate of chemical attack [14].

The aluminum oxide coating on the as-rolled surfaces of the AA2024 microtruss was also characterized by XRD using Cu-K\(\alpha\) X-rays (\(\lambda = 0.154\) nm). Figure 3.9 shows a characteristic amorphous structure with peaks of relatively strong intensity matching those for reference Al. During anodization, two types of anodic oxides are formed: barrier oxide films and porous anodic oxide films. Barrier oxide films prepared by
anodization have been shown to be amorphous in nature with the underlying Al surface exhibiting strong diffraction lines [15]. This observation was consistent with the results of the present study.

Figure 3.8. Representative SEM micrographs of a AA2024 microtruss strut, with perforation geometry $0.9w_o$ with a hard anodized with $\text{Al}_2\text{O}_3$ coating ($t_{\text{Al}_2\text{O}_3} = 40 \, \mu\text{m}$) showing (a) overall cross-section and (b) close-up of microtruss edge.
3.3 MECHANICAL TESTING

3.3.1 Confined Compression Testing

To simulate the mechanical performance that stand-alone microtruss cores would demonstrate in a sandwich structure during uniaxial compression [16], microtruss cores were tested using confinement plates that restrict the lateral movement of the nodes in order to induce inelastic buckling failure, Figure 3.10 [17]. The confinement plates were made by machining an orthogonal array of 1.2 mm deep by 3.5 mm wide channels into a 5 mm thick plain carbon steel plate. The channel depth was 167% of the copper strut thickness and 150% of the aluminum sheet thickness, which in both cases was sufficient to keep the microtruss core nodes mechanically locked into place. Confinement plates generally eliminate edge effects, allowing microtruss properties to be determined from small-scale samples, as small as 2 x 2 unit cells where the peak load per strut has been shown to be independent of the number of struts in the sample [17]. Uniaxial confined

Figure 3.9. X-ray diffraction pattern of the aluminum oxide coating on a AA2024 alloy microtruss using Cu-Kα radiation, λ = 0.154 nm.
compression was performed at a cross-head displacement rate of 1 mm/min. The compressive strain was estimated from the crosshead displacement [18-22]. Five samples were tested for each material condition; mechanical properties are reported as the mean plus/minus the standard deviation. Note that each test in itself represents the average behaviour of 40 individual struts.

Figure 3.10. Confinement plates used to simulate the mechanical performance a microtruss core would exhibit when used as the interior of a sandwich structure [17].

3.3.2 Microhardness

Samples were mounted longitudinally in epoxy to expose the strut length cross-section and prepared using standard metallographic techniques. C11000 was prepared using a 5-step procedure with 320 grit SiC to achieve a planar surface followed by 6 µm diamond suspension, 3 µm diamond suspension, 1 µm diamond suspension and polished using 0.05 µm alumina suspension. AA2024 was prepared using a 4-step procedure using 320 grit SiC to achieve a planar surface followed by 6 µm diamond suspension, 3 µm diamond suspension and polished using 0.05 µm alumina suspension.

Microindentation hardness measurements were taken using a MHV 2000 microhardness tester with a 0.98 N applied load and 10 s dwell time according to ASTM standard E384-11e1 [23]. The hardness number is based on the size of the indentation
made by a square-based pyramidal-shaped diamond indenter formed in the surface of the
mounted and polished strut, where the Vickers hardness (HV) number is based on the
average size of the two indentation diagonals. Indentations were made along the strut
profile and spaced at least 2.5 times the diagonal length of the diamond indenter used in
the Vickers hardness test [24]. A typical indentation profile is shown below, Figure 3.11.

![Microhardness profile in a cross-sectional profile of a C11000 microtruss strut.](image)

**3.4 MICROSCOPY**

In order to study the failure mechanisms during uniaxial compression, a subset of
samples was pre-loaded to characteristic strain values, unloaded and examined by
scanning electron microscopy (SEM) and/or optical stereomicroscopy. Optical
microscopy was used in order to examine architectural collapse with a high depth of
focus over the entire sample. Profile and fracture surface characterization was performed
using a SM2800 stereomicroscope with Nikon Olympus SC30 (camera) and D.E.
Olympus software. The challenge with SEM was having a combination of the largest
possible working distance and lowest magnification in order to investigate the structural coating failure mechanisms over the entire length of the strut. Regions of interest, i.e. zones of crack formation, were investigated at higher magnification, and imaged using secondary electrons. Adobe Photoshop CS6 was used to stitch together SEM images taken in a series for a single sample. A single strut of Al₂O₃/AA2024 was also sectioned and mounted onto carbon tape on a stage and subsequently gold-coated and characterized by a Hitachi S-4500 field emission scanning electron microscope. Profile characterization of the microtruss coating and core of Al₂O₃/AA2024 was prepared using standard metallographic practices described in the Section 3.3.2, carbon coated and viewed using a Hitachi S-570 scanning electron microscope in backscattered mode.

3.5 REFERENCES


4 COPPER MICROTRUSSES

4.1 STRETCH-BEND FABRICATION

4.1.1 First Cycle

Figure 4.1 presents a typical forming curve for C11000 in the as-received condition. This forming curve can be subdivided into two stages: bending-dominated and stretching-dominated, see Figure 4.1. During the bending-dominated stage the nodes plastically deform around the tip of the forming pins. A bending force forming parameter $F_B$ was used to define the onset of plastic deformation during stretch-bend fabrication [1]. The method requires a small offset based from the initial elastic region of the curve; a 0.05 mm offset (approximately 5% of starting sheet thickness) was selected for this study. During the stretching-dominated stage, the struts are elongated as the relative

![Figure 4.1. Stretch-bend fabrication force as a function of displacement for an as-received starting sheet of C11000.](image)
displacement of pins on opposite sides of the press increases. The maximum forming force, $F_{MAX}$, represents the forming limit of the material and corresponds to the maximum perforation stretching displacement, $d_{MAX}$, which occurs before strut failure. In the as-received state for the case of one sample core, fracture at a single node caused the forming limit to be reached, pictured in Figure 4.2. The onset of fracture defines the upper limiting range of architectures that can be fabricated by stretch bending, determining the minimum relative density that can be achieved and the maximum internal truss angle.

The relative density of the pyramidal microtruss core is calculated using the following equation:

$$\rho_R = \frac{1}{1 + \left(\frac{d_M}{t_o}\right)(1 - \phi)}$$  \hspace{1cm} (4.1)

where $t_o$ is the initial starting sheet thickness, $d_M$ is the perforation-stretching displacement, and $\phi$ is the open area fraction. The angle between the horizontal and inclined strut determines the truss angle, $\omega$

$$\omega = \arctan\left(\frac{d_M}{l_o}\right)$$  \hspace{1cm} (4.2)

and is pictured in Figure 4.3. The minimum relative density and truss angle for the as-received state, shown in Figure 4.1, was 5.91 ± 0.18% and 27.1 ± 0.8° respectively.
Figure 4.2. As-received C11000 microtruss formed to failure showing (a) failure at node and, (b) no failure at node.

Figure 4.3. Schematic diagram of the strut parameters of a microtruss showing the perforation stretching displacement $d_M$ and strut geometry showing thickness, initial length and truss angle corresponding to $t_0$, $l_0$ and $\omega$, respectively.

In order to determine a suitable pre-annealing treatment that would allow a greater range of microtruss architectures to be fabricated, the as-received perforated starting sheet of C11000 copper alloy underwent tensile testing and was compared to coupons that were heat treated for a range of times: 15 min to 4 h at 400°C and 30 s to 30 min at 500°C. The goal of heat treatment was to increase the formability. Representative
tensile curves are shown in Figure 4.4 and the tensile properties are summarized in Table 4.1. The as-received coupon had significantly less (up to 64%) total strain until failure compared to the coupons that were heat treated. When comparing the 400˚C treatment to the 500˚C, the samples heat treated at the higher temperature resulted in the most total strain achievable.

Figure 4.4. Representative tensile curves for C11000 non-standard coupons in the as-received condition and heat treated for 15 min and 4 h at 400˚C and 30 s and 30 min at 500˚C.

Figure 4.5 plots the tensile properties as a function of annealing time. Overall, increasing the treatment temperature softens the starting sheet and increases the total strain to failure. Increasing the heat treatment time results in a decrease in the 0.2% yield strength (Figure 4.5a) from 194 MPa in the as-received state to 118 MPa after 15 min at 400˚C and 60 MPa after 4 hrs at 400˚C, with a similar reduction to ~60 MPa occurring after annealing at 500˚C. Total strain to failure is increased with increasing heat treatment time with the greatest amount of strain to failure occurring with coupons treated for 30
min implying that this length of heat treatment time will allow the greatest amount of formability in the stretch-bend fabrication of microtrusses. An increase in heat treatment time, past 30 min, resulted in no significant gain in elongation to failure.

Table 4.1. Summary of the tensile testing properties including yield strength at 0.2% offset ($\sigma_{YS}$), ultimate tensile strength ($\sigma_{UTS}$), and total failure strain ($\varepsilon_T$), with standard deviation for each, for the C11000 alloy in the as-received condition and heat treatment times ranging from 15 min to 4 h at 400°C and 30 s to 30 min at 500°C. Sample size was 5 tensile coupons per set.

<table>
<thead>
<tr>
<th>Heat Treatment Time (s)</th>
<th>Yield Strength, $\sigma_{YS}$ (MPa)</th>
<th>Ultimate Tensile Strength, $\sigma_{UTS}$ (MPa)</th>
<th>Total Strain, $\varepsilon_T$ (mm/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-received</td>
<td>194 ± 8</td>
<td>235 ± 2</td>
<td>0.21 ± 0.02</td>
</tr>
<tr>
<td>400°C</td>
<td>118 ± 12</td>
<td>231 ± 1</td>
<td>0.50 ± 0.02</td>
</tr>
<tr>
<td>900</td>
<td>95 ± 3</td>
<td>232 ± 3</td>
<td>0.53 ± 0.02</td>
</tr>
<tr>
<td>1800</td>
<td>101 ± 5</td>
<td>235 ± 1</td>
<td>0.50 ± 0.02</td>
</tr>
<tr>
<td>3600</td>
<td>60 ± 2</td>
<td>235 ± 1</td>
<td>0.54 ± 0.02</td>
</tr>
<tr>
<td>14400</td>
<td>60 ± 2</td>
<td>235 ± 1</td>
<td>0.54 ± 0.02</td>
</tr>
<tr>
<td>500°C</td>
<td>59 ± 2</td>
<td>208 ± 2</td>
<td>0.56 ± 0.04</td>
</tr>
<tr>
<td>30</td>
<td>66 ± 6</td>
<td>208 ± 2</td>
<td>0.53 ± 0.01</td>
</tr>
<tr>
<td>60</td>
<td>57 ± 7</td>
<td>206 ± 2</td>
<td>0.53 ± 0.03</td>
</tr>
<tr>
<td>300</td>
<td>61 ± 3</td>
<td>205 ± 2</td>
<td>0.55 ± 0.03</td>
</tr>
<tr>
<td>900</td>
<td>64 ± 6</td>
<td>204 ± 1</td>
<td>0.58 ± 0.02</td>
</tr>
</tbody>
</table>

Figure 4.5. Properties from tensile testing of C11000 non-standard coupons in the as-received and heat treated condition at 400°C and 500°C showing (a) 0.2% offset yield strength, $\sigma_{YS}$ and (b) the total strain at failure, $\varepsilon_T$. 

63
Figure 4.6 shows the effect of heat treatment on the forming curves of the perforation-stretched C11000 copper alloy. Generally, heat treatment significantly increases the formability of the alloy, allowing the formation of architectures with higher truss angles and lower relative densities. A summary of these forming parameters are shown in Table 4.2 summarizing the bending force, maximum stretch force, maximum stretch displacement, truss angle ($\omega_M$) and relative density ($\rho_R$). The addition of a heat treatment at 400˚C for 30 min increases the stretch displacement from 5.44 mm (as-received) to 7.97 mm (Figure 4.6a). Increasing the annealing temperature to 500˚C could result in a ~15% increase in stretch forming displacement (Figure 4.6a), with a correspondingly greater architectural space accessed. The heat treatments also had the effect of more uniformly distributing the final strut fracture. For example, Figure 4.7 shows the case of a typical annealed microtruss (30 min at 500˚C) which had four struts fracture simultaneously at the fabrication limit.

![Figure 4.6](image)

Figure 4.6. Stretch-bend fabrication curves showing the effect of heat treatment for (a) the as-received starting sheet, 15 min to 4 h at 400˚C and, (b) the as-received starting sheet, 30 s and 30 min at 500˚C on the maximum stretch displacement $d_{MAX}$ for the C11000 copper alloy.
Figure 4.7. 30 min heat treated microtruss formed to failure showing (a-d) failure at nodes.

Table 4.2. Summary of the bending force ($F_B$), maximum stretch force ($F_{\text{MAX}}$), maximum stretch displacement ($d_{\text{MAX}}$), truss angle ($\omega$) and relative density ($\rho_R$), with standard deviation for each, for C11000 copper alloy as-received starting sheet and heat treated at 400$^\circ$C and 500$^\circ$C. Sample size was 5 microtruss cores per set.

<table>
<thead>
<tr>
<th>Heat Treatment Time (s)</th>
<th>Bending Force, $F_B$ (kN)</th>
<th>Maximum Stretch Force, $F_{\text{MAX}}$ (kN)</th>
<th>Maximum Stretch Displacement, $d_{\text{MAX}}$ (mm)</th>
<th>Truss Angle, $\omega$ (°)</th>
<th>Relative Density, $\rho_R$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-received</td>
<td>0.59 ± 0.04</td>
<td>10.8 ± 0.8</td>
<td>5.44 ± 0.16</td>
<td>27.1 ± 0.8</td>
<td>5.91 ± 0.21</td>
</tr>
<tr>
<td>400$^\circ$C</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>900</td>
<td>0.61 ± 0.07</td>
<td>13.7 ± 0.3</td>
<td>7.83 ± 0.27</td>
<td>37.4 ± 1.1</td>
<td>3.70 ± 0.33</td>
</tr>
<tr>
<td>1800</td>
<td>0.36 ± 0.09</td>
<td>13.6 ± 0.3</td>
<td>7.97 ± 0.16</td>
<td>38.0 ± 0.5</td>
<td>3.64 ± 0.21</td>
</tr>
<tr>
<td>3600</td>
<td>0.19 ± 0.02</td>
<td>14.1 ± 0.3</td>
<td>8.26 ± 0.17</td>
<td>39.0 ± 0.5</td>
<td>3.53 ± 0.20</td>
</tr>
<tr>
<td>14400</td>
<td>0.15 ± 0.01</td>
<td>14.0 ± 0.1</td>
<td>8.48 ± 0.08</td>
<td>39.8 ± 0.2</td>
<td>3.44 ± 0.14</td>
</tr>
<tr>
<td>500$^\circ$C</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>30</td>
<td>0.15 ± 0.07</td>
<td>13.9 ± 0.1</td>
<td>8.48 ± 0.52</td>
<td>40.1 ± 1.9</td>
<td>3.80 ± 0.2</td>
</tr>
<tr>
<td>60</td>
<td>0.15 ± 0.07</td>
<td>14.2 ± 0.2</td>
<td>8.91 ± 0.12</td>
<td>41.7 ± 0.4</td>
<td>3.60 ± 0.1</td>
</tr>
<tr>
<td>300</td>
<td>0.11 ± 0.04</td>
<td>13.9 ± 0.6</td>
<td>8.72 ± 0.44</td>
<td>41.0 ± 1.6</td>
<td>3.69 ± 0.4</td>
</tr>
<tr>
<td>900</td>
<td>0.07 ± 0.05</td>
<td>13.9 ± 0.2</td>
<td>9.13 ± 0.25</td>
<td>42.4 ± 0.8</td>
<td>3.52 ± 0.2</td>
</tr>
<tr>
<td>1800</td>
<td>0.09 ± 0.03</td>
<td>14.5 ± 0.5</td>
<td>9.21 ± 0.19</td>
<td>42.7 ± 0.6</td>
<td>3.49 ± 0.1</td>
</tr>
</tbody>
</table>
Figure 4.8 illustrates the effect of heat treatment time on $F_B$, $F_{MAX}$ and $d_{MAX}$.

Figure 4.8a shows that the bending force rapidly decreases at longer heat treatment times. Figure 4.8b shows that $F_{MAX}$ is lower in the as-received sample but increases substantially and remains relatively constant with the length and temperature of heat treatment. In Figure 4.8c it is seen that the maximum stretch displacement increases with heat treatment time at 500°C. Figure 4.9 shows the relationship between maximum stretch displacement during fabrication and the total strain to failure during tensile testing for the sample conditions considered in the present study. Accordingly, a pre-annealing treatment of 30 min at 500°C was selected as the basis of fabricating and forming the C11000 alloy microtrusses in the remainder of this thesis.

Figure 4.8. Stretch-bend fabrication properties of C11000 alloy for (a) bending force, $F_B$; (b) maximum stretch force, $F_{MAX}$; (c) maximum stretch displacement, $d_{MAX}$, as a function of heat treatment time at 400°C and 500°C.
4.1.2 Multi-Cycle

In order to increase the range of accessible cellular architectures, a two-step forming process was also developed. Samples were first stretched to 80% of $d_{\text{MAX}}$, annealed for 30 min at 500°C, and then stretch formed again until failure. Figure 4.10 presents the deformation forming curves with the stretching force shown as a function of stretch displacement. Figure 4.10a shows a complete 1-step forming curve indicating the maximum stretch displacement of 9.4 mm and the point of 0.8$d_{\text{MAX}}$. Figure 4.10b shows representative forming curves for 2-step forming where the microtruss is first formed to 0.8$d_{\text{MAX}}$ followed by heat-treatment and subsequent second forming step, resulting in a greater stretch displacement overall to $d_{\text{MAX,2}}$. This 2-step forming method opens up a broader range of accessible architecture space at decreased relative density.
Figure 4.10. Stretch force as a function of stretch displacement for (a) one cycle of deformation, and (b) two forming cycles with an intermediate heat treatment step for the stretch-bend deformation forming of C11000 pyramidal microtruss.

Both the 1-step and 2-step forming methods have been shown to make low density cellular copper using a simple stretch-bend fabrication method to produce a cellular copper with good repeatability. Figure 4.11 shows the effect of perforation stretching displacement on truss angle and relative density. The maximum perforation-stretching limit of C11000 is plotted to show that a maximum truss angle of 43° and minimum relative density of 3.5% is achievable for this alloy and geometry when using the 1-step forming method and the maximum truss angle can be further increased to 48° and a relative density of 2.9% when 2-step forming method is used. Overall then, the cellular copper produced by stretch bend fabrication for the first time in this study is able to match or exceed the low densities achieved by the cellular copper synthesis methods discussed in Chapter 2.
Figure 4.11. Relative density and truss angle changing as a function of perforation stretch-bend displacement, showing the maximum stretch limit using the one-step and two-step forming method to fabricate C11000 copper alloy pyramidal microtruss.
4.2 UNIAXIAL COMPRESSION TESTING

4.2.1 Effect of Architecture

The effect of architecture on the mechanical properties was studied using samples that had undergone a final heat treatment to remove the fabrication induced work hardening, providing a set of microtruss cores having a range of truss angles, but essentially the same internal metallurgical state. This was verified by taking microhardness measurements (0.98 N applied load and 10 s dwell time) at the mid-point of the struts. With the exception of one sample type (the two-step microtruss samples having the lowest internal strut angle), the microhardness of the annealed samples ranged from 54 ± 4 HV to 62 ± 3 HV. Note that optical microscopy characterization of the low angle two-step sample (having a hardness of 74 ± 3 HV) indicated that it had undergone incomplete recrystallization during the post-fabrication annealing stage.

Figure 4.12 shows representative compressive stress-strain curves for $\omega = 38^\circ$ and $\omega = 47^\circ$ microtruss cores. There is an initial period of elastic loading, after which point the struts began to fail by inelastic buckling. Since the modulus of the loading curve can be reduced during the initial compressive deformation due to bedding-in of the microtruss into the compression plates [2,3], a reload modulus was defined as the maximum slope of the compressive reload stress-strain curve obtained when the microtruss was first loaded to 75% of the peak stress, unloaded and subsequently reloaded, following the same method used for metallic foams [4,5], a representative curve of the compression testing pathway is shown in Appendix A. Two strength properties can be taken from these curves: the maximum initial stress supported by the architectures (peak strength, $\sigma_P$) and the minimum stress supported during collapse.
(valley strength, $\sigma_v$). The densification energy, $J_D$, is defined as the energy absorbed between stress levels of one half the peak strength ($\frac{1}{2}\sigma_P$) and twice the peak strength ($2\sigma_P$) [6]. The $E_{\text{Reload}}$, $\sigma_P$, $\sigma_V$, and $J_D$ parameters from uniaxial compression testing of annealed microtruss cores are summarized in Table 4.3 (see Appendix A for an illustration of these parameters).

Figure 4.12. Representative compressive stress-strain curves for annealed C11000 alloy microtrusses with truss angle, $\omega = 20^\circ$, $38^\circ$ (fabricated using the 1-step method), and $\omega = 47^\circ$ (fabricated using the 2-step method).

The mechanical properties ($E_{\text{Reload}}$, $\sigma_P$, $\sigma_V$, and $J_D$) as a function of truss angle are shown in Figure 4.13. The reload modulus, Figure 4.13a, increases at a nearly constant rate from 18.4 to 64.4 MPa, as the truss angle increases from $21^\circ$ to $47^\circ$. However, with the exception of one sample set, Figure 4.13b, the peak compressive strengths of the annealed cores are within the comparatively narrow range of 0.57 to 0.71 MPa over the
same truss angle range. It should be noted that the two-step annealed microtruss sample having the lowest strut angle (38 ± 1.9°) had a higher peak strength of 0.98 ± 0.09 MPa than any of the other annealed core microtrusses, which is consistent with the higher mid-strut hardness value seen for this partially recrystallized sample type mentioned earlier.

In terms of the post-buckling collapse mechanism, the valley strength of the 1-step cores was relatively constant ranging from 0.55 to 0.50 MPa while the valley strength of the 2-step cores decreased with increasing truss angle, from 0.65 to 0.32 MPa, Figure 4.13c. Similarly, the energy absorbed during architectural collapse, illustrated by the densification energy, Figure 4.13d, remained approximately constant over the same range of truss angles, with the exception of the two-step annealed microtruss sample that had experienced incomplete recrystallization.

Table 4.3 Summary of one-step annealed (1-ANN) and two-step annealed (2-ANN) C11000 alloy truss core compression testing showing reload modulus of elasticity ($E_{\text{Reload}}$), peak strength ($\sigma_P$), valley strength ($\sigma_V$), densification energy ($J_D$), truss angle ($\omega$) and relative density ($\rho_R$) with standard deviation for each. Sample size was 5 microtruss cores per set.

<table>
<thead>
<tr>
<th>Fabrication Pathway</th>
<th>Reload Modulus, $E_{\text{Reload}}$ (MPa)</th>
<th>Peak Strength, $\sigma_P$ (MPa)</th>
<th>Valley Strength, $\sigma_V$ (MPa)</th>
<th>Densification Energy, $J_D$ (MJ/m$^3$)</th>
<th>Truss Angle, $\omega$ (°)</th>
<th>Relative Density, $\rho_R$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1-ANN</td>
<td>18.4 ± 1.17</td>
<td>0.57 ± 0.05</td>
<td>0.55 ± 0.01</td>
<td>0.42 ± 0.02</td>
<td>21.2 ± 0.24</td>
<td>7.49 ± 0.08</td>
</tr>
<tr>
<td></td>
<td>24.3 ± 0.65</td>
<td>0.65 ± 0.04</td>
<td>0.55 ± 0.02</td>
<td>0.39 ± 0.02</td>
<td>25.4 ± 0.93</td>
<td>6.31 ± 0.22</td>
</tr>
<tr>
<td></td>
<td>29.1 ± 1.04</td>
<td>0.66 ± 0.02</td>
<td>0.60 ± 0.03</td>
<td>0.45 ± 0.03</td>
<td>29.7 ± 0.84</td>
<td>5.37 ± 0.16</td>
</tr>
<tr>
<td></td>
<td>42.3 ± 0.95</td>
<td>0.73 ± 0.04</td>
<td>0.52 ± 0.06</td>
<td>0.32 ± 0.04</td>
<td>34.5 ± 0.24</td>
<td>4.26 ± 0.32</td>
</tr>
<tr>
<td></td>
<td>45.2 ± 0.67</td>
<td>0.71 ± 0.05</td>
<td>0.54 ± 0.02</td>
<td>0.30 ± 0.04</td>
<td>37.3 ± 0.34</td>
<td>4.16 ± 0.01</td>
</tr>
<tr>
<td></td>
<td>46.1 ± 0.09</td>
<td>0.71 ± 0.10</td>
<td>0.50 ± 0.02</td>
<td>0.35 ± 0.03</td>
<td>38.5 ± 0.23</td>
<td>4.00 ± 0.07</td>
</tr>
<tr>
<td>2-ANN</td>
<td>48.7 ± 2.30</td>
<td>0.98 ± 0.09</td>
<td>0.65 ± 0.04</td>
<td>0.56 ± 0.08</td>
<td>37.9 ± 1.94</td>
<td>4.08 ± 0.38</td>
</tr>
<tr>
<td></td>
<td>59.0 ± 1.71</td>
<td>0.70 ± 0.04</td>
<td>0.41 ± 0.02</td>
<td>0.34 ± 0.01</td>
<td>43.1 ± 1.27</td>
<td>3.38 ± 0.06</td>
</tr>
<tr>
<td></td>
<td>64.4 ± 1.70</td>
<td>0.70 ± 0.02</td>
<td>0.32 ± 0.01</td>
<td>0.35 ± 0.01</td>
<td>46.5 ± 0.81</td>
<td>3.08 ± 0.08</td>
</tr>
</tbody>
</table>
Figure 4.13. (a) Reload modulus, $E_{\text{Reload}}$; (b) peak strength, $\sigma_P$; (c) valley strength, $\sigma_V$; and (d) densification energy, $J_D$ are shown as a function of truss angle for 1-step and 2-step annealed C11000 microtruss cores.

The post-buckling softening behaviour of the microtruss cores is an important consideration in terms of the microtruss core’s energy absorption characteristics and can be evaluated in terms of the fractional load drop after the peak strength. The fractional load drop is one measure of the deviation of the microtruss from an idealized energy absorber that fails at a constant plateau stress [4]. Figure 4.14 presents the stress drop from peak to valley divided by the peak stress as a function of truss angle. The increasing
fractional load drop with increasing truss angle can be explained in terms of the microtruss collapse mechanism shown in Figure 4.15. With increasing compressive strain, primary weakening of the strut (Figure 4.15a) occurs, which corresponds to the formation of a plastic hinge at the buckled middle of the strut (Figure 4.15b). The valley strength coincides with the onset of secondary strengthening (Figure 4.15c), i.e. when the buckled strut makes contact with the compression surface. The increased load drop from peak to valley with increasing truss angle seen in Figure 4.14 occurs because of the prolonged primary hinge weakening stage for microtrusses having higher internal truss angles and consequently larger strut lengths, which delays the onset of contact between the buckled strut and the compression surface.

Figure 4.14. The fractional load drop as a function of truss angle for 1-step and 2-step annealed microtrusses.
Figure 4.15. Schematic diagram illustrating microtruss collapse mechanisms during uniaxial compression showing: (a) as-formed microtruss strut, (b) primary weakening of the microtruss strut via inelastic buckling, and (c) secondary strengthening of the strut as it makes contact with the compression plates.

The increased strut length with increasing internal truss angle also affects the peak strength, but in two counteracting ways. On the one hand, there is a tendency for the compressive strength of a microtruss core to increase with increasing truss angle because of the increasingly efficient load resolution with respect to the loading axis. This can be seen in the relationship:

$$\sigma_{\text{truss}} = N\sigma_{\text{strut}} \left( \frac{A_{\text{strut}}}{A_{\text{truss}}} \right) \sin \omega$$  \hspace{1cm} (4.3)

where $\sigma_{\text{truss}}$ represents the overall stress supported by the truss, $N$ is the number of struts in the microtruss unit cell, $\sigma_{\text{strut}}$ is the axial stress in a given strut, $A_{\text{strut}}$ is the cross-sectional strut area and $A_{\text{truss}}$ is the area of the microtruss unit cell [7]. On the other hand,
as the forming displacement (and correspondingly the internal truss angle, \( \omega \)) increases, the microtruss struts become both longer and narrower. Accordingly, the dimensions of strut length, width and thickness change with forming displacement. The change in length can be described as:

\[
  l = l_o / \cos \omega = l_o \sec \omega
\]  

(4.4)

where \( l \) and \( l_o \) are the final and initial strut length respectively. Assuming a constant strut volume and a constant reduction in cross-section dimension, the width \( (w) \) will change according to:

\[
  w = w_o \sqrt{\cos \omega}
\]  

(4.5)

where \( w_o \) is the initial strut width, and likewise the strut thickness \( (t) \) can be calculated using:

\[
  t = t_o \sqrt{\cos \omega}
\]  

(4.6)

where \( t_o \) is the initial thickness. The slenderness ratio \( (L/r, \text{ where } L \text{ is the length of the column and } r \text{ is the radius of gyration}) \) can be expressed in terms of the strut length and thickness as:

\[
  \frac{L}{r} = \frac{l_o \sqrt{12}}{t}
\]  

(4.7)

Substituting Equations 4.4 and 4.6 into 4.7, the expression for the slenderness ratio becomes:

\[
  \frac{L}{r} = \frac{l_o \sqrt{12}}{t_o \left( \cos \omega \right)^{3/2}}
\]  

(4.8)

which simplifies to:
\[ \frac{L}{r} = \sqrt{12 \left( \frac{l_o}{t_o} \right) \sec^2 \omega} \]  

(4.9).

The relationship between the slenderness ratio and truss angle is illustrated in Figure 4.16 for the samples considered in the present study. The samples formed to a truss angle of 21° have a slenderness ratio of 50, while the samples formed to a truss angle of 47° have a slenderness ratio of 78.

![Figure 4.16](image_url)

Figure 4.16. Slenderness ratio, \( L/r \), as a function of truss angle, \( \omega \), for the starting sheet geometry used in the present study.

The stress at which the microtruss struts become structurally unstable (the critical buckling stress) can be expressed in terms of the slenderness ratio [8] as:

\[ \sigma_{cr} = \frac{k^2 \pi^2 E_t}{(L/r)^2} \]  

(4.10)

where \( k \) describes the rotational constraints of the strut ends (\( k = 1 \) for pin-jointed struts and \( k = 2 \) for rigid-jointed struts) and \( E_t \) is the tangent modulus (\( E_t = d\sigma/d\varepsilon \)). For columns with very high slenderness ratios \( E_t \) is equivalent to \( E \) resulting in elastic buckling. At
intermediate slenderness ratios, the Ramberg-Osgood model [9] can be used to describe the non-linear elastic to plastic transition in the stress-strain curve:

\[ \varepsilon = \frac{\sigma}{E} + \varepsilon_o \left( \frac{\sigma}{\sigma_{YS}} \right)^N \]  

(4.11)

where \( E \) is the modulus of elasticity of the material, \( \sigma_{YS} \) is the 0.2\% offset yield strength and \( N \) is a strain hardening exponent. Figure 4.17 shows a column curve (critical buckling stress as a function of slenderness ratio) for an annealed temper of C11000. Two lines are shown in Figure 4.17, the first represents the calculated elastic buckling curve, i.e. from Equation 4.10 using the Young’s modulus of Cu \( (E = 115 \text{ GPa}) \). The second line represents the predicted critical buckling strength based on the Ramberg-Osgood constitutive model. This second line deviates from the first line for slenderness ratio less than \( L/r \approx 350 \), i.e. as buckling becomes increasingly inelastic with increasing slenderness ratio. Finally, the critical buckling strength is limited by the annealed temper’s yield strength (horizontal line at \( \sigma_{CR} = \sigma_{YS} = 47 \text{ MPa} \)). Note that the samples of the present study have strut slenderness ratios falling within the range of 50 to 78.
Figure 4.17. Column curves giving the critical stress as a function of slenderness ratio, $L/r$, using the Ramberg-Osgood constitutive relationships of annealed C11000.

Figure 4.18 plots the experimental compressive peak stresses and analytical critical stress predictions (calculated using Equations 4.9-4.11) for the upper ($k=2$) and lower ($k=1$) limits with the exception of one data point, the experimental data fall within the upper and lower bounds. The relative insensitivity of the peak strength to the changing truss angle for the samples of the present study is therefore controlled by the trade-off between the counteracting effects of increased load efficiency and decreased strut stability, leading to a maximum in the peak strength at a truss angle of approximately 38°. Note that the incompletely recrystallized sample, which was formed to the lowest angle using the 2-step fabrication method, falls above the upper bound and is suggestive of the effect that work hardening can have in contributing to the peak compressive strength, as will be further discussed in the next section.
Figure 4.18. Experimental and predicted analytical compressive strength (using \( k = 1 \) and \( k = 2 \)) for 1-step and 2-step annealed C11000 microtrusses as a function of truss angle.

Finally, while Figure 4.18 shows that the peak strength begins to decrease for truss angles above \(~40^\circ\)\), from the perspective of specific strength it is beneficial to continue increasing the internal truss angle because of the continuously decreasing relative density (from \(7.5\%\) at \(21^\circ\) to \(3.1\%\) at \(47^\circ\)). The specific peak strength as a function of density is shown for 1-step and 2-step annealed samples in Figure 4.19. Note that this trend of increasing specific peak strength with decreasing density is the opposite of what one would typically expect for a cellular metal. In metallic foams, for example, the strength typically scales with density to the power of \(3/2\) [4], meaning that the strength decreases at a faster rate than the rate with decreasing density because of the bending-dominated nature of the architecture. Accordingly, microtruss architectures are much more attractive than conventional metallic foams.
Figure 4.19. Property map showing the specific peak compressive strength as a function of absolute density for 1-step and 2-step annealed cores.
4.2.2 Effect of Work Hardening

The relative significance of work hardening as a strengthening mechanism was studied by comparing annealed and as-fabricated microtrusses from both the 1-step and 2-step deformation forming processes. Figure 4.20 shows representative compressive stress-strain curves of C11000 pyramidal microtruss cores of as-fabricated and annealed microtruss cores fabricated to an internal truss angle of 38°. The as-fabricated microtruss core is almost three times stronger (2.06 ± 0.10MPa) than the annealed core (0.71 ± 0.10 MPa).

![Figure 4.20. Representative compressive stress-strain curves of 1-step as-fabricated and annealed microtruss cores, showing the relative significance of architecture (truss angle, \( \omega \), and relative density, \( \rho_r \)) and microstructure.](image)

Figure 4.21 presents the mechanical properties (\( E_{\text{Reload}} \), \( \sigma_p \), \( \sigma_y \), and \( J_D \)) as a function of truss angle for both the annealed and as-fabricated microtruss cores; experimental values are summarized in Table 4.4. While the annealed cores demonstrated...
a nearly linear increase in reload modulus with increasing truss angle, the reload modulus of the as-fabricated cores initially increased more quickly with truss angle but then plateaued at a value of ~80 MPa over the angular range of 33 to 44°, Figure 4.21a. The fact that work hardening has an effect on the reload modulus suggests that the strut deformation is not entirely axial, and that the finite node size of the microtruss may induce a small degree of bending before the initial peak stress is reached. For example, a 47° as-fabricated microtruss had a reload modulus of 108 ± 7 MPa, which is ~70% higher than a 46° annealed microtruss with a reload modulus of 64 ± 2 MPa. Figure 4.21b shows the peak strength as a function of truss angle. It can be seen that work hardening introduced an approximately 2x increase in compressive peak strength (from 0.57 ± .05 MPa to 1.2 ± 0.07 MPa) for the least-formed truss angles of (ω ~ 20°) and an approximately 3x increase in compressive peak strength (from 0.76 MPa to 2.1 MPa) for the largest 1-step truss angle (ω ~ 38°). The increasing peak strength with increasing truss angle is largely due to the accumulation of fabrication-induced work hardening with forming displacement and will be discussed in detail over the remainder of the chapter. In contrast to the peak strength, the valley strength of the as-fabricated cores remained in a relatively constant range, between 0.95-1.12 MPa, which is ~50% higher than the valley strength of the corresponding annealed cores, Figure 4.21c. The valley strength of the 2-step as-fabricated cores was lower than the valley strength of the 1-step as-fabricated cores, repeating the trend seen for the peak strength. Finally, the densification energy of the as-fabricated cores increased with increasing truss angle for both the 1-step and 2-step samples; each set was considerably higher than the densification energy of the
annealed cores which were relatively constant over the range of truss angles studied, Figure 4.21d.

Table 4.4. Summary of one-step as-fabricated (1-AF) and two step as-fabricated (2-AF) cores showing reload modulus \( (E_{\text{Reload}}) \), peak strength \( (\sigma_P) \), valley strength \( (\sigma_V) \), densification energy \( (J_D) \), truss angle \( (\omega) \), and relative density \( (\rho_R) \) with standard deviation for each. Sample size was 5 microtruss cores per set.

<table>
<thead>
<tr>
<th>Fabrication Pathway</th>
<th>Reload Modulus, ( E_{\text{Reload}} ) (MPa)</th>
<th>Peak Strength, ( \sigma_P ) (MPa)</th>
<th>Valley Strength, ( \sigma_V ) (MPa)</th>
<th>Densification Energy, ( J_D ) (MJ/m(^3))</th>
<th>Truss Angle, ( \omega ) (°)</th>
<th>Relative Density, ( \rho_R ) (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1-AF</td>
<td>30.9 ± 1.27</td>
<td>1.21 ± 0.07</td>
<td>0.95 ± 0.01</td>
<td>0.83 ± 0.07</td>
<td>22.5 ± 0.08</td>
<td>7.21 ± 0.08</td>
</tr>
<tr>
<td></td>
<td>47.1 ± 0.18</td>
<td>1.33 ± 0.09</td>
<td>0.92 ± 0.07</td>
<td>0.64 ± 0.08</td>
<td>24.6 ± 0.07</td>
<td>6.59 ± 0.19</td>
</tr>
<tr>
<td></td>
<td>61.7 ± 2.62</td>
<td>1.55 ± 0.10</td>
<td>1.07 ± 0.03</td>
<td>0.87 ± 0.03</td>
<td>26.9 ± 0.10</td>
<td>5.96 ± 0.02</td>
</tr>
<tr>
<td></td>
<td>69.3 ± 0.60</td>
<td>1.74 ± 0.01</td>
<td>1.04 ± 0.02</td>
<td>0.99 ± 0.01</td>
<td>28.7 ± 0.35</td>
<td>5.58 ± 0.23</td>
</tr>
<tr>
<td></td>
<td>76.5 ± 1.78</td>
<td>2.05 ± 0.08</td>
<td>1.14 ± 0.02</td>
<td>1.20 ± 0.01</td>
<td>32.2 ± 0.28</td>
<td>4.93 ± 0.05</td>
</tr>
<tr>
<td></td>
<td>76.6 ± 1.30</td>
<td>2.01 ± 0.10</td>
<td>1.15 ± 0.05</td>
<td>1.50 ± 0.09</td>
<td>34.8 ± 0.13</td>
<td>4.51 ± 0.02</td>
</tr>
<tr>
<td></td>
<td>85.0 ± 1.24</td>
<td>2.06 ± 0.05</td>
<td>1.12 ± 0.04</td>
<td>1.37 ± 0.06</td>
<td>38.1 ± 0.25</td>
<td>4.05 ± 0.03</td>
</tr>
<tr>
<td>2-AF</td>
<td>69.2 ± 0.69</td>
<td>1.36 ± 0.09</td>
<td>0.60 ± 0.06</td>
<td>0.83 ± 0.07</td>
<td>40.6 ± 0.51</td>
<td>3.78 ± 0.35</td>
</tr>
<tr>
<td></td>
<td>81.7 ± 1.75</td>
<td>1.58 ± 0.06</td>
<td>0.70 ± 0.01</td>
<td>1.04 ± 0.01</td>
<td>43.4 ± 1.51</td>
<td>3.61 ± 0.09</td>
</tr>
<tr>
<td></td>
<td>108 ± 7.05</td>
<td>1.73 ± 0.04</td>
<td>0.68 ± 0.01</td>
<td>1.12 ± 0.08</td>
<td>47.3 ± 0.43</td>
<td>2.98 ± 0.06</td>
</tr>
</tbody>
</table>

Figure 4.21 indicated that the fabrication-induced work hardening had a significant effect on each of the four mechanical properties studied. In order to understand these effects better, microhardness measurements were made at mid-strut for each sample. These values (and those of the annealed core samples as reference) are plotted in Figure 4.22. There is a surprisingly good correspondence between Figures 4.22 (hardness) and 4.21b (peak strength) despite the large range of architectures studied, and the fact that one property is measuring the resistance to penetration of a diamond pyramidal indenter into the middle of a given strut, while the other is measuring the maximum load that the full set of struts can support during inelastic buckling. While higher truss angles and therefore lower relative densities can be achieved using the two-step fabrication route, less work hardening could be introduced into the microtruss struts.
during the second forming step. For example, the mid-strut hardness of the highest truss angle sample produced in a single step ($\omega = 38 \pm 0.2^\circ$) was 94 ± 1 HV, while the mid-strut hardness of the highest angle two-step formed samples ($\omega = 47 \pm 0.4^\circ$) was ~7% lower at 87 ± 2 MPa.

![Figure 4.21.](image)

Figure 4.21. (a) Reload modulus, $E_{\text{Reload}}$, (b) peak strength, $\sigma_p$, (c) valley strength, $\sigma_v$; and (d) densification energy, $J_D$; shown as a function of truss angle for 1-step and 2-step annealed (ANN) and as-fabricated (AF) cores.
While stretch-bend fabrication imparts progressively more plastic strain to the microtruss struts as the forming displacement increases, this strain is not uniformly distributed along the struts. The hardness gradients along each strut can be subdivided into the three regions shown in the optical microscopy images of Figure 4.23: node (Region I), hinge (Region II), and strut (Region III). The non-uniform reduction in cross sectional thickness illustrates the localized nature of deformation during fabrication. Microhardness profiles along the top and bottom of the strut cross-section of the 38° ANN, 46° ANN, 38° AF and 47° AF samples are shown in Figure 4.24. The greatest amount of plastic deformation was induced at the hinge, with Region II of 1-step and 2-step annealed microtrusses having an average hardness of 66.2 ± 2.3 HV and 64.7 ± 0.9 HV respectively and 1-step and 2-step as-fabricated microtrusses having an average hardness of 94.4 ± 4.9 and 82.8 ± 3.7 HV respectively. Region I (top of the node), undergoes the least deformation when considering the entire microtruss core and has the
lowest hardness values. Despite the gradient in work hardening, the microhardness along the strut itself (Region III) was relatively constant at a value of 82 ± 3.6 HV. The post-fabrication annealed microhardness profile was lower than that of the as-fabricated sample by approximately 25 HV. (Note that the microhardness of the annealed sheet material prior to fabrication was 53 ± 1.9 HV).

Figure 4.23. Optical microscopy images showing microhardness inner and outer indentation profiles of a microtruss strut for (a) hinge region and (b) along strut length showing the subdivision into node (I), hinge (II) and strut (III) regions. There is non-uniform plastic deformation input during plastic deformation in terms of bending around the pin head and stretching along the strut length.
Figure 4.24. Microhardness profiles for 1-step and 2-step annealed and as-fabricated microtruss copper along the hinge and strut length. The mid-point of deformation occurred at a profile distance of ~8 mm. Regions I-III are described in Figure 4.23.

The increased buckling resistance that could be obtained from work hardening can be seen by plotting the column curves for different work hardened tempers of C11000. The Ramberg-Osgood constitutive relationship was fit to published ASM true stress-strain curves for different tempers of C11000: hard, half hard, quarter hard and annealed [10] (Appendix B) and used to generate the column curves shown in Figure 4.25. The benefit of increased work hardening on enhancing buckling-resistance is clearly evident. Since the material properties of the as-fabricated microtruss cores are continuously varying with increasing forming displacement, there is no direct way to connect the predicted C11000 column curves from Figure 4.25 with microstructure strength. Instead, an analytical model based on the assumption of uniform plastic strain imparted during stretch-bending and an evolving set of Holloman material parameters [11] was used to predict the peak strength of the microtruss struts, shown in Figure 4.26.
(details of the model are given in Appendix C). Note that the sharp drop in the predicted strength at 38° is from the intermediate annealing step. Overall, there is relatively good agreement between the experimental measurements of the present study and the predicted peak strength of the as-fabricated cores with the experimental values falling just below the upper ($k = 2$) predicted boundary.

Figure 4.25. Critical buckling stress as a function of slenderness ratio, $L/r$, using the Ramberg-Osgood constitutive relationships of C11000 for the following tempers: hard, half hard, quarter hard and annealed. The horizontal profiles indicate the temper’s yield strength which is taken as the upper limit to the critical buckling strength.
Figure 4.26. Experimental and predicted analytical compressive strength (using $k=1$ and $k=2$) for 1-step and 2-step annealed and as-fabricated C11000 microtrusses as a function of truss angle.

Finally, the specific peak strength as a function of density is shown in Figure 4.27 for all of the as-fabricated samples along with the annealed samples from Section 4.2.1 as a comparison. While the 38° 1-step as-fabricated samples had a higher peak strength than the 41°, 43°, and 47° 2-step as-fabricated samples, the 2-step samples had a lower density and were thus able to achieve higher specific strengths ($6.5 \pm 0.4$ MPa·m$^3$/kg at $\omega = 47 \pm 0.4^\circ$ compared to $5.7 \pm 0.5$ MPa·m$^3$/kg at $\omega = 38 \pm 0.2^\circ$). Even though the work hardened microtruss cores exhibited quite a favourable performance compared to the annealed core samples, it may be possible to further improve their performance by better controlling the distribution of plastic strain within the microtruss strut. This could be achieved by changing the perforation geometry [12] and/or by modifying the pin diameter and
curvature [13] in order to more effectively concentrate the plastic distribution along the strut length.

Figure 4.27. Specific peak strength is shown as a function of absolute density for annealed (1-step and 2-step in the green and blue bubbles, respectively) and as-fabricated (1-step and 2-step in the grey and red bubbles, respectively) C11000 microtruss cores.
4.3 NANOCRYSTALLINE NICKEL-IRON COATING

The next part of the study examined the mechanical properties of periodic cellular pyramidal copper microtrusses reinforced by an electrodeposited nanocrystalline nickel-iron sleeve. This part of the study combined two microstructural strengthening design strategies: work hardening and nanocrystalline sleeve electrodeposition. The pyramidal truss core microtrusses were fabricated following the method outlined in Chapter 3. The pre-form cores had a final height of 10.1 ± 0.21 mm and a truss angle of 45 ± 1.2°, corresponding to a relative density of 3.2 ± 0.2%.

Two sets of samples were produced. In the first case, the microtruss cores were annealed after stretch bend fabrication, but before electrodeposition. The second set of samples was electrodeposited in the as-fabricated condition. Mechanical properties from confined compression testing are summarized in Table 4.5. Figure 4.28 shows representative stress strain curves for the uncoated copper microtruss and hybrid nanocrystalline NiFe microtrusses of varying coating thicknesses ranging from 33 µm to 99 µm from the annealed set (Figure 4.28a) and as-fabricated set (Figure 4.28b). Both annealed and as-fabricated microtruss cores underwent the same overall inelastic buckling failure mechanism as the uncoated samples.
Figure 4.28. Representative stress-strain compression curves for uncoated and coated (a) annealed and (b) as-fabricated C11000 cores electrodeposited with a nanocrystalline NiFe coating.
In each case, there was a large effect of coating thickness on the mechanical properties ($E_{\text{Reload}}$, $\sigma_P$, $\sigma_V$, and $J_D$), shown as a function of coating thickness in Figure 4.29. In general, the reload modulus increases with increasing coating thickness for both annealed and as-fabricated cores with the exception of the as-fabricated cores with the thickest (~100 µm) coating. The decrease in reload modulus can be attributed to the decrease in coating integrity resulting in premature failure before peak strength. In the case of the heat treated cores, the peak strength (Figure 4.29b) increased by up to a factor of 8 from 0.55 ± 0.01 MPa to 4.4 ± 0.84 MPa. In the case of the as-formed samples, the peak strength increased by a factor of up to ~3.5 from 1.5 ± 0.23 MPa to 5.4 ± 1.7 MPa. It should be noted that although the as-fabricated cores were always stronger than the annealed core samples, the amount of peak strength increase with coating thickness was approximately the same, as shown in Figure 4.29b. Both annealed and as-fabricated cores demonstrated increasing valley strength with increasing coating thickness, Figure 4.29c, with the exception of the annealed core with the thickest coating (~100 µm) which showed a higher valley strength than its coated as-fabricated counterpart. Figure 4.29d shows that the densification energy for both annealed and as-fabricated cores increases with increasing coating thickness demonstrating the benefits of the energy absorption character of a structural coating.
Figure 4.29. Uniaxial compression testing properties as a function of coating thickness for annealed and as-fabricated NiFe/Cu microtruss cores showing (a) reload modulus, (b) compressive peak strength, (c) valley strength, and (d) densification energy.
Table 4.5. Summary of annealed (ANN) and as-fabricated (AF) C11000 alloy pyramidal truss cores coated with nanocrystalline NiFe coating (n-NiFe) showing reload modulus of elasticity ($E_{\text{Reload}}$), peak strength ($\sigma_P$), valley strength ($\sigma_V$), and densification energy ($J_D$), with standard deviation for each, of C11000 alloy pyramidal microtruss cores. Sample size was 5 microtruss cores per set.

<table>
<thead>
<tr>
<th>Fabrication Pathway</th>
<th>Coating Thickness, $t$ (mm)</th>
<th>Peak Strength, $\sigma_P$ (MPa)</th>
<th>Reload Modulus, $E_{\text{Reload}}$ (MPa)</th>
<th>Valley Strength, $\sigma_V$ (MPa)</th>
<th>Densification Energy, $J_D$ (MJ/m$^3$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>nNiFe/Cu – ANN core</td>
<td>-</td>
<td>0.55 ± 0.01</td>
<td>55.2 ± 1.7</td>
<td>0.35 ± 0.02</td>
<td>0.21 ± 0.04</td>
</tr>
<tr>
<td></td>
<td>0.030 ± 0.01</td>
<td>1.30 ± 0.25</td>
<td>64.4 ± 4.6</td>
<td>0.79 ± 0.33</td>
<td>0.58 ± 0.04</td>
</tr>
<tr>
<td></td>
<td>0.057 ± 0.01</td>
<td>2.59 ± 0.32</td>
<td>69.9 ± 2.4</td>
<td>1.44 ± 0.03</td>
<td>0.68 ± 0.03</td>
</tr>
<tr>
<td></td>
<td>0.077 ± 0.04</td>
<td>3.45 ± 0.55</td>
<td>75.9 ± 2.4</td>
<td>2.08 ± 0.08</td>
<td>1.72 ± 0.08</td>
</tr>
<tr>
<td></td>
<td>0.100 ± 0.02</td>
<td>4.39 ± 0.84</td>
<td>87.4 ± 1.4</td>
<td>2.59 ± 0.09</td>
<td>2.11 ± 0.19</td>
</tr>
<tr>
<td>nNiFe/Cu – AF core</td>
<td>-</td>
<td>1.47 ± 0.23</td>
<td>73.4 ± 1.1</td>
<td>0.80 ± 0.06</td>
<td>0.44 ± 0.03</td>
</tr>
<tr>
<td></td>
<td>0.025 ± 0.01</td>
<td>2.23 ± 0.22</td>
<td>87.8 ± 4.4</td>
<td>1.19 ± 0.07</td>
<td>0.89 ± 0.07</td>
</tr>
<tr>
<td></td>
<td>0.044 ± 0.03</td>
<td>2.71 ± 0.51</td>
<td>104 ± 4.1</td>
<td>1.41 ± 0.03</td>
<td>1.09 ± 0.05</td>
</tr>
<tr>
<td></td>
<td>0.074 ± 0.06</td>
<td>4.50 ± 0.27</td>
<td>132 ± 3.7</td>
<td>2.05 ± 0.06</td>
<td>1.85 ± 0.07</td>
</tr>
<tr>
<td></td>
<td>0.104 ± 0.01</td>
<td>5.43 ± 1.20</td>
<td>127 ± 7.8</td>
<td>2.36 ± 0.07</td>
<td>2.10 ± 0.05</td>
</tr>
</tbody>
</table>

The strength increase provided by the nanocrystalline electrodeposition can be approximated by the predicted behaviour of a hollow tube microtruss made of nanocrystalline NiFe [14]. This approach is based on the assumption that the coating contributes its inelastic buckling resistance to the composite microtruss, which is a reasonable assumption provided that the coated and uncoated microtrusses undergo the same form of inelastic buckling failure [14]. The moment of inertia of the nanocrystalline sleeve can be described as,

$$I = \frac{(w+2t)(b+2t)^3 - wb^3}{12} \quad (4.12)$$

where $w$ is the strut width, $b$ is the strut thickness and $t$ is the coating thickness. Area, $A$, is defined as the cross-sectional area of the nanocrystalline tube, described as,

$$A = (w+2t)(b+2t) - wb \quad (4.13)$$

Equations 4.9-4.11 were used to predict the buckling strength of a hollow nanocrystalline tube with varying sleeve thickness. The predicted buckling strength of the nanocrystalline
sleeve can be seen in Figure 4.30 which presents the measured increase in peak compressive strength as a function of nanocrystalline NiFe coating thickness for the experimentally measured annealed and as-fabricated samples. Also shown on this curve are the predicted boundary conditions of pin-jointed end constraints \((k=1)\) and rigid-jointed end constraints \((k=2)\). The experimentally measured values fall within the predicted boundaries for coating thickness of 70 µm and below, while the peak strength increase of the thickest samples was less than the analytical prediction.

![Figure 4.30](image-url) Figure 4.30. Measured increase in peak compressive strength as a function of coating thickness for nanocrystalline NiFe coated copper alloy microtruss pre-forms for the annealed (ANN) and as-fabricated (AF) samples, and predicted peak compressive strength as a function of n-NiFe thickness from the analytical model using the boundary conditions, \(k = 1\) (pin-joint) and \(k = 2\) (rigid joint).

In order to examine the reason for the lower than predicted peak strength increase in the thickest samples, the tangent modulus of the compressive stress-strain curves was investigated. Figure 4.31 shows representative tangent modulus-strain curves for the \(t_{NiFe} = 22\) µm and \(t_{NiFe} = 98\) µm as-fabricated core samples. The tangent modulus is indicative
of the coating integrity with sharp load drops representing potential coating/strut failure [13]. The thinnest ($t_{NiFe} = 22 \, \mu m$) and thickest coating, $t_{NiFe} = 98 \, \mu m$, showed numerous load drops of varying rates over the strain interval between the maximum initial slope and the peak stress, indicating that coating failure had been initiated before the peak load was reached. However, the load drops are larger and occur earlier for the thicker coating. Cracks were observed in all cases but were most pronounced for the thickest coatings.

Figure 4.31. Tangent modulus as a function of strain for uncoated and coated Cu microtrusses electroplated with nanocrystalline NiFe with an as-fabricated core.

SEM characterization, shown in Figures 4.32-4.34, was used to investigate the detailed failure mechanisms in the hybrid microtrusses after pre-loading to compressive strains of ~0.3. Delamination between the n-NiFe sleeve and the copper pre-form was observed near the hinge region of the composite struts. Figure 4.32a presents an SEM image of an as-fabricated nanocrystalline NiFe coated copper microtruss showing a zone of plastic wrinkling at the strut node, Figure 4.32a(i), and the onset of crack initiation at the hinge region just after peak stress, $\epsilon \sim 0.2$, Figure 4.32a(ii). Figure 4.32b is a detail
showing crack initiation at the node with evidence of delamination. A similar observation to the location of hinge fracture in n-Ni/Al microtruss cores was made by Bele et al. [15]. In that study, sleeve delamination and fracture was explained by the development of
Figure 4.32. Scanning electron microscopy images of a n-NiFe/Cu microtruss core pre-loaded to a compressive strain of \( \varepsilon \sim 0.13 \) (i.e. just after the peak stress) showing (a) mid-strut plastic bending with sleeve wrinkling shown in (i) and fracture in the hinge region in (ii). A detail of (ii) is shown in (b).
tensile stresses across the sleeve/core interface because of the local shell buckling. Delamination of the coating results in the development of local tensile stresses within the sleeve before the buckling stress is reached. In the case of n-Ni reinforcement of aluminum [15], the tendency towards local shell buckling and fracture increased with increasing coating thickness, which is consistent with the results of the present study. As deformation continued ($\varepsilon \sim 0.3$), wrinkling could be seen at mid-strut on the compression side of the buckled strut (Figure 4.33), while fracture can be seen on the tensile side (Figure 4.34). The higher magnification image in Figure 4.34b shows the crack extending fully across the strut and exposing the underlying copper pre-form core. Overall then, SEM characterization of partially failed samples indicated that those zones experiencing the highest local compressive stresses tend to wrinkle and those that experience the highest tensile stresses tended to fracture.

Figure 4.30 suggests that the achievable strength for a coating thickness of 100 $\mu$m is $\sim$6.6 MPa, while experimentally, the strength of an annealed microtruss core with 101 ± 2 $\mu$m coating was 3.84 ± 0.84 MPa and the strength of an as-fabricated microtruss core with 104 ± 14 $\mu$m coating was 3.96 ± 1.17 MPa. In order to achieve the predicted strength, there are two possible microtruss strut design strategies that could be taken to mitigate coating delamination and to delay the onset of sleeve fracture. The first strategy is to improve the adhesion between the coating and core. It may be possible to increase the coating adhesion by surface roughening, intermediate strike coatings and displacement films [16]. The second approach is to increase the ductility of the nanocrystalline coating by broadening the grain size distribution in order for the coating to have more complementary plastic deformation to match the core. It has been
Figure 4.33. Scanning electron microscopy image of as-fabricated n-NiFe coated C11000 copper microtruss showing wrinkling at mid-strut and crack initiation at the hinge region.
shown that increasing the grain size distribution into the range of ~20-200 nm can significantly increase the tensile elongation to fracture, e.g. up to ~15-20% with a yield strength penalty on the order of ~100-200 MPa [17,18]. Sacrificing some measure of yield strength in order to prolong the integrity of the nanocrystalline coating may offer enhanced energy absorption and perhaps even a peak strength increase.

Despite the fracture seen for the thickest reinforcing sleeves of nanocrystalline NiFe, electrodeposition was able to provide a significantly greater strength increase than work hardening. However, it was also accompanied by a considerable weight penalty. The relative merit of the two strengthening approaches can be considered by plotting the specific peak strength as a function of density for all sample types, Figure 4.35a. For all coating thicknesses tested, the weight penalty associated with n-NiFe was more than compensated for by the combination of ultrahigh strength and large second moment of area in the sleeve. It is worth restating that although the as-fabricated cores were always stronger than the annealed core samples, the amount of peak strength increase with nanocrystalline coating was approximately the same for each system (Figure 4.30). This is a key point because it indicates that over the range of architectures considered in the present study, the two strengthening mechanisms are effectively additive. This point will be considered in more detail in Chapter 6.

Figure 4.35b shows how the peak strength results from this study compare to previous copper cellular studies. This study achieved absolute densities ranging from 0.3-0.7 Mg/m³ while other studies produced copper foam with absolute densities ranging from 1.3-7.2 Mg/m³. The copper microtrusses also outperformed in terms of specific peak strength, having 33% greater peak strength than the strongest copper foam.
Figure 4.34. Scanning electron microscopy images of a n-NiFe/Cu microtruss core pre-loaded to a compressive strain of $\varepsilon \sim 0.3$ (i.e. near the valley stress), showing the overall strut failure (a) and crack formation at the tensile region of the middle of the strut (b).
Figure 4.35. Specific peak strength is shown as a function of absolute density for C11000 alloy pyramidal microtruss cores for (a) uncoated annealed (green bubble), uncoated as-fabricated (grey bubble), n-NiFe with annealed core (purple bubble) and coated n-NiFe as-fabricated core (blue bubble), and (b) the results from this study (as shown in (a)) compared to previous copper foam studies made by solidification and sintering [19-21].
4.4 REFERENCES


5 ALUMINUM MICROTRUSSSES

5.1 STRETCH-BEND FABRICATION

5.1.1 First Cycle

The effect of perforation geometry was investigated in the AA2024 system in order to better distribute the plastic strain imparted during stretch-bend fabrication in a comparatively low formability alloy. The designations “0.6\(w_o\)”, “0.7\(w_o\)” and “0.9\(w_o\)” refer to the reduction in the mid-strut starting width (\(w_o\)) that was used to generate a rounded square configuration (Figure 3.5). The forming limits for 0.6\(w_o\), 0.7\(w_o\) and 0.9\(w_o\) were investigated comparing a solutionizing treatment (30 min at 493°C [1]) versus an O-temper treatment (3 h at 395°C) (Figure 5.1). In the case of a 0.9\(w_o\) O-temper, the onset of plastic deformation occurs at \(F_{Bl}\) \(\approx\)0.45 kN. With increasing force, the sheet stretches and the struts elongate (increasing displacement), subsequently reaching the forming limit of the material. The maximum force and maximum displacement occur at the forming limit, in this case, \(~9.2\) kN and \(~4.8\) mm, respectively. Using a solutionizing heat treatment, a sample of the same perforation geometry had an increased stretch displacement of \(~5.9\) mm, which is 20% greater than when an O-temper is used, indicating that this heat treatment allows for increased formability. For the case of 0.6\(w_o\) both the bending force and the maximum force were lower, however, a greater maximum stretch displacement was attained at 6.4 mm. Overall, the solutionizing heat treatment resulted in increased formability of the alloy, while a decreasing strut width also resulted in increased formability. A summary of the forming parameters are given in Table 5.1 and plotted as a function of strut width in Figure 5.2. The stretch-bend forming parameters as a function of strut width are shown in Figure 5.2, for the annealed (O-
temper) and solution heat treated conditions. The maximum stretch displacement increased with decreasing strut width, Figure 5.2a, showing that the modified perforation geometry does allow larger forming displacements to be achieved. Accordingly, there is an increase in the maximum internal truss angle (Figure 5.2b) and a decrease in the minimum relative density (Figure 5.2c) with decreasing strut width. Likewise, it should be noted that a heat treated sample has increased formability over its annealed counterpart allowing for both increased truss angle and lower relative density.

Figure 5.1. Perforation stretch-forming curves for AA2024 aluminum alloy showing the effect of (a) temper treatment and (b) perforation geometry for 0.6\(w_o\), 0.7\(w_o\), and 0.9\(w_o\) where failure of the microtruss core occurs at the maximum stretch force.
Table 5.1. Data table showing bending force, $F_B$, maximum stretch force, $F_{\text{MAX}}$, maximum stretch displacement, $d_{\text{MAX}}$, truss angle, $\omega$, and relative density, $\rho_R$, with standard deviation for each, for the first cycle of forming of AA2024 aluminum alloy pyramidal microtruss cores with perforation geometries $0.6w_o$, $0.7w_o$ and $0.9w_o$ in the annealed (O-temper) and solution heat treated condition. Sample size was 5 per set.

<table>
<thead>
<tr>
<th>Strut width, $w$</th>
<th>Bending Force, $F_B$ (kN)</th>
<th>Maximum Stretch Force, $F_{\text{MAX}}$ (kN)</th>
<th>Maximum Stretch Displacement, $d_{\text{MAX}}$ (mm)</th>
<th>Truss Angle, $\omega$ (°)</th>
<th>Relative Density, $\rho_R$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.6</td>
<td>0.35 ± 0.01</td>
<td>7.26 ± 0.21</td>
<td>5.34 ± 0.10</td>
<td>35.6 ± 1.2</td>
<td>7.27 ± 0.9</td>
</tr>
<tr>
<td>0.9</td>
<td>0.46 ± 0.02</td>
<td>9.15 ± 0.33</td>
<td>4.85 ± 0.12</td>
<td>25.0 ± 0.9</td>
<td>10.8 ± 1.1</td>
</tr>
<tr>
<td>First Cycle: O-temper</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.6</td>
<td>0.40 ± 0.03</td>
<td>8.93 ± 0.33</td>
<td>6.37 ± 0.27</td>
<td>41.2 ± 1.1</td>
<td>6.09 ± 1.1</td>
</tr>
<tr>
<td>0.7</td>
<td>0.41 ± 0.01</td>
<td>9.44 ± 0.64</td>
<td>6.21 ± 0.16</td>
<td>37.2 ± 2.3</td>
<td>6.74 ± 1.3</td>
</tr>
<tr>
<td>0.9</td>
<td>0.57 ± 0.02</td>
<td>12.2 ± 0.38</td>
<td>5.86 ± 0.08</td>
<td>30.7 ± 0.8</td>
<td>8.96 ± 1.0</td>
</tr>
<tr>
<td>First Cycle: Solutionized</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Figure 5.2. Forming results for annealed and solutionized AA2024 pyramidal microtrusses as a function of strut width, $w_o$, showing (a) maximum stretch displacement, $d_{\text{MAX}}$, (b) maximum truss angle, $\omega_{\text{MAX}}$, and (c) relative density, $\rho_R$. 
Perforation geometry affects the amount and location of plastic strain distribution along the strut length, with rounded square geometries tending to spread the localization of strain away from the point of contact between pin and sheet in the node region [2]. In the current study, it was observed that the location of strut failure depended on the perforation geometry. Figure 5.3 shows SEM images of failed struts for the range of perforation geometries studied. For the thinnest width of $0.6w_o$, failure occurred in the reduced cross-section region of the strut (Figures 5.3a and b). With only a 10% increase in strut width, the failure point shifted to the point of curvature transition, much closer to the node (Figure 5.3c). Finally, with the widest struts studied, $0.9w_o$, failure occurred in the node itself, at the point of contact between the pin and the sheet (Figure 5.3d). The location of strut fracture is indicative of the region of highest plastic strain once a critical load is reached. Ideally failure will occur at mid-strut and all struts will fail simultaneously. For the $0.6w_o$ geometry, multiple struts had failed in each sample, while for the $0.7w_o$ and $0.9w_o$ geometries, the fabrication limit was determined by only single strut failure.
Figure 5.3. Scanning electron microscopy images of failed AA2024 pyramidal microtruss struts for (a) and (b) 0.6\textsubscript{wo}, (c) 0.7\textsubscript{wo} and (d) 0.9\textsubscript{wo} where the position of the strut failure is influenced by the perforation geometry (i.e. strut width).

### 5.1.2 Multi-Cycle

Due to the relatively low formability of AA2024, a single step of stretch-bending was insufficient to reach the target internal truss angle of 45°; starting from a solutionized state the internal truss angle was limited to 31° (0.9\textsubscript{wo}) and 41° (0.6\textsubscript{wo}). The next step in this study therefore involved investigating the potential of multi-cycle stretch-forming to achieve more efficient architectures. Figure 5.4 illustrates the multi-step forming sequence for the 0.9\textsubscript{wo} perforation geometry where samples were formed to 0.8 \textsubscript{dMAX} in each intermediate step. Note that the use of 80\%\textsubscript{dMAX} forming threshold (taken in order to
ensure that the as-fabricated struts are neck free) limited the amount of accessed/available architecture space. The forming parameters for the intermediate cycles are summarized in Table 5.2 for the $0.9w_o$ perforation geometry as well as for the $0.7w_o$ and $0.6w_o$ geometries.

It should be noted that with each passing forming cycle, the amount of forming force required increases while the amount of stretch displacement decreases. In the case of $0.9w_o$, the maximum stretch displacement decreased with increasing cycle number while the truss angle increased due to strut lengthening. The increased truss angle also corresponds to a decrease in relative density and by the fourth cycle, the microtruss cores had a relative density of 2.9% for $0.6w_o$, 3.4% for $0.7w_o$ and 4.7% for $0.9w_o$ at failure.

The location of strut failure was tracked through each forming cycle. By the fourth cycle, the failure locations had shifted away from the node for the $0.7w_o$ and $0.9w_o$ perforation geometries (the failure location for the $0.6w_o$ geometries remained in the reduced cross-section region for all forming cycles). Figure 5.5 shows the failure location after the fourth forming cycle for all three geometries, with the $0.7w_o$ location having shifted to the reduced cross-section region (similar to the $0.6w_o$ geometry after Cycle 1), and the $0.9w_o$ location having shifted to the point of curvature transition (similar to the $0.7w_o$ geometry after Cycle 1). This progression of strut fracture location away from the node and onto the strut is determined by the complex interaction of plastic strain accumulation and changing initial geometry for each forming cycle.
Figure 5.4. Representative multi-cycle forming curves for $0.9w_o$ AA2024 pyramidal microtrusses showing the stretch force as a function of stretch displacement after two steps (a), three steps (b), and four steps (c), with intermediate heat treatment between each forming step.
Table 5.2. Summary of bending force, $F_B$, maximum stretch force, $F_{MAX}$, maximum stretch displacement, $d_{MAX}$, truss angle, $\omega$, and relative density, $\rho_R$, with standard deviation for each, for the second, third and fourth forming cycles of AA2024 aluminum alloy pyramidal microtruss cores with perforation geometries $0.6w_o$, $0.7w_o$ and $0.9w_o$ heat treated to be in the solutionized state. Sample size was 5 per set.

<table>
<thead>
<tr>
<th>Strut width, $w_o$</th>
<th>Bending Force, $F_B$ (kN)</th>
<th>Maximum Stretch Force, $F_{MAX}$ (kN)</th>
<th>Maximum Stretch Displacement, $d_{MAX}$ (mm)</th>
<th>Truss Angle, $\omega$ (˚)</th>
<th>Relative Density, $\rho_R$</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Second Cycle</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.6</td>
<td>3.69 ± 0.05</td>
<td>16.1 ± 0.63</td>
<td>4.40 ± 0.87</td>
<td>53.9 ± 1.14</td>
<td>4.46 ± 0.88</td>
</tr>
<tr>
<td>0.7</td>
<td>4.56 ± 0.03</td>
<td>10.4 ± 0.16</td>
<td>3.57 ± 0.91</td>
<td>47.4 ± 1.07</td>
<td>5.41 ± 0.92</td>
</tr>
<tr>
<td>0.9</td>
<td>5.31 ± 0.03</td>
<td>9.51 ± 0.46</td>
<td>2.97 ± 0.92</td>
<td>38.4 ± 1.00</td>
<td>7.64 ± 0.93</td>
</tr>
<tr>
<td><strong>Third Cycle</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.6</td>
<td>4.36 ± 0.21</td>
<td>15.8 ± 0.66</td>
<td>3.02 ± 0.84</td>
<td>59.6 ± 1.71</td>
<td>3.58 ± 0.85</td>
</tr>
<tr>
<td>0.7</td>
<td>5.04 ± 0.12</td>
<td>10.8 ± 0.85</td>
<td>3.01 ± 0.54</td>
<td>54.6 ± 1.75</td>
<td>4.17 ± 0.55</td>
</tr>
<tr>
<td>0.9</td>
<td>7.09 ± 0.23</td>
<td>10.5 ± 0.19</td>
<td>2.95 ± 0.09</td>
<td>46.8 ± 1.01</td>
<td>5.70 ± 0.10</td>
</tr>
<tr>
<td><strong>Fourth Cycle</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.6</td>
<td>5.05 ± 0.14</td>
<td>15.8 ± 0.46</td>
<td>2.98 ± 0.52</td>
<td>64.3 ± 2.23</td>
<td>2.94 ± 0.53</td>
</tr>
<tr>
<td>0.7</td>
<td>6.51 ± 0.09</td>
<td>11.6 ± 0.58</td>
<td>2.80 ± 0.57</td>
<td>59.8 ± 2.32</td>
<td>3.42 ± 0.57</td>
</tr>
<tr>
<td>0.9</td>
<td>8.12 ± 0.09</td>
<td>9.04 ± 0.12</td>
<td>2.52 ± 0.33</td>
<td>52.1 ± 0.42</td>
<td>4.71 ± 0.34</td>
</tr>
</tbody>
</table>

The incremental increase per forming cycle for each perforation geometry is shown in Figure 5.6. In Figure 5.6a, the new forming displacement gained in each cycle is plotted. For the $0.9w_o$ geometry the displacement decreases from approximately 5.86 mm, to 2.97 mm, to 2.95 mm, to 2.52 mm (Cycles 1 to 4 respectively), while for the $0.6w_o$ geometry, the displacement decreased from approximately 6.37 mm, to 4.40 mm, to 3.01 mm to 2.98 (again, Cycles 1 to 4, respectively). There is therefore progressively less and less benefit to forming with each subsequent deformation cycle, with the targeted 45° internal truss angle (at less than $0.8d_{MAX}$) being reached within 3 cycles for $0.6w_o$, 4 cycles for $0.7w_o$ and 4 cycles for $0.9w_o$.

Overall then, it was determined that with multi-cycle forming all three perforation geometries could be fabricated to beyond the target truss angle of 45°. This is an important contribution because the first study of stretch-bend fabricated microtrusses had concluded that only the highest formability alloys could be produced by this method [3].
In contrast, the present study has demonstrated that even relatively low formability alloys such as AA2024 can in fact be used for stretch bend fabrication given the correct sheet geometry and intermediate annealing steps.

Figure 5.5. Scanning electron microscopy images of failed AA2024 pyramidal microtruss struts at the fourth stage of multi-cycle stretch-bend forming for (a) $0.6w_o$, (b) $0.7w_o$ and (c) $0.9w_o$ where the position of the strut failure is influenced by the perforation geometry (i.e. strut width).
Figure 5.6. Incremental increase per forming cycle for solutionized AA2024 pyramidal microtruss cores shown as a function of cycle number for the (a) maximum stretch displacement, $d_{MAX}$, (b) difference in maximum truss angle, $\omega_{MAX}$, and (c) difference in minimum relative density, $\rho_{R}$. 
5.2 **UNIAXIAL COMPRESSION TESTING**

Two precipitation hardening treatments were tested to find the most advantageous route for strengthening the AA2024 microtrusses: age hardening (T6) and thermomechanical treatment (T8) [1]. The T6 treatment consisted of solution heat treating the as-fabricated microtruss cores at 493°C for 30 min followed by artificial ageing at 191°C for 3-6 h [1]. The T8 thermomechanical treatment would take advantage of the plastic strain imparted to the microtruss during fabrication in order to enhance the precipitation kinetics during artificial ageing (also at 191°C) [1]. In other words, the processing difference between the T8 and the T6 microtruss cores was the absence of the solution heat treatment (30 min at 493°C) in the case of T8. While the T8 condition may allow higher strengths to be achieved, the kinetics and particle size distribution of the precipitates during the final ageing step depend on the prior plastic deformation. The uncertainty with respect to applying this precipitation strategy to AA2024 microtruss cores centers on two factors: the non-uniformity of plastic strain imparted to the microtruss struts during deformation, and the limited overall amount of plastic strain imparted during the final stretch-bending cycle (i.e. when the internal truss angle increased from only 47° to 52°).

The mechanical properties of cores having the T6 and T8 treatments were compared after heat treatment times of 3 h, 6 h, and 12 h (representative stress-strain curves shown in Figure 5.7). Following the initial period of elastic loading, the microtruss struts failed by inelastic buckling. In all cases, T8 outperformed T6 by having a greater initial peak strength: 27% at 3 hrs, 19% at 6 hrs, and 10% at 12 hrs. Despite the non-uniform and limited plastic deformation during the final fabrication step, sufficient
plastic strain was generated within the microtruss struts in order to increase the precipitation kinetics and achieve a higher peak strength over a more conventional T6 processing route. The T8 processing route is also advantageous in that it completely eliminates one high temperature heat treatment.

![Graphs showing compressive stress-strain curves for T6 and T8 AA2024 alloy pyramidal microtruss cores for artificial ageing times.](image)

Figure 5.7. Representative compressive stress-strain curves for 0.7% solutionized T6 and T8 AA2024 alloy pyramidal microtruss cores for artificial ageing times of (a) 3 h, (b) 6 h and (c) 12 h.

For the rest of this chapter, the high strength version of the AA2024 microtruss cores will be taken as the T8 temper, while the low strength version will be the O-temper (obtained by heat treating the as-fabricated samples at 385°C for 3 h). Using these two
metallurgical states, the effect of perforation geometry and hard anodizing was considered. The mechanical properties ($E_{\text{Reload}}, \sigma_p, \sigma_v$, and $J_D$) for $0.6w_o$ and $0.9w_o$ perforation geometries are summarized in Table 5.3 and illustrated in Figure 5.8.

Table 5.3. Mechanical properties from uniaxial compression testing of AA2024 pyramidal microtruss cores with perforation geometry $0.6w_o$ and $0.9w_o$ for both the O-temper and T8 heat treatments summarizing the reload modulus ($E_{\text{Reload}}$), peak strength ($\sigma_p$), valley strength ($\sigma_v$), and densification energy ($J_D$) with standard deviation for each. Sample size was 5 per set.

<table>
<thead>
<tr>
<th>Heat Treatment</th>
<th>Reload Modulus, $E_{\text{Reload}}$ (MPa)</th>
<th>Peak Strength, $\sigma_p$ (MPa)</th>
<th>Valley Strength, $\sigma_v$ (MPa)</th>
<th>Densification Energy, $J_D$ (Mg/m$^3$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.6w_o</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>O-temper</td>
<td>30.45 ± 3.97</td>
<td>0.76 ± 0.02</td>
<td>0.43 ± 0.09</td>
<td>0.61 ± 0.17</td>
</tr>
<tr>
<td>T8</td>
<td>32.28 ± 4.30</td>
<td>1.36 ± 0.05</td>
<td>0.71 ± 0.13</td>
<td>0.77 ± 0.18</td>
</tr>
<tr>
<td>0.9w_o</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>O-temper</td>
<td>42.10 ± 2.11</td>
<td>1.33 ± 0.07</td>
<td>0.74 ± 0.06</td>
<td>0.59 ± 0.12</td>
</tr>
<tr>
<td>T8</td>
<td>77.37 ± 2.43</td>
<td>2.64 ± 0.18</td>
<td>0.94 ± 0.06</td>
<td>1.32 ± 0.06</td>
</tr>
</tbody>
</table>

Figure 5.8 shows typical stress-strain curves for the T8 and O tempers having the $0.6w_o$ and $0.9w_o$ perforation geometries. The $0.9w_o$ curves shown in Figure 5.8a are comparable to what had been seen previously for the C11000 microtrusses, with an initial period of elastic loading followed by a peak stress. In contrast, the $0.6w_o$ samples (Figure 5.8b) exhibited a more complex transition from elastic loading to buckled state. In fact, the $0.6w_o$ T8 sample exhibited a minor initial peak at ~1.1 MPa after which the stress dropped to ~0.8 MPa before increasing again to the major peak strength of 1.4 MPa. In order to investigate this different behaviour, $0.6w_o$ and $0.9w_o$ samples were pre-loaded to just after the main peak stress, corresponding to a strain of $\varepsilon \approx 0.15$, and compared to the original undeformed starting sheet (Figure 5.9). Like the C11000 microtrusses, the struts of the $0.9w_o$ sample have buckled in only one direction (Figure 5.9a), meaning that a projection of the partially collapsed microtruss along the height dimension essentially
corresponds to the geometry of the undeformed starting sheet. In contrast, the buckling mode is considerably more complex for the $0.6w_o$ samples with the struts failing in both the thickness and width directions (Figure 5.9b). An explanation for the off-axis buckling observed in $0.6w_o$ can be found in the difference between the second moment of area in the width and thickness orientations for the two perforation geometries. While the second moment of area is lower in the thickness direction than in the width direction, the ratio between the two orientations in the as-fabricated buckling condition drops from ~13 at $0.9w_o$ to ~6 at $0.6w_o$, Figure 5.10. Given its smaller second moment of area, strut failure would have been first initiated through the thickness orientation in the $0.6w_o$ samples. But as the first (weakest) strut(s) failed, it would have resulted in a load imbalance at the nodes which could have triggered failure in the width direction as well. Overall then, the $0.6w_o$ samples still fail by inelastic buckling, but the particular mode of buckling is more complex given the similarity between width and thickness direction second moment of areas.

![Figure 5.8](image)

Figure 5.8. Representative compressive stress-strain curves comparing AA2024 O-temper and T8 cores for (a) $0.9w_o$ and (b) $0.6w_o$ perforation geometries.
Figure 5.9. AA2024 microtruss core with (a) $0.6w_o$ and (b) $0.9w_o$ perforation geometries each with an undeformed reference sheet underneath the failed cores and a superimposed line indicating the orientation of buckling.
The effect of perforation geometry (and metallurgical state) on the mechanical properties of the AA2024 microtruss cores are summarized in Table 5.3 and illustrated in Figure 5.11, with the density normalized properties given in Figure 5.12. Overall, the T8 temper results in significantly enhanced peak strength, valley strength and densification energy. The peak compressive strength of the T8 core was almost double the strength of the O-temper core for both perforation geometries. In terms of the post-buckling collapse mechanism, the valley strength of the cores was likewise increased with the T8 temper and even larger performance increases were seen for the densification energy. Normalizing the mechanical property data in Figure 5.12 brought performance of the $0.6w_o$ samples closer to that of the $0.9w_o$ samples, but the $0.9w_o$ samples were still generally superior.
Figure 5.11. Summary of mechanical properties from uniaxial compression testing of AA2024 alloy pyramidal microtruss cores with 0.6\(w_o\) and 0.9\(w_o\) perforation geometry and O-temper or T8 cores showing (a) reload modulus, (b) peak strength, (c) valley strength, and (d) densification energy.
Figure 5.12. Summary of normalized mechanical properties from uniaxial compression testing of AA2024 alloy pyramidal microtruss cores with 0.6\(w_o\) and 0.9\(w_o\) perforation geometry and O-temper or T8 cores showing (a) reload modulus, (b) peak strength, (c) valley strength, and (d) densification energy.

The critical buckling stress was determined in terms of the slenderness ratio using Equation 4.10 to take into consideration strut architecture and microstructure in terms of its geometry, strut end constraints and the Ramberg-Osgood model, Equation 4.11, to describe the transition from elastic to plastic deformation in the stress-strain curve. ASM true stress-strain curves were used to approximate the Ramberg-Osgood parameters for the O-temper and T8 tempers of AA2024 [4] (see Appendix B for the fitting of AA2024 true stress-strain curves to obtain the Ramberg-Osgood parameters). Figure 5.13 shows the critical buckling stress as a function of slenderness ratio for Ramberg-Osgood
constitutive relationship of the O and T8 tempers, where their yield strength becomes the upper boundary of the critical buckling strength. At higher slenderness ratios, i.e. $L/r > 125$, the O and T8 tempers fail at the same critical buckling stress. At intermediate slenderness ratios, i.e. $30 \leq L/r \leq 125$, the critical buckling stress of the two tempers diverge as both tempers gradually transition from elastic to inelastic buckling. The correlation between the experimental compressive peak stresses and analytical critical stress predictions (determined using Equations 4.7-4.9) for the lower ($k=1$) and upper limit ($k=2$), for slenderness ratios of 47 and 64 for $0.6w_o$ and $0.9w_o$ respectively, are shown in Table 5.4. In the case of $0.9w_o$, for both tempers, the experimental strength falls within the predicted values. For $0.6w_o$, the experimental strength fell below the predicted lower limit by $\sim 40\%$, possibly because the analytical model does not account for the more complex form of off-axis buckling.

Figure 5.13. Critical stress as a function of slenderness ratio for the Ramberg-Osgood constitutive relationships of AA2024 for the O-tempor ($\sigma_{YS} = 180$ MPa) and T8 temper ($\sigma_{YS} = 441$ MPa).
Table 5.4. Experimental ($\sigma_{P,\text{Expt}}$) and analytical ($\sigma_{P,\text{Predicted}}$) compressive peak strength for 0.6$w_o$ and 0.9$w_o$ AA204 microtrusses with O-temper and T8.

<table>
<thead>
<tr>
<th>Heat Treatment</th>
<th>$\sigma_{P,\text{Expt}}$ (MPa)</th>
<th>$\sigma_{P,\text{Predicted}}$ (MPa)</th>
<th>$\sigma_{P,\text{Expt}}$ (MPa)</th>
<th>$\sigma_{P,\text{Predicted}}$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>$k = 1$</td>
<td>$k = 2$</td>
<td>$k = 1$</td>
</tr>
<tr>
<td>O-temper</td>
<td>1.33</td>
<td>1.14</td>
<td>2.28</td>
<td>0.76</td>
</tr>
<tr>
<td>T8</td>
<td>2.64</td>
<td>1.94</td>
<td>3.69</td>
<td>1.36</td>
</tr>
</tbody>
</table>

Overall, both perforation geometry and thermomechanical core treatment influence the compressive peak strength. Figure 5.14 plots the peak strength against the density for the AA2024 microtruss cores used in this study. Also shown in Figure 5.14 are the peak strengths of AA3003 microtruss cores [2] having a comparable architecture to the microtrusses of the present study. The T8 0.9$w_o$ AA2024 exhibited the highest peak strength out of comparable aluminum microtruss core studies completed to date [2]. Likewise, the T8 0.6$w_o$ AA2024 cores had a higher peak compressive strength than AA3003 1.0$w_o$ cores having a higher density. The small amount of plastic strain imparted in the final forming cycle is sufficient to give an aluminum alloy microtruss with enhanced mechanical properties compared to the more favourable AA3003 microtrusses considered to date.
Figure 5.14. Property map showing peak compressive strength as a function of density for AA2024 and AA3003 alloy pyramidal microtrusses with strut width $0.5w_o$-$1.0w_o$. 
### 5.3 ALUMINUM OXIDE COATING

The next part of the study investigated the effect of structural coatings on AA2024 pyramidal microtrusses. The pyramidal pre-form microtruss cores were fabricated following the method described in Chapter 3. The pre-form cores had a final height of 10.1 ± 0.1 mm and a truss angle of 45 ± 1°, corresponding to a relative density of 3.2 ± 0.1%. Four sets of coated samples were fabricated: perforation geometries of 0.6\(w_o\) and 0.9\(w_o\) with two core tempers each (O-temper and T8 temper). Mechanical properties (\(E_{\text{reload}}, \sigma_p, \sigma_v\), and \(J_D\)) from uniaxial compression testing are summarized in Table 5.5 and plotted in Figure 5.15. While there was a small increase to the reload modulus (an increase of ~6%), it should be noted that testing irregularities prevented the full number of repeats to be collected for this parameter and consequently no standard deviation has been given. On the other hand, the peak strength of both tempers is substantially increased (Figure 5.15a). The oxide coating has comparatively less effect on the valley strength (Figure 5.15b). For the 0.6\(w_o\) samples in particular there was no measurable effect of the oxide. This may be due to the combination of off-axis buckling that was seen for this perforation geometry (Figure 5.9) and the lack of oxide reinforcement on the thickness face (sides) of the struts (Figure 3.8). Likewise, the oxide had little effect on the densification energy of the 0.6\(w_o\) samples (Figure 5.15c). When strut buckling was restricted to just one axis (0.9\(w_o\) perforation geometry), the oxide had a considerably larger effect on the densification energy.
Table 5.5. Mechanical properties from uniaxial compression testing of Al₂O₃/AA2024 microtruss cores with perforation geometry 0.6wₒ and 0.9wₒ and with the O-temper or T8 heat treatment summarizing the reload modulus (E_{Reload}), peak strength (σ_p), valley strength (σ_v), and densification energy (J_D) with standard deviation for each (with the exception of reload modulus). Sample size was 5 per set.

<table>
<thead>
<tr>
<th>Heat Treatment</th>
<th>Reload Modulus, E_{Reload} (MPa)</th>
<th>Peak Strength, σ_p (MPa)</th>
<th>Valley Strength, σ_v (MPa)</th>
<th>Densification Energy, J_D (MJ/m³)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.6wₒ</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>O-temper</td>
<td>3.55</td>
<td>0.99 ± 0.05</td>
<td>0.49 ± 0.05</td>
<td>0.57 ± 0.05</td>
</tr>
<tr>
<td>T8</td>
<td>27.8</td>
<td>1.53 ± 0.05</td>
<td>0.64 ± 0.21</td>
<td>0.63 ± 0.10</td>
</tr>
<tr>
<td>0.9wₒ</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>O-temper</td>
<td>66.1</td>
<td>1.82 ± 0.10</td>
<td>0.92 ± 0.06</td>
<td>0.91 ± 0.03</td>
</tr>
<tr>
<td>T8</td>
<td>72.8</td>
<td>2.94 ± 0.24</td>
<td>1.31 ± 0.23</td>
<td>1.60 ± 0.08</td>
</tr>
</tbody>
</table>

Figure 5.15. Summary of mechanical properties from uniaxial compression testing of coated Al₂O₃/AA2024 alloy pyramidal microtruss cores with 0.6wₒ and 0.9wₒ perforation geometry and O-temper or T8 cores showing (a) peak strength, (b) valley strength and (c) re-load modulus.
Figure 5.16 illustrates the increase in peak strength with coating reinforcement for each of the configurations. The strength increase for the $0.6w_o$ architectures, at $\sim 0.2$ MPa, was half the value of the strength increase seen for the $0.9w_o$ architectures at $\sim 0.4$ MPa. While the O-temper microtrusses may have been strengthened slightly more than the T8 microtrusses, the error bars overlapped in each case. To a first approximation then, strengthening mechanism additivity for Al$_2$O$_3$/AA2024 is dependent on architecture and nearly independent of temper.

![Graph](image)

Figure 5.16. The increase in peak strength from the addition of an Al$_2$O$_3$ coating on a AA2024 microtruss.

In order to gain further insight into the composite strut failure, the tangent modulus of the compressive stress-strain curves was investigated. Figure 5.17 shows representative compressive stress-strain curves with corresponding tangent modulus-strain curves for $0.9w_o$. The addition of an aluminum oxide coating on an O-temper core increases the peak strength by over 35% (Figure 5.17a), however when the tangent modulus is plotted as a function of strain, there is a large initial load drop for the coated
core (Figure 5.17b) compared to the uncoated case. A similar effect was observed with the thicker coatings of nanocrystalline NiFe on copper, Figure 4.31.

Figure 5.17. Representative compressive stress-strain curves of AA2024 alloy pyramidal microtruss cores comparing (a) coated and uncoated 0.9\(w_o\) perforation geometry with O-temper core and corresponding rate of change shown in (b).

It was comparatively more difficult to extract meaningful information from the tangent modulus of the 0.6\(w_o\) curves because the more complex buckling failure modes lead to load irregularities before the peak strength even for the uncoated samples, Figure 5.18. However, it was still possible to see a significant difference between the behaviour of the coated and uncoated samples in the post buckling regime, where large fluctuations in the tangent modulus are evident.
Figure 5.18. Representative compressive stress-strain curves of AA2024 alloy pyramidal microtruss cores comparing (a) coated and uncoated $0.6w_o$ perforation geometry with O-temper core and corresponding rate of change shown in (b).

Some insight into the progression of strut failure is shown in Figure 5.19 with optical microscopy images of $0.6w_o$ struts compressed to $\varepsilon = 0.7$, which is equivalent to the strain at twice the peak strength, $2\sigma_P$. See Appendix D, section D1 for a series of optical microscopy images showing the progression of strut failure for strains from $\varepsilon = 0.15$ to 0.7. The uncoated O-temper core, Figures 5.19a, shows extensive plastic deformation and some off-axis buckling. Surprisingly, the oxide coating is nearly intact along the length of the failed O-temper composite strut (Figure 5.19b). It should be noted that the $0.6w_o$ struts exhibited various buckling wavelengths, which is quite different from the case of the C11000 and n-NiFe/C11000 struts which almost exclusively failed at the strut mid-point and can be explained by the decreased strut width allowing more complex forms of off-axis buckling to occur. An uncoated T8 core with $0.6w_o$ perforation geometry is shown in Figure 5.19c, which shows two points of highly localized curvature in addition to the initiation of a fracture on the strut side that
experiences tension. Similar behaviour was seen in the presence of the coating (Figure 5.19d).

Figure 5.19. Optical microscopy images of $0.6w_o$ cores compressed to $\varepsilon = 0.7$ showing (a) uncoated annealed, (b) coated $\text{Al}_2\text{O}_3$ annealed, (c) uncoated T8, and (d) coated $\text{Al}_2\text{O}_3$ T8.

Optical microscopy images were also obtained for the $0.9w_o$ microtrusses compressed to $\varepsilon = 0.7$. Similar to the $0.6w_o$ cores, ductile strut buckling failure is observed with the uncoated O-temper core, Figure 5.20a. The presence of an aluminum oxide coating contributes to more localized curvature with some fracture (Figure 5.20b). An uncoated T8 core is shown in Figure 5.20c with a coated counterpart in Figure 5.20d; the addition of the ceramic sleeve on a T8 core again results in more significant strut fracture for this temper. It is worth noting that the overall failure progression of the
Al₂O₃/AA2024 struts is different than the failure mechanism seen previously for Al₂O₃/AA3003 composites [5]. In that case, coating thicknesses above ~5 µm led to local shell buckling and progressive hinge fracture that moved down the strut length (with a complete absence of global buckling seen for the thickest oxide coating of ~40 µm [5]). This difference in behaviour is likely due to the greater slenderness ratio of the struts in the present study (64 and 47 for the 0.6w₁₀ and 0.9w₁₀ Al₂O₃/AA2024 struts compared to 28 for the Al₂O₃/AA3003 struts), where the increased slenderness ratio in the present study would have favored global buckling over hinge failure.

Figure 5.20. Optical microscopy images of 0.9w₁₀ cores compressed to ε = 0.7 showing (a) uncoated annealed, (b) coated Al₂O₃ annealed, (c) uncoated T8, and (d) coated Al₂O₃ T8.
To further inspect the coating failure of the coated T8 struts, SEM micrographs of failed strut profiles were collected (see Appendix D, section D2 for SEM micrographs showing failed strut profiles for $0.6w_o$ and $0.9w_o$ at strain, $\varepsilon = 0.7$). Figure 5.21 shows SEM images of coated microtrusses with $0.6w_o$ and $0.9w_o$ perforation geometry and T8 core.
Figure 5.21. SEM micrographs of failed strut profiles at strain, $\varepsilon \approx 0.7$ for Al$_2$O$_3$/AA2024 with T8 core for (a) 0.6$w_o$, and (b) 0.9$w_o$. 
Failure in the form of both Al$_2$O$_3$ coating fracture and AA2024 core fracture is observed for both perforation geometries. With 0.6$w_o$, fracture occurs on the side of the bending strut that experiences tension (Figure 5.21a), while at the same point of compressive strain $\varepsilon = 0.7$, the fracture appears to have nearly extended fully through the entire strut of the 0.9$w_o$ sample (Figure 5.21b).

Further insight into the effect of core temper was obtained by using SEM to examine partially failed strut cross-sections for 0.6$w_o$ and 0.9$w_o$ Al$_2$O$_3$/AA2024 in the T8 and O tempers at a strain of $\varepsilon \approx 0.7$ (see Appendix D, section D3 for SEM micrographs of partially failed strut cross-sections for 0.6$w_o$ and 0.9$w_o$ Al$_2$O$_3$/AA2024 at strain, $\varepsilon = 0.7$). Figure 5.22 presents SEM micrographs of partially failed strut cross-sections for 0.6$w_o$ Al$_2$O$_3$/AA2024 composites in the T8 and O-temper. In the case of an O-temper core, Figure 5.22a, the coating is shown to be of uniform thickness encapsulating the strut and remains continuous around the strut perimeter and buckling occurs with gradually tapering curvature. For a T8 core, Figure 5.22b, despite the uniform coating along stretches of no failure, there is a discontinuity in the regions experiencing tensile stresses with large cracks progressing from the coating deep into the AA2024 core. These fractures increase the sharpness of the local strut curvature and are quite different from the behaviour of the uncoated T8 struts. This crack propagation from coating to core can be seen more clearly in Figure 5.23. At a strain of $\varepsilon \approx 0.7$, the T8 strut has nearly completely fractured.
Figure 5.22. SEM microscopy of partially failed strut cross-sections of 0.6\textsubscript{w/o} Al\textsubscript{2}O\textsubscript{3}/AA2024 at strain, $\varepsilon = 0.7$ with an (a) O-temper and, (b) T8 core.
A higher magnification SEM micrograph of a coated T8 core with 0.6$w_o$ perforation geometry compressed to $\varepsilon = 0.7$ is shown in Figure 5.23. The detail shows crack initiation of a deep crevice on the strut side that experienced tension, and fracture plus spalling on the side of the strut that experienced compression. The tensile fracture extends into the AA2024 core, as seen in the cross-section image of Figure 5.22b. A detail of the fracture profile, Figure 5.24, emphasizes the loss of coating integrity at the location of failure on each side of the strut that experienced tensile and compressive forces.

Figure 5.23. SEM micrograph of a 0.6$w_o$ Al$_2$O$_3$ coated T8 core compressed to $\varepsilon = 0.7$. 
Overall, the addition of an Al$_2$O$_3$ coating was able to provide strength increases on each perforation geometry with somewhat greater strength increases seen for the O-temper core than the T8 temper core. The benefits of the two strengthening methods can be considered by plotting peak strength against density for all sample types, Figure 5.25. Due to the low density of the ceramic coating and the loss of mass during the pre-anodizing cleaning steps, there is virtually no weight penalty associated with its use (as compared to the weight penalty from the n-NiFe/C11000 coatings, Figure 4.39). Chiefly, Al$_2$O$_3$/AA2024 microtrusses had lower absolute densities compared to a previous study of Al$_2$O$_3$/AA3003 [5] and showed the additive nature of the strengthening mechanisms. Further, perforation shape, temper and a coating treatment all contribute to the mechanical performance of a microtruss core and that both microstructure and perforation
geometry influence material properties. The additive nature of strengthening mechanisms and the effect of geometry and how each contributes to overall microtruss strength will be further discussed in Chapter 6.

![Property map showing specific peak strength as a function of density for coated Al₂O₃/AA2024 alloy pyramidal microtrusses with 0.6w and 0.9w perforation geometry and O-temper or T8 cores and coated Al₂O₃/AA3003 alloy pyramidal microtrusses.]

Figure 5.25. Property map showing specific peak strength as a function of density for coated Al₂O₃/AA2024 alloy pyramidal microtrusses with 0.6w and 0.9w perforation geometry and O-temper or T8 cores and coated Al₂O₃/AA3003 alloy pyramidal microtrusses.

### 5.4 REFERENCES


6 MODELING MICROTRUSS SYSTEMS

6.1 PARENT MATERIAL STRENGTH INCREASES

In order to address the question of strengthening mechanism additivity in microtrusses, the case of an idealized material system in the presence of work hardening was considered first. The C11000 copper system was examined using strut geometries similar to the microtruss systems examined in Chapter 4, with the assumption of elastic-perfectly plastic behaviour. Idealized stress-strain curves for C11000 were constructed based on a Young’s Modulus of 115 GPa and yield strengths corresponding to the annealed ($\sigma_{YS} = 47$ MPa), quarter hard ($\sigma_{YS} = 208$ MPa), half hard ($\sigma_{YS} = 271$ MPa), and hard ($\sigma_{YS} = 324$ MPa) tempers, Figure 6.1. The difference in yield strength, $\Delta\sigma_{YS}$, was defined as the difference in strength between a work hardened temper and the annealed baseline.

![Figure 6.1. Elastic-perfectly plastic stress-strain curves for C11000 based on the following tempers: annealed ($\sigma_{YS} = 47$ MPa), quarter hard ($\sigma_{YS} = 208$ MPa), half hard ($\sigma_{YS} = 271$ MPa), and hard ($\sigma_{YS} = 324$ MPa).]
Figure 6.2 presents the column curves for the five C11000 tempers giving the critical stress as a function of slenderness ratio. Three distinct zones can be seen, in Zone 1 ($L/r < 59$) the critical stress of elastic buckling exceeds the yield strength for each of the four tempers. In Zone 3 ($L/r > 155$) all four of the tempers fail by elastic buckling. Finally, Zone 2 ($59 \leq L/r \leq 155$) represents the intermediate case where the critical buckling stress progressively exceeds the yield strength with increasing amount of work hardening as the slenderness ratio decreases. Similar to the case for yield strength, the difference in critical buckling strength, $\Delta \sigma_{CR}$, was defined as the difference in strength between the work hardened temper in question and the annealed baseline.

![Figure 6.2. Column curves giving the critical stress as a function of slenderness ratio, $L/r$, using the elastic-perfectly plastic idealization from Figure 6.1.](image)

The relationship between the increase in critical failure stress ($\Delta \sigma_{CR}$) and the increase in yield strength ($\Delta \sigma_{YS}$) is plotted in Figure 6.3. Slenderness ratios below $L/r = 59$ follow the same path in which there is a one-to-one increase between critical buckling
stress and yield strength. In other words, whatever strength increase has been provided to the parent material via work hardening is seen as an equivalent increase to the failure strength of the microtruss strut for $L/r < 59$. As the slenderness ratio increases, progressively less of this work hardening can be used to strengthen the strut. For example, at $L/r = 90$, a strut in the annealed state can only have its critical buckling strength increased by ~100 MPa even if the parent material yield strength is increased by 200 or 300 MPa. Once the slenderness ratio exceeds 155, there is no benefit to work hardening. No matter how much the yield strength of the parent material is increased, the critical buckling strength remains constant.

![Graph showing the difference in critical buckling strength as a function of increase in yield strength for the case of elastic-perfectly plastic behavior of C11000 for a range of slenderness ratios, $59 \leq L/r \leq 155$.](image)

Figure 6.3. The difference in critical buckling strength as a function of increase in yield strength for the case of elastic-perfectly plastic behavior of C11000 for a range of slenderness ratios, $59 \leq L/r \leq 155$.

The same analysis was repeated using the Ramberg-Osgood model fits to the experimental stress-strain curves (see Appendix B). The column curves for the four C11000 tempers, giving the critical stress as a function of slenderness ratio, is shown in
Figure 6.4. Similar to the case for elastic-perfectly plastic behaviour, the difference in yield strength, $\Delta \sigma_{YS}$, was defined as the difference in strength between a work hardened temper and the annealed baseline. In this case, the upper boundary of Zone 1 shifted from $L/r < 59$ to $L/r < 18$ while the lower boundary of Zone 3 was increased from $L/r > 155$ to $L/r > 300$.

Figure 6.4. Column curves giving the critical stress as a function of slenderness ratio, $L/r$, using the Ramberg-Osgood constitutive relationships of C11000 for the following tempers: hard, half hard, quarter hard and annealed.

The correlation between the increase in critical failure stress ($\Delta \sigma_{CR}$) and increase in yield strength ($\Delta \sigma_{YS}$) based on the more accurate Ramberg-Osgood constitutive model material properties is shown in Figure 6.5. In this case, every slenderness ratio below 18, will have a one-to-one relationship between critical buckling stress and yield strength, which confirms that whatever strength increase is provided to the parent material in short, stocky struts via work hardening is seen as an equivalent increase to the failure strength.
of the microtruss strut. When the slenderness ratio is increased above $L/r = 18$, there is less potential for work hardening to increase the strut strength. For example, using the same $L/r = 90$ slenderness ratio as described previously for the elastic-perfectly plastic case, a strut in the annealed state can again only have its critical buckling strength increased by $\sim 100$ MPa. When the slenderness ratio is greater than 300, there is no benefit to work hardening. The critical buckling strength will not increase despite an increase in yield strength of the parent material.

![Figure 6.5. The difference in critical buckling strength as a function of yield strength for C11000 with a range of slenderness ratios $18 \leq L/r \leq 300$ using column curves based on material properties from Figure 6.4.](image)

The difference between the two material models in terms of the predicted buckling strength is shown in Figure 6.6 where both idealized elastic-perfectly plastic and Ramberg-Osgood stress-strain curves for the half hard temper are plotted along the intersections of the critical stress from the buckling equation (Equation 4.8). At a
slenderness ratio of $L/r = 65$, the critical buckling stress intersects with the elastic-perfectly plastic model at the yield stress, while for the Ramberg-Osgood model, the intersection occurs approximately 40 MPa below. The elastic-perfectly plastic material model assumption therefore minimizes the breadth of the Zone 2 slenderness ratio range, because the proportional limit is effectively increased up to the yield strength. This can be seen in the schematic diagram of Figure 6.7, which illustrates $\sigma_{CR}$ vs $\sigma_{YS}$ space. The difference between the elastic-perfectly plastic case and the Ramberg-Osgood case is that Zone 2 is expanded from $59 \leq L/r \leq 155$ to $18 \leq L/r \leq 300$. In Chapter 4, the C11000 microtrusses were formed to internal truss angles ranging from $21^\circ$ to $47^\circ$, which corresponded to internal slenderness ratios of 45 to 70. These samples would have all therefore fallen into the Zone 2 regime.

Figure 6.6. Idealized elastic-perfectly plastic and Ramberg-Osgood stress-strain curves of the C11000 half hard temper showing the computation of the critical buckling strength intersection of the critical stress from the inelastic buckling equation, Equation 4.8) for slenderness ratios of $L/r = 21$ and 65.
As would be expected for both material models, the slenderness ratio at which the increase in critical buckling stress exceeds the increase in yield stress decreases with increasing material strength. These values are summarized in Table 6.1 and plotted as function of yield strength in Figure 6.8. The case where there is more than one strengthening mechanism active in the base material of the core would be addressed in similar terms to the consideration given above for work hardening. For example, in the AA2024 system, strengthening is typically achieved either through precipitation hardening or through a combination of precipitation and work hardening (as discussed in Section 5.2). Having an additional strengthening mechanism active in the microtruss parent material would increase the yield strength and push the yielding/buckling
boundary to lower slenderness ratios. In other words, an extrapolation of a trend such as that seen in Figure 6.8 where the slenderness ratio at the transition between yielding and buckling decreases with increasing yield strength. This effect can be rationalized by considering that the higher strength of the parent material allows higher compressive stresses to be reached in the strut, which increases the susceptibility to buckling failure therefore lowering the buckling/yielding transition slenderness ratio.

![Figure 6.8](image)

**Figure 6.8.** The slenderness ratio at which the critical buckling stress equals the yield strength plotted as a function of yield strength for idealized elastic-perfectly plastic and Ramberg-Osgood models.

**Table 6.1.** Slenderness ratio, $L/r$, at which the inelastic buckling strength exceeds the yield strength according to C11000 temper for idealized elastic-perfectly plastic column curves (Figure 6.2) and Ramberg-Osgood column curves (Figure 6.4).

<table>
<thead>
<tr>
<th>Temper</th>
<th>Idealized</th>
<th>Ramberg-Osgood</th>
</tr>
</thead>
<tbody>
<tr>
<td>Annealed</td>
<td>155</td>
<td>32</td>
</tr>
<tr>
<td>Quarter hard</td>
<td>74</td>
<td>25</td>
</tr>
<tr>
<td>Half hard</td>
<td>65</td>
<td>21</td>
</tr>
<tr>
<td>Hard</td>
<td>59</td>
<td>18</td>
</tr>
</tbody>
</table>
6.2 STRUCTURAL COATING STRENGTH

The question of strengthening mechanism additivity in the presence of a structural coating was examined in more detail using the n-NiFe/C11000 example as a model system. This system was chosen over Al₂O₃/AA2024 because of the larger effect that the coating has on the mechanical properties of the overall composite and the ability of composite inelastic buckling models to predict the strength. To model the case of a copper microtruss coated with nanocrystalline nickel-iron (n-NiFe), composite stress-strain curves with elastic-perfectly plastic behavior were generated based on an annealed C11000 core with varying area fractions of nanomaterial, Figure 6.9. The composite stress-strain curves show two limiting regions: an initial elastic region and a final perfectly plastic region. The initial slope is derived using the rule of mixtures based on the modulus of elasticity from the core and coating, and assumes the core and coating both act elastically. At the yield strength of the core, the core becomes perfectly plastic while the coating remains elastic and the composite microtruss’s stiffness decreases to the weighted average of the coating material’s modulus. At the yield strength of the n-NiFe coating, both the core and coating behave plastically. This behaviour is repeated for increasing volume fractions of n-NiFe coating on the microtruss.
Figure 6.9. Stress-strain curves for C11000 based on elastic-perfectly plastic tensile behaviour with increasing volume fraction ($f_{\text{n-NiFe}}$) of n-NiFe coating on an annealed core ($\sigma_{YS} = 47$ MPa).

A more accurate representation can be obtained using the Ramberg-Osgood model to analyze how the material properties of the core and coating contribute to the composite properties. Composite stress-strain curves were generated for the case of increasing n-NiFe area fraction for each of the C11000 core tempers; the example of an annealed core temper is shown in Figure 6.10. The yield strength for the n-NiFe/Cu composites was defined using the 0.2% offset method and values are summarized in Table 6.2.
Figure 6.10. Stress-strain curves for C11000 based on Ramberg-Osgood material properties with increasing volume fraction of n-NiFe on for an annealed core.

Table 6.2. 0.2% yield strength, $\sigma_{YS}$, for composite n-NiFe/Cu with annealed temper core.

<table>
<thead>
<tr>
<th>$f_{n-NiFe}$</th>
<th>$\sigma_{YS}$, Annealed core</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.2</td>
<td>147</td>
</tr>
<tr>
<td>0.4</td>
<td>280</td>
</tr>
<tr>
<td>0.6</td>
<td>557</td>
</tr>
<tr>
<td>0.8</td>
<td>770</td>
</tr>
</tbody>
</table>

Figure 6.11 shows the construction of a strength additivity map for pin-jointed ($k=1$) C11000 struts of slenderness ratio 78 (corresponding to the slenderness ratios in Section 4.3) reinforced with n-NiFe sleeves. Figure 6.11a plots the critical buckling stress (determined from the Ramberg-Osgood based column curves in Section 6.1) for the annealed, quarter hard, half hard and hard tempers at a slenderness ratio of 78. Figure 6.11b shows a linear interpolation between these same data points and a delineation of the additivity map between zones of inelastic and elastic core buckling. Finally, Figure 6.11c
plots the predicted critical buckling stress of $L/r = 78$ columns reinforced with 10 $\mu$m thick coatings of n-NiFe.

Figure 6.11. Critical buckling strength as a function of yield strength for C11000 using slenderness ratios, $L/r = 78$ and $k = 1$, showing (a) individual data points for uncoated core, (b) elastic and inelastic regimes, and (c) the addition of a 10 $\mu$m coating applied to the core.

Figure 6.12 plots the full additivity map with coating thicknesses of up to 200 $\mu$m in 10 $\mu$m thick increments. There are several key issues to note. First, the amount of the strength increase seen over the elastic core buckling regime (i.e. for core yield strengths $> 200$ MPa) was independent of core yield strength. In other words, a 10 $\mu$m thick coating on an elastically buckling quarter hard C11000 core provides the same strength increase
as a 10 µm thick coating on a half hard or hard core. Second, the amount of strength increase for each 10 µm thick increment of n-NiFe coating thickness decreases with increasing coating thickness. For example, the first 10 µm of coating reinforcement provides a strength increase of 1.44 MPa/µm, while the incremental coating thickness increase from 190 µm to 200 µm provides only a strength increase of 1.27 MPa/µm. Third, the incremental strength increase is higher in the zone of inelastic core buckling. For example, a 10 µm thick coating on an annealed temper core provides a strength increase of 2.09 MPa/µm compared to the 1.44 MPa/µm seen for the elastically buckling tempers. Just like the case for the elastically buckling tempers, however, the incremental strength increase decreases with increasing coating thickness so that the final 10 µm of coating thickness on a 200 µm thick coating provides an incremental increase of only 1.58 MPa/µm.

![Figure 6.12. Critical buckling strength as a function of yield strength with increasing nanocrystalline NiFe coating thickness ranging from 0 to 200 µm for C11000 using slenderness ratios, L/r = 78 and k=1.](image)
The end constraint of the buckling column can also have an effect on the strengthening mechanism additivity. Figure 6.13 plots the additivity map for the same sleeve core geometries but now with a rigid end constraint (i.e. \( k=2 \)). The same overall trend is seen for \( k=2 \) as with \( k=1 \). The incremental strength increase is considerably larger for the case of rigid end constraints (e.g. 3.83 MPa/\( \mu \)m compared to 2.09 MPa/\( \mu \)m for the first 10 \( \mu \)m increment on an annealed copper core).

![Critical Buckling Stress vs Yield Strength](image)

Figure 6.13. Critical buckling strength as a function of yield strength with increasing nanocrystalline NiFe coating thickness ranging from 0 to 200 \( \mu \)m for C11000 using slenderness ratios, \( L/r = 78 \) and \( k = 2 \).

The extent of strengthening mechanism additivity in composite microtruss materials is therefore a complex function of material, geometric and boundary condition effects. Material effects act through the constitutive relationships of the components making up the microtruss struts and are particularly sensitive to the elastic-plastic transition in the stress-strain curves. Geometric effects act through the slenderness ratio of the strut, the area fraction of the composite fractions and the form of their distribution.
over the strut cross-section (having the high strength nanocrystalline material on the outside of the microtruss strut is much more effective than having the configuration reversed). Finally, boundary condition effects, while less significant than the other two categories, nevertheless play a role in determining the overall strengthening additivity.
7 CONCLUSIONS AND FUTURE WORK

Microtruss metals are an attractive materials domain to develop because they are efficient load-supporting structures that can also meet environmental needs in terms of material and energy savings. To achieve this goal, architectural and microstructural effects need to be studied in order to develop different strategies for strengthening these structures. The present study investigated potential strengthening mechanisms in two material systems, examining the nature of their strength additivity.

Pyramidal C11000 copper alloy microtrusses were fabricated using a multi-cycle stretch-bend method for the first time. The effect of architecture and microstructure were investigated over a range of truss angles. The increase in mechanical performance with decreased relative density makes copper microtrusses an attractive option for multifunctional applications to take advantage of the excellent thermal conductivity. The effects of pre-annealing time and temperature on the starting sheet material were studied to achieve the greatest formability (stretch forming displacement) and to access a wider range of architectural space. The selected pre-annealing conditions increased the uniformity of final failure with multiple strut fractures occurring at the failure limit. The cellular copper produced by stretch bend fabrication for the first time in this study is able to match or exceed the low densities achieved by the cellular copper synthesis methods reported to-date.

It was shown that increasing the truss angle affects the mechanical performance in two counteracting manners: the compressive strength of a microtruss core increases because of the increasingly efficient load resolution with respect to the loading axis, however, the microtruss struts also become both longer and narrower and thus, more
susceptible to buckling. The relative insensitivity of the peak strength to the changing truss angle was therefore controlled by the trade-off between the counteracting effects, leading to a maximum in the peak strength at a truss angle of approximately 38°. Work hardening was found to significantly improve the mechanical performance of each mechanical property measured: reload modulus, peak strength, valley strength and densification energy. Even though the work hardened microtruss cores exhibited quite a favourable performance compared to the annealed core samples over the same range of truss angles (for example, up to a factor of three increase in strength), it may be possible to further improve their performance by better controlling the distribution of plastic strain to the microtruss strut during the stretch-bend fabrication. Future work could include microstructural characterization by electron backscatter diffraction in order to more thoroughly examine the plastic distribution imparted to the microtruss struts during fabrication.

By electrodepositing high strength sleeves of nanocrystalline NiFe, two complementary strengthening mechanisms in cellular copper were combined for the first time: a structural coating and fabrication induced work hardening. The structural nanocrystalline coating gave up to ~250% specific strength increase, and the two strengthening mechanisms were found to be additive over the range of pyramidal architectures studied. The failure mode of the composite microtrusses was considerably more complex, however, with sleeve delamination and fracture observed by SEM. These observations were explained by the development of tensile stresses across the sleeve/core interface during local shell buckling at the hinge. Delamination and cracking of the coating resulted in the development of local tensile stresses within the sleeve before the
buckling stress was reached; partially failed samples indicated that those zones experiencing the highest local compressive stresses tended to wrinkle and those that experienced the highest tensile stresses tended to fracture. Future work could involve Finite Element analysis to examine the evolution of stresses across the sleeve/core interface during architectural collapse.

Aluminum microtrusses of AA2024 alloy offer a lightweight load-bearing option with less weight penalty than copper microtrusses. The interaction of three microstructural strengthening mechanisms was studied using this system: work hardening, precipitation hardening and a structural hard anodized ceramic Al₂O₃ sleeve coating. However, given the lower formability of this alloy, the effects of both perforation geometry and starting temper on stretch-bend fabrication were examined.

The best combination of formability came from a solutionizing heat treatment and a decreasing strut width. At the forming limit, it was observed that the location of strut failure depended on the perforation geometry where the location of strut fracture was indicative of the region of highest plastic strain once a critical load was reached.

Several cycles of deformation forming and annealing were required in order to reach the desired internal truss angle. With each passing forming cycle, the amount of forming force required increases while the amount of stretch displacement decreases. There is therefore progressively less and less benefit to forming with each subsequent deformation cycle, with the targeted 45° internal truss angle (at less than 0.8d_{MAX}) being reached within 3 cycles for 0.6w_o, 4 cycles for 0.7w_o and 4 cycles for 0.9w_o. This study has therefore demonstrated that even relatively low formability alloys such as AA2024
can in fact be used for stretch bend fabrication given the correct sheet geometry and intermediate annealing steps.

Another important contribution was despite the fact that non-uniform strain was imparted during the final fabrication step, sufficient plastic strain was generated within the microtruss struts in order to increase the precipitation kinetics using a T8 temper, allowing higher peak strengths over a more conventional T6 processing route. The T8 processing route was also advantageous in that it completely eliminated one high temperature heat treatment step. Future work could include Finite Element analysis to determine the optimal pin and sheet geometries for stretch-bend fabrication.

In addition, while modifying the perforation geometry allowed higher internal truss angle architectures to be fabricated, it also led to more complex strut failure during compression testing. Off-axis buckling was observed in 0.6wₒ due to the smaller second moment of area in the width direction. Overall, both perforation geometry and thermomechanical core treatment influence the compressive peak strength and AA2024 microtrusses could be fabricated having enhanced mechanical properties compared to previously studied AA3003 microtrusses.

The AA2024 microtrusses could also be reinforced with an anodized coating of aluminum oxide. To a first approximation, strengthening mechanism additivity for Al₂O₃/AA2024 is dependent on architecture and nearly independent of temper (T8 vs O-temper core). The overall failure progression for Al₂O₃/AA2024 struts was different than the failure mechanism seen previously for Al₂O₃/AA3003 composites and was likely due to the greater slenderness ratio of the struts in the present study (64 and 47 for the 0.6wₒ and 0.9wₒ Al₂O₃/AA2024 struts compared to 28 for the Al₂O₃/AA3003 struts). The
increased slenderness ratio in the present study would have favored global buckling over hinge failure. SEM characterization emphasized the loss of coating integrity at the location of failure on each side of the strut that experienced tensile and compressive forces. Future work could address the question of aluminum oxide coating uniformity and the effect of applying post-anodizing coating treatments, such as sealing.

Further insight into the nature of strengthening additivity was addressed by considering two model systems: elastic-perfectly plastic and Ramberg-Osgood constitutive relationships. The effect of geometry and material properties were studied using the two models where zones of slenderness ratios were defined. The relationship between the increase in critical buckling stress ($\Delta \sigma_{CR}$) and the increase in yield strength ($\Delta \sigma_{YS}$) due to work hardening was a one-to-one relationship at the lower limit of slenderness ratio. In this zone, whatever strength increase has been provided to the parent material via work hardening is seen as an equivalent increase to the failure strength of the microtruss strut. In the intermediate zone of slenderness ratios, the benefit of work hardening could be realized, while at high slenderness ratios, struts behave similarly in an elastic fashion despite their temper. Similar zones were defined for each system; however, in the case of the Ramberg-Osgood model, the range of slenderness ratio was expanded.

Finally, strengthening additivity was also considered in the presence of a structural coating. The amount of the strength increase seen over the elastic core buckling regime was shown to be independent of core yield strength and the amount of strength increase for a given increment of coating thickness decreased with increasing coating thickness. Furthermore, the incremental strength increase was higher in the case of
inelastic core buckling. The end constraint of the buckling column was also seen to have an effect on the strengthening mechanism additivity. Strengthening mechanism additivity in composite microtruss materials is therefore a complex function of material, geometric and boundary condition effects and the inelastic buckling models used in the present study can be used to take these factors into account.
Appendix A: Microtruss Mechanical Properties from Uniaxial Compression Testing and their Definitions

Microtruss mechanical properties were obtained from the data generated by uniaxial compression tests following conventional mechanical testing methods used for metallic foams [1]. In this study, the mechanical properties from a uniaxial compression test of a microtruss, pictured in Figure A1, are defined as follows:

- **Peak strength** ($\sigma_P$): the initial maximum stress, the same as for metallic foams [1]
- **Valley strength** ($\sigma_V$): the minimum stress following the peak stress
- **Reload modulus** ($E_{\text{Reload}}$): the maximum slope on the reload path (see Figure A2) when the microtruss is first loaded to $\frac{3}{4}\sigma_P$, unloaded and subsequently reloaded, following the conventional mechanical testing used for metallic foams [1]
- **Densification energy** ($J_D$): the area under the curve between strain at $\frac{1}{2}\sigma_P$ and $2\sigma_P$ [1-3].
Figure A1. Representative compressive stress-strain curve showing the mechanical properties obtained from a typical compression test: reload modulus ($E_{\text{Reload}}$), peak strength ($\sigma_P$), valley strength ($\sigma_V$), and densification energy ($J_D$).

The densification strain has been selected after Olurin et al. [2] to account for the bedding-in effects where: $\varepsilon_D = \varepsilon(\frac{1}{2}\sigma_P) - \varepsilon(2\sigma_P)$ and relates to the measured densification energy, $J_D$, during microtruss collapse [1]. This area under the curve between strain at $\frac{1}{2}\sigma_P$ and $2\sigma_P$ is calculated as an upper-bound integral using a Riemann sum. The calculation of $J_D$ as a Riemann sum can be calculated over the entire strain interval $i$ using [4]:

$$J = \sum_i \sigma_{\varepsilon} \cdot (\varepsilon_i - \varepsilon_{i-1}) \quad (A.1)$$

Since the modulus of the loading curve can be reduced during the initial compressive deformation due to bedding-in of the microtruss into the compression plates [5-6], a reload modulus was defined as the maximum slope of the compressive reload stress-
strain curve obtained when the microtruss was first loaded to 75% of the peak stress, unloaded and subsequently reloaded, following the same method used for metallic foams [1,7], (one sample was tested fully to obtain $\sigma_P$ with subsequent samples tested to determine parameters based on $\sigma_P$), a representative curve of the compression testing pathway is shown in Figure A2.

Figure A2. Detail of the low strain region in a typical compressive stress-strain curve of a work hardened C11000 alloy pyramidal microtruss. The sample was first loaded to $\frac{3}{4}\sigma_P$, unloaded, and subsequently reloaded to beyond the peak strength in order to obtain the reload modulus.

References


Appendix B: Fitting C11000 and AA2024 stress-strain curves to obtain Ramberg-Osgood Parameters

C11000 stress strain values were obtained from published ASM stress-strain curves [1], Figure B1, and fitted to determine the Ramberg-Osgood strain hardening parameters listed for each temper in Table B1 below. The Ramberg-Osgood model [2] can be used to describe the non-linear elastic to plastic behaviour on the stress-strain ($\sigma$-$\varepsilon$) curve:

$$\varepsilon = \frac{\sigma}{E} + \varepsilon_0 \left( \frac{\sigma}{\sigma_{YS}} \right)^N$$

where $E$ is the modulus of elasticity of the material, $\sigma_{YS}$ is the 0.2% offset yield strength and $N$ is a strain hardening exponent. Each temper’s yield strength was taken using the 0.2% offset method.

![Experimental C11000 stress-strain curves](image)

Figure B1. Experimental C11000 stress-strain curves (open symbol) from [1] fit to the Ramberg-Osgood constitutive model (solid lines).
Table B1. Summary of C11000 Ramberg-Osgood fitting parameters: Young’s modulus ($E$), 0.2% offset yield strength ($\sigma_{YS}$), plastic strain taken at $\sigma_{YS}$ ($\varepsilon_o$), and strain hardening exponent, ($N$).

<table>
<thead>
<tr>
<th>Temper Condition</th>
<th>$E$ (GPa)</th>
<th>$\sigma_{YS}$ (MPa)</th>
<th>$\varepsilon_o$ (mm/mm)</th>
<th>$N$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Annealed</td>
<td>115</td>
<td>47</td>
<td>0.002</td>
<td>4.1</td>
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<tr>
<td>Quarter hard</td>
<td>115</td>
<td>208</td>
<td>0.002</td>
<td>6.8</td>
</tr>
<tr>
<td>Half hard</td>
<td>115</td>
<td>271</td>
<td>0.002</td>
<td>10</td>
</tr>
<tr>
<td>Hard</td>
<td>115</td>
<td>324</td>
<td>0.002</td>
<td>13</td>
</tr>
</tbody>
</table>

AA2024 stress strain values were obtained from published ASM stress-strain curves [1], Figure B2, and fitted to determine the Ramberg-Osgood strain hardening parameters listed for each temper in Table B2 below.

![Stress-strain curves](image)

Figure B2. Experimental AA2024 stress-strain curves (open symbol) from [1] fit to the Ramberg-Osgood constitutive model (solid lines).
Table B2. Summary of AA2024 Ramberg-Osgood fitting parameters: Young’s modulus ($E$), 0.2% offset yield strength ($\sigma_{YS}$), plastic strain taken at $\sigma_{YS}$ ($\varepsilon_o$), and strain hardening exponent, ($N$).

<table>
<thead>
<tr>
<th>Temper Condition</th>
<th>$E$ (GPa)</th>
<th>$\sigma_{YS}$ (MPa)</th>
<th>$\varepsilon_o$ (mm/mm)</th>
<th>$N$</th>
</tr>
</thead>
<tbody>
<tr>
<td>O-temper*</td>
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<td>0.002</td>
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<td>T86</td>
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<td>493</td>
<td>0.002</td>
<td>26.1</td>
</tr>
</tbody>
</table>

*O-temper properties from [3].

The definition of aluminum temper designations is as follows [4]:

O: annealed to obtain lowest strength temper.

T3: solution heat-treated, cold worked, and naturally aged to a substantially stable condition.

T4: solution heat-treated and naturally aged to a substantially stable condition.

T8: solution heat-treated, cold worked and artificially aged.

References


Appendix C: Derivation of Holloman Parameters for Analytical Strength Predictions

The following section provides a brief overview of the model used to predict the critical buckling stress in work hardened C11000 microtruss cores (Section 4.2.2). It is based on the work of H.M. Yu, a colleague in the Hybrid Materials Design group at the University of Toronto [1].

The Holloman power law, used to describe the plastic behaviour from a stress-strain curve, is:

$$\sigma = Ke^n$$  \hspace{1cm} (C.1)

Where $\sigma$ is the stress, $K$ is a strain hardening co-efficient, $\varepsilon$ is the strain and $n$ is a strain hardening exponent. It can also be expressed as:

$$\ln \sigma = \ln K + n \ln \varepsilon$$  \hspace{1cm} (C.2)

which describes a straight line, and when graphed, $\ln(K)$ is the y-intercept, and $n$ is the slope of the line. The strain hardening exponent, $n$, will decrease with increasing yield strength. The tensile curve can further be described in two parts based on a material’s stress-strain behaviour. The initial part describes the elastic behaviour using Hooke’s Law ($\sigma = E\varepsilon$), where $E$ is Young’s Modulus. Following the pathway on the stress-strain curve, Hooke’s law transitions to the Holloman equation at the proportional limit.

Combining the two laws, the following equation is derived:

$$\sigma = K \left( \frac{\sigma_{YS}}{E} \right)^n$$  \hspace{1cm} (C.3)

where $\sigma_{YS}$ is the yield strength. This relationship can be further expanded to:

$$\ln K = n \ln E + (1 - n) \ln \sigma_{YS}$$  \hspace{1cm} (C.4)
In the present study investigating microtrusses, Holloman parameters were derived from experimental tensile stress-strain curves to understand the relationship between the truss (i.e. forming) angle and the mechanical performance properties. For a microtruss with an annealed temper, Holloman parameters \( K \) and \( n \) remain constant because they have a uniform internal metallurgical state. For work hardened microtrusses, the truss angle is a function of the degree of work hardening input into the struts during fabrication. Stretch-bend fabrication lengthens the struts as the struts are displaced out of plane by alternating pins, this corresponds to a reduced cross-section at mid-strut. The final length of the strut, \( l \), can be calculated using:

\[
l = l_o / \cos \omega = l_o \sec \omega \tag{C.5}
\]

where \( l_o \) and \( \omega \) are the initial strut length and truss angle respectively. Assuming a constant strut volume and a constant reduction in cross-section dimension, the width \( w \) will change according to:

\[
w = w_o \sqrt{\cos \omega} \tag{C.6}
\]

where \( w_o \) is the initial strut width, and likewise the strut thickness \( t \) can be calculated using:

\[
t = t_o \sqrt{\cos \omega} \tag{C.7}
\]

As the yield strength increases, the strain hardening parameters, \( K \) and \( n \), decrease (based on a decreasing tangent modulus past the proportional limit). Since Holloman is function of yield strength and plastic strain, the total plastic true strain is related to a microtruss core’s truss angle by the following relation:

\[
\varepsilon_{\text{trus}}^{\text{pl}} = \ln(\sec \omega). \tag{C.8}
\]
The above analytical model, Equation C.8, allows the prediction of inelastic buckling strength of a work hardened microtruss as a function of truss angle.

References

Appendix D: Microscopy of AA2024 and Al₂O₃/AA2024 Showing Strut Failure Progression during Uniaxial Compression Testing

The following sections, D1–D3 are microscopy images of AA2024 and Al₂O₃/AA2024 struts taken using: optical microscopy (Section D1), SEM micrographs showing failed strut profiles (Section D2) and SEM of partially failed strut cross-sections (Section D3) showing their failure progression at strain values ε ≈ 0.15 (equivalent to the strain directly following peak strength, σₚ), ε ≈ 0.3 (equivalent to the strain directly prior to valley strength, σᵥ) and ε ≈ 0.7 (equivalent to the strain at 2x peak strength, 2σₚ).
Section D1: Optical Microscopy Images of AA2024

<table>
<thead>
<tr>
<th>Strain, $\varepsilon$</th>
<th>Uncoated AA2024</th>
<th>Coated $\text{Al}_2\text{O}_3$/AA2024</th>
</tr>
</thead>
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<tr>
<td>0.2</td>
<td><img src="image1" alt="Image" /></td>
<td><img src="image2" alt="Image" /></td>
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<tr>
<td>0.3</td>
<td><img src="image3" alt="Image" /></td>
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<tr>
<td>0.7</td>
<td><img src="image5" alt="Image" /></td>
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</tbody>
</table>
### $0.6w_o$, T8 core

<table>
<thead>
<tr>
<th>Strain, $\varepsilon$</th>
<th>Uncoated AA2024</th>
<th>Coated Al$_2$O$_3$/AA2024</th>
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<tr>
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<tr>
<td>0.3</td>
<td><img src="image3" alt="Image" /></td>
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<tr>
<td>0.7</td>
<td><img src="image5" alt="Image" /></td>
<td><img src="image6" alt="Image" /></td>
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### 0.9\textit{w}_p, annealed core

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<th>Coated Al$_2$O$_3$/AA2024</th>
</tr>
</thead>
<tbody>
<tr>
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<td><img src="image1.png" alt="Image" /></td>
<td><img src="image2.png" alt="Image" /></td>
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<tr>
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<td><img src="image3.png" alt="Image" /></td>
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<tr>
<td>0.7</td>
<td><img src="image5.png" alt="Image" /></td>
<td><img src="image6.png" alt="Image" /></td>
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</table>
### 0.9\( w_o \), T8 core

<table>
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<th>Uncoated AA2024</th>
<th>Coated Al(_2)O(_3)/AA2024</th>
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Section D2: SEM Micrographs showing Failed Strut Profiles

<table>
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<th>$0.6\omega_o$, annealed core</th>
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<td></td>
<td>Uncoated AA2024</td>
</tr>
<tr>
<td>0.7</td>
<td><img src="image1.png" alt="Micrograph" /></td>
</tr>
<tr>
<td></td>
<td>Coated Al$_2$O$_3$/AA2024</td>
</tr>
<tr>
<td></td>
<td><img src="image2.png" alt="Micrograph" /></td>
</tr>
<tr>
<td>Strain, $\varepsilon$</td>
<td>0.6$w_o$, T8 core</td>
</tr>
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<td>-----------------</td>
<td>------------------</td>
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<tr>
<td></td>
<td>Uncoated AA2024</td>
</tr>
<tr>
<td>0.7</td>
<td>Coated $\text{Al}_2\text{O}_3$/AA2024</td>
</tr>
<tr>
<td>Strain, $\varepsilon$</td>
<td>$0.9w_o$, annealed core</td>
</tr>
<tr>
<td>----------------------</td>
<td>-------------------------</td>
</tr>
<tr>
<td></td>
<td>Uncoated AA2024</td>
</tr>
<tr>
<td>0.7</td>
<td>Coated Al$_2$O$_3$/AA2024</td>
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</tbody>
</table>

![Image of uncoated AA2024 sample](image1)

![Image of coated Al$_2$O$_3$/AA2024 sample](image2)
<table>
<thead>
<tr>
<th>Strain, ( \varepsilon )</th>
<th>0.9( \nu ), T8 core</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Uncoated AA2024</td>
</tr>
<tr>
<td>0.7</td>
<td>Coated Al(_2)O(_3)/AA2024</td>
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</tbody>
</table>

1 mm
### Section D3: SEM Micrographs of Partially Failed Strut Cross-Sections

<table>
<thead>
<tr>
<th>Strain, $\varepsilon$</th>
<th>0.6$w_{0.5}$ Coated Al$_2$O$_3$/AA2024</th>
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<tbody>
<tr>
<td></td>
<td>Annealed core</td>
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<tr>
<td>0.7</td>
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<tr>
<td></td>
<td>T8 core</td>
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<tr>
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<td><img src="image2.png" alt="Image" /></td>
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<tr>
<td></td>
<td>1 mm</td>
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<tr>
<td>Strain, $\varepsilon$</td>
<td>0.9$\omega$, Coated Al$_2$O$_3$/AA2024</td>
</tr>
<tr>
<td>----------------------</td>
<td>-----------------------------------------</td>
</tr>
<tr>
<td>0.7</td>
<td><strong>Annealed core</strong></td>
</tr>
</tbody>
</table>

![Image of annealed core](image1)

| 0.7                  | **T8 core**                             |

![Image of T8 core](image2)