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Numerical and experimental modeling of multiple pass radial forging of Alloy 718

Domblesky, Joseph Paul, Ph.D.
The Ohio State University, 1994
NUMERICAL AND EXPERIMENTAL MODELING OF MULTIPLE PASS
RADIAL FORGING OF ALLOY 718

Dissertation

Presented in Partial Fulfillment of the Requirements for the Degree
Doctor of Philosophy in the Graduate School of The Ohio State
University

By

Joseph P. Domblesky, B.S., M.S.

* * * * *

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1994

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To My Wife and Son
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CHAPTER I

INTRODUCTION

Superalloys are generally considered to comprise a group of Ni-based alloys which were developed specifically for use at elevated temperatures where high strength and corrosion resistance are required. The term superalloy was first used after World War II to describe alloys which were primarily developed for use in aircraft turbine engines (Bradley, 1988). Since their inception, a number of superalloys have been developed for a variety of applications. In terms of commercial importance, Alloy 718 remains the most widely used and at present accounts for almost 50% of the total superalloy tonnage produced in North America. Aircraft turbine engines comprise the single largest area of application for Alloy 718, especially as a disk material in the combustor and exhaust stages. Alloy 718 is a nickel-iron based alloy which is employed at temperatures up to 650 degrees Celsius. Typical chemical composition for Alloy 718 is listed in Table 1.1

Table 1.1. Nominal chemical composition of Alloy 718 in weight percent (after Muzyka, 1979).

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Fe</th>
<th>Ti</th>
<th>Al</th>
<th>B</th>
<th>Mo</th>
<th>Ob</th>
</tr>
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<tr>
<td>0.04</td>
<td>19.0</td>
<td>52.5</td>
<td>18.71</td>
<td>0.90</td>
<td>0.50</td>
<td>0.005</td>
<td>3.05</td>
<td>5.30</td>
</tr>
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1.1 Alloy 718 Processing and Microstructure

Alloy 718 has an austenitic FCC matrix which contains a number of secondary phases to improve mechanical properties at elevated temperatures. Alloy 718 is strengthened primarily by precipitation hardening ($\gamma'$, $\gamma''$) and to a limited extent by solid solution strengthening. While gamma prime ($\gamma'$) is present, the main source of precipitation hardening is due to gamma double prime ($\gamma''$) as a result of high Cb content. Gamma double prime is a body centered tetragonal metastable phase which strengthens the matrix via coherency strains. Delta ($\delta$) is an orthorhombic metastable phase (Ni$_3$Cb) which precipitates along the grain boundaries where it acts to pin the boundaries and inhibit grain boundary migration at elevated temperatures for increased creep strength. The phase-solvus temperatures for $\gamma''$ and $\delta$ as a function of weight percent Cb is shown in Figure 1.1. Depending on the microstructure to be produced, Alloy 718 is generally forged between 954-1066 degrees Celsius. Forging is typically performed in the $\gamma'' + \delta$ two phase or the $\delta$ single phase fields. Due to the strengthening characteristics of $\gamma''$, forging below the $\gamma''$ solvus may result in excessive forging loads.
Because, Alloy 718 is heavily alloyed and was specifically developed for high-temperature strength, it has a narrow processing window and is difficult to process. In order to improve the material uniformity and mechanical properties, it is necessary to destroy the cast structure and form a new, refined microstructure through hot-working. Such hot-working processes which are designed to reduce ingot or billet to standard mill products such as plate, sheet, or bar are called primary working processes (Dieter, 1986). By subjecting the cast ingot to a combination of elevated temperature ($T > 0.5 \ T_m$) and plastic deformation, atomic diffusion becomes significant and the
microstructure is transformed through recovery and recrystallization. Wrought billets generally possess superior dynamic fracture properties at temperatures up to 535 degrees Celsius over cast material of the same composition (Bradley, 1988). The wrought microstructure generally possesses superior performance and workability over cast structures as shown in Figure 1.2. Primary processing of Alloy 718 is typically performed using a two step sequence: forging and post-forging heat treatment. During deformation, the billet microstructure, and hence mechanical properties, are defined by the simultaneous application of deformation and elevated temperature. In the post-forging heat treatment, the billet is solution treated and aged to increase the strength through precipitation hardening.

![Figure 1.2. Comparison of formability for as-cast and wrought microstructures for PK 31 nickel alloy (after Turner, 1984).](image-url)
1.1.1 Alloy 718 Processing Methods

There are three basic processing approaches which are used to forge Alloy 718 depending on the microstructure, and hence properties which are required. The three approaches are termed standard, high-strength, and direct aged (DA) processes (Barker et al., 1985). The primary difference in properties is the result of the grain size and microstructure which is produced due to the different forging temperatures employed. All of these methods rely on the presence or absence of stable δ phase for grain size control. Typical grain sizes produced are ASTM 4-6 for the standard processed Alloy 718, whereas average grain sizes of ASTM 8 and 10 are produced in high-strength and DA processed Alloy 718 respectively (Barker et al., 1985).

1.2 Issues in Primary Processing

The principal microstructural variables which are of primary concern to the forger are grain size and shape. This is due to the major influence that grain size has on static and dynamic mechanical properties. Due to the wide variety of microstructures which may be produced in Alloy 718, it is imperative to develop an optimum thermomechanical processing sequence which will not only achieve the desired microstructure and properties but also be the most economical in terms of manufacturing cost.
At the present time, the design of processing schedules for Alloy 718 remains empirical and experience based. In order to obtain the desired billet microstructure, typical forging practice is to run a series of full-scale production trials until a suitable pass schedule is found. There are number of problems inherent in such an approach. The first is that this is a very inefficient, and expensive procedure. A second problem is that such an approach does not permit a great deal of manufacturing flexibility. A third problem is that due to the development of highly advanced design techniques, the performance limits of Alloy 718 are rapidly being approached (Loria, 1991). This necessitates that increasingly tighter process controls be imposed and highly sophisticated processing techniques be used to meet design specifications.

To solve the above problems, there has been a strong desire in the forging industry to move towards a more scientifically based methodology to understand the relationships between processing and microstructure (Howson and Couts, 1989; Malas et al., 1981). This change has been mainly driven by the high materials and processing costs associated with superalloys where there is a strong impetus to "get it right the first time". This necessitates having a means for analyzing and optimizing the process to predict the process response and product properties prior to forging.

Because deformation and temperatures during incremental deformation are generally non-uniform, the potential for inhomogeneity in the refined microstructure is high. To control the process, there are two possible
strategies which may be selected. The first is to monitor variations in the workpiece or process parameters and adjust the process control variables when an out-of-bounds condition is encountered. However, monitoring is a reactive response and internal conditions can not easily be checked in the billet during forging. Therefore, most efforts have been focused on mathematically modeling the forging process. This permits a more predictive approach to be taken and enables a better understanding of the process.

In order to predict microstructure, it is necessary to develop a new approach which integrates principles of materials science and continuum mechanics to link the microstructure and the process parameters. While some work has been performed in analyzing the primary processing stage (Isogawa et al., 1990; Boyko et al., 1991; Jackman et al., 1992), most research efforts have been aimed at controlling and predicting the thermomechanical history of the billet during forging. In terms of process-property relations, very little work has been done to study the development of microstructure during multiple pass primary processing in processes such as radial forging.

1.3 Approach

The main objective of this study is to develop and validate a numerical and experimentally based method which can be used to assist the process engineer in designing and optimizing radial forging schedules to achieve a
predetermined grain size in the as-forged billet. In order to achieve this, the following procedure was followed:

1. Develop a model to predict thermomechanical history of the billet during multiple pass radial forging.

2. Validate the strain, strain rate, and temperature predictions of the process model by experiments on a radial forging machine.

3. Develop quantitative relationships for as-forged grain size as a function of primary process parameters by simple, controlled upset experiments and their simulation.

4. Apply these relationships to the radial forging process in order to assist the forging engineer in process planning.

1.3.1 Validation of the Numerical Model

Prior to using the numerical model to predict the thermomechanical history, it is necessary to verify that the predictions are sufficiently accurate. An axisymmetric deformation model based on the finite element method (FEM) is normally preferred to predict strain, strain rate, and temperature in radial forging of superalloys due to its ability to consider metal flow in three directions. To validate the FEM model's predictions, internal and surface temperatures, metal flow, and forging loads were measured during a
controlled forging trial performed at an industrial facility and will be compared to the FEM predictions.

1.3.2 Procedure for the Simulation of Multiple Pass Forging

The most practical way to determine how a workpiece will react to a proposed thermomechanical path is to simulate the proposed deformation and heating schedule using a representative sample of the material in a laboratory test procedure. Such an approach is often used to evaluate material hot-workability. While a variety of tests and testing methods are available to perform this type of physical simulation, axial compression testing most closely approximates the stress-state and deformation rates experienced by the material during radial forging. The compression test also enables large strains to be achieved without inducing material instabilities such as necking.

To generate the microstructural data which was used to develop the metallurgical model, compression tests were performed using representative material from billet slices. To perform the compression tests, a statistically designed, full factorial study will be carried out. The effect of strain per pass, temperature, time per pass, and number of passes was considered. These parameters were chosen based on the fact that they are specified and controlled by the pass designer.
1.3.3 Model Building

To develop quantitative relationships between the as-forged grain size and forging parameters, multi-variate regression techniques were used initially. Such techniques enabled quantitative relationships to be developed between a dependent variable and one or more independent variables. From the resultant equations, mathematical predictions regarding the response of the independent variable were made provided that the orthogonality (independence) requirement was met.

1.4 Organization of the Dissertation

In this dissertation, the work performed in developing and validating the combined method is presented in detail. Development of the numerical and metallurgical models are presented along with the results obtained.

Chapter 2 is a survey of the literature pertaining to FEM modeling of the radial forging process. Chapter 3 presents a survey of the approaches which have been proposed in the literature to predict process-property relations in forging. Chapter 4 details the development of the FEM process model used to simulate multiple pass radial forging sequences in this study. Validation of the process model and the experimental techniques used to measure strain and temperatures are presented in Chapter 5. Development of the metallurgical model and experimental simulation methodology is presented in Chapter 6. The development of the metallurgical model for predicting as-
forged grain size is described in Chapter 7. Discussion and validation of the process-property behavior is described in Chapter 8. Future work and a summary of the results are presented in Chapter 9.
CHAPTER II

RADIAL FORGING TECHNOLOGY REVIEW

2.1 Principles of the Radial Forging Process

Radial forging which is often referred to as, and confused with, rotary forging is a relatively new forging technology. The term rotary forging is employed due to the fact that in some operations the workpiece is periodically rotated during the forging operation. Radial forging, appears to be widely used in Europe to describe the process while both radial and rotary forging are commonly used terms in the United States. The basic configuration of the radial forging machine was initially developed by Dr. Bruno Kralowetz in Steyr, Austria in 1946 (Hojas, 1988). However, due to the limitations of control technology, it was not until 1960 that the first working model of a four hammer machine was introduced commercially.

A schematic showing the primary components of a radial forging machine is shown in Figure 2.1. The basic premise of the machine is that four radially arranged dies simultaneously press the workpiece surface. Radial forging is essentially an open-die forging process which uses two, instead of one, pairs
of opposing dies. The primary features of the machine are the vertical forging box containing the forging tools and two synchronous chuckheads, which are track bound manipulators, that act to center and index the workpiece during forging. During forging, movement of the chuckheads is intermittent and synchronized with that of the dies to prevent twisting of the workpiece. During the time that the dies are in contact with the workpiece, the chuckheads remain stationary. When forging round products, the workpiece is rotated and axially indexed in a simultaneous motion between blows to produce a uniform forged surface as depicted in Figure 2.2.

The radial forging machine is essentially a mechanically driven short stroke press where rotary motion is transferred into reciprocating linear motion of the forging tools. Each die is mounted on a connecting rod which is driven by an adjustable eccentric shaft which enables variable stroke lengths. The eccentric shafts are driven via gears which synchronize the die travel. Both round and rectangular cross-sections may be forged as die travel may be adjusted either in sets or in unison via gears in the forging box.

A distinct characteristic of the radial forging process is that it has a very high stroking rate due to the fact that the hammers are spaced very close to the workpiece; requiring only a short period of time between subsequent strokes. It is important to make the distinction that the dies actually press on the workpiece though it would appear that with the high stroking rate that the dies actually operate as hammers. However, the deformation rates in radial
forging are less than $254.0 \text{ mm/mm/sec.}$ which is within the range typically encountered for mechanical presses.

The advantage of using four hammers is that the probability of edge cracks developing from secondary tensile stresses is quite small as the surface is under a compressive state of stress. The chief mode of deformation in radial forging is compression accompanied by considerable spreading in the axial direction. Due to the opposing motion of the hammers, no forces are transmitted to the machine itself. Therefore, the machine foundation remains virtually free from stress and vibration (Hojas, 1976). Because the workpiece is incrementally deformed, only a small portion of the workpiece is being worked at one time and the forging forces are relatively low.

Heat loss from the workpiece is minimized and in many cases, with proper selection of process parameters, the temperature may increase from deformation as the hammers contact the workpiece between 150-200 times per minute. This is advantageous as it eliminates the need to reheat the billet between passes. The need for die changes is also eliminated with the radial forging machine as the amount of reduction may be varied by changing the stroke length of the hammers. On many machines this is accomplished by simply reprogramming the control software. Another advantage of the radial forging process is that the internal soundness of cast billets is improved and internal voids are closed due to the compressive hydrostatic stresses which are generated.
Figure 2.1. Schematic of a four die radial forging machine with a mechanical drive.
Figure 2.2. Kinematics of the radial forging process.
One common feature of most radial forging machines is that integral, automatic stock handling by robotic manipulators is incorporated into the machine to provide coordinated rotation and feed of the stock during forging. For the case where the workpiece must be forged over its entire length in one heat, two chuckheads are normally used (Hojas, 1976). When two manipulators are used, one manipulator serves to feed the workpiece while the second manipulator holds the workpiece and maintains centering. The stock manipulators serve to provide both longitudinal and rotational movement to the workpiece between strokes as described above. An advantage of radial forging is that the manipulators constantly maintain the workpiece center position coincident with the center of the forging box regardless of the billet cross-section and the reduction being taken.

Many radial forging installations have completely automated forging, heat treatment, and automatic stock handling integrated into flexible manufacturing systems (FMS). An example of such an installation is the in-line operation at Carpenter Technology's forge plant in Reading, Pennsylvania which is depicted in Figure 2.3. Another example of a fully automated and integrated radial forging line is located at the U.S. Army's Watervliet Arsenal in Watervliet, New York.
Figure 2.3. Layout of the flexible forging cell for radial forging at Carpenter Technology's forge plant in Reading, Pennsylvania.
2.2 Variables in Radial Forging

A list of significant process variables which influence the radial forging process is shown in Table 2.1. The effect of the most significant process variables which must be considered in pass design and their effect on microstructural development of Alloy 718 are reviewed below. However, it must be noted that due to the interaction of a large number of variables, it is necessary to consider their combined effect when designing a forging pass sequence. While the radial forging process is routinely used to convert billet, there is very little available data which describes microstructural features based on process parameters in the hot-working regime (Livesay and Sellars, 1983). This is in contrast to the case for recrystallization of cold-worked metals where the effects of strain and temperature are qualitatively summarized in the six laws proposed by Burke and Turnbull in 1952 (Cotterill and Mould, 1976).
Table 2.1. Significant variables in the radial forging process.

<table>
<thead>
<tr>
<th>Category</th>
<th>Variables</th>
</tr>
</thead>
<tbody>
<tr>
<td>1. Material Variables:</td>
<td>• Material Flow Stress</td>
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<tr>
<td></td>
<td>• Starting Microstructure</td>
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<tr>
<td></td>
<td>• Chemistry</td>
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<td></td>
<td>• Material Physical Properties</td>
</tr>
<tr>
<td>2. Machine and Tooling Variables</td>
<td>• Die Geometry</td>
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<tr>
<td></td>
<td>• Die Material and Properties</td>
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<td></td>
<td>• Mandrel Properties and Design (Tube Forging Only)</td>
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<td></td>
<td>• Use of Lubricant on Mandrel (Tube Forging Only)</td>
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<td></td>
<td>• Stroking Rate</td>
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<td></td>
<td>• Length of Stroke</td>
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<td></td>
<td>• Rated Machine Tonnage</td>
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<td></td>
<td>• Axial and Rotational Feed Rates</td>
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<tr>
<td>3. Process Variables</td>
<td>• Billet Temperature</td>
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<tr>
<td></td>
<td>• Number of Forging Passes</td>
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<td></td>
<td>• Reduction Taken per Pass</td>
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<tr>
<td></td>
<td>• Feed Rates (Axial and Rotational)</td>
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<td></td>
<td>• Quenching Media</td>
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<td></td>
<td>• Furnace Homogenization</td>
</tr>
<tr>
<td></td>
<td>• Holding Time</td>
</tr>
</tbody>
</table>
2.2.1 Machine Variables

In practice, many of these variables are built in on the machine by the builder and must be considered as fixed. Among the forging machine variables which need to be considered are: allowable forging tonnage, die velocity (as a function of stroking rate), and die geometry. All of these have an indirect effect in influencing the as-forged microstructure due to their effect on strain, strain rate, and temperature in the billet. While these variables are fixed, they do differ as a function of machine size and from manufacturer to manufacturer.

The maximum forging load will affect the amount of deformation which may be taken per pass. For a machine with a low rated load capacity, the amount of deformation is limited and may not reach the workpiece core on larger workpieces. This ultimately affects the maximum strain which can be imposed during each pass and its uniformity.

Die velocity as a function of forging stroke will affect the strain rate experienced by the workpiece material and hence the recrystallization behavior. The die velocity for each forging tool will follow that of a mechanical press as a harmonic or sinusoidal function as given by equation (2.1):

\[
V = \frac{\pi nh}{30} \sqrt{\frac{s}{h} - 1}
\]  

(2.1)
where: 

\[ V = \text{instantaneous die velocity (mm/sec.)} \]

\[ n = \text{machine stroking rate (strokes/min.)} \]

\[ h = \text{distance from Top Dead Center (TDC)} \]

\[ s = \text{total tool stroke (38.1 mm)} \]

A typical stroke-velocity curve for mechanically driven radial forging machines operating at 160, 180, 200, and 220 strokes per minute are shown in Figure 2.4. From Figure 2.4, it may be seen that the velocity is continuously decreasing during contact with the billet. The hit of a mechanical press is more like a squeeze than a hammer blow and thus prevents deformation from being concentrated in the material surface layers (Dieter, 1986).
The die velocity will affect the average strain rate as given by the ratio of $V/d$ where $V$ is the die velocity and $d$ is the diameter of the deforming material. Depending on the strain rate imposed, this will affect the critical strain at which dynamic recrystallization (DRX) occurs. The strain rate affects the DRX grain size through the Zener-Holloman parameter, $Z$. 

Figure 2.4. Comparison of stroke-velocity curves for a mechanically driven radial forging machine operating at 160, 180, 200, and 220 strokes per minute with an effective stroke of 38.1 mm.
The DRX grain size decreases with an increase in $Z$ (McQueen et al., 1983). With increased strain rate, the material will achieve a smaller steady-state grain size. This is reflected by the dependence of equilibrium or steady-state grain size, $D_s$, to the Zener-Holloman parameter, $Z$:

$$D_s^{-1} = k \log Z$$ (2.3)

However, conflicting results were reported by Mosser et al. (1989) who found that at a temperature of 1050 degrees Celsius there was no apparent dependence of recrystallized grain size with strain rate over typical forging ranges for Alloy 718. Possible reasons for this deviation offered by the authors included: significant adiabatic heating for large strains and the possibility of metadynamic recrystallization during the delay in quenching.

