ABSTRACT

Thermal diffusivity/conductivity can be used to assess pre-densification sintering phenomena in injection molding powders, indicating the onset of component strength during heating. Laser flash analysis can provide in-situ information on interparticle bonding, giving a sense of when handling strength emerges. Combined with strength evolution and densification data, a model can predict points in the sintering cycle that are vulnerable to damage mechanisms. Initial test results indicate that neck growth directly corresponds to the increase in thermal diffusivity once sintering has begun, but further investigation is needed in the transition from powder to the presintered condition.

INTRODUCTION

Powder metallurgy offers the intriguing capability of creating net-shaped components while tailoring materials and microstructure to obtain desirable properties. Careful attention to both the shaping and sintering processes can improve quality and consistency of the final product. Die compaction allows simply shaped parts to be formed at a relatively low cost, but undesirable density gradients are a common by-product of this forming process. Alternatively, metal injection molding allows more shape complexity with a homogeneous, but less dense, green structure, which results in a more fragile part going in to the sintering process. During sintering, as the compact gains strength from the growth of interparticle bonds, density gradients and thermal gradients can generate stresses, which potentially distort, warp, or crack the component. Insights into strength evolution and heat transfer during the sintering process are critical to predicting and avoiding such damage. Data obtained in situ during the sintering process provide the most direct assessment of mechanisms that are active in the development of strength during sintering.

The objective of this research is to demonstrate the connection between strength evolution and thermal diffusivity during sintering, and use this relationship to predict vulnerabilities in the material during the sintering cycle and avoid potentially damaging stresses. The work will build on past strength evolution research, and use laser flash analysis as a non-contact technique capable of assessing in situ thermal diffusivity evolution of injection molded systems. Mechanical properties and microstructural evolution
will be investigated to reinforce and complement the proposed relationship between the thermal
diffusivity and strength of the material. This paper will focus on initial results for strength evolution and
thermal diffusivity evolution of the candidate system.

BACKGROUND

A variety of techniques have been proposed and explored as avenues to monitor the sintering process.
One approach to describing the types of proposed monitors is to categorize them as direct versus indirect
techniques. Direct techniques such as quantitative microscopy, SEM hot stage, and video imaging
provide visual data to quantify neck growth and densification. Physical measurements of surface area via
BET evaluation, and densification via dilatometry also are considered direct techniques, and combining
such complementary techniques can offer physical insight into separate sintering mechanisms—for
example neck growth caused by surface diffusion with no contribution to densification. Indirect
methods, on the other hand, study the initiation and growth of interparticle bonds by assessing other
properties of the material, and inferring an interparticle neck size or density. For example, the electrical
and thermal conductivities of a compact increase during the early stages of sintering, therefore measuring
conductivity could show the relative change in interparticle contact area. The elastic properties, such as
bulk modulus, also change, which allow ultrasonic evaluation of the sintering process. Mechanical
evaluations, such as hardness and transverse rupture testing, also indicate the extent of interparticle
bonding. In situ mechanical testing has been the basis of key work on strength evolution testing, but this
type of test provides only one piece of data per test run, and the more fragile PIM compacts are difficult
to evaluate in situ with accuracy. Since many of the sintering monitors mentioned above are functions of
the same variables (interparticle bond area, particle packing, and grain size, to name a few), a hypothesis
is proposed that one technique, in situ conductivity, may be used to evaluate another, in situ strength.

Models describing conductivity, either electrical or thermal, of a two-phase material usually assume a
continuous network of one of the phases. In the most basic sense, a rule of mixtures model may be
applied, using effective load bearing area or minimum solid area to represent the average cross-section
of interconnected solid in a porous material:

\[
\lambda_{\text{eff}} = \lambda_m (1 - \varepsilon)
\]  

(1)

where \(\lambda_{\text{eff}}\) is the effective conductivity of the porous material, \(\lambda_m\) is the conductivity of the fully dense
material, and \(\varepsilon\) is the porosity (0<\(\varepsilon\)<1.0). This representation assumes zero conductivity in the pores and
solid conductivity equivalent to that of a wrought bar the same size as the effective load bearing area.
This simple model can provide an upper limit for conductivity, but particularly at higher porosities and
early stages of sintering, pore shape, pore size distribution, and interparticle contacts limit conductivity to
much lower values. In order to establish a valid lower bound for thermal conductivity evolution, we must
start with a bed of granular material.

The most basic model for conduction in a bed of loose powder assumes a simple cubic stacking of mono-
sized spherical particles, with effective conductivity a function of the number of contact points on a
single particle, also known as the coordination number. Models for more densely packed structures, as
well as random packing, may be extrapolated by increasing the coordination number. As a lower bound,
the particles may be considered in point contact, although this is not strictly possible in the mathematical
sense. Batchelor and O'Brien's analysis of this problem is widely referenced for this application. For a
bed of solid particles in a gas, conductivity may be described by:

\[
\frac{\lambda_{\text{eff}}}{\lambda_g} = V_s N_c * 2 \ln \left( \frac{\lambda_m}{\lambda_g} \right) - C
\]  

(2)
where $\lambda_{\text{eff}}$ is the effective conductivity of the porous material, $\lambda_s$ is conductivity of the solid, $V_s$ is the solid volume fraction, $N_c$ is the coordination number, $\lambda_g$ is conductivity of the gas in the pore and $C$ is an empirical constant equal. This relationship is not practical for application in electrical conductivity of a solid in a gas, since the electrical conductivity of gas is effectively zero, but it is a common reference point for thermal conductivity of a solid/gas granular bed. Once particle contact is considered to have dimensions of area, a basic unit cell is usually modeled with particle contacts described as cylinders in end-to-end contact. Such a configuration can be described as an electrical (or thermal) resistance element:

$$R = \frac{L}{\lambda A}$$

where $R$ is the resistance, $L$ is the cylinder height, $\lambda$ is the conductivity, and $A$ is the cross-sectional area.

The same field equations and models used for electrical conductivity also hold for thermal conductivity. A sample of these relations is shown in Table 1. However, the presence of gas in the pores has more impact on thermal conductivity than on electrical conductivity. Particularly at lower densities (higher porosity), conductivity at the interparticle contact is low enough that the conductivity of the gas phase is no longer negligible. Howard and Kozack\textsuperscript{10} used Maxwell’s isolated pore model as an upper limit and the Kunii-Smith loose powder model as a lower limit to explore density and pore conductivity effects. This work demonstrated that for a granular material, not only does effective conductivity increase with fractional density, but process gas also has a surprisingly significant impact on effective conductivity at low density—a particle bed with a hydrogen process gas is five times as conductive as a nitrogen process gas\textsuperscript{10}.

**Table 1. Sample of models for electrical conductivity, $\lambda$, and thermal conductivity, $\kappa$, in porous materials.**

<table>
<thead>
<tr>
<th>Model</th>
<th>Author, [Reference]</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\lambda_{\text{eff}} = \lambda_o (1 - \varepsilon)$</td>
<td>Hirschhorn, [7]</td>
<td>Rule of mixtures; $\lambda_o$ is the wrought conductivity</td>
</tr>
<tr>
<td>$\lambda_{\text{eff}} = \lambda_o (1 - 2\varepsilon)$</td>
<td>Boccinni, [11]</td>
<td>Porosity less than 20%</td>
</tr>
<tr>
<td>$\lambda_{\text{eff}} = \lambda_o (1 - \varepsilon)^3$</td>
<td>Ondracek, [12]</td>
<td>Upper and lower bounds for porous, isotropic material</td>
</tr>
<tr>
<td>$\lambda_{\text{eff}} = \frac{1 - \varepsilon}{1 + \chi \varepsilon^2}$</td>
<td>Koh and Fortini, [13]</td>
<td>$\chi$ represents pore distribution and shape—typically about 11.</td>
</tr>
<tr>
<td>$\lambda_{\text{eff}} = \lambda_o \left[ \frac{N_c}{1 - \varepsilon} \right]^{1/3}$</td>
<td>Jernot, [14]</td>
<td>Incorporates different packing schemes by changing coordination number, $N_c$.</td>
</tr>
<tr>
<td>$\lambda_{\text{eff}} = V_s N_c \star 2 \ln \left( \frac{\lambda_s}{\lambda_g} \right) - C$</td>
<td>Batchelor and O’Brien, [9]</td>
<td>Loose powder in point contact; $\lambda_g$ and $\lambda_s$ are conductivity of the gas and of the solid, respectively; $C$ is an empirical constant (11 for random packing).</td>
</tr>
<tr>
<td>$\kappa_{\text{eff}} = \kappa_g \left[ \varepsilon + \frac{\beta(1 - \varepsilon)}{\phi + \gamma(\kappa_g / \kappa_s)} \right]$</td>
<td>Kunii and Smith, [15]</td>
<td>Heat transfer in loose powder; $\beta$ and $\gamma$ usually 1 and 2 respectively; $\phi$ represents heat transfer mechanism</td>
</tr>
</tbody>
</table>
Recent work on thermal conductivity evolution, focused on initial stage sintering of ceramics, models neck growth based on surface diffusion only (pre-densification)\(^1\). This work begins with a pre-existing bond between the two particles, and does not address any change in the nature of the quality of the bond. Initial stage sintering of the two-sphere model as described by Coble is used to represent interparticle contact growth. Calculations of the effective thermal conductivity of a powder compact using a 3-d numerical model show a linear relationship between thermal conductivity ($\kappa$) and neck size ratio ($X/R$).

Several different methods are available to measure thermal conductivity. R. F. Speyer gives a review of basic techniques\(^1\). While thermal and electrical conductivities are steady state measurements, thermal diffusivity is a dynamic measurement. The laser flash method for measuring thermal diffusivity was first proposed by Parker, Jenkins, Butler, and Abbott\(^1\) in the early 1960’s and has developed into an ASTM standardized technique for assessing thermal diffusivity of a material\(^1\). This technique overcomes the difficulties of surface heat losses and contact resistance at heat sources and sinks, and sometimes complex or large sample size requirements encountered in other thermal conductivity techniques. The flash method uses a small disk of sample material (12 mm in diameter, 1-3 mm thick), irradiates one face of the disk with an “instantaneous” burst of energy from a xenon flash, laser, or electron beam, and measures the temperature rise on the other face of the disk using an infrared sensor. The short timeframe minimizes heat loss, and the non-contact energy input and measurement minimizes thermal contact problems. In combination with specific heat and density data, thermal conductivity of a material may be calculated according to the relationship:

$$\kappa = \alpha C_p \rho \tag{4}$$

where $\kappa$ is thermal conductivity (W/m-K), $\alpha$ is thermal diffusivity (m\(^2\)/s), $C_p$ is the specific heat of the sample material (J/kg-K), and $\rho$ is the density (kg/m\(^3\)). A comparative method may also be used to extract thermal conductivity from the laser flash technique. By comparing diffusivity measurements of both a sample and reference material whose specific heat and density data are known, conductivity may be calculated without a separate specific heat measurement.

Thermal diffusivity measurement via laser flash is a convenient capability for several reasons:
1. It is non-contact, therefore alleviating some of the problems associated with electrical conductivity measurements;
2. It allows multiple measurements to be made in a single sintering cycle;
3. Existing commercial equipment allows measurements to be made “in situ”, in desired sintering temperature and process gas environments.

This technique is quite promising for the examination of neck growth, but it is not without its own degree of experimental uncertainty. Some of the potential sources of error are:
1. The porous nature of the material creates a complex combination of heat transfer mechanisms that must be considered in any results;
2. Any surface roughness varies the depth of penetration of the incident laser beam, which then impacts the effective thickness of the material;
3. Shrinkage in a sintering sample will change the thickness and density of the sample material during measurement.
4. The high energy input on the front face of the sample disk may affect the microstructure or sintering characteristics of the material

**MODEL DEVELOPMENT**

The basic thermal circuit model will be used to predict and interpret thermal diffusivity data. The assumptions of the model, along with potential avenues for further refinement, are described below.
An injection-molded compact can be modeled as a bed of monosized spheres in a regular, periodic arrangement. A simple cubic unit cell structure was chosen as a starting point, based on its ease for modeling one-dimensional heat flow, and its packing density of 52%, which was fairly close to the green density of the sample material (55%). Interparticle contacts were modeled as cylinders stacked end-to-end, with neck size (cylinder diameter) increasing as sintering occurred. Since the interparticle contact is not very efficient at conduction when there is a small neck size and large contact resistance, conduction of heat across the gas in the pore is an important contributor to heat transfer during the early stages of sintering. The bulk of the spherical particle of diameter ‘D’ away from the cylindrical neck was modeled as a square layer of DxD cross-section and 0.25D thickness. A sketch of this configuration is shown in Figure 1.

Three mechanisms of heat transfer were considered: conduction, convection, and radiation. During the early stage of sintering, which is the area of interest (<700°C), radiation was considered negligible. Previous studies have demonstrated that convection may be ignored for porous media with small pore size. The remaining mechanism, conduction, was considered as a one-dimensional flow, which may be modeled as a circuit of thermal resistors, analogous to an electrical circuit. Each segment of the model has the resistance described by Equation (3) in the previous section:

\[ R = \frac{L}{\kappa A} \]  

where L is the length of the segment, A is the cross-sectional area, and \( \kappa \) is the thermal conductivity of the material. The neck-interface-neck portion of the model was a series of resistors, which are added together:

\[ R_1 = R_N + R_I + R_N \]  

The remainder of the unit cell is divided into halves represented as the shoulder-pore-shoulder structure, modeled as two identical series circuits:

![Figure 1. Thermal circuit model of two particles in contact](image-url)
The effective thermal resistance of the unit cell is shown as the three summed resistors in parallel, shown as:

$$\frac{1}{R_{\text{eff}}} = \frac{1}{R_1} + \frac{1}{R_2} + \frac{1}{R_3}$$  \hspace{1cm} (11)

Oxides, surface contaminants, and surface roughness will be modeled as a layer of glass with a thickness of 1% of the particle diameter at the beginning of the modeling cycle, shrinking to the width of a grain boundary at higher temperatures. Thermal conductivity of gas is a function of atomic size, pressure, and temperature. Pressure will be held as a constant at 1 bar (atmospheric), and conductivity for a given process gas is solely a function of temperature.

Many of the strength models traditionally focus on higher density, lower porosity materials, since these materials are the most common in application. At lower densities, which better represent the early sintering of an injection-molded material, the size of the interparticle bonds dominates the strength of a sintered material. A strength model incorporating this dependence has been described by Xu and German23 as:

$$\sigma_s = \frac{\sigma_o V_s N_c}{K \pi} \left( \frac{X}{D} \right)^2$$  \hspace{1cm} (5)

where $\sigma_s$ is the wrought strength as a function of temperature, $V_s$ is a fractional density, $N_c$ is the coordination number, and $K$ is a stress concentration factor related to neck curvature. In considering this relation, $K$ may be approximated using a model of a tensile bar with circular cross-section and a U-shaped groove4, the curvature of which may described in terms of neck size ratio:

$$p = \frac{1}{8} \left( \frac{X}{D} \right)$$  \hspace{1cm} (6)

where $p$ is the radius of curvature of the groove. Handbook values24 for this configuration may be fitted to a curve and used in the strength model. Xu’s strength equation is appropriate for both room temperature and high temperature application, as long as thermal softening behavior of the bulk material is known, and demonstrates an increase in strength due to neck growth until thermal softening decreases the bulk material strength at high temperatures.

**EXPERIMENTAL PROCEDURES**

Nickel was chosen as a baseline system due to its stable nature—no solid phase changes, and resistance to persistent oxidation. The candidate powder, Novamet 4SP-10, is a carbonyl nickel powder with relatively low surface roughness (versus the spiky surface normally associated with carbonyl nickel) and spherical shape. The powder is screened for a narrow particle-size distribution--80% of the powder is between 4 and 11 micrometers, with a median particle size of 7 micrometers. A basic sintering profile of 10 °C/min heating rate to 1100 °C (80% of nickel’s absolute melting temperature) was used, with a reducing atmosphere of 100% hydrogen gas.

Sintering events were initially profiled with a dilatometry run on loose powder in an Anter Unitherm vertical pushrod dilatometer. The powder was poured into an alumina crucible, then tapped until no further settling was observed. A second dilatometry test was run on nickel powder tapped and presintered in the alumina crucible, then removed to obtain a more precise measurement of initial sample height.
To produce small batches of components of various shapes, a quasi-injection-molded shaping concept was employed. The nickel powder was mixed and lightly compacted on a Carver hand press with 2% by weight ethylene bis-stearamide (EBS, trade name Acrawax®) at very low pressures to a green density of approximately 55%. Transverse rupture tests were selected to evaluate strength evolution, since this technique accommodates even fragile specimens without generating damage due to pressure required to grip a tensile specimen. Transverse rupture bars (31.7 mm x 12.7 mm x 5-7 mm) were pressed, thermally debound and presintered (3 °C/min to 400 °C, 120 minute hold). The bars were then sintered and water quenched at 100 °C increments of the heating portion of the baseline sintering cycle (500-1100 °C). The same technique was also used to make disks for laser flash testing, 12.25 mm in diameter by 2-3 mm thick. Disks were presintered for thermal diffusivity analysis. Density of the quenched bars was determined using the Archimedes water immersion technique on a portion of the broken bar after destructive testing.

Transverse rupture tests were performed on a Sintech 20/D mechanical load frame with an MTS 1000 lb load cell in standard 3-point bend configuration. Test execution and data management were controlled using Testworks software. Crosshead movement was displacement-controlled at a constant rate of 3 mm/minute. Data acquisition was performed by a digital controller at 1000 Hz.

Thermal diffusivity measurements were made using the laser flash technique on an Anter Flashline 5000 system. Two 12 mm presintered samples were supported side by side in an alumina holder, encased in a Kanthal radiant furnace, one cycle with a flowing nitrogen atmosphere, and the second with flowing hydrogen. A programmed heating rate of 10 °C/minute was used, although the requirement for a stable temperature when the measurements are taken necessitated a much lower rate (<1 °C/min) as the furnace approached each target temperature. Measurements were made from room temperature to 600 °C, then back down to room temperature. At each of these setpoints, an Nd/glass laser projected a beam through a quartz window into the furnace, illuminating the top face of the 12 mm sample disk. A Nb/Sb infrared sensor viewed the bottom face through a sapphire window and mirror relay, recording the equilibrium temperature, and the subsequent temperature rise at a rate of 15 kHz. Three data points were taken on each sample at each temperature increment.

EXPERIMENTAL RESULTS

Experimental and predicted linear shrinkage results for the loose powder during a 10 C/min constant heating rate are shown in Figure 2. An in-house initial stage sintering software, SintWin25, was programmed with the powder and material characteristics, shown in Table 2. The current version of SintWin generates a neck size ratio resulting from all sintering mechanisms active under the given sintering conditions, so a direct conversion to shrinkage via the relationship:

$$\frac{\Delta L}{L_o} = \left[ \frac{X}{2D} \right]^2$$

is not totally accurate. For example, surface diffusion is active at lower sintering temperatures, but neck growth due to surface diffusion will not result in shrinkage. However, as a starting point, this comparison provided an interesting baseline. The dilatometry results indicated that densification was active as low as 400 °C, and that significant densification (> 3%) was occurring by 700 °C. Based on these observations, 400 °C was chosen as a presintering temperature, with the intent of just barely establishing an interparticle bond to allow handling but still give visibility into the initial development of the bond, and the strength evolution prior to significant densification. The upper limit on the primary range of interest for this study, therefore, was 700 °C.
Table 2. Sintering parameters for SintWin computer simulation for 4SP-10 Ni powder²⁶.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Theoretical Density (kg/m³)</td>
<td>8.9</td>
</tr>
<tr>
<td>Atomic Volume (m³)</td>
<td>8.18 E-30</td>
</tr>
<tr>
<td>Surface Diffusion Frequency Factor (m²/s)</td>
<td>0.02</td>
</tr>
<tr>
<td>Activation Energy for Surface Diffusion (kJ/mol)</td>
<td>164</td>
</tr>
<tr>
<td>Volume Diffusion Frequency Factor (m²/s)</td>
<td>0.00014</td>
</tr>
<tr>
<td>Activation Energy for Volume Diffusion (kJ/mol)</td>
<td>276</td>
</tr>
<tr>
<td>Grain Boundary Diffusion Frequency Factor (m³/s)</td>
<td>4 E-16</td>
</tr>
<tr>
<td>Activation Energy for Grain Boundary Diffusion (kJ/mol)</td>
<td>108</td>
</tr>
<tr>
<td>Activation Energy for Evaporation (kJ/mol)</td>
<td>375</td>
</tr>
<tr>
<td>Surface Energy (J/m²)</td>
<td>1.86</td>
</tr>
<tr>
<td>Particle Diameter (micrometer)</td>
<td>7</td>
</tr>
<tr>
<td>Green Density (kg/m³)</td>
<td>4895</td>
</tr>
</tbody>
</table>

Strength evolution versus peak sintering temperature, along with predicted strength using Xu’s model with neck growth predicted by SintWin, are shown in Figure 3. Xu’s model predicts ultimate tensile strength. Since transverse rupture strength is typically 1.6 to 2.1 times greater than the ultimate tensile strength of a P/M material²⁰, the model’s results were multiplied by a factor of 2. At lower temperatures (<800 °C), the material behaved in a brittle manner, deforming elastically until sudden failure. At higher sintering temperatures, excessive plastic deformation occurred prior to failure, indicating that the elastic assumptions made in the test development were certainly violated, and resulting values likely reflect lower strength than the true values. Samples sintered at these temperatures exhibited enough toughness to withstand the grip pressures of a tensile test configuration, which would be a more accurate measurement of stress-strain conditions for a plastically deforming material.
Figure 3. Room temperature transverse rupture strength after constant heating rate of 10°C/min in H2, water quench.

Figure 4. In situ thermal diffusivity for quasi-PIM presintered Ni disks in argon atmosphere.

In-situ thermal diffusivity versus test temperature results, along with bulk material behavior, are shown in Figure 4. The porous material follows the same general trend as the bulk material—showing decreasing thermal diffusivity with the initial increase in temperature, followed by a slight increase above the Curie point (358 °C). These results also show a higher thermal diffusivity in the cooling cycle versus the heating cycle, indicating a possible increase in sinter bonding.
Figure 5 shows the same results normalized by the bulk material behavior, effectively removing the influence of the material properties. These results indicate relatively no change in diffusivity to 400 °C (the pre-sintering temperature) and a large increase from 400 to 600 °C. During the cooling cycle, a small increase in thermal diffusivity from 600 to 400 °C may indicate a continued increase in sinter bonding in this temperature range, followed by an unchanged state below 400 °C. A comparison of measured results versus the thermal circuit model is shown in Figure 6. The thermal circuit model under-predicts the material behavior initially, then over-estimates diffusivity during the cooling cycle.

Figure 5. Normalized in situ thermal diffusivity during heating/cooling cycle

Figure 6. Calculated versus measured in situ thermal diffusivity
DISCUSSION

The ability to effectively tie together strength and thermal diffusivity/conductivity depends on the ability to understand and model the nature of the interface—cross-sectional area, interface quality, number of contacts, etc. It is useful to use literature-based values for sintering parameters in the initial stage sintering equations/simulations, but the practical application of the simulation to a particular system often necessitates some iterative tailoring to match the observed behavior of the system. Initially, the large discrepancy between predicted and actual linear shrinkage indicated that sintering was occurring more readily than the literature based parameters would imply. A qualitative assessment of bond formation via scanning electron microscope images was performed, with sample images illustrating the ruptured bonds at each temperature shown in Figure 7.

![Figure 7. SEM images of as received powder and fracture surfaces showing ruptured bonds; samples sintered at 10 °C/min at a constant heating rate in 100% flowing hydrogen, and quenched in water.](image)

These images reinforced the indication that sintering was active in this system at lower temperatures than those implied by the literature-based activation energies. Small modifications to surface diffusion activation energies (lowering the values from 164 kJ/mol to 145 kJ/mol) more accurately reflected the neck growth observed. This modified simulation is the input into Xu’s strength model and the thermal circuit model as shown in Figures 3 and 6. Modifications to grain boundary and volume diffusion parameters resulted in no change, indicating that surface diffusion is the dominant mechanism for this system at the temperature region of interest.

The comparison of predicted versus measured thermal diffusivity clearly indicates that further refinement of the physical model is needed. However, it is interesting to note that contrary to Howard and Kozak’s calculations, no significant difference was observed between hydrogen and nitrogen atmospheres. One potential source of this discrepancy may lie in the experimental procedure. Typically, the sample chamber is evacuated, then backfilled with an inert gas, prior to the beginning of hydrogen flow. This
would fill the pores with the inert gas (nitrogen), which would only be replaced with hydrogen by diffusion acting to ease a concentration gradient. Alternative procedures are under investigation.

The normalized behavior shown in Figure 5 clearly shows the activation of sintering above the presintered condition (400 °C), and the cessation of sintering during the cooling portion of the cycle. This result demonstrates the capability of this technique to provide insight into early stage sintering, but the transition between powder and presintered material is still of key importance. Currently, powders and liquids are tested via laser flash in sapphire crucibles. This avenue of testing would potentially complete the data assessment between loose powder and pre-sintered material, and will be evaluated in future research.

CONCLUSIONS

The initial exploration of the use of laser flash analysis to directly assess thermal diffusivity and indirectly assess strength evolution during sintering follows general predictions based on neck growth, but indicates a need for significant refinement of the strength and thermal models as applied to this system. Further investigation of microstructural evolution as well as thermal and mechanical properties will be required to support the hypothesis of inferring in situ strength from thermal diffusivity testing. If the hypothesis can be successfully demonstrated, the results will have a two-fold benefit. Firstly, the prediction of in situ strength can help avoid damage during sintering, but also the thermal conductivity values themselves are key inputs in constitutive modeling to better predict and control final shapes and properties of a sintered component.
References

22. J. S. Agapiou and M. F. DeVries, “An Experimental Determination of the Thermal Conductivity of a