The effect of die design has been shown to have a strong influence on forging loads in the radial forging process (Lahoti and Altan, 1980; Raghupathi et al., 1980-1; Lahoti and Altan, 1976-1). The most important aspect of die design as it relates to microstructural development is under the inlet angle where most deformation occurs in the forging zone. The die inlet angle will also greatly affect the deformation efficiency. In radial forging it is desirable to obtain uniform deformation throughout the workpiece to ensure that the forged product will have uniform mechanical properties and response to
subsequent heat treatment. If deformation is concentrated near the surface, a higher dislocation density will develop and recrystallization will tend to occur preferentially in that region.

In order to achieve uniform deformation, a small inlet angle is generally required. It has been shown analytically and experimentally that maximum penetration is achieved using a small inlet angle (Lahoti and Altan, 1980; Raghupathi et al., 1980-1; Lahoti and Altan, 1976-2). The effect of a large inlet angle will be to concentrate deformation near the surface. The reason for this may be understood by considering the ratio $r/L$ where $h$ is the preform radius and $L$ is the length of die contact. When the ratio of $r/L$ is $> 1$, deformation will not reach the workpiece core (inhomogeneous deformation occurs). Only when the ratio is less than or equal to 1 will deformation reach the core and will be more homogeneous (Schey, 1987). In radial forging it is obvious that $L$ will be less for a large inlet angle in comparison to a small inlet angle, all other factors held equal. The die design is usually fixed in practice by the machine manufacturer due to the rapid increase in tangential or axial die loads with increasing inlet angle which causes premature wear of the guide bushings (Zogas, 1991).

The die material used to construct the forging tools has a very limited impact on microstructural development. The influence of die material on tool wear is, however, very significant in terms of influencing the process economics.
2.2.2 Process Variables

This group of variables consists of the parameters which generally must be
selected by the forging engineer during pass design. These are the variables
which need to be optimized and studied using simulation techniques. This
includes axial and rotational feed rate, reduction per pass, number of passes,
and billet temperature.

Axial feed rate has been shown to be important in influencing the workpiece
temperature during forging (Hojas, 1988; Lahoti and Altan, 1980; Raghupathi
et al., 1980-1; Lahoti and Altan, 1976-2; Domblesky et al., 1993) and must be
considered as a variable in the model. The effect of increasing feed rate is to
impose greater deformation per unit time, increasing the workpiece
temperature through deformation heating. Because deformation takes place
much faster than heat transfer, the adiabatic temperature rise due to the
conversion of mechanical energy to heat is calculated using the following
relation:

$$\Delta T_D = \frac{\beta \sigma \dot{\varepsilon} \Delta t}{Jc \rho}$$

(2.4)

where:

- $\beta$ = Proportion of Mechanical Energy Converted to Heat
- $\sigma$ = Effective Flow Stress
- $\dot{\varepsilon}$ = Effective Strain Rate
- $\Delta t$ = Time Increment for Deformation
- $J$ = Conversion Factor for Mechanical to Heat Energy
\[ c = \text{Heat Capacity} \]
\[ \rho = \text{Material Density} \]

For an increased axial feed rate, the strain rate, \( \dot{\varepsilon} \), will increase (Domblesky et al., 1993), all other factors being held constant. Therefore, it is apparent from equation (2.4) that the increase in temperature should be a linear function of strain rate which is itself a function of axial feed rate. As a result, the workpiece temperature will be controlled by the axial feed rate and the effect of surface chilling offset by deformation heating. If the proper feed rate is chosen, the workpiece surface temperature has been observed to be almost "isothermal" during forging (Raghupathi et al., 1980-1; Marsh, 1982) implying that the bulk workpiece temperature remains essentially constant during forging.

While the effect of rotational feed has not been considered in any study to date, it is highly probable that it will also affect temperature and deformation during forging in a manner similar to the axial feed rate. This is due to the fact it also controls the amount of material being deformed during each stroke and hence influences strain distribution. The rotational feed rate is typically fixed by the machine builder. For GFM radial forging machines, the rotational feed is fixed at 15 degrees per stroke.

The reduction or strain imposed per pass is critical for developing the as-forged microstructure. While the maximum strain per pass is limited by the rated load capacity of the machine, reductions of 10 to 25% in cross-sectional
area are commonly used. If insufficient strain is imposed during a pass, then
the material will not be deformed sufficiently to recrystallize and the
material will undergo recovery and grain growth. For moderate amounts of
reduction, a duplex or necklaced structure will form. For heavy reductions,
the material will undergo full recrystallization. While it is recognized that
increased strain will result in a more highly refined grain size, very little
quantitative information is available in the literature on this.

The billet temperature will have a significant impact on the as-forged
microstructure and billet formability. Generally, a decrease in the working
temperature cause the critical strain at which DRX commences to increase
and the resulting grain size to decrease. This is thought to be due to a higher,
more uniform dislocation density and increased strain energy (Cotterill and
Mould, 1976). The billet temperature will also determine the rate at which
the transformation kinetics would occur. Depending on the forging
temperature, δ phase precipitates may be present to pin the grain boundaries
and limit the maximum grain size. If the temperature is too low, the billet
will be difficult to forge due to the presence of γ" which is a precipitation
strengthenener. It has been reported in the literature that an increase in the
working or forging temperature will increase the DRX grain size (Guimares
and Jonas, 1981). This may also be seen by the temperature term in the Zener-
Holloman parameter.

The preform dimensions while being purely geometric quantities,
nonetheless can influence the deformation efficiency. The main factors to be
considered are the billet diameter and length. The billet diameter will determine whether deformation is homogeneous or non-homogeneous depending on the tool contact length as described above. Development of a deformation gradient in the billet may lead to a distinct grain size gradient (Mataya and Matlock, 1989). The billet length will also affect the total time required for each pass for a given feed rate. This will control the elapsed time between successive deformation for a material point and determine the effect of static transformation processes.

2.2.3 Workpiece Variables

Under material properties there are a number of variables which must be considered as they directly impact the workpiece response to forging. Material composition; flow stress as a function of temperature, strain, and strain rate; homogeneity of chemistry and microstructure; ductility; melting and solutioning temperatures; surface properties (scaling); and thermal properties must all be considered when the goal is to optimize any microstructural feature.

The effect of material composition is what determines the materials response to forging conditions. In essence, this is what sets the material's behavior in processing and service. The effect of varied chemical compositions over the workpiece is a result of non-equilibrium solidification in a casting. This compositional gradient causes different regions of the workpiece to react differently to the same set of deformation conditions. To eliminate this effect
in forging, the casting is usually subjected to a pre-forging anneal which allows the compositional gradients to be eliminated by diffusion processes. Normally the workpiece is considered to be of a homogeneous chemical composition during process modeling.

The surface finish will affect the friction conditions and thermal losses at the workpiece surface. During heating, an oxide coating or "scale" generally forms on the billet surface. Formation of scale tends to increase emissivity and hence, radiation losses, at hot-working temperatures. The presence of scale also acts to increase the friction factor and may affect deformation under certain process conditions.

Thermal properties of the material have a major effect on heat transfer in the workpiece and losses to the environment. This will significantly influence the degree of thermal, and hence microstructural, gradients in the as-forged billet. The two most significant modes of heat transfer in the workpiece are radiation to the environment and conduction in the workpiece. Due to the relatively low thermal conductivity of Alloy 718, it is likely that a significant thermal gradient along the radial direction will be maintained in the billet during forging which will lead to a variation in grain size. This has been observed in forged billet where a difference of 1 ASTM grain size has been observed between the surface and the center (Jackman et al., 1989).
The starting material or grain size will probably have some influence though very little has been written on this topic. The work by Ainak et al. (1988) in powder metallurgy of René 95 and Nielsen et al. (1993) in ring rolling of Alloy 718 indicate that this may not be a significant process variable. This is based on the experimental observation that the DRX of the as-forged microstructure is independent of the starting grain size (Ainak et al., 1988; Nielsen et al., 1993). This finding would indicate that deformation data carried out under a certain set of conditions would have a extended range of applications. This is in contrast to studies where iron samples have been deformed at room temperature and it is thought that there is a strong dependency of grain size to dislocation density which causes difference in recrystallized grain size (Cotterill and Mould, 1976).

2.3 Radial Forging Process Models

A number of models based on the mathematical theory of plasticity have been developed to analyze the radial forging process. However, most of these methods, such as the slab and slip line field, are better suited for stress analysis and are of limited utility in analyzing time dependent thermomechanical behavior. Because of the complex flow behavior and strong coupling between deformation and temperature, the resulting field equations which describe the system behavior are complex and must be solved numerically. For analyzing such problems, the FEM has proven to be a valuable tool in forging process analysis.
2.3.1 Lagrangian Based Process Models.

The most common approach to modeling the radial forging process has been based on updated Lagrangian based models due to the wide availability of such codes and their general acceptance by the metal forming community. When examined from a Lagrangian reference frame, the radial forging process may be seen to be a combination of steady-state and transient analyses. If the heat transfer to the tools is neglected, heat transfer boundary conditions are essentially independent of time and heat transfer is essentially reduced to a one-dimensional steady-state problem. Due to the cyclical contact by the forging tools, deformation will be transient in character.

The common strategy used by several workers to model the radial forging process was based on a "stroke-by-stroke" approach assuming a 2-D axisymmetric deformation model. The underlying principle used was to model each tool stroke as a separate FEM simulation. The workpiece is indexed by an amount equal to the axial feed rate between each simulation. Because a Lagrangian formulation is used, detailed information on strain, strain rate, and load as a function of stroke or time may be obtained. Due to the rotational symmetry of the problem, only one half of the workpiece is generally modeled to reduce computational requirements. Due to use of an two-dimensional deformation model, it is necessary to introduce the following assumptions which are implicit in the models presented:
1) Small clearances between each tool are neglected
2) Rotational feed component is neglected
3) Zero (fixed velocities) are prescribed at one end of the workpiece to model the chuckhead restraint.
4) Deformation is followed by a dwell period for heat transfer at the end of each pass

Isogawa et al. (1990) at Daido Steel Company in Japan have modeled the radial forging process of Alloy 718 using a plasticine model and a two-dimensional FEM program. The stated objective of their study was to build a base from which a model could be developed to predict the microstructure of Alloy 718 after multiple pass forging. Their main interest appeared to be in predicting the temperature distributions which developed during forging as a function of feed rate. This was intended as a precursor to developing a model incorporating recrystallization kinetics for microstructure prediction. The process variables in the model were axial feed, entry temperature of the billet, and percent area reduction. Verification was based on surface temperatures and as-forged microstructure from actual forging trials.

The conditions assumed by the authors in their simulations were that the process was axisymmetric, and that heat losses were due to radiation only. Friction effects between the tool and workpiece were ignored. Flow stress data at a constant strain rate of 680.7 mm/mm/sec. was used during the simulation. This assumed value appears to be somewhat higher than the strain rates typically found in radial forging. Work-hardening and strain rate
Based on their simulation study, they concluded that the temperature gradient between the workpiece surface and center was very sensitive to axial feed rate. For the two different feed rates simulated (2, 4 m/min.), the lower feed rate was predicted to decrease the temperature gradient between the surface and workpiece center whereas the higher feed rate caused the gradient to become steeper. This was qualitatively verified in forging trials by the presence of circumferential cracking which was predicted to occur at the higher feed rate due to the reduced formability at the higher temperature. This would indicate that the feed rate could be optimized in order to minimize the temperature gradient in the billet. This would result in more uniform conditions for recrystallization along the radial axis of the workpiece.

Based on the plasticine modeling tests, the authors reported that the axial strain, $\varepsilon_z$, was uniform as a function of radial position. The axial strain was found by:

$$\varepsilon_z = \ln \frac{\text{final element length}}{\text{initial element length}} \quad (2.5)$$

This is an interesting result as effective strain has been found to vary with radial position in FEM studies of radial forging (Boyko et al, 1991; Dombresky et al., 1994). However, their finding may be due to the billet geometry used.
Transverse shear strain measured over the plasticine billet cross-section was found to vary as a function of radial position (Figure 2.5). The transverse shear strain was initially zero at the surface and was found to increase to a maximum value at a distance which was approximately 30% of the workpiece radius ($r/R = 0.7$). After the maximum shear strain was reached, the shear strain was observed to drop rapidly to zero at $r/R = 0.5$ and remain at zero in the remainder of the billet. An interesting result the authors noted during forging trials was that the region of maximum shear strain corresponded to the region where circumferential cracking occurred during forging trials. Due to the reduced formability, the billet material could not sustain a high level of deformation and consequently failed.

Boyko et al. (1991) modeled the radial forging and open die processes to study the effect of process parameters on metal flow in Alloy 718. Temperature behavior was not considered in detail in their study. In their study DEFORM v1.2 was used to model transverse (plane-strain) and longitudinal (axisymmetric) sections. Among the parameters studied were die shape and curvature, friction factor, and bite ratio. The study appeared to have been based on the characteristics of the SMS-Hasenclever radial forging machine which is hydraulically driven. In their model, a 533 mm long workpiece and 178 mm long die were used to simulate deformation conditions. The axial feed rates used were generally on the order of the die length. While these dimensions will facilitate numerical modeling, they are not typical of the actual process dimensions and the results may not accurately reflect the physical process. In addition, the velocity-stroke profile for a hydraulically
driven machine such as the SMS-Hasenclever is totally different from that of a mechanically driven machine.

![Normalized Billet Radius (r/R) vs. Transverse Shear Strain](image)

Figure 2.5. Distribution of transverse shear strains in a rotary forged workpiece simulated by plasticine modeling. Based on data from the study by Isogawa et al. at Daido Steel (1990).

Results of their work showed that deformation was higher near the surface than at the center of the workpiece based on a plane strain simulation. However, it must be noted that due to the fact that area and volume constancy is preserved in a plane strain model, no axial or longitudinal flow is assumed to occur. A comparison of flat and convex shaped dies was also
studied and showed that the flat die tended to impart more deformation into the workpiece. For low bite ratios which was defined as:

\[
\text{Bite Ratio} = \frac{\text{axial feed per stroke}}{\text{workpiece radius}}
\]  

(2.6)

For a bite ratio of 0.24 their simulations showed that a significant change in the friction factor from 0.3 to 0.8 made very little difference in metal flow for the geometry modeled. For a larger bite ratio, the effect of changing the friction factor was also reported to have a minor effect. Based on their simulations they also found that the characteristic "cone" or "fishtail" generated at the end of the workpiece was affected by the bite ratio. Increasing the bite ratio tended to decrease the size of the cone. Chilling at the workpiece surface was attributed by the authors to cause increased deformation into the interior of the workpiece on the basis of plane strain simulations. This was attributed to the cooler surface material being more rigid than the hotter core material and in effect acting as an extension of the die.

2.3.2 Eulerian Based Process Models

A special purpose FEM code based on the Eulerian formulation was developed by Thompson et al. at Colorado State University (Thompson et al., 1992; Jackman et al, 1992). In using the Eulerian formulation, deformation is assumed to be equivalent to a fluid flow problem where attention is fixed on flow behavior within a simple boundary. The program appears to be limited
to analyzing round billets of constant cross-section with no restraint due to the chuckheads. The program does not appear to offer the capability of following metal flow, material spread, or load during the stroke.

The code is reported to be three-dimensional and de-couples the deformation and thermal analyses. Due to the fact that the radial forging process is inherently cyclical, a "cycle averaged" method is used. In cycle averaging, the process is modeled on a steady-state basis.

In order to use the steady-state formulation for deformation, the authors made the following assumptions regarding the process (Thompson et al., 1992; Jackman et al., 1992):

1) Deformation heating is equivalent during each forging stroke.
2) Amount of material deformed and heat generation is larger than the axial feed per stroke.

Based on these two assumptions, the authors concluded that due to the fact that no fluctuations were observed in surface temperature during forging that the above assumptions were valid. If the first assumption is false, large fluctuations in the surface temperature would result along the length of the billet during forging. If the second assumption is incorrect, then circumferential fluctuations in the surface temperature would be observed in the workpiece between strokes.
Due to the partitioning of the workpiece into sections for analysis purposes, the following assumptions must be used to ensure compatibility between the individual sections after modeling. The use of an average die velocity is due to the cycle averaging method used in the analysis (Thompson et al., 1992).

1) No flux takes or material flow occurs between individual sections
2) Plane sections remain plane during deformation
3) An "average" die velocity is used

The predicted results from the simulation of a small diameter billet generally agreed with previously reported results (Boyko et al., 1991; Domblesky et al., 1994) which showed that strain was maximum at the surface and minimum at the workpiece core. One interesting prediction was that the center of a small diameter billet was predicted to be strain-free after forging. This is not likely for a small diameter billet and may be the result of the model formulation.
CHAPTER III

REVIEW OF METHODS FOR PREDICTING MICROSTRUCTURE AND MATERIAL BEHAVIOR IN FORGING

A survey of the literature was undertaken to review the approaches which have been proposed to predict microstructure for various hot-working processes. Due to the complexity of metal forming processes and the need to consider local variations in deformation and temperature history, a number of the methods reviewed employ the finite element method as the basis for predicting the thermomechanical history of the workpiece. Because the entire workpiece history must be considered to predict microstructure, computer based FEM programs are thus well suited to this type of iterative analysis. All of the forging related methods found in the literature were for predicting the as-forged grain size and the main difference found was in the methodology used for predicting the microstructure after forging. A review of each method is given below.

3.1 Materials Based Methods

3.1.1 Dynamic Material Modeling

Dynamic Material Modeling or DMM was first proposed by Prasad et al. (1984) for predicting material behavior in thermomechanical terms and then extended
by Gegel et al. (1986) at Wright-Patterson Air Force Base to predict workability in forging. While DMM was developed for evaluating material workability rather than microstructural prediction, the method is useful as an illustration of how macro (continuum) and micro (atomistic) phenomena may be coupled for evaluating material response to processing conditions. An example of such a map developed by DMM for Ti-6242 is shown in Figure 3.1.

The foundation of the method is to quantitatively define how the material will dissipate the applied instantaneous power and how this will affect the flow behavior (Dwivedi and Balakrishnan, 1990). In the DMM method, material flow is considered in a macroscopic as well as a microscopic sense in developing a process map which delineates regions of stable and unstable flow. Stable flow regions are defined as regions where maximum energy is converted by the lattice into stored strain energy. Stable behavior also implies in this case that the energy applied is being dissipated at a rate equal to the external work being done. Unstable flow is the converse where fracture and flow instabilities such as shear bands occur.

In order to formulate/describe the material behavior using DMM, it is necessary to have the constitutive relation for the material. In the DMM Method, the instantaneous power per unit volume, $P$, absorbed by the workpiece during deformation is given by (Gegel et al., 1986):

$$P = \bar{\sigma} \dot{\varepsilon}$$  \hspace{1cm} (3.1)
Figure 3.1. Processing map developed using DMM methodology for Ti-6242 material (after Gegel, 1986).

The instantaneous power, \( P \), may be considered to be comprised of two components, \( G \) and \( J \) defined as follows (Gegel et al., 1986):

\[
P = G + J = \int \sigma \dot{\varepsilon} + \int \varepsilon \dot{\sigma}
\]  
(3.2)
It may be seen from equation (3.2) that G is simply the strain energy and J is its complement. G is assumed to be the power dissipated by plastic work while J is assumed to represent microstructural changes which dynamically disperse energy (Gegel et al., 1986). One of the key assumptions made in DMM is that fracture and instability processes are associated with G while microstructural effects which dissipate energy are associated with J (Dwivedi and Balakrishnan, 1990).

By formulating material constitutive equations as Liapunov functions, a criterion may be formulated describing the system's stability and by evaluating the following differentials (Gegel et al., 1986). It should be noted that it is not necessary to actually determine the exact solution to the differential equations. The Liapunov method enables system stability to be inferred directly from the equations without exact knowledge of the solution.

\[
\frac{\partial \eta}{\partial \log \dot{\varepsilon}} \quad \text{where} \quad \eta = \frac{2m}{m+1} \tag{3.3}
\]

\[
\frac{\partial S}{\partial \log \dot{\varepsilon}} \quad \text{where} \quad S = -\frac{1}{T} \left. \frac{\partial \log \bar{\sigma}}{\partial \frac{1}{T}} \right|_{\dot{\varepsilon}} \tag{3.4}
\]

Contour maps may be drawn based on equations (3.3) and (3.4) which show the most stable regions for processing. Stable behavior merely implies that the system fulfills the Liapunov stability requirements. It does not imply that the microstructure is evolving in the desired manner.
For Alloy 718 which has good formability, definition of stable and efficient processing regimes are of secondary importance to achieving a specified microstructure (Howson and Couts, 1989). If a suitable relationship could be found which would link workability or the stability parameters to microstructure, then perhaps the method could be extended for microstructural prediction. Due to the dependence of the DMM equations on $\eta$ and material constitutive relations, perhaps these parameters could be related to some microstructural feature such as grain size for a particular material.

### 3.1.2 Materials Modeling

Another materials based method called MATMOD was proposed by Miller (1976-1, 1976-2) and subsequently extended and modified by Miller and others (Lowe and Miller, 1984-1; Lowe and Miller, 1984-2; Lowe and Miller, 1986; Henshall and Miller, 1990). The underlying principle of MATMOD is to employ one unified constitutive equation to model the material behavior for a wide range of complex conditions and loading histories. While purported to be a materials model, it is heavily based on phenomenological behavior and much of the data needs to be generated from standard mechanical tests. The unified approach employed by Miller is based on a representation of the material's physical behavior by internal state variables representing the materials structure (Miller and Tanaka, 1988). The form of the unified equation is:

$$
\varepsilon = f \left( \frac{\sigma - R}{D} \right)
$$

(3.5)
where $R$ and $D$ are the history variables for the material. The resulting form of the function is as follows:

$$
\dot{\varepsilon} = B\theta \left[ \sinh \left( \left( \frac{|\sigma - R|}{D} \right)^{1.5} \right) \right]^n \text{sgn}(\sigma - R)
$$

where $R$ and $D$ are formulated as time derivatives which are evaluated and integrated at each time step of the designated loading history. The influence and basis on creep equations is readily apparent from equation (3.6).

While MATMOD is reported to be a unified constitutive equation, it is of extremely limited utility in the analysis and simulation of forging. This is due to the fact that forging takes place entirely under plasticity conditions rather than creep and the microstructure is characterized by recrystallization whereas MATMOD can only handle recovery. The formulation of MATMOD is extremely complex in comparison to other constitutive relations and requires between 20-40 constants to be evaluated for use in the analysis. While the history of the material is considered, the actual change in microstructure can not be followed or related to the stress-strain history.
3.2 FEM Based Methods

3.2.1 Incremental Prediction and Updating Techniques

A simple approach was proposed by Evans (1987) for incorporating an FEM model and to experimentally based kinetics equations to predict the steady-state grain size and update the fraction volume recrystallized at each simulation step. Using a rigid-viscoplastic FEM code and the relation between the dynamically recrystallized grain size (DRX), $M$, and the Zener-Holloman parameter, $Z$, Evans began by noting that the DRX grain size may be related to $Z$ in terms of the ASTM grain size:

$$M = C \varepsilon \exp\left(\frac{Q}{RT}\right)$$  \hspace{1cm} (3.7)

where $C$ is a constant and $Q$, $R$, and $T$ are the activation energy, Gas Constant, and temperature respectively.

By noting that DRX does not commence until the peak or critical strain, $\varepsilon_C$ is reached, several simplifying assumptions may be made. With the exception of material which is strained very close to the peak strain, material which is strained past the peak strain will achieve a constant grain size as given by equation (3.7) above. Evans assumed that material strained less than the peak strain would undergo recovery only and that nucleation during DRX would be isotropic and randomly distributed.
Based on log-log plots of experimental data, an empirical relationship showing that the volume fraction recrystallized was linear with respect to temperature and strain rate at a constant strain of 0.5. The rate of change of volume fraction recrystallized material with respect to strain, as a function of strain rate and temperature, was then defined as:

\[
\dot{r} = \frac{dr}{d\varepsilon} = D\dot{\varepsilon}\exp\left(\frac{G}{RT}\right)
\]  

(3.8)

and it was deduced that \( \dot{r} \) was a function of strain only. Because the volume fraction recrystallized is a function of strain only where:

\[
\dot{\varepsilon} = \dot{\varepsilon} \Delta t
\]  

(3.9)

the amount of material recrystallized at a given strain may then be found by multiplying \( \dot{\varepsilon} \) and \( \dot{r} \).

To predict the microstructure, the above relations were incorporated into the FEM code to predict the volume fraction recrystallized at each time step by the following relation:

\[
f_2 = 1 - (1 - f_1)\exp\left(-\dot{r}\dot{\varepsilon} \Delta t\right)
\]  

(3.10)

Equation (3.10) is simply the volume recrystallized in an increment of time \( \Delta t \) using a forward Euler scheme written in terms of \( f_1 \). By using equations (3.7)
and (3.10), it is possible to update the grain size and volume recrystallized during each time step in the workpiece based on the FEM solution for the strain rates. This is a simple approach which may easily be incorporated into a FEM code for closed die forging where deformation is accomplished during a very short time period with no cyclical deformation assumed to occur.

A more sophisticated FEM based approach has been suggested by Hewitt et al. (1980) and Ainiak et al. (1987) based on isothermal forging of René 95. This approach also involves updating the microstructure or grain size at the same time that the FEM solution is achieved. Their premise is based on the fact that flow stress at hot-working temperatures is primarily a function of strain rate, $\dot{\varepsilon}$, and microstructure, $S$, if temperature variations are negligible (i.e. isothermal forging) then the following is true:

$$\bar{\sigma} = f(\dot{\varepsilon}, S)$$

In the incremental approach to solving FEM problems, the velocities for each nodal point are calculated after a specified time interval, $\delta t$, and new strain rates obtained. After convergence has been obtained, the strain rate for that step may be calculated. The microstructure at that step can then be predicted if the relationship between the rate of change of the microstructure and the strain rate and current structure are known. This procedure is repeated at every time step until the simulation is complete at which time the final microstructure may be found. While not actually implementing this approach they did demonstrate
that this approach is feasible for superalloys produced by powder metallurgy methods under isothermal conditions.

The advantage of the method is that the entire history for each node is considered, however the method still requires considerable data collection over a wide range of conditions for each material which is to be modeled. Not only does the usual data for FEM modeling need to be generated, but in addition it is necessary to develop data describing the rate of change of grain size for each strain rate used and structure condition. If the approach is extended to non-isothermal conditions then the same data would need to be collected for each temperature thus squaring the number of data points to be collected.

Another advantage of their method is that hot-working loads are more accurately predicted due to the consideration of structure at each time step. In comparing experimentally obtained load-stroke curves for Ti-6V-4Al with those obtained from FEM predictions, Burte (1988) found that there was a significant difference between the two curves. This is not surprising in view of the fact that most FEM analyses use constitutive relations based on continuum behavior and do not consider the effect of changing microstructure during processing.

An example of where FEM was combined with an analytical rule to interpret/predict microstructural changes was presented by Malas et al. (1981). In their study, Malas et al. (1981) applied a phase rule to FEM predictions to predict the amount of phase change which occurred during upset forging of α and α-β Ti 6242 alloys based on the predicted TMP history of the workpiece.
Based on the ALPID predictions of strain rate and hydrostatic stress state, they applied Kumazawa's phase rule to predict the regions where $\alpha$ and $\beta$ would transform and where voids would form. Their results showed very good correlation between predicted and actual results validating the use of FEM as a tool in predicting microstructure.

3.2.2 Microstructural Prediction Using Expert Systems

Another computer-based approach reviewed involved use of an expert system as a post-processor for the FEM analysis. This approach employed metallurgical rules embedded in an expert system to make predictions regarding grain size for a friction welding process using a Ti-6Al-4V alloy (Meltsner, 1991; Meltsner, 1989). For a given set of process conditions, an FEM program was used to generate the predicted deformation conditions and to create a separate file containing the entire process history for each node. The expert system then post-processes the information for each nodal point to make rule-based predictions about how the microstructure could have evolved during forming based on cooling rates, maximum temperature, strains, etc.

In order to implement such a method it is necessary to develop both the "expert system shell" and to codify all possible knowledge about the material behavior into a logical set of rules. These rules must be able to handle a wide variety of processing situations without any ambiguity. While this approach works well for a material for which a great deal of information is known, it still requires that
complete knowledge of the processing-property relationships be known beforehand.

While the methods proposed by Hewitt et al. (1980) and Meltsner (1991, 1989) can provide a detailed picture regarding how the microstructure evolves, they require extensive information regarding the detailed interrelationships between the process conditions and the grain size for each material. In most cases it is also necessary to develop or modify the software required for implementation as no general purpose programs of either type are presently available.

3.3 Experimental Based Methods

3.3.1 Deformation Maps

Application of "deformation or microstructural maps" to forging process analysis were the outcome of deformation mechanism maps first developed by Frost and Ashby (1982). Deformation mechanism maps are two-dimensional plots in stress and temperature space, with strain rate superimposed, which show the regions where particular deformation mechanisms are rate controlling. These maps are generated by solving the constitutive relations for various deformation mechanisms and then determining the regions where a particular mechanism is rate-controlling (Hertzberg, 1989). An example of such a deformation map for pure Ni is shown in Figure 3.2. While being useful for component design and service life prediction, these maps are not in a form which is very useful to the forging engineer for predicting microstructure or
properties during forging. Knowledge of the rate controlling mechanism is inconsequential as forging is performed under high strain rate deformation conditions where dislocation glide is prevalent.

By changing the deformation mechanism map to reflect as-forged grain size as a function of the primary forging control variables, the map is transformed into a useful application tool for forging process design. The usual method of generating deformation maps to predict as-forged microstructure is to plot steady-state grain size as a function of strain rate and temperature. Strain is assumed to be at a sufficient level to form steady-state grain size. Due to the lack of available relations relating grain size to processing conditions, this information must be generated by conducting a series of iso-strain rate compression tests at various forging temperatures. After each test, the resulting steady-state DRX grain size is plotted on axes of strain rate and temperature. An example of a deformation map for Alloy 718 with uniform, starting ASTM 6.5-7 grain size is shown in Figure 3.3. The map shown delineates the regions where uniform recrystallized grains, necklaced grains, and shear bands will be formed (Howson and Couts, 1989). To use this method it is necessary to know beforehand what the expected range of forging conditions will be in order to carry out the testing. This may prove to be very costly if a wide range of conditions are to be encountered, necessitating a large test matrix.
Figure 3.2. Deformation mechanism map for fine grain pure nickel with a 10μm grain size (after Frost and Ashby, 1982).

The deformation map is the simplest of the methods which have been used to predict microstructure and have been most commonly applied to closed-die forging (Howson and Couts, 1989; Mosser et al., 1989). To make effective use of the deformation map, it is generally necessary to predict the local deformation
and temperature conditions using a numerical or experimental method. For this purpose, FEM is most often used to predict the local deformation and temperature conditions in the workpiece during forging. After simulating the forging process, a deformation map and/or empirical based rules are applied manually or by a post-processor to predict the expected microstructure in each region based on the resulting strain, strain rate, and temperature iso-contours.

Development and application of deformation maps to closed-die forging has been discussed by Howson and Couts (1989). In their work, compression testing was used to develop a map showing the DRX steady-state grain size generated and any resulting mechanical instabilities as a function of processing conditions. The intended use of the map was to guide forging design. While not extensively reporting on the results from actual forging, this approach is used by many of the larger forging companies such as GE and SNECMA for closed die forging.

While deformation maps have been successfully applied in closed die forging, it is apparent that a number of drawbacks are associated with using such an approach to predict microstructure in radial forging. The main problems are that much of the dynamics and history of the material which affect the microstructure are ignored. The deformation map considers the material to follow a state diagram behavior where the microstructure is assumed to be due to the final status of the system or material. However, deformation maps are simple to apply and are based on practical data which is typically used in the forge plant.
Another drawback in applying this method to radial forging is that a rather wide range of conditions may be present in the workpiece and deformation may not be steady-state. The matter of accurately predicting grain size after each pass is not an easy task as discussed above. This task is made more difficult by the fact that the grain size is continually evolving both dynamically and statically during the forging process and the deformation map typically provides
information on the microstructure during steady-state conditions during DRX. Moreover the dwell time between passes plays a significant role and its effect on microstructural development must be considered (Mataya et al., 1987). Additionally the deformation map does not provide a complete description of the recrystallized microstructure. To obtain information regarding percent recrystallized, it is necessary to apply a recrystallization-temperature-time surface map.

3.3.2 Analytical and Experimental Methods

A number of purely analytical attempts at modeling recrystallization and grain growth have been published (Anderson and Rollet, 1989; Sellars, 1990) but most of these have focused on the atomistic level and are generally not well suited for application to actual forging problems. These will not be considered in great detail but the method will be summarized to illustrate the general principles involved in the approach.

In attempting to describe how a microstructure will evolve and what the steady-state or equilibrium microstructure will be, a great deal of attention has been focused on computer-based numerical simulation. The need to employ numerical techniques to study evolving microstructures is similar to the need to employ FEM in solid mechanics where few analytical or exact solutions have been developed. This is due to the complex interactions of various physical phenomenon and an imperfect understanding of the processes involved.
The most common technique employed in materials science studies has been Monte Carlo (MC) simulation. MC is the term used to describe a class of simulation approaches where probability distributions are used to describe the likelihood that a system parameter will take on a particular value (Fabrycky and Thuesen, 1980). Due to the probabilistic nature of the solution technique, it is generally not possible to develop a closed-form solution for a given problem. An additional characteristic of the technique is that it is often necessary to perform a large number of simulation runs under identical initial conditions to determine how the solution will converge on an "average" basis. Such approaches where the average of the observations are used to generate the mean distribution are termed "crude" MC techniques (Hillier and Lieberman, 1980).

The general procedure used to employ MC simulation in studying transforming microstructures is to discretize the microstructure onto a discrete, geometric lattice. The evolution of the microstructure is considered by randomly selecting a lattice site and orientation and computing the resultant energy change. Each lattice site is assigned an index corresponding to the particular grain orientation, sites with more than one unlike neighbors are considered to be high energy sites (e.g. grain boundaries) while sites with no unlike neighbors are considered to be zero energy (e.g. grain interior) sites (Holm et al., 1990). The grain energy, $E$, for a boundary site is computed using a Hamiltonian function of the form:

$$E = J \sum_i \sum_j \left(1 - \delta_i \delta_j \right)$$

(Rollet et al., 1990) (3.12)
If the energy change is less than or equal to zero, the site is changed to the new orientation corresponding to a reduction in free energy as a result of grain growth or recrystallization. The simulation is performed until no further change is detected in the microstructure which requires approximately 1,000,000 simulation steps (Holm et al., 1990).

Physical simulation was carried out by Mataya and Matlock (1989) to simulate and study microstructural evolution during radial forging of Alloy 718 as-cast ingots. To simulate the radial forging process, Mataya and Matlock used a compression test to deform specimens at a constant strain rate of 1.0 sec\(^{-1}\). Cylindrical specimens were axially compressed using a hydraulic load frame and held at constant temperatures of 950, 1050, and 1150 °C to simulate the strain-dwell cycles experienced by material points at the workpiece ends and mid-section. Strain, strain rate, and temperature were assumed to be uniform throughout the workpiece during the simulation. Grain size was measured at the end of each simulated pass to study the microstructural evolution and identify the transformation mechanism.

There are two potential problems with the method used by Mataya and Matlock (1989). The first is the assumption that strain, strain rate, and temperature being uniform in the workpiece is generally not true. FEM simulations and experimental measurements (Domblesky and Shivpuri, 1992; Domblesky et al., 1993) of the first pass used by Mataya and Matlock (1989) have shown that while strain and temperature are uniform along the axial direction, they can vary in the radial direction. Strain rates were found to be highest at the workpiece
center and to generally increase with distance from the surface. However, the average value of strain rate of 25.4 m/m/sec. used by Mataya and Matlock is within the correct magnitude of strain rates encountered during radial forging and is not expected to adversely affect their simulation results. The second problem is that due to the gradients of deformation and temperature, it is also very likely that the forged microstructure will also demonstrate gradients in grain size. Based on their assumption of uniform deformation and temperature conditions, it will not be possible to predict those gradients using compression testing alone.
CHAPTER IV

FINITE ELEMENT MODELING

4.1 Development of the Finite Element Model

To perform the numerical or FEM simulations, DEFORM v3.1 was chosen because it has been developed specifically for the simulation of metal flow problems in forging. DEFORM (formerly ALPID) is a commercial software package developed at Battelle Columbus Labs (now supported by Scientific Forming Technologies Corporation) for the simulation of two-dimensional plane strain and axisymmetric metal forming processes using arbitrarily shaped dies and workpieces. Both isothermal or non-isothermal forging conditions may be simulated. DEFORM is based on the Updated Lagrangian formulation with rigid-thermoviscoplastic material behavior. The deformation and heat transfer analyses are coupled and solved in a staggered manner for non-isothermal simulations.

The DEFORM software is equipped with integrated pre- and post processing as well as the FEM simulation engine. The pre-processing module allows the user to generate and define the geometry, boundary, and initial conditions of the problem. Parameters may be input as a constant value or as a function of time, pressure, or temperature.
4.1.1 Axisymmetric and Plane Strain Models

To perform process modeling with DEFORM, the user is provided with two options for the process description. These are "plane strain" and "axisymmetry". The assumptions inherent in each of the deformation models are as follows. For plane strain, plastic flow is assumed to occur only in one plane and is independent of the distance from the plane (Kobayashi et al., 1989). The strains and strain rates in the direction perpendicular to the plane in which flow is occurring are assumed to be zero. Mathematically the strains are related as follows:

\[ d\varepsilon_x = -d\varepsilon_y \]  
\[ d\varepsilon_z = 0.0 \]  

(4.1a)  
(4.1b)

In an axisymmetric model the workpiece and dies are assumed to be generated by rotating a plane about the axis of symmetry. Additionally, the die is assumed to fully envelop the workpiece around the outer diameter. Metal flow is assumed to occur in the radial, circumferential, and axial directions (cylindrical coordinates). The strain relations may be written as follows:

\[ d\varepsilon_r = d\varepsilon_\theta = -\frac{1}{2}d\varepsilon_z \]  

(4.2)

The axisymmetric model was employed in the present study due its ability, albeit implicit, of being able to portray the drawing out of the billet and the
variation of surface temperature along the length of the workpiece. Axisymmetric and plane strain representations of radial forging are shown in Figure 4.1. In DEFORM, area and volume conservation are enforced (DEFORM User's Manual, 1992). Hence in the plane strain model, as noted above, the metal flow is restricted to the x-y plane which does not take into account flow along the longitudinal direction or third axis. The axisymmetric model also diverges from physical reality in that the assumption that the dies are totally concentric with the outer diameter of the workpiece. Furthermore, the small clearances between the dies at the end of the forging stroke are neglected. The amount of clearance for round dies may be found by calculating the envelopment ratio which is defined as:

\[ \frac{r_{w}}{r_{t}} = 1 \text{ (for axisymmetric models)} \]  (4.3)

where \( r_{w} \) is the workpiece radius and \( r_{t} \) is the tool radius. Therefore to use the axisymmetric model to simulate four die radial forging, the following assumptions need to be made:

1) Small clearances between the dies are neglected
2) Rotational feed is neglected
3) Each stroke of the forging dies may be modeled as one simulation
4) The flat surfaces on the forging tools may be approximated by semi-circular arcs
Figure 4.1. Cross-sectional views of (a) axisymmetric and (b) plane strain deformation models for a radial forging machine with four tools. Only a quarter section is shown due to symmetry.
Figure 4.2. Longitudinal view of the axisymmetric model for four die radial forging in DEFORM. Only one-half of the workpiece is shown due to symmetry.
4.1.2 Mesh Discretization

Prior to performing the simulation, the workpiece and dies must be discretized into a mesh of "finite elements". The workpiece was discretized using four node quadrilateral elements. The Automatic Mesh Generator Module (AMG) in DEFORM was used to generate the mesh. Due to the original mesh becoming highly distorted during the numerical simulation it was necessary to perform remeshing whenever the analysis module detected a negative Jacobian. Negative Jacobian refers to a numerical instability resulting whenever the mesh becomes severely deformed. DEFORM v3.1 contains provisions which enable remeshing to be performed automatically during the program execution.

4.1.3 Simulation Controls

The simulation controls available to the user include maximum load capacity of the press, maximum stroke, number of solution steps, maximum allowable strain per step, etc., allowing a wide variety of forming conditions to be modeled. DEFORM program execution is controlled primarily via the selected time or stroke increment of each simulation step. Because the tool velocity varies non-linearly as a function of time, the simulation was controlled using maximum stroke increment per step. Past experience with DEFORM has shown that solution accuracy and speed of convergence improves for each time step with smaller time or stroke increments, albeit at a penalty of increased CPU times. For the simulations in the study, a stroke increment of the order 0.125 mm
was used in all simulation runs. This value was observed to give good convergence and reasonable execution times.

4.1.4 Material Properties

Material properties for the billet being modeled include such variables as thermal data; flow stress as a function of strain, strain rate, and temperature; density; etc. One of the characteristics of Alloy 718 is that it is strain-softening during hot-working, i.e. after a critical strain is reached the material becomes softer with increased deformation as a result of dynamic recrystallization.

The flow stress data used in the FEM simulation was based on a billet with a starting microstructure having a grain size of ASTM 6-7. To account for the temperature and strain rate sensitivity of the flow stress, a third order polynomial (equation 4.4) proposed by Srinivasan et al. (EMTEC, 1990) was used as the constitutive equation for Alloy 718.

\[
\log \sigma = A + Bx + Cx^2 + Dx^3 \tag{4.4}
\]

where:

\[
x = \log_{10} \varepsilon \tag{4.5}
\]

and \(A, B, C, \) and \(D\) are coefficients which are a function of the forging temperature.
4.1.5 Thermal Properties

The modes of heat transfer operating during radial forging are conduction, convection, and radiation. During radial forging, thermal losses occur at free surfaces and at die-workpiece interfaces. Thermal transfer from free surfaces has been extensively studied and quantified. Heat conduction between the workpiece and dies is extremely difficult to study and has not been as well characterized due to experimental difficulties. It has been shown experimentally that the interface heat transfer coefficient for most metals is constant above a threshold pressure and is an order of magnitude lower for zero load conditions (Semiatin et al., 1978). For the model the interface heat transfer coefficient was assumed to have a constant value. It was assumed that the pressure build-up at the die interface would take place instantaneously in comparison to the time under load. This was based on results of a load-stroke curve measured at Watervliet Arsenal (Raghupathi et al., 1980). For the study the value used for the interface heat transfer coefficient was specified as 5.0 kW/m²/°C. The interface heat transfer loss at the chuckhead/workpiece interface was neglected due to the insulating characteristics of the ceramic inserts used to grip the billet.

The universal forging dies used in radial forging of Alloy 718 are usually H-13 base metal with a welded overlay of Udimet 520. Since the bulk of the dies are H-13, the dies were modeled as being totally rigid and having thermal characteristics of H-13. Moreover, heat transfer to the dies is controlled by the interface heat transfer coefficient and the assumption of H-13 as die material is reasonable. Thermal data for H-13 die steel was taken from the literature (ASM
International, 1985). The die temperature was initially prescribed as 426.0 °C for the simulation which is consistent with radial forging practice for Alloy 718.

4.2 Kinematic Boundary Conditions

To model each pass, only one chuckhead was included in the model. This was necessary to prevent the workpiece material from bulging ahead of the tool as a result of being overly restricted in the axial direction by two chuckheads. At the start of each pass simulation, the gripper was positioned at the trailing end of the workpiece. The position of the chuckhead was then switched at the midpoint of the current pass and re-positioned at the leading edge of the workpiece as illustrated in Figure 4.3.

In the physical process, the grippers securely hold the billet surface at four points. During forging, each chuckhead acts as a spring which allows the workpiece to move slightly as material in the deformation zone is axially displaced. In addition to maintaining the workpiece position, the grippers also apply front and back pull to reduce the total forging load required.
Direction of Axial Feed

(a) First Half of Pass Simulation

Chuckhead Initial Position

Mid-Pass

(b) Second Half of Pass Simulation

Chuckhead After Repositioning

Figure 4.3. Stroke-by-stroke model representation of radial forging showing the chuckhead and gripper position during the (a) first half of the pass simulation and repositioning during the (b) second half of the pass simulation.
To eliminate the transition region in the strain contours resulting from the fixed nodes boundary condition in the sequential step model, the chuckhead was initially modeled as an elastic "die" having sticking friction. The gripper was positioned as a stationary "die" perpendicular to the billet end. To allow for limited movement of the workpiece during deformation, the "die" was prescribed to be composed of an elastic material having a Young's Modulus on the order of 2100 MPa. This was felt to allow sufficient workpiece restraint while allowing a small axial movement of the workpiece. However, this was not successful due to poor convergence and the excessive CPU time required for each simulation. As a result of the die deflection, the simulation time increased dramatically and each simulation required between 45-55 minutes to perform on a VAX Station 4100 computer running under VAX/VMS Version 5 at 8-10 MIPS.

Re-evaluating the gripper representation showed that four point axial symmetry existed between the billet and both the grippers and the forging tools. In both cases the billet surface is almost enclosed at the points of contact as shown in Figure 4.4. A modified approach demonstrated that the chuckhead restraint could be modeled as an axisymmetric ring surrounding the perimeter of the workpiece with sticking friction at the gripper/workpiece interface to constrain the billet. Workpiece restraint in the axial direction was obtained by assuming sticking friction equal to the billet shear stress \( m = 1.0 \) at the interface as defined by:

\[
\tau_i = m \frac{\bar{\sigma}}{\sqrt{3}} \quad \text{(Von-Mises definition)}
\]

(4.6)
where: \( \tau_i = \text{Interfacial shear stress} \)
\( \overline{\sigma} = \text{Material flow stress} \)
\( m = \text{Friction factor (constant)} \)

By modeling the chuckhead as a rigid ring with sticking friction, a substantial reduction in the CPU time required for each simulation was achieved and was found to be equivalent to the time required for an analysis using zero axial velocity boundary conditions. The resultant strain contours also did not display the strain transient at mid-pass which was previously observed when the chuckhead position was reversed in the sequential step model.

The effect of front and back pull on the billet was also modeled by alternately imposing force and pressure boundary conditions on the workpiece ends. The front and back pulls initially used in the model were 1,740 and 2,640 kN taken from data measured by Lahoti and Altan (1976-2). However, these attempts were unsuccessful as convergence could not be obtained in any of the simulations attempted. The lack of convergence was due to the force imbalance resulting from the large axial force component generated by the forging tools during the simulation. In order to obtain a solution, DEFORM requires that quasi-static equilibrium be maintained at each time step in the simulation.
Figure 4.4. Rotational symmetry of the (a) workpiece grippers and (b) forging tools at the billet surface.
4.3 Inter-stroke Heat Losses

During radial forging, thermal losses occur at free surfaces and to a lesser extent at the die-workpiece interface. Heat generation will result from deformation (adiabatic) heating of the billet material according to equation (4.7). The relevant values for Alloy 718 are included in parentheses. Thermal data for Alloy 718 conductivity and heat capacity were taken from the data of Isogawa et al. (1990).

\[ \Delta T = \beta \frac{\sigma e}{Jcp} \]  

(4.7)

where:

- \( \beta \) = Proportion of mechanical energy converted to heat (0.90-0.95)
- \( J \) = Conversion factor (= 1.0 for SI units)
- \( c \) = Heat capacity (434.75 J/kg·°K)
- \( \rho \) = Density (8.23 g/cm³)

Due to the problem geometry which consisted of a large length to diameter ratio, heat transfer from the billet is essentially a one-dimensional, transient problem with internal conduction, convection, and radiation losses in the radial direction. For the present model, the billet was assumed to undergo radiation losses only. Based on the temperatures typically used in forging of Alloy 718, it was found that radiation losses are the dominant mode of heat transfer at the billet surface. Because the convective losses are very small in comparison to the total heat loss, this component may be neglected in the analysis. Convective losses are linearly related to the difference between the billet surface and
ambient temperature while radiation losses are proportional to the fourth power. The total heat flux at the billet surface may be partitioned into two components:

\[ q_{\text{total}} = q_{\text{convective}} + q_{\text{radiation}} \]  

(4.8)

The convection and radiation terms may also be written as follows:

\[ q_{\text{total}} = hA(T_s - T_\infty) + \varepsilon \alpha A(T_s^4 - T_\infty^4) \]  

(4.9)

where:
- \( h = \text{convection coefficient} = 0.052 \text{ W/m}^2\cdot\text{°K} \)
- \( A = \text{Surface Area of the Billet} \)
- \( T_s = \text{Temperature at Billet Surface} = 1066 \text{ °C initially} \)
- \( T_\infty = \text{Ambient Temperature} = 18 \text{ °C} \)
- \( \varepsilon = \text{Total Emissivity} = 0.4 \)
- \( \alpha = \text{Boltzman Constant} = 5.670 \times 10^{-8} \text{ W/m}^2/\text{K}^4 \)

Based on the problem geometry and natural convection for air, the convection coefficient for a horizontal cylinder will be \( h = 0.052 \text{ W/m}^2\cdot\text{°K} \). The total emissivity for the Alloy 718 billet was assumed to be 0.4 based on surface temperatures taken during the forging experiment at Carpenter Technology. Taking the ratio of convection and radiation terms:

\[ \frac{hA(T_s - T_\infty)}{\varepsilon \alpha A(T_s^4 - T_\infty^4)} = \frac{0.004}{0.0444} = 0.0901 \]  

(4.10)
From equation (4.10) it may be seen that convective losses account for less than 10% of the total heat loss from the billet. Because the convective losses are very small in comparison to the total, this component may be neglected in the analysis. While radiation is a function of the surface temperature and tends to decrease with temperature, the billet temperature during radial forging does not vary considerably. Thus, the assumption of neglecting convection during forging is valid.

In contrast to other hot forming operations, die chilling is generally considered to be negligible in radial forging due to the short contact time between the billet and forging tools. It has been estimated that the tools are in contact with the workpiece for less than 10-15% of each forging cycle (Nagpal and Lahoti, 1980). For a machine operating at a stroking rate of 180-200 strokes per minute, the non-contact time between each stroke will be on the order of 0.3 seconds. While the amount of heat loss occurring during the short non-contact time between each stroke is insignificant, the accumulated loss after each pass is very important and must be considered in the analysis.

Because the billet is deformed incrementally, the surface will not remain at a uniform temperature during each pass. Based on this, it is desirable to model the heat loss during each stroke in the corresponding simulation. The simplest approach to account for the non-contact time is to model the complete tool stroke from Top Dead Center (TDC) to Bottom Dead Center (BDC). However, there are two problems inherent with such an approach. The first problem is that it is very inefficient from a computational standpoint to simulate the entire
stroke. Even though no deformation is occurring and the die is undergoing rigid body translation only, the solution procedure must still be performed at each step. The second problem is that the average and limiting strain rates at each time step in DEFORM are adjusted based on the deformation conditions from the previous time step. If rigid body translation is occurring, the average and limiting strain rates will drop to negligible values. Once the die comes into contact with the workpiece surface, the initial guess for the current time step will be too low and the solution will not converge. The variational form of DEFORM is based on a large, positive penalty constant to satisfy the incompressibility constraint and may be written as (after Kobayashi, 1989):

\[
\delta \Pi = \int_V \dot{\varepsilon} \delta \dot{\varepsilon} dV + K \int_V \dot{\varepsilon}_v \delta \dot{\varepsilon}_v - \int_{F_I} \delta u_i dS = 0 \tag{4.11}
\]

where:

- \( \delta \dot{\varepsilon} \) = virtual strain rate
- \( K \) = penalty constant
- \( \dot{\varepsilon}_v \) = volumetric strain rate
- \( \delta \dot{\varepsilon}_v \) = virtual volumetric strain rate
- \( F_I \) = surface traction
- \( \delta u_i \) = virtual velocity

For a material which is not deforming plastically (i.e. rigid or nearly rigid), the first term in equation (4.11) may be written as (after Kobayashi et al., 1989):

\[
\int_V \left( \frac{\dot{\varepsilon}}{\varepsilon_0} \right) \delta \dot{\varepsilon} dV \tag{4.12}
\]
where \( \dot{\varepsilon}_0 \) is the limiting strain rate below which the material is assumed to behave as a rigid body.

Now as \( \dot{\varepsilon}_0 \) becomes increasingly smaller with each subsequent time step, the first term in equation (4.11) will become very large. As a result, the penalty constant \( K \) will also become very large and the solution will not converge.

However, the user is provided with an option to describe the stroke-velocity curve as a set of discrete points or as a FORTRAN subroutine. To account for the non-contact time of the tools between each stroke, an implicit method was used to introduce a non-contact period equivalent to the dwell time of the forging tools between each forging or loading cycle at the beginning of each simulation. By adjusting the first step in each simulation such that the die has a sufficiently small velocity, the simulated time step can be made to match the amount of non-contact time between forging cycles. Once the tool has come into contact with the workpiece surface, the normal stroke-velocity curve is followed as calculated from equation (2.1). An example of the modified stroke-velocity curve used in the simulation is shown in Figure 4.5. Due to the small time increments involved, the heat transfer for each stroke dwell was performed in a single time step.
Figure 4.5. Modified stroke-velocity curve of the forging tools used in the DEFORM FEM simulation of radial forging to account for heat losses occurring between forging strokes.

4.4 FEM Simulation Procedure

A significant problem in trying to simulate the radial forging process is that an excessive amount of user interface time is needed due to the large number of simulations which are required. In a typical forging pass, the tools will contact the workpiece 100-300 times necessitating that a corresponding number of FEM simulations be performed. Due to the need to update the die position and the
boundary conditions between each simulation, it is necessary for the user to perform a prescribed set of pre-processing tasks prior to continuing the next simulation. Because of the one-to-one correspondence between the number of tool strokes and simulations, manual preprocessing was found to be impractical for modeling a complete pass. Based on an average of three minutes to pre-process each simulation, a cumulative total of 5.35 man-hours are required to simulate a single pass comprised of 107 tool strokes, excluding the simulation time.

To facilitate the simulations, an automated routine was written using VAX/VMS command language to update the boundary conditions and resubmit the job for the next simulation using pre-supplied input data. A flow diagram is shown in Figure 4.6. This routine is designed to run jobs sequentially with no user interface time required. By running in the background, execution of the routine is transparent to the user. The main control routine is a very flexible program which may be edited by the user to accommodate different pass conditions. The required user inputs to the routine are:

1) Name of database file to be used
2) Starting and ending nodes along the workpiece centerline
3) r and z translations of the die
4) Number of time steps in each simulation
5) Maximum stroke allowed
6) Number of simulations to be performed
The routine was originally developed to run in a VAX/VMS operating environment. The routine was later converted to operate on IBM X-Windows workstations running in a UNIX environment. Use of the routine has resulted in a reduction of user interface time from 5.35 hours to less than 1 hour for each pass. By running the routine on the IBM workstations, the simulation time has also been reduced by a factor of 10 over the VAX Station 4100. With minimal alteration to the input file, the routine can also be adapted to simulate large diameter tube forging sequences.

4.4.1 Simulation of the Process Sequence

The steps used to model a pass sequence consisted of the same series of events which occurred during the actual forging process as listed in Table 4.1. From a simulation standpoint, it may be seen that there are two types of analysis events: non-isothermal deformation and transient heat transfer. The primary differences between the two are that for a heat transfer analysis, only time-dependent heat losses are considered. For a non-isothermal analysis, the coupling effect between deformation and heat transfer is considered at each time step. While the dwell periods may be modeled as non-isothermal events, it is more efficient from a computational standpoint to employ a heat transfer analysis (i.e. no deformation analysis). Billet transfer and inter-pass dwell periods may also be simulated by performing transient heat transfer analyses. Due to the requirement to simultaneously consider deformation heating and heat losses, each pass was modeled using the non-isothermal mode in DEFORM.
Input FEM Model Data and Mesh Objects

Manually Perform DEFORM Simulations Until Steady-State Deformation is Achieved

Invoke Automatic Pre-Processing and Simulation Routine

Specify Pre-Processing and Simulation Parameters

Simulations Completed?

Yes

Perform Post-Processing and Interpretation of FEM Results

No

Count = Count + 1

Figure 4.6. AUTODEFORM 2.1 flow diagram for automated pre-processing of DEFORM radial forging simulations.
At the beginning of each simulation sequence, the billet was assumed to be at a uniform temperature equal to that of the furnace. A transient heat transfer analysis was then performed corresponding to the time required to transfer the billet from the furnace to the start of forging. During each pass, deformation was modeled under non-isothermal conditions to obtain the thermomechanical history. Heat transfer occurring between consecutive forging cycles was accounted for using the method described in Section 4.3. Inter-pass dwell times were also simulated using a transient heat transfer analysis. The temperature distribution and cumulative strain in the billet at the end of each pass were then carried over and used as the initial temperature and strain distributions for the subsequent pass simulation. Heat losses resulting during the billet transfer from the radial forging machine to the cut-off saw were also modeled using a transient heat transfer analysis.

Table 4.1. Chronology of events during billet conversion.

<p>| | |</p>
<table>
<thead>
<tr>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>1.</td>
<td>Heat Alloy 718 billet in rotary hearth furnace to uniform temperature.</td>
</tr>
<tr>
<td>2.</td>
<td>Transfer the billet from the furnace to the rotary forge (50-60 seconds).</td>
</tr>
<tr>
<td>3.</td>
<td>Perform first pass reduction (pass times range from 30-200 seconds).</td>
</tr>
<tr>
<td>4.</td>
<td>Reposition billet for second pass (typically 10-20 seconds).</td>
</tr>
<tr>
<td>5.</td>
<td>Perform second pass reduction (between 30-200 seconds).</td>
</tr>
<tr>
<td>6.</td>
<td>Reposition billet for third pass (typically 10-20 seconds).</td>
</tr>
<tr>
<td>7.</td>
<td>Perform third pass reduction (between 30-200 seconds).</td>
</tr>
<tr>
<td>8.</td>
<td>Transfer billet to abrasive cut-off saw to crop ends.</td>
</tr>
</tbody>
</table>
The FEM representation of the radial forging process is shown in Figure 4.2. A summary of each object used in the FEM simulation is listed in Table 4.2.

Table 4.2. Summary of object meshes used in the simulation.

<table>
<thead>
<tr>
<th>Object</th>
<th>Elements</th>
<th>Nodes</th>
<th>Type</th>
</tr>
</thead>
<tbody>
<tr>
<td>Billet</td>
<td>1100</td>
<td>922</td>
<td>Plastic</td>
</tr>
<tr>
<td>Tool</td>
<td>351</td>
<td>310</td>
<td>Rigid</td>
</tr>
<tr>
<td>Chuckhead</td>
<td>75</td>
<td>48</td>
<td>Rigid</td>
</tr>
</tbody>
</table>

The friction factor for the die/workpiece interface was taken as $m = 0.6$ which is typical of hot-working conditions for Ni-based alloys where no lubrication is used (Altan et al., 1983).
5.1 Model Validation Procedure

Previously reported FEM studies of radial forging (Isogawa et al., 1990; Jackman et al., 1992) have based the model validation on comparisons of surface temperatures and as-forged microstructure from forged billet. While this type of comparison does provide a measure of agreement between the model and physical process, it is highly empirical. This is typified by temperature predictions which may agree well with the measured surface temperatures while correlation with the billet interior remains unknown. Therefore, it is necessary to base the comparison on direct measurements of billet temperature and effective strain in order to fully validate the process model.

To validate the FEM model described in Chapter IV, a controlled forging experiment was designed and carried out at Carpenter Technology’s Steel Division in Reading, PA. The main objectives of the forging experiment were to measure the billet temperature and strain profiles in the as-forged billet. Due to the lack of existing methods available to accomplish these tasks, new experimental methods were developed. The experimental methodologies which
were developed to measure the billet temperature and effective strain after forging are described below.

The test case, a three pass sequence, was simulated and then compared to experimentally measured results which were obtained from the controlled forging performed at Carpenter Technology. To forge the workpiece, a GFM model SX-65 radial forging machine rated at 14,000 kN per tool and operating at 180 strokes per minute was used. The billet material was Pyromet 718 which is a nickel-iron based superalloy having a composition identical to Alloy 718. The steps used to prepare the billet for the forging experiment are listed in Table 5.1. The starting billet was 4.12 meters in length with a 254 mm diameter which was reduced to 198.12 mm using the pass schedule listed in Table 5.2. The billet temperature was monitored during furnace heating via thermocouples mounted at center, mid-radius, and surface positions at mid-length of the workpiece. The furnace temperature was set at 1066 °C and the billet was heated for a total of eight hours prior to forging.

Table 5.1. Processing sequence used to prepare the Alloy 718 billet for the radial forging experiment at Carpenter Technology.

<table>
<thead>
<tr>
<th>Step</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>1)</td>
<td>VIM/VAR 508 mm Ø Ingot</td>
</tr>
<tr>
<td>2)</td>
<td>Homogenizing Heat Treatment</td>
</tr>
<tr>
<td>3)</td>
<td>Rotary Forge to 264 mm Ø Billet</td>
</tr>
<tr>
<td>4)</td>
<td>Machine to 254 mm Ø Round</td>
</tr>
</tbody>
</table>
Table 5.2. Pass schedule used in the forging experiment at Carpenter Technology's Rotary Forge Shop.

<table>
<thead>
<tr>
<th>Pass Number</th>
<th>Percent Reduction in Area</th>
<th>Average Feed Rate</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>10.7</td>
<td>6.86 m/min.</td>
</tr>
<tr>
<td>2</td>
<td>15.2</td>
<td>4.85 m/min.</td>
</tr>
<tr>
<td>3</td>
<td>19.6</td>
<td>3.79 m/min.</td>
</tr>
</tbody>
</table>

5.1.1 Temperature Measurement

The standard method for measuring the thermal profile is to use thermocouples embedded in the billet. However, this is not practical for cases where the billet will undergo deformation and the thermocouples have a high probability of being destroyed. While touch probes have been used, these were ruled out due to safety considerations and the high temperatures which were involved. Thus a non-contact method of temperature measurement was used.

Surface temperatures were measured during each pass at 1.0 second intervals using infrared pyrometers mounted at the entry and exit of the forging box. Surface temperatures are normally measured during forging and permit direct comparison with the FEM model predictions for nodal temperatures at the surface. To measure the billet temperature profile after forging, an infrared pyrometer was positioned adjacent to the abrasive cut-off saw as shown in Figure 5.1. Immediately after the cut-off operation was completed and the slug end removed, the newly generated surface was scanned to record the thermal
profile. While some heat may have been generated during cutting, this was assumed to be negligible in comparison to the bulk billet temperature. Due to the possibility of the abrasive wheel fracturing during cutting, the pyrometer system was triggered from a remote location. The temperature profile was then stored on a magnetic tape using a Hitachi data acquisition system to electronically record the temperature profile.

5.1.2 Strain Measurement

Most deformation studies in forging have employed grid networks (rectangular or polar) to measure deformation directly or made use of metallographic techniques to indicate the presence of non-uniform deformation (Dieter, 1986). However, grids are mainly suitable for two-dimensional studies and metallographic correlation is limited by the absence of process-property relations. Process models based on plasticine have also been employed but the high cost of developing a scale model machine and problems with ensuring similitude between the model and the billet material often limit their use in forging studies.
Figure 5.1. Pyrometer arrangement at the cut-off saw used for measuring the billet temperature profile after forging Carpenter Technology.
To measure the effective strain in the billet after forging, two 25.4 mm diameter Alloy 718 (Pyromet) rods which had previously been doped with tungsten powder particles were press fit into the billet at 90 degrees to one another. To prevent the rods from shifting during forging, the free surfaces of both inserts were welded to the billet surface. In order to eliminate transient effects due to deformation at the billet ends, the rods were placed at 1.10 and 1.22 m from the leading workpiece end. After forging and cooling to room temperature, the billet was immersed in a water tank and ultrasonically scanned to locate the position of the rods (Figure 5.2). The difference in density between the Pyromet material matrix and the coarse grained tungsten particles caused the ultrasonic wave to be attenuated and prematurely reflected. The reflected elastic wave was then used to locate the deformed rods. The segment containing the rods was then removed from the billet and sectioned to characterize the deformed shape of the plugs and to measure the average effective strain as a function of radial position. The hoop and longitudinal strain components were calculated from the major and minor dimensions of the deformed Pyromet 718 rods (Figure 5.3) as follows:

\[ \varepsilon_\theta = \ln\left(\frac{d_2}{d_0}\right) \]  

(5.1)

\[ \varepsilon_z = \ln\left(\frac{d_1}{d_0}\right) \]  

(5.2)
and by constancy of volume:

\[ \varepsilon_r + \varepsilon_\theta + \varepsilon_z = 0.0 \]  \hspace{1cm} (5.3)

The effective strain was calculated using cylindrical coordinates:

\[ \varepsilon = \sqrt{\frac{2}{3} \left( \varepsilon_r^2 + \varepsilon_\theta^2 + \varepsilon_z^2 \right)} \]  \hspace{1cm} (5.4)

where:

- \( \varepsilon_r \) = Radial Plastic Strain Component
- \( \varepsilon_\theta \) = Circumferential Plastic Strain Component
- \( \varepsilon_z \) = Axial Plastic Strain Component
Figure 5.2. Schematic representation of the ultrasonic immersion technique used to detect the location of the tungsten doped Pyromet rods after forging.
5.1.3 Forging Load Measurement

The forging load during each pass was measured by pressure transducers which are used as part of the overload protection system of the radial forging machine. While forging load signature analysis is often used as a method of verifying FEM model accuracy, it was not feasible as a comparison in the present study for the following reasons. The first is that elastic deformation in the sizing zone is neglected in the model due to the rigid-viscoplastic formulation used in DEFORM. As a result, the significant contribution of deformation from the sizing zone to the total forging load in the model can not be directly obtained. Secondly, due to the use of an axisymmetric deformation model, the correlation of the predicted load to the measured load will be a function of what proportion
of the forging tool lands are assumed to be in contact with the billet surface at
the end of each stroke. A third problem is that the actual forging load is sensitive
to the effect of the front and back pull applied by the chuckheads which could
not be incorporated into the model for the reasons described earlier in Chapter
4.

However, the forging load was analyzed using the slab method in a previous
study (Domblesky et al., 1992) and the method was found to yield results which
were within 80% of the actual values. The relevant equations used in the slab
analysis for radial forging are derived in Appendix C.

5.2 Results

The FEM simulations were performed on a Digital Equipment Corporation VAX
Station 4100 system running under VAX/VMS Version 5.5 Operating System
and an IBM 340 RISC Station running under AIX (UNIX). Each simulation
required between 12-15 and 6-8 minutes of CPU time to perform on the VAX
and IBM systems respectively.

5.2.1 Deformed Mesh

The workpiece was remeshed at the beginning of the second pass simulation
due to a single element at the trailing workpiece end becoming distorted. This
may have been eliminated through the use of a finer mesh density at the
workpiece ends than was used in the simulation. The deformed mesh during
each pass showed that deformation was essentially homogeneous with negligible shearing in the axial direction and very little effect from friction. If friction had been a significant factor in deformation, the elements near the surface should have exhibited macroscopic shearing along the axial direction relative to the core material. However, the deformed mesh was observed to remain relatively uniform after each pass simulation for friction factors of $m = 0.3$ and $m = 0.6$.

5.2.2 Effective Strain

The Von-Mises or effective true strain along the radial direction for the location corresponding to the Pyromet insert which was initially at 1.10 meters from the leading end is shown in Figure 5.4 for each of the three passes simulated. The highest strains were predicted to be near the billet surface while the lowest strains were predicted in the billet core. The accumulated strain in the billet after the third pass is shown in Figure 5.5. With the exception of the billet ends, the variation in effective strain along the billet length was predicted to be negligible.

A comparison of the DEFORM predictions and the measured strains after the third pass are shown in Figure 5.6. The strains in both Pyromet rods were found to be equivalent at each radial location. From Figure 5.6 it may also be seen that the DEFORM model over predicts strain at the billet center and under predicts strain at the surface. This result is to be expected given the use of an axisymmetric deformation model where the material spread between the tools is
neglected due to the fact that only two of the three plastic strains are independent.

Figure 5.4. Effective strains in the billet predicted by DEFORM for the three pass radial forging simulation.
Figure 5.5. Predicted effective true strain in the billet after three passes. All strain values are cumulative.
Based on a comparison of the shape between the undeformed and deformed rods, deformation was observed to be aligned along the \( r \), \( \theta \), and \( z \) principal axes. The deformed cross-section of the rods are illustrated in Figure 5.7. Based on the elliptical shape of the deformed rods, it is apparent that significant secondary tensile strains are generated in the axial directional while secondary compressive strains are generated in the radial and hoop directions. To verify the absence of shear strains, a 9.5 mm diameter hole was drilled parallel to one of the rods in the billet section (Figure 5.8) at Carpenter. By comparing the
difference in centerline positions, $\Delta x_j$, between the drill hole and the rod in each of the sections, it was found that the centerline locations were parallel in each section confirming a lack of shear deformation in the transverse plane ($r\theta$ plane). A possible reason for this behavior may be found by considering the inherent symmetry of the process and the restraint imposed by the material surrounding each rod. The lack of shear deformation was also inferred by the DEFORM model which predicted that metal flow was primarily the result of homogeneous deformation with negligible contribution from interface friction.

Figure 5.7. A schematic representation of the undeformed and deformed cross-sections of the Pyromet 718 inserts.
Based on the plasticine modeling work performed by Isogawa, et al. (1990) at Daido Steel, it was expected that significant shear strains would be present in the subsurface layers in the r-z and r-θ planes. However, no shear strains were observed in the deformed Pyromet rods. The reason for the large shear strains observed in their work may have been the result of secondary tensile forces causing the plasticine layers to separate and shift during deformation. A second, more likely, possibility is that the friction conditions in their plasticine model
may not have matched the actual conditions. Based on the deformed shape of the plasticine billet which they reported (Figure 5.9), the material at the center of the billet appears to have been axially displaced by a greater amount than the surface material. This would result in the center of the billet bulging out which is not observed under actual forging conditions.

(a) Undeformed Plasticine Billet

(b) Deformed Plasticine Billet

Figure 5.9. A schematic representation of the longitudinal cross-section of the deformed plasticine billet obtained in the Daido Steel simulation of radial forging (after Isogawa et al., 1990).

The radial distribution of shear strains in the r-θ plane which they reported (Figure 2.5) showed a steep gradient in the sub-surface material. If such localized shear strains were generated in the billet during hot forging, it is highly probable that the billet would not be able to support such a steep strain
gradient and would tear internally along the circumferential direction. Moreover, the stress state required to generate shear strains in the r-θ plane would likely result in the billet being twisted in torsion.

5.2.2.1 Effect of Tool Overhang on Deformation

Because it is known that the strain distribution in rolling and radial forging is sensitive to the ratio of billet diameter or radius and the tool contact length, two additional sets of simulations were performed to compare the effect of tool overhang (Figure 5.10). Alternatively, this may be considered to be equivalent to an increase or decrease in the axial feed rate. By increasing the axial feed rate, the amount of undeformed material present under the inlet angle of the tools will also increase, resulting in a higher contact length. Two cases corresponding to a feed rate which was 2 times the original feed rate and a second case which was 0.5 times the original feed rate were simulated for each pass of the forging experiment.

The result of increasing the tool overhang was found to increase the predicted effective strain upwards with little change in the radial strain gradient as shown in Figure 5.10. A decrease in the overhang was found to lower the strain distribution. The difference in the strain distribution was correlated to the degree of bulging which was observed at the ends. For larger feed rates, the greatest amount of bulging was observed.
Figure 5.10. Effect of tool contact length on the effective strain distribution in the billet for the radial direction after the third forging pass. All values are accumulated strain.

5.2.3 Temperature

To check the validity of the thermal data used in the process model, the billet heating cycle was simulated using DEFORM. The results from the simulation were then compared against the thermocouple readings which were taken during billet heating to validate the model. The initial billet temperature was 18 °C while the furnace temperature was set at 1066 °C. The thermocouple data generated during heating showed that the billet achieved a uniform temperature
of 1066 °C at the surface, mid-radius, and center locations at 50% of the heating cycle. The DEFORM simulation predicted that the billet temperature at the center was within 4.5 °C of the actual temperature at 50% completion of the heating cycle. As seen in Figure 5.11, the predicted billet temperature is nearly uniform and agrees very well with the thermocouple data. This agreement confirms that the thermal data and the DEFORM model are capable of accurately simulating heat transfer conditions in radial forging.

The predicted temperature contours in Figure 5.12 showed a significant gradient between the surface and sub-surface layers was initiated during the 2.5 minute period when the billet was transferred from the furnace to the GFM. The DEFORM model predicted that the sub-surface gradient would be maintained throughout all three passes. Due to the low thermal conductivity and high heat capacity of Alloy 718, the steep temperature gradient near the surface is to be expected. This prediction is supported by the difference in grain size between the billet surface and interior which is typically found in forged billet. This result indicates that transfer times from the furnace to the rotary forging machine should be minimized to reduce the temperature gradient, and hence the difference in grain size, in the billet.
Figure 5.11. Predicted temperature profile of the Alloy 718 billet used in the forging experiment after 50% completion of the furnace heating cycle.
Figure 5.12. Predicted billet temperature profile entering the GFM forging machine. Temperature increases correspond to deformation of the material point while temperature losses correspond to dwell and hold times.
Axial feed rate has generally been considered to be the controlling factor in surface temperature during radial forging. However, the results show that percent reduction of area may also be a controlling factor. In the forging experiment, it may be seen that the maximum rise in surface temperature was in pass 2 which does not correspond to the maximum axial feed rate. Temperatures along the radial direction were observed to increase during each subsequent pass due to the increased reductions which were employed. The temperature profile in the radial direction during forging is shown in Figure 5.13. With the exception of the workpiece ends, little variation in temperature was predicted along the length of the billet.
Figure 5.13. Predicted radial time-temperature history of the billet during the three pass radial forging experiment.

5.2.3.1 Surface Temperatures

A comparison of the predicted surface temperatures with the average surface temperatures entering and exiting the forging box is shown in Figure 5.14. All nodal temperatures were taken from the assumed location of the Pyromet 718 inserts and were found to be within reasonable engineering accuracy. The model was also found to correctly predict the gradient in surface temperature from the
leading end to the trailing end of the billet which is a consequence of incremental deformation.

Figure 5.14. Predicted surface temperatures at the first Pyromet 718 insert during the three pass radial forging experiment.

The difference between the predicted and measured surface temperatures for the second and third passes appears to be the result of a decrease in the total emissivity of the billet after deformation. For the simulation, the total emissivity was assumed to remain constant at 0.4. The apparent change in emissivity of the
The predicted rise in temperature at the surface, $\Delta T_p$, due to adiabatic heating was observed to differ slightly from the measured temperature rise, $\Delta T_m$. For the three forging passes, $\Delta T_m$ was found to be 15.0, 38.0, and 12.0 °F respectively. From the simulations, $\Delta T_p$ was predicted to be 16.0, 24.0, and 22.0 °F. If the adiabatic heat gain for each pass is calculated from equation (2.2) based on the strain difference $\Delta \varepsilon$ between each pass, the difference may be the result of the model underpredicting and overpredicting surface strains in passes 2 and 3 respectively.

5.2.3.2 Billet Temperatures

A comparison of the predicted and the measured billet temperatures are shown in Figure 5.15. The DEFORM predictions were taken after simulating the 2 minute transfer period which was required to transport the billet from the rotary forge to the cut-off saw. A comparison of temperatures at the surface, mid-radius, and center shows that the correlation between the predicted and measured temperatures is not good. At the billet core, the difference in temperature is observed to be over 167 °C. This discrepancy may have been caused by the 15-30 second delay incurred in scanning the exposed surface as a result of the slug remaining in place after the cut-off operation. An additional
factor was that the thermal emissivity setting on the pyrometer was set at 0.8 which appears to be too high for the surface condition of the billet used.

<table>
<thead>
<tr>
<th>Pt.</th>
<th>Pyrometer</th>
<th>DEFORM</th>
<th>% Error</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>890.0</td>
<td>1061.0</td>
<td>16.1</td>
</tr>
<tr>
<td>B</td>
<td>852.0</td>
<td>1028.0</td>
<td>17.1</td>
</tr>
<tr>
<td>C</td>
<td>840.0</td>
<td>903.0</td>
<td>6.9</td>
</tr>
</tbody>
</table>

Figure 5.15. Comparison of billet temperatures predicted by DEFORM and measured by the infrared pyrometer. Temperatures were measured after radial forging at the cut-off saw.

In comparing the thermal data obtained from the billet surface before, during, and after forging, it was noted that there was a significant difference in the time rate of change in surface temperatures. During the initial transfer from the furnace to the GFM, the billet surface was observed to drop by $0.43 \, ^{\circ}C$ per
second. For the second transfer period corresponding to the billet transport from the GFM to the cut-off saw, the billet surface was found to decrease by more than 1.40 °C per second. The latter value may be seen to be 3.3 times higher. The large increase in cooling rate does not appear to be physically reasonable and indicates the presence of a large experimental error component in the pyrometer readings taken at the cut-off saw. This may have been the result of a change in emissivity of the newly generated surface, incorrect emissivity pyrometer setting, or delay in scanning the surface.

To test the possibility that the model was overpredicting temperatures in the core, the workpiece was simulated under pure heat transfer conditions for a period equivalent to the forging trial and transfer time. During the simulation, no deformation was assumed to occur. The temperature contours from the simulation showed that the DEFORM predictions remained significantly higher than the pyrometer readings. Hence, it must be concluded that the pyrometer readings are in error. Additional support for the accuracy of the DEFORM predictions is based on metallurgical kinetics observed in radial forging of Alloy 718. After forging, it has been found that the temperature is sufficiently high enough in the billet interior to allow continued static grain growth in the billet during air cooling (Jackman, 1993). Based on the temperature of 890 °C in the center of the billet measured by the pyrometer, significant grain growth after forging would not be feasible at this low temperature.
CHAPTER VI

PHYSICAL SIMULATION

6.1 Introduction

Microstructural prediction during deformation is a difficult science because it depends on the link between the phenomenological approach to plasticity, which has the scale length of feet, and the material microstructure, which is based on a scale length of a few microns. Because of the need to quantitatively link Alloy 718 microstructure to radial forging process variables, it was necessary to employ an experimentally based approach to physically simulate the radial forging process in real-time. By modeling multiple pass forging sequences, the evolution of Alloy 718 microstructure in response to radial forging process parameters could be simulated and used to build process-property relations.

In closed die forging processes; strain, strain rate, and temperature are the primary forging variables which need to be directly related to the microstructural features and mechanical properties of interest. However, for open-die forging processes such as radial forging, the effect of dwell time and multiple deformation cycles must also be considered. Based on this, it is
necessary to modify the single stroke compression test which is commonly used to study microstructural evolution.

6.2 Objectives

The purpose of performing the physical simulation was to achieve the following:

1) Generate a database for developing process-property relations in multiple pass radial forging of Alloy 718
2) Follow the evolution of grain size during multiple pass forging
3) Experimentally determine the effects of specific process parameters on grain size development

For the simulation study, the following process parameters were selected as the independent variables: strain per pass, number of passes, billet temperature, and time per pass. Based on discussions with forging industry personnel, these variables were previously identified as being significant process parameters which could be controlled during simulation. The range of parameters used in the study were based on currently used commercial processing windows for Alloy 718.

Because of the wide variety of microstructures which may be produced in Alloy 718 which include, uniform, duplex, necklaced, and warm-worked, a single microstructural feature such as grain size is generally not adequate to
represent the as-forged microstructure. Therefore, it is necessary to use recrystallized grain size and percent recrystallized to describe the microstructure. To characterize how mechanical properties are influenced by each process parameter, the microhardness of each sample was also measured.

To design the simulation study, a full $3^n$ factorial experimental design was selected. A factorial design refers to an experiment in which multiple factors are controlled and their effects at each of two or more levels are investigated (Natrella, 1985). A full factorial design was necessary due to the need to determine the nature of the relationship between the dependent and independent variables. A second reason for the factorial design was that it was also desired to statistically analyze each factor's effect and to determine if any multi-factor interactions were important. Taguchi techniques, while being more efficient and robust, can only be used to identify parameter settings or to screen different experimental factors. While a $2^n$ design will lead to a significant reduction in the number of design points required, the reduced data set will only provide information regarding linear trends with a resultant lack of information about higher order effects (Mead, 1988).

6.3 Experimental Setup

The simulations were conducted using a Gleeble 1500 dynamic thermo-mechanical testing machine. The Gleeble 1500 materials simulation system is comprised of a servo-hydraulic tensile/compression test frame equipped
with ram speed and resistance heating controls (Figure 6.1). The Gleeble may be programmed for sequential or simultaneous heating and deformation which enables simulation to be carried out under pre-programmed and controlled strains, strain rates, and temperatures. The ability to conduct compression tests at strain rates and temperatures typical of industrial processing conditions also make the Gleeble 1500 a useful tool for simulating multiple pass radial forging. Due to resistance heating of the specimen, a uniform temperature profile is produced in the longitudinal and transverse directions of the specimen and maintained during testing.

Table 6.1. Specifications of the Gleeble 1500 used to perform the radial forging simulations.

<table>
<thead>
<tr>
<th>Specification</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Maximum Load Capacity</td>
<td>80,064 N</td>
</tr>
<tr>
<td>Maximum Heating Rate</td>
<td>10,000 degrees C/sec.</td>
</tr>
<tr>
<td>Maximum Jaw Travel</td>
<td>101.6 mm</td>
</tr>
<tr>
<td>Maximum Velocity</td>
<td>3 cm/sec. (Low Speed)</td>
</tr>
</tbody>
</table>
Figure 6.1. Schematic of the Gleeble 1500 dynamic simulation system used to simulate microstructural evolution of Alloy 718 during multiple pass radial forging sequences.
The die stack consisted of one set of tungsten carbide platens (19 mm in diameter and 25.4 mm in height) mounted on 24 mm diameter Cu sheet. The die stack was mounted in 304 Stainless Steel jaws which were securely seated on the test frame. Use of the Cu sheets was necessary to ensure good electrical contact between the anvils and the electrical current supply. To minimize inhomogeneous deformation or barreling induced by friction, two sheets of graphite foil were placed on the tungsten carbide anvils to act as a lubricant. The effective value of the friction factor, $m$, found by FEM simulation, was found to be 0.3 which is typical for graphite based lubricants. While powdered glass is generally used as a lubricant during hot forging, it could not be used in the present study due to the requirement to maintain electrical conductivity between the anvils and the specimen. Because the small change in chemical composition as a result of carbon diffusion near the specimen ends was inconsequential, Ta sheets were not used to provide a diffusion barrier.

The specimens used in the Gleeble simulations were EDMed from mid-radius of each billet slice such that the compression axis was parallel to the longitudinal axis of each billet slice. The cylindrical specimens used during testing were 10 mm diameter and 10 mm in height. The diameter of the specimens was limited to 10 mm due to the 80,064 N maximum rated load capacity of the Gleeble 1500. To minimize scatter in the starting grain size, all samples were machined from a 50 mm band located at mid-radius of each billet slice (Figure 6.2). In order to prevent the compression specimens from
buckling during deformation, it was necessary to maintain an aspect ratio (height/diameter) of less than 1.5.

Figure 6.2. Schematic showing the location of the billet slice from which the compression specimens were machined.

6.4 Experimental Procedure

6.4.1 Characteristics of the Billet Material

For the simulation study, Alloy 718 billet material supplied by Carpenter Technology and Teledyne Allvac Vasco was used. The nominal chemical composition of each billet slice which was used is shown in Table 6.2. From
the chemical compositions listed, it may be noted that while there are slight differences in the nominal compositions for each billet slice, the chemistry is within acceptable limits for Alloy 718.

Table 6.2. Nominal chemical compositions of the (a) Carpenter Technology Pyromet 718 and (b) Teledyne Allvac Vasco Alloy 718 billet slices (in wt. %).

(a) Carpenter Technology Pyromet 718

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Mo</th>
<th>Co</th>
<th>Al</th>
<th>Ti</th>
<th>Cb</th>
<th>Fe</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.024</td>
<td>18.45</td>
<td>2.98</td>
<td>0.35</td>
<td>0.51</td>
<td>0.97</td>
<td>5.36</td>
<td>17.15</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

(b) Teledyne Allvac Vasco Alloy 718

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Mo</th>
<th>B</th>
<th>Al</th>
<th>Ti</th>
<th>Cb</th>
<th>Fe</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.035</td>
<td>17.80</td>
<td>2.90</td>
<td>0.004</td>
<td>0.70</td>
<td>1.00</td>
<td>5.40</td>
<td>18.30</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

The billet slice from Teledyne Allvac Vasco was representative of the billet preform material as it enters the rotary forging machine. The material from Carpenter Technology had been forged and heat treated and it was necessary to perform an additional 25 minute heat treatment at 1038 °C to obtain a starting grain size of ASTM 6.

The microstructure of the Teledyne Allvac billet slice consisted of a uniform recrystallized grains from a 355.6 mm diameter preform (Figure 6.3). An
examination of the as-received microstructure revealed that the grain size at the center was ASTM 5 while the grain size at mid-radius and the surface locations was ASTM 6. This difference is thought to be the result of thermal gradients formed in the ingot during preforming.

The as-received microstructure of the Carpenter Technology billet slice consisted of uniform, equiaxed, recrystallized grains of ASTM 8 throughout the slice. The microstructure of the compression specimens after heat treatment is shown in Figure 6.4.

![As-received microstructure of the Teledyne Allvac billet slice.](image)

Figure 6.3. As-received microstructure of the Teledyne Allvac billet slice.
6.4.2 Testing Procedure

Prior to testing, each specimen was first cleaned to remove the thin as-cast layer resulting from EDM machining and any oxide present on the surface from heat-treatment. This was necessary to ensure that a good junction would be formed between the specimen and thermocouple wires. Chromel-Alumel (Type K, 0.254 mm diameter) thermocouples having an error limit of 0.75% were percussion welded to the specimen at mid span perpendicular
to the compression axis to provide a closed loop to the Gleeble for temperature control. To prevent a thermal path between the thermocouple wires, which would result in a low temperature reading, the thermocouple wires were separated by 1-3 mm.

Each specimen was placed between the anvils and a small compressive force between 1 - 5 kg applied to hold the specimen in place during set-up and to ensure good electrical contact during heating. The test chamber was then evacuated to create a soft vacuum of 13.33 N/m² to prevent excessive oxidation of the tungsten carbide anvils during the simulation. Prior to performing each test, the specimen was heated to testing temperature over a three minute period and then held at temperature for seven minutes to equilibrate the specimen microstructure. Each test was then performed under computer control using a pre-programmed thermal and deformation sequence at a constant strain rate of 1.0 mm/mm/sec. All control programs were written in Gleeble Programming Language (GPL). Strain rate was controlled by specifying the amount of crosshead travel and time for each deformation cycle. Specimen temperature was controlled to within ± 6.0 °C. Following the simulation run, a three minute cool-down to room temperature was specified. In actuality the specimen cooled to 500-560 °C when the test was terminated. The average cooling rate was 6.0 °C/sec. during the cooling. However, once the temperature dropped below the δ solvus (968 °C), the precipitation of δ phase along the grain boundaries locked in the as-forged microstructure and inhibited grain growth.
6.4.3 Simulation of Multiple Pass Forging Sequences

In order to reproduce the as-forged microstructure, it was necessary to simulate the billet transfer and forging sequence in real time. The homogenization of the preform in the furnace was represented by heating the specimen for 10 minutes allowing the microstructure to equilibrate. The billet transfer is typically 60 seconds and was directly incorporated into the simulation sequence. To simulate microstructural evolution during radial forging, it was necessary to consider that the billet is incrementally forged and the time between successive deformation cycles may vary depending on the axial location of the material point (Figure 6.5). However, the material in mid-billet tends to experience nearly constant dwell times and was considered to be representative of the bulk conditions in the billet during forging. Therefore, it was decided to simulate the radial forging process using conditions representative of those at mid-length of the billet. Metallographic examination of billets converted by radial forging have shown that, with the exception of the billet ends, the variation in properties and grain size is minimal for billets under 30 feet in length (Jackman, 1991), supporting the approximation. To keep the number of simulations to within an acceptable level, the effect of post-forging cooling was neglected.
The assumptions which were used to simulate each multiple pass radial forging sequence were as follows:

1) deformation occurred under a constant strain rate of 1.0 mm/mm/sec.
2) effect of deformation (adiabatic) heating was negligible
3) pass reductions did not vary in each pass or between passes
4) chemical composition of the billet material is homogeneous
5) the time between deformation cycles remained constant
It has been found from FEM simulations and by calculating the average strain rate using equation (6.1) that strain rates during radial forging will typically be between 1-5 mm/mm/sec. during deformation.

\[
\dot{\varepsilon} = \frac{V}{h}
\]  

(6.1)

where \(V\) is the instantaneous die velocity calculated from equation (2.1) and \(h\) is the deforming height (radius) of the billet material. In the absence of significant die chilling, the deforming height, \(h\), will be equal to the instantaneous height of the deforming specimen. During each simulation, the strain rate was held constant by controlling the crosshead travel and time during each deformation step.

Due to the difficulty of correlating the cooling rates for the small samples used in the tests to that of the larger billet and the large number of heat-treatment/aging schedules used for Alloy 718, the study was limited to consideration of the as-forged microstructure to keep the number of tests to within a reasonable level.

The microstructural test methodology was validated by simulating a five pass sequence with post forging cooling based on data supplied by Teledyne Allvac Vasco in Monroe, North Carolina. The pass sequence is shown in Table 6.3. The actual grain size in the forged billet was found to be ASTM 6 while the grain size from the validation test was also found to be ASTM 6 confirming the validity of the assumptions and simulation methodology.
Table 6.3. Five pass sequence from Teledyne Allvac Vasco used to validate the Gleeble simulation methodology. The time between passes was 60 seconds.

<table>
<thead>
<tr>
<th>Pass #</th>
<th>Percent Reduction</th>
<th>Time/pass (seconds)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>15%</td>
<td>53</td>
</tr>
<tr>
<td>2</td>
<td>15%</td>
<td>60</td>
</tr>
<tr>
<td>3</td>
<td>20%</td>
<td>71</td>
</tr>
<tr>
<td>4</td>
<td>20%</td>
<td>74</td>
</tr>
<tr>
<td>5</td>
<td>20%</td>
<td>87</td>
</tr>
</tbody>
</table>

6.4.4. Experimental Conditions Used in the Study

To perform the metallurgical study, each independent variable was simulated at three equal levels (low, medium, and high) with the exception of the number of forging passes which was simulated at four levels. The level of each independent variable is listed in Table 6.4.

Table 6.4. Experimental conditions used in the Gleeble simulation of multiple pass radial forging.

Temperature, T: 954, 1010, and 1066 °C
Strain per pass, \( \Delta \varepsilon \): 0.0, 0.1, and 0.2 mm/mm/sec.
Number of passes, \( N_p \): 1, 2, 3, and 4
Time per pass, \( t_p \): 30, 60, and 90 seconds
Due to the problems in temperature control for specimens which had been severely deformed, it was necessary to limit the pass simulations to four passes with a maximum strain of 0.2 per pass imposed. In attempting to simulate sequences with a cumulative strain > 0.8, the specimen was observed to experience large temperature fluctuations after the fifth deformation cycle.

The timing sequence used during each simulation is listed in Table 6.5 showing the dwell times between sequential deformation cycles.

Table 6.5. Dwell time history between deformation cycles used in the Gleeble simulations. Each interpass dwell time includes the 10 second dwell period required to index the billet between passes.

<table>
<thead>
<tr>
<th>Block #</th>
<th>Transfer to GFM</th>
<th>Enter Pass 1</th>
<th>InterPass Dwell 1-2</th>
<th>InterPass Dwell 2-3</th>
<th>InterPass Dwell 3-4</th>
<th>Exit Pass</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>60</td>
<td>15</td>
<td>40</td>
<td>40</td>
<td>40</td>
<td>15</td>
</tr>
<tr>
<td>2</td>
<td>60</td>
<td>30</td>
<td>70</td>
<td>70</td>
<td>70</td>
<td>30</td>
</tr>
<tr>
<td>3</td>
<td>60</td>
<td>45</td>
<td>90</td>
<td>90</td>
<td>90</td>
<td>45</td>
</tr>
</tbody>
</table>
6.5 Metallographic Analysis

Each of the deformed Alloy 718 samples were sectioned, mounted, ground, polished, and etched prior to taking micrographs for use in grain size rating. Each of the samples were sectioned along a transverse plane at mid span perpendicular to the compression axis. The grain size rating was performed by comparing the sample microstructure from the micrographs to a known ASTM grain size standard.
CHAPTER VII

MATHEMATICAL MODEL BUILDING AND STATISTICAL ANALYSIS

7.1 Introduction

While the response of the microstructure to each of the forging parameters may be generally surmised from the Gleeble simulation data, no mathematical relationships may be directly inferred. At the present time, little information is presently available to link microstructure and process parameters together in a quantitative manner or to indicate the nature of such relationships in radial forging. In the absence of such information, the most rational approach is to use statistical methods to investigate the nature of the relationships and then attempt to build the model from a metallurgical foundation.

In developing quantitative relations based on experimental data, the statistical techniques known as Analysis of Variance (ANOVA) and Multiple Regression are useful tools for establishing empirical relations. An additional technique known as Response Surface Methodology is also useful in generating graphical contour plots to locate an optimum point. However, the intent of the present study was to characterize the metallurgical response variable for Alloy 718 over a wide range of processing conditions. Determining the optimum pass design...
was not considered to be a primary objective due to the range of billet microstructures which are forged in Alloy 718.

7.1.1 Previous Statistical Studies Based on Alloy 718

Nielsen et al. (1993) analyzed the significance of time, temperature, cooling rate, and deformation for simulated ring rolling process parameters for Alloy 718. To simulate the ring rolling mill, high temperature compression tests were performed using a Gleeble 1500 thermomechanical testing machine. The intent of their study was to identify the most important independent variables and their effect on the forged microstructure. To identify which process variables were significant, Taguchi techniques were used to analyze the experimental data. While Taguchi methods enable a reduced data set to be used, the main drawback of using such an analysis is that interaction terms are not explicitly considered (Montgomery, 1991). An additional consideration is that, depending on the particular design selected, the burden of determining which second order interactions are to be explicitly included in the analysis must be made prior to data collection. This may result in important interaction terms being neglected.

The results of their analysis showed that the significant factors influencing the recrystallized grain size at a 99% confidence level were the starting temperature and the interaction between starting temperature and cooling rate during deformation. Cooling rate was found to be significant at a 90% confidence level. For percent recrystallized, they reported that the starting temperature and percent reduction were significant at 99% and 90% confidence levels.
respectively. For Rockwell C hardness (HRC), their analysis showed that cooling rate was the only significant variable at a 99% confidence level.

7.2 Analysis of Variance and Regression Analysis

7.2.1. Background

To analyze the experimental data from the Gleeble simulations, the statistical technique known as Analysis of Variance (ANOVA) was used. ANOVA broadly refers to a collection of statistical techniques which have been developed for the analysis of quantitative responses from experimental situations (Devore, 1987). In analyzing experimental responses, ANOVA performs two primary functions. The first is that it separates the variation due to individual experimental factors into components and secondly, allows inferences to be made regarding the effects of each factor based on the variation between units (Mead, 1988). Knowledge of the total variation and the ability to partition it into components allow tests to be made regarding the validity of two opposing hypotheses concerning the experimental data by means of an F-Test. The null hypothesis under consideration states that the population means for all factors are equal (i.e. none of the factors under consideration are statistically significant in affecting the experimental response or output). The opposing hypothesis states that one or more of the factors does cause a statistically significant affect on the response variable.
In order to examine the amount of deviation that each experimental factor causes, it is necessary to calculate the Sum of Squares. For a four factor experiment with factors $i, j, k,$ and $l$ with $m$ replications, the total Sum of Squares is defined to be:

\[
SST = \sum_i \sum_j \sum_k \sum_m \left( y_{ijklm} - \bar{y}_{....} \right)^2
\]  

(7.1)

where $y_{ijklm}$ is the observation taken at the "$i$ th level of factor A", "$j$ th level of factor B", etc. and $\bar{y}_{....}$ is the "grand mean".

From equation (7.1), it may be seen that the total Sum of Squares (SST) is the sum of the difference between each data point and the grand mean, squared. The SST for a factorial experimental design may be further partitioned into components due to treatments (SSTR), replicates (SSR), and error (SSE):

\[
\sum_i \sum_j \sum_k \sum_m \left( y_{ijklm} - \bar{y}_{....} \right)^2 = \sum_i \sum_j \sum_k \sum_m \left( y_{ijkl} - \bar{y}_{...} \right)^2 + \sum_i \sum_j \sum_k \sum_m \left( y_{ijkl} - \bar{y}_{m} - \bar{y}_{...} \right)^2 + \sum_i \sum_j \sum_k \sum_m \left( y_{ijkl} - \bar{y}_{ijkl} - \bar{y}_{m} - \bar{y}_{...} \right)
\]  

(7.2)

The statistical model corresponding to the above four-factor factorial experiment is:

\[
y_{ijkl} = \mu + \alpha_i + \beta_j + \gamma_k + \delta_l + (\alpha\beta)_{ij} + (\alpha\gamma)_{ik} + (\alpha\delta)_{il} + (\beta\gamma)_{jl} + (\beta\delta)_{jl} + (\gamma\delta)_{kl} \\
+ (\alpha\beta\gamma)_{ijk} + (\alpha\beta\delta)_{ijl} + (\beta\gamma\delta)_{jkl} + (\alpha\gamma\delta)_{ikl} + (\alpha\beta\gamma\delta)_{ijkl}
\]  

(7.3)
The Sum of Squares for each term (factor and interaction) is found by partitioning the SST as follows:

\[ SSA = SST - SSB - SSC - \ldots - SSABCD \]  
\[ SSB = SST - SSA - SSC - \ldots - SSABCD \]  

(7.4)  
(7.5)

The individual Sum of Squares may also be calculated by an equation of the form:

\[ SSA = \frac{1}{bcdr} \sum_{i=1}^{a} T_{i}^{2} - C \]  

(7.6)

where \( T_{i} \) is defined as the total of all observations at a given level of factor \( i \) and where \( C \) is defined as:

\[ C = \frac{T_{i}^{2}}{abcdr} \]  

(7.7)

where a, b, c, and d are the number of levels for factors i, j, k, and l and r is the number of replications.

To test for the amount of variability in the data, it is necessary to perform an F-test. An estimate of the variation of each observation from the sample mean is obtained by calculating the Sum of Squares for each factor and dividing by the degrees of freedom. For Factor A, the mean squares is calculated by:
\[ MSA = \frac{SSA}{n - 1} \]  

where \( n \) is the total number of observations in the sample. By comparing the ratio of the Mean Squares (MS) for a particular factor to the Mean Square Error (MSE), a statistic is generated which can be used as the basis for conducting a F-Test to compare variability. If the F-ratio/statistic is found to be greater than the standard value for random experimental error obtained from a F-table, then the variation observed in the experiment can not be attributed to the inherent randomness in the data. Based on this, it must then be concluded that a given factor causes a statistically significant deviation from the sample mean.

Knowledge of each factor’s effect and the existence of significant interaction terms is of importance in developing process-property relations and deciding which factors need to be included. In testing for the presence of interaction terms, it is possible to determine if an individual factor is sensitive to or conditioned by the levels of other experimental factors. This information is also necessary for selecting the type of mathematical model (linear or non-linear) to be used as well as determining which terms should be included in the model.

Various relationships between the response or output variables and the input parameters are illustrated for a simple two factor experiment in Figure 7.1. It may be seen for the first case (Figure 7.1 (a)) that a change in either or both inputs does not change the response of the output (neither factor significant). In the second case (Figure 7.1 (b)) it may be seen that the response will change with factor A but is insensitive to any change in factor B. For the third case (Figure 7.1
(c) it may be seen that the response surface remains a plane and is sensitive to changes when one or both factors are changed (both factors significant). For the fourth case (Figure 7.1 (d)) where the response is assumed to be a function of both factors as well as a non-linear interaction term (higher order term), the response will be a contoured surface rather than a plane. Furthermore, it may be seen that the response of factor A is sensitive to or conditioned by the level of factor B as well as the converse.
Figure 7.1. Effects of individual factors and interaction terms on the shape of the response surface for a two factor experiment.
7.2.2 Regression Analysis

Once the significant factors have been identified through ANOVA, empirical relations may be generated through regression analysis.

For a four-factor experiment, the response, $y$, will be a linear combination of the experimental parameters:

$$y = f(x_1, x_2, x_3, x_4) \quad (7.9)$$

For a non-linear model a general quadratic of the form in equation 7.10 may be formed:

$$y = \beta_0 + \sum_i \beta_i x_i + \sum_i \beta_i x_i^2 + \sum_{ij} \beta_{ij} x_i x_j \quad (7.10)$$

To solve for the regressor coefficients, $\beta_0$, it is necessary to solve a system of equations of the form in equation (7.10).

7.2.3 Data Analysis

To perform the analysis of variance and regression calculations, the commercial statistical analysis program, Statistical Analysis System or SAS, was used (SAS Users Manual, 1989). To perform the data analysis, the general linear model (GLM) was selected. Selection of the GLM permitted the use of an unbalanced
model where the data points for the sub-solvus temperatures could be neglected. Due to the negligible possibility of significant high order terms, the effect of the third and fourth order interaction terms were neglected and pooled with the MSE term.

The four independent variables under consideration were: temperature \((T)\), strain per pass \((\Delta \varepsilon)\), number of passes \((N_p)\), and time between successive deformations \((t_p)\). The response variables obtained from the simulations were recrystallized grain size, percent recrystallized, and microhardness measurements.

In performing the analysis, it was assumed that all of the data populations are normally distributed, have equal variances, and each experimental observation is independent (Miller and Freund, 1977). However, the ANOVA methodology is also robust (Miller and Freund, 1977). The implication of this is that if either the assumption of normality or equality of variances is moderately violated, the analysis results will not be adversely impacted.

Prior to performing the ANOVA for each response variable, the adequacy of the assumptions regarding the data was checked. To verify the assumption of normally distributed means, a normal probability plot of the cumulative residuals was constructed. The residual for each data point refers to the unexplainable variation remaining after subtracting the treatment and block effects. The residual for each data, \(r_{ijkl}\) point was calculated using equation (7.11):
\[ r_{ij} = y_{ij} - \bar{y}_{i..} - \bar{y}_{..j} + \bar{y}_{..} + y_{..} \]  \hspace{1cm} (7.11)

A normal probability plot of the residuals for recrystallized grain size and microhardness were found to produce a straight line inferring that the normality assumption was correct. Percent recrystallization was found to produce a non-linear plot indicating a severe violation of the normality assumption.

To test for the equality of variances, a plot of residuals versus predicted values for each response variable was made. With the exception of percent recrystallized which was found to be highly correlated, recrystallized grain size and hardness was found to produce random plots. While the data for percent recrystallized could have been transformed to obtain equal variances, this would have resulted in the ANOVA results being applicable only for the transformed data. Because of the severe violation of both assumptions, the percent recrystallized data could not be analyzed using the ANOVA technique.

7.2.3.1 Analysis of Recrystallized Grain Size

In order to perform the analysis of variance for the recrystallized grain size, it was necessary to consider that the specimens machined from the Allvac billet slice, which were used in the simulation, contained a duplex grain structure. The duplex structure, consisting of 50% ASTM 8.5 and 50% ASTM 6 grains, was the result of a press forging operation used to upset the billet and develop a preform for radial forging. To perform the ANOVA analysis, it is necessary to represent
the grain size data as a single scalar value. While the microstructure may be represented by taking a weighted average of the grains present in each specimen, this may lead to a non-realistic representation of the microstructure (Peyroutou and Honnorat, 1991). Therefore, it was decided to use the recrystallized grain size as the response variable rather than a composite grain size. The presence of a duplex microstructure was most prevalent in the specimens which were simulated at the sub-solvus temperature of 954 °C. Due to the low temperature level at which sub-solvus deformation took place, little transformation or refinement in the grain size was observed. As a result, many of the specimens in Blocks 2 and 3 which had duplex starting microstructures retained these structures during the simulation. The samples simulated at super-solvus temperatures developed a uniform grain size during heating. Due to the possibility of confounding the data because of the two different temperature regimes (sub and super-solvus) and the mixed starting structure, the sub-solvus data points were eliminated from further consideration in the ANOVA.

The ANOVA results for the recrystallized grain size at super-solvus temperatures are shown in Table 7.1. The average recrystallized grain size for the simulated specimens was found to be ASTM 6.89. All significant F-values are denoted by a bullet (•). The F-values used in testing significance were obtained from a standard table of F-values for a confidence level of 0.99. For F-values which had denominators which were between standard tabulated values, \( F_\alpha(x,y) \) was obtained by linear interpolation where \( \alpha \) is the confidence level and \( x \) and \( y \) are the degrees of freedom in the numerator and denominator of the F-
value respectively. F-values for $F_{.99}(1, 40)$, $F_{.99}(2, 40)$, $F_{.99}(3, 40)$, $F_{.99}(4, 40)$, and $F_{.99}(6, 40)$ were found to be 7.31, 5.18, 4.31, 3.83, and 3.29 respectively.

Table 7.1. ANOVA table for recrystallized grain size data from the Gleeble simulation. All entries have been rounded to two decimal places.

<table>
<thead>
<tr>
<th>Source</th>
<th>DOF</th>
<th>Sum Squares</th>
<th>Mean Square</th>
<th>F-Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>tp</td>
<td>2</td>
<td>2.97</td>
<td>1.48</td>
<td>12.56*</td>
</tr>
<tr>
<td>T</td>
<td>1</td>
<td>46.72</td>
<td>46.72</td>
<td>395.76*</td>
</tr>
<tr>
<td>$\Delta \varepsilon$</td>
<td>2</td>
<td>104.22</td>
<td>52.11</td>
<td>441.38*</td>
</tr>
<tr>
<td>$N_p$</td>
<td>3</td>
<td>1.53</td>
<td>0.51</td>
<td>4.31*</td>
</tr>
<tr>
<td>$T/\Delta \varepsilon$</td>
<td>2</td>
<td>0.05</td>
<td>0.02</td>
<td>0.21</td>
</tr>
<tr>
<td>$T/N_p$</td>
<td>3</td>
<td>2.86</td>
<td>0.95</td>
<td>8.08*</td>
</tr>
<tr>
<td>$\Delta \varepsilon/N_p$</td>
<td>6</td>
<td>1.87</td>
<td>0.31</td>
<td>2.64</td>
</tr>
<tr>
<td>$tp/T$</td>
<td>2</td>
<td>4.34</td>
<td>2.17</td>
<td>18.38*</td>
</tr>
<tr>
<td>$tp/\Delta \varepsilon$</td>
<td>4</td>
<td>1.31</td>
<td>0.33</td>
<td>2.76</td>
</tr>
<tr>
<td>$tp/N_p$</td>
<td>6</td>
<td>0.53</td>
<td>0.09</td>
<td>0.75</td>
</tr>
<tr>
<td>Error</td>
<td>40</td>
<td>4.72</td>
<td>0.12</td>
<td></td>
</tr>
<tr>
<td>Total</td>
<td>71</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

The results of the F-test indicate that temperature and strain per pass are the most significant factors in affecting the recrystallized grain size by two orders of magnitude when compared to the remaining factors. Time per pass and the number of passes were also found to be significant at the 99% confidence level.
but not to the same degree as temperature and deformation. While temperature and number of passes are significant, the presence of the interaction term is also an important result. This indicates that the as-forged grain size will be affected by both temperature and the number of passes and that the effect of each factor is conditioned by the level of the other. This is possibly the result of the pass sequences simulated at moderate reductions and 1010 °C. Similar comments apply for the interaction between temperature and the time per pass.

The most interesting result is the apparent independence between deformation and temperature which was indicated by the ANOVA results. This may be explained by considering that in order for recrystallization to occur, a sufficient number of recrystallization nuclei must be generated through slip or deformation which is a function of the strain per pass only. The grain size after recrystallization is primarily a function of the density of recrystallization nuclei (Jonas and McQueen, 1975). Because the nuclei form from dislocation subgrains, the recrystallized grain size will be dependent on the degree of deformation applied. As a result, increasing the amount of deformation will lead to an increased number of sites for nuclei to form. The effect of increasing the temperature will be to increase the recrystallization kinetics via thermal energy but will not alone be sufficient to cause recrystallization to occur. The process of nucleation and growth are thermally activated and their rates increase with temperature (Jonas and McQueen, 1975).

SAS was also used to perform a multiple regression analysis to fit a second order polynomial model to the recrystallized grain size data used in the ANOVA
analysis (Table 7.1). To develop the mathematical model, only the significant terms from the ANOVA analysis in Table 7.1 were included. Tests for the normality and equality of variances assumptions were found to hold for the data used. A slight degree of correlation between the predicted grain size and residuals was detected during the equality of variance test but this will not significantly affect the results due to the robustness of the regression method.

To minimize the possibility that the time and temperature terms would dominate the regression equation, the input values were transformed such that the values were limited to between 0 and 3. The resulting relation for the recrystallized grain size (ASTM), $d_{gs}$, as a linear combination of the first and second order terms is shown in equation 7.12:

$$
d_{gs} = 5.753 + 1.063t^* - 0.014T^* + 1.427\Delta\varepsilon^* + 0.644N_p - 0.222T^*N_p - 0.521t^*T^* 
$$

where $t^*_p$ and $T^*$ are the transformed variables and:

$$
t^*_p = \frac{t_p}{30.0} \tag{7.13}
$$

$$
T^* = \frac{T - 1650.0}{100.0} \tag{7.14}
$$

The model was found to have an $R^2$ coefficient of 0.89 indicating that the model has good fit and accounts for 89% of the total variation in the data.
Further review of the data and ANOVA results suggested that a metallurgical based mathematical model for static recrystallization and grain growth would result in a more compact representation of the recrystallized grain size for supersolvus temperatures. The ANOVA results in Table 7.1 show that temperature and strain per pass were more significant by two orders of magnitude in influencing the grain size. Based on the dominance of temperature and strain, the remaining terms were neglected from consideration. It was assumed that the recrystallized grain size was a function of the strain per pass, $\Delta \varepsilon$, and the starting grain size for a given forging temperature. As indicated by the grain size data, the recrystallized grain size was assumed to be independent of the number of passes taken. While the effect of starting grain size was not explicitly considered in the statistical analysis, this was also determined to be an important factor. A smaller grain size will have a greater grain boundary surface area and a larger number of sites which will preferentially form recrystallization sites.

Based on these considerations, several equations were considered and a relation of the form shown in equation (7.15) was found to represent the steady-state recrystallized grain size (ASTM) for each experimental block. Good agreement was obtained after considering that the constant $a$ was also a function of the time per pass as well as the temperature

$$\log d_g = a \sqrt{\varepsilon_{\text{pass}}} + \log s$$  \hspace{1cm} (7.15)
where $s$ is the starting grain size (ASTM) of the specimen and $a$ is a constant which is a function of temperature. The form of equation (7.15) is similar to that proposed by Channon and Walker (1953) based on their study of the effect of grain size on the recrystallization of Cu and α-brass. By re-arranging equation (7.15) and eliminating logarithms, equation (7.16) may be seen to be similar to the classical equation for time-dependent grain growth in a single phase field for metals.

$$d_{ss} = se^{\left(\frac{a}{\sqrt{t_{mm}}}\right)}$$

(7.16)

A comparison of the predicted and experimental grain size results are shown in Figures 7.2 through 7.4 based on equation (7.15) using average values of $a$ which were calculated from the experimental results. Values of $a$ obtained from the study are shown in Table 7.2. From Figures 7.2 through 7.4, it may be seen that the relation provides a good fit with the experimental data. Equation (7.15) shows that the recrystallized grain size during radial forging is primarily sensitive to the strain per pass and moderately sensitive to the time per pass at a given temperature. Equation (7.15) was found to give good results for billet material which maintains a steady-state recrystallized grain size after the first or second pass. A second comparison based on the data of Mataya and Matlock (1989) for as-cast Alloy 718 showed that the equation does not yield good predictions for pass schedules where the grain size is continually being refined during each pass as is typical during conversion of ingots containing large, as-cast grains in the range of ASTM 0 and ASTM 1.
Table 7.2. Values of $a$ for each experimental block and temperature at supersolvus temperatures.

<table>
<thead>
<tr>
<th>Temperature</th>
<th>Block 1</th>
<th>Block 2</th>
<th>Block 3</th>
</tr>
</thead>
<tbody>
<tr>
<td>1010 °C</td>
<td>0.43</td>
<td>0.41</td>
<td>0.44</td>
</tr>
<tr>
<td>1066 °C</td>
<td>0.46</td>
<td>0.60</td>
<td>0.61</td>
</tr>
</tbody>
</table>

Figure 7.2. Comparison of predicted and experimentally measured grain sizes from the Gleeble tests for 30 second simulated pass times at 1010 °C and 1066 °C. The actual grain sizes were taken after simulating four passes.
Figure 7.3. Comparison of predicted and experimentally measured grain sizes from the Gleeble tests for 60 second simulated pass times at 1010 °C and 1066 °C. The actual grain sizes were taken after simulating four passes.
Figure 7.4. Comparison of predicted and experimentally measured grain sizes from the Gleeble tests for 90 second simulated pass times at 1010 °C and 1066 °C. All grain sizes were taken after simulating four passes.
7.2.3.2 Microhardness Analysis

Tests of the normality and equality of variance assumptions showed that both assumptions were valid for the microhardness readings. The ANOVA table for the room temperature microhardness data converted to Rockwell C Hardness (HRc) is shown in Table 7.3. The mean hardness value was found to be 18.121 HRc. All significant F-values in Table 7.3 are denoted by a bullet (•). F-values for $F_{.99}(2, 68)$, $F_{.99}(3, 68)$, $F_{.99}(4,68)$, and $F_{.99}(6, 68)$ were found to be 4.97, 4.12, 3.64, and 3.11 respectively.

The results from Table 7.3 indicate that the time per pass, temperature, and strain are significant at a 99% confidence level in influencing the microhardness in Alloy 718 during radial forging. The interaction between time per pass and strain per pass was also found to be significant at a 99% confidence level. Tests of the normality and equality of variance assumptions showed that both assumptions were valid for the data.
Table 7.3. ANOVA table for room temperature microhardness data converted to Rockwell C from the Gleeble simulation. All entries have been rounded to two decimal places.

<table>
<thead>
<tr>
<th>Source</th>
<th>DOF</th>
<th>Sum Squares</th>
<th>Mean Square</th>
<th>F-Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>$t_p$</td>
<td>2</td>
<td>983.87</td>
<td>491.93</td>
<td>25.90*</td>
</tr>
<tr>
<td>$T$</td>
<td>2</td>
<td>1217.58</td>
<td>608.79</td>
<td>32.06*</td>
</tr>
<tr>
<td>$\Delta \varepsilon$</td>
<td>2</td>
<td>1467.34</td>
<td>733.67</td>
<td>38.63*</td>
</tr>
<tr>
<td>$N_p$</td>
<td>3</td>
<td>97.48</td>
<td>32.49</td>
<td>1.71*</td>
</tr>
<tr>
<td>$T/\Delta \varepsilon$</td>
<td>4</td>
<td>138.30</td>
<td>34.57</td>
<td>1.82</td>
</tr>
<tr>
<td>$T/N_p$</td>
<td>6</td>
<td>87.21</td>
<td>14.54</td>
<td>0.77</td>
</tr>
<tr>
<td>$\Delta \varepsilon/N_p$</td>
<td>6</td>
<td>253.38</td>
<td>42.23</td>
<td>2.22</td>
</tr>
<tr>
<td>$t_p/T$</td>
<td>4</td>
<td>391.95</td>
<td>97.99</td>
<td>5.16*</td>
</tr>
<tr>
<td>$t_p/\Delta \varepsilon$</td>
<td>4</td>
<td>106.38</td>
<td>26.59</td>
<td>1.40</td>
</tr>
<tr>
<td>$t_p/N_p$</td>
<td>6</td>
<td>114.33</td>
<td>19.06</td>
<td>1.00</td>
</tr>
<tr>
<td>Error</td>
<td>68</td>
<td>1291.40</td>
<td>18.99</td>
<td></td>
</tr>
<tr>
<td>Total</td>
<td>107</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
The resulting form of the regression equation for the hardness was found to be:

\[ HR_c = 40.453 - 20.774t^* + 5.401T^* + 8.338\Delta\varepsilon^* + 0.989T^*\Delta\varepsilon^* \]
\[ -0.058T^* N_p - 2.340t^*_p T^* + 6.070\left(t^*_p\right)^2 + 1.300\left(T^*\right)^2 \]  

(7.16)

The \( R^2 \) value was found to be 0.647 indicating that the model was only moderately successful at predicting the microhardness and is able to account for only 65% of the variation in the data. Initially it was thought that the variability in the data may have been caused by readings being taken on different grain sizes for duplex microstructures. However, this was eliminated as a potential cause of the observed variation as a number of samples with uniform ASTM 6 grain sizes also displayed a large amount of variability in hardness readings between samples. In order to eliminate this problem, future work should make use of readings based directly on Rockwell measurements rather than microhardness readings. Additionally, the effect of cooling rate will also need to be included as a process parameter in future studies.
CHAPTER VIII

METALLOGRAPHIC EVALUATION AND DEVELOPMENT OF FORGING GUIDELINES

8.1 Introduction

After completing the simulation study and rating the as-forged grain size and percent recrystallized for each specimen, the metallographic data was evaluated to ascertain the effect of each independent variable on the as-forged microstructure. The emphasis in the metallurgical analysis was on interpreting the effect that each process parameter considered in the study had on the as-forged grain size. Because of the pronounced effect that second phase particles such as δ phase have on the grain refinement kinetics, the evaluation was divided into sub-solvus and super-solvus forging. Sub- and super-solvus forging refers to whether deformation was carried out in the presence or absence of δ phase.

Based on the grain size and percent recrystallized results obtained from the metallographical analysis, guidelines for designing multiple pass radial forging sequences for Alloy 718 preforms were developed.
8.1.1 Review of Previous Work

While some work has been performed in simulating microstructural evolution during radial forging (Weis et al., 1989; Mataya and Matlock, 1989; and Mataya et al., 1987), the range of conditions studied considered ingot microstructure and needs to be extended to include a smaller starting grain size and the effect of varied dwell times. Additionally, little work has been done to examine the response of mixed microstructures in Alloy 718 preforms after radial forging. Such microstructures are common in billets which have been press forged prior to radial forging to produce a preform.

Multiple stroke compression testing was performed at the Colorado School of Mines by Weis et al. (1989), Mataya and Matlock (1989), and Mataya et al. (1987) to study grain refinement in cast Alloy 718 ingots. The objectives of their study were to examine the flow curve behavior during each simulated forging cycle and the transformation phenomena which were involved in multiple pass radial forging. Due to the lack of strain softening evidenced by the flow curves, they reported that static, rather than dynamic, recrystallization was responsible for grain refinement during radial forging. Based on the grain sizes obtained from simulated billet positions at the leading, trailing, and mid-length, they reported that the most important parameters were temperature, dwell time, and strain per pass in descending order.
8.2 Metallographic Evaluation

A review of the recrystallized grain size and percent recrystallized data obtained from the Gleeble simulations indicated that the study results were consistent with observations made during production forging of Alloy 718 billet (Jackman, 1993). For the range of process conditions simulated, Alloy 718 was found to exhibit a high degree of formability with none of the samples being observed to fail or fracture. All metallographic observations were taken from a plane lying perpendicular to the compression axis at mid-span. The grain size ratings (in ASTM) and percent recrystallized for each simulated pass sequence is summarized in Tables 8.1 through 8.3 and Appendix B.

The effect of forging temperature and deformation was found to have a major influence on the as-forged grain size and percent recrystallized in all of the simulated sequences. Deformation was found to be the primary factor which was responsible for the as-forged grain size while temperature dictated the recrystallization kinetics. The time per pass and number of passes simulated was generally found to have a less significant effect on the as-forged microstructure. The effect of the temperature/strain interaction was also found to be important as it influenced the point at which steady-state grain size occurred. The temperature/time per pass interaction was significant in that increased time and higher temperature was more noticeable in causing grain coarsening at higher temperatures. These findings are all in agreement with the results which were predicted by the ANOVA test.
Table 8.1. As-forged grain sizes after simulating multiple pass radial forging for a 30 second pass time. The starting microstructure was uniform ASTM 6 taken from the Carpenter Technology billet slice after heat treating. An asterisk (*) denotes that the material was unrecrystallized after Gleeble testing.

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Low Reduction (0.0 strain per pass)</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>954°C</td>
<td>6*</td>
<td>6*</td>
<td>6*</td>
<td>6*</td>
</tr>
<tr>
<td>1010°C</td>
<td>6*</td>
<td>6*</td>
<td>6*</td>
<td>6*</td>
</tr>
<tr>
<td>1066°C</td>
<td>6*</td>
<td>5</td>
<td>5</td>
<td>5</td>
</tr>
<tr>
<td><strong>Medium Reduction (0.1 strain per pass)</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>954°C</td>
<td>6*</td>
<td>6*</td>
<td>6*</td>
<td>6*</td>
</tr>
<tr>
<td>1010°C</td>
<td>6.5</td>
<td>8.5</td>
<td>8.5</td>
<td>8.5</td>
</tr>
<tr>
<td>1066°C</td>
<td>7</td>
<td>7</td>
<td>6.5</td>
<td>7</td>
</tr>
<tr>
<td><strong>High Reduction (0.2 strain per pass)</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>954°C</td>
<td>6*</td>
<td>6*</td>
<td>9</td>
<td>9</td>
</tr>
<tr>
<td>1010°C</td>
<td>8.5</td>
<td>9</td>
<td>9</td>
<td>9</td>
</tr>
<tr>
<td>1066°C</td>
<td>8</td>
<td>8</td>
<td>8</td>
<td>8</td>
</tr>
</tbody>
</table>
Table 8.2. As-forged grain sizes after simulating multiple pass radial forging for 60 second pass time. The starting microstructure was duplex ASTM 6.0 and 8.5 grains taken from the as-received Teledyne Allvac billet slice. An asterisk (*) denotes that the material was unrecrystallized after Gleeble testing.

<table>
<thead>
<tr>
<th></th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Low Reduction (0.0 strain per pass)</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>954 °C</td>
<td>6, 8.5 *</td>
<td>6, 8.5 *</td>
<td>6, 8.5 *</td>
<td>6, 8.5 *</td>
</tr>
<tr>
<td>1010 °C</td>
<td>6</td>
<td>6</td>
<td>6</td>
<td>6</td>
</tr>
<tr>
<td>1066 °C</td>
<td>4</td>
<td>4</td>
<td>4</td>
<td>4</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th></th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Medium Reduction (0.1 strain per pass)</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>954 °C</td>
<td>6, 8.5 *</td>
<td>6, 8.5 *</td>
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<td>1066 °C</td>
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<table>
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<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
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</thead>
<tbody>
<tr>
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<tr>
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</tr>
<tr>
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<td>1066 °C</td>
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</tbody>
</table>
Table 8.3. As-forged grain sizes after simulating multiple pass radial forging for a 90 second pass time. The starting microstructure was duplex ASTM 6.0 and 8.5 grains taken from the as-received Teledyne Allvac billet slice. An asterisk (*) denotes that the material was unrecrystallized after Gleeble testing.

<table>
<thead>
<tr>
<th></th>
<th>Pass 1</th>
<th>Pass 2</th>
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<tr>
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<td>6, 8.5*</td>
<td>6, 8.5*</td>
<td>5</td>
<td>6, 8.5*</td>
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<tr>
<td>1010 °C</td>
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<td>6.5</td>
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<tr>
<td>1066 °C</td>
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<tr>
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<td>6, 8.5*</td>
<td>6, 8.5*</td>
<td>6, 8.5*</td>
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<tr>
<td>1010 °C</td>
<td>6</td>
<td>8.5</td>
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<tr>
<td>1066 °C</td>
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<td><strong>High Reduction (0.2 strain per pass)</strong></td>
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<tr>
<td>1066 °C</td>
<td>7</td>
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</tbody>
</table>
8.3 Sub-solvus Forging Simulations

When forging in the sub-solvus regime below 980 °C, both δ and γ" phase may be present in the microstructure. The main effect of δ phase in Alloy 718 is to limit the maximum grain size by pinning grain boundaries. Little mechanical strengthening results from δ phase and it may actually be detrimental to rupture strength if present in acicular form (Wilkinson, 1989). The presence of δ phase acts to limit the maximum grain size by exerting a drag force against the moving grain boundary. This drag force is in opposition to the force which gives rise to grain growth (Shewmon, 1983). In fact, this limiting behavior is exploited as the basis for mini-grain processing where a fine dispersion of δ phase is used to produce a very fine-grained microstructure. The effect of second phase particles in limiting grain size and growth is illustrated in Figure 8.1.

Figure 8.1. Effect of second phase particles in limiting grain size in polycrystalline metals (after Porter and Easterling, 1981).
While a finer grain size may be achieved by forging at a sub-solvus temperature, as opposed to forging at a super-solvus temperature, if the temperature is too low, $\gamma''$ will also be present in this temperature regime. Due to its coherence with the matrix and high-temperature strengthening characteristics, $\gamma''$ will result in much higher forging loads.

Due to the reduced driving force for recrystallization at sub-solvus forging temperatures, the engineering laws of static recrystallization predict that a greater amount of strain will be required for recrystallization to commence. This shift may be illustrated by considering the effect that temperature has on the critical strains which are required to initiate and to allow complete recrystallization to occur respectively. Based on the temperature-deformation-recrystallization curves for Alloy 718 shown in Figure 8.2, it may be seen that as the deformation temperature decreases, all other conditions held constant, the critical or minimum strain to initiate recrystallization ($T_{0\%}$) will increase. The amount of strain required to complete recrystallization ($T_{100\%}$) will also increase as temperature is decreased.
As the maximum strain per pass is limited by the machine capacity and is typically ≤ 0.25 for large diameter billets, it is unlikely that conditions will be favorable for dynamic recrystallization to occur. However, for multiple pass forging sequences where there is a significant dwell time between successive deformations, the effect of static recovery and recrystallization must be considered. Static recovery will cause well defined dislocation substructures to form through dislocation rearrangement and annihilation and consequently reduce the material's ability to recrystallize. No apparent change will be
observed in the grain size. Static recrystallization will transform the deformed grains into new, strain free grains via nucleation and growth processes.

8.3.1 Effect of Deformation at Sub-solvus Temperature Forging

Based on the simulations which were conducted, the number of passes did not have a significant impact by itself at sub-solvus forging temperatures. This is based on the observation that a critical level of strain needed to be accumulated prior to recrystallization and for completing recrystallization after each pass. Therefore, it is more relevant to consider the total or accumulated true strain which is a function of the strain per pass and the number of passes.

For samples which were not deformed (0.0 strain per pass), but were held at temperature only, no change in the microstructure was observed for samples from either billet slice. Due to the lack of grain growth evident during heating, thermal effects may be neglected during sub-solvus forging as grain growth will not occur due to the insufficient driving force necessary to overcome the pinning effects of δ phase. This indicates that the starting microstructure will be retained until a sufficient level of strain has been accumulated and recrystallization commences. Under these circumstances, differences in the starting billet microstructure will influence the as-forged microstructure as predicted by Burke and Turnbull's fourth law of recrystallization.

For moderate strain levels (0.1 strain per pass), recrystallization was not observed in any of the pass sequences which were simulated. Due to the
relatively low level of strain applied during each pass simulation, the accumulated strain was insufficient and remained below the critical strain for recrystallization in all of the cases considered. For strain levels below 0.2, little change was evident in the grain shape. At strains greater than 0.3, the grains began to demonstrate slight elongation in the working direction.

For heavy reductions (0.2 strain per pass), recrystallization was not observed for Blocks 1 (30 seconds per pass) and 2 (60 seconds per pass) until the third pass had been simulated which corresponded to a total strain of 0.6. Recrystallization was not observed until after the fourth pass in Block 3 (90 seconds per pass) when the specimen had been deformed to 0.8 total strain. Elongation of the grains in the working direction was also evident at the start of recrystallization (Figures 8.3 and 8.4). Typical grain size of the recrystallized material was found to be ASTM 9-10 though full recrystallization was not observed in any of the pass sequences which were simulated. This indicates that additional strain would be required to obtain 100% recrystallization of the material.

Examining the micrographs in Figures 8.3 and 8.4, it may be seen that a necklace structure consisting of fine recrystallized grains surrounding the coarser unrecrystallized grains was formed. The recrystallized grains were found to form preferentially at the boundaries of the existing grains. The finding that a significant amount of recrystallization occurred at sub-solvus temperatures in the present study is in contrast to the results reported by Mataya and Matlock (1989). In their study where ingot material with a starting grain size of ASTM 1.0 was used, it was reported that only 6% of the specimen had recrystallized after
simulating four passes at a strain of 0.23 per pass and a 60 second pass time. The significant amount of recrystallization observed in the present study indicates that the starting grain size will have a large influence on the feasibility of microstructural refinement when forging at sub-solvus temperatures. The effect of the smaller grain size will be to have a much greater percentage of grain boundaries which have a high free energy and thus are favorable sites for recrystallization nuclei to form.

Figure 8.3. Partially recrystallized microstructure after simulating a three pass forging sequence at 954 °C and an accumulated strain of 0.6 and 30 second dwell time between passes. The recrystallized grain size is ASTM 9.0 with 40% unrecrystallized ASTM 6.0.
Figure 8.4. Partially recrystallized microstructure after simulating a three pass forging sequence at 954 °C and an accumulated strain of 0.8 and 90 second dwell time between passes. The recrystallized grain size is ASTM 10.0 with 60% unrecrystallized ASTM 6.0.

The micrographs also reveal that once the starting material had recrystallized, the refined grains did not change with additional deformation which is consistent with static recrystallization. Due to the small amount of strain which was imposed during each simulated pass and the observation that the critical strain for Alloy 718 is between 0.25 and 0.3 for 954 °C at a strain rate of 1.0 sec⁻¹, it is unlikely that recrystallization could have occurred as the result of dynamic recrystallization.
8.3.2 Time per Pass at Sub-solvus Temperature

The effect of increasing the time per pass, or alternatively to increase the time between successive deformation cycles, is expected to enhance the effect of recovery. Due to the lower activation energy required, static recovery becomes operable at lower temperatures than static recrystallization. As a result of dislocation annihilation and rearrangement which occurs during recovery, the driving force for recrystallization will be reduced and the recrystallized grain size will increase.

While the effect of additional strain in subsequent passes was found to increase the percentage of material which had recrystallized, it is also necessary to consider the effect of recovery between each simulated deformation cycle. This is due to the fact that recovery and recrystallization can operate concurrently. Mataya and Matlock (1989) while finding no evidence of recrystallization in their simulation of ingot conversion, attributed the softening, $S$, in the yield stress between simulated passes to static recovery where $S$ is defined as:

$$S = \frac{\sigma_f - \sigma_r}{\sigma_f - \sigma_i} \times 100$$ (8.1)

and:  
- $S$ is the percentage of maximum softening which could be realized  
- $\sigma_f$ is the final stress prior to unloading  
- $\sigma_r$ is the 0.2% offset yield stress upon reloading after a given dwell time  
- $\sigma_i$ is the yield stress for the initial loading cycle
The largest amount of recrystallized material was observed after the fourth pass for the 30 second pass time (80%) and the minimum recrystallized amount was found to be for the 90 second pass time where only 40% of the material was found to have recrystallized after the fourth pass. No change in the grain size due to grain growth was observed for any of the pass times simulated.

The observation that the percent grain recrystallized increased as the time per pass, \( t_p \), decreases, neglecting the slight difference in starting microstructure in Block 1, indicates that static recovery and recrystallization are occurring simultaneously in the Alloy 718 specimens during the simulated passes. If static recrystallization alone were responsible, then the percent recrystallized should have been observed to increase with increasing time per pass. In the absence of static recovery, the volume fraction transformed will increase according to the classical sigmoidal curve shown in Figure 8.5. Under such conditions, the 30 second pass time would have demonstrated the minimum amount of percent recrystallized and the 90 second pass time would have demonstrated the maximum amount of recrystallization due to the longer dwell time. Therefore, static recovery must be occurring to reduce the total percent of material undergoing static recrystallization as the time per pass is increased.
Based on the above discussion, the primary consideration for sub-solvus forging is to specify a sufficient number of passes at the highest possible reductions to ensure that full recrystallization will occur in the billet. The number of passes required will also be a function of the starting grain size as well as the time per pass. Therefore, to ensure that full recrystallization takes place, it is desirable to minimize pass time in order to reduce the total number of forging passes which will be required to achieve a fully recrystallized microstructure.

8.4 Super-solvus Forging Simulations

When forging at temperatures above the upper solvus limit, the material will consist entirely of $\gamma$ phase. As a result, grain growth is not restricted by $\delta$ phase and recrystallization becomes more dominant than recovery. An additional
consequence is that the recrystallization kinetics are accelerated by the increased thermal energy. Grain growth will also become significant in the super-solvus regime.

After the Alloy 718 specimens had been heated to temperature and simulated under no deformation pass conditions, the grain size prior to deformation could be determined and the effect of hold time at each temperature could be ascertained. Knowledge of the starting microstructure prior to forging is necessary to ensure predictability and full control over the as-forged microstructure at each step.

For the super-solvus temperature of 1010 °C, the microstructure was found to consist of uniform ASTM 6 to 6.5 grains for all of the simulated hold times. Because δ phase is fully solutioned at 1010 °C, rapid grain growth is expected above this temperature due to the lack of restraint from δ and accelerated grain boundary mobility resulting from the increased thermal energy. Due to the reduction in free energy from minimizing the total grain boundary area and increase in average grain diameter, the smaller ASTM 8.5 grains in the Allvac billet slice coarsened to ASTM 6.0.

As the temperature was increased to 1066 °C, it was expected that larger grains would develop as a result of accelerated grain growth. With the exception of the first pass simulation in Block 1 which retained an ASTM 6.0 grain size, the remaining hold times in Block 1 were found to result in an ASTM 5.0 grain size. All simulated hold times in Blocks 2 and 3 were observed to produce a grain
size of ASTM 4.0. This demonstrates that a preform which initially contains a duplex microstructure from the press forging operation will coarsen to a uniform grain size. The slight difference in grain size between Block 1 and Blocks 2 and 3 may have been the result of slight compositional differences in the two billet slices.

8.4.1 Effect of Number of Passes

In comparing the effect of each pass sequence on the as-forged microstructure, it was found that the grain size did not vary appreciably once it had initially recrystallized. This invariance of the grain size with the number of passes is shown in Figures 8.6 - 8.8. Once the microstructure had fully recrystallized, the grain size was generally maintained throughout the remaining passes. Maintenance of the as-forged grain size is probably the result of repeated static recrystallization occurring (Mataya and Matlock, 1989; Weis et al., 1989). Given the high percentage of refinement for the low strain levels imposed, the results appear to confirm that a static transformation process was responsible for grain refinement. The critical strain for dynamic recrystallization to occur under the conditions simulated will be between 0.15 and 0.2 and full recrystallization will not be achieved until a total strain of 0.4-0.6 is reached. Full dynamic recrystallization at 0.2 strain can only take place at very high forging temperatures, typically in excess of 1166 °C (Garcia et al., 1990).
Figure 8.6. Effect of the number of simulated forging passes on grain refinement in Alloy 718 for super-solvus forging at 1010 and 1066 °C with 30 second dwell time between passes.
Figure 8.7. Effect of the number of simulated forging passes on grain refinement in Alloy 718 for super-solvus forging at 1010 and 1066 °C with 60 second dwell time between passes.
Figure 8.8. Effect of the number of simulated forging passes on grain refinement in Alloy 718 for super-solvus forging at 1010 and 1066 °C with 90 second dwell time between passes.
The results in Figure 8.6 indicate that the number of passes has very little effect on the as-forged grain size in multiple pass forging sequences when forging above the δ solvus. The percent recrystallized was not found to be affected by the number of simulated passes.

The number of passes which are necessary to obtain a uniform recrystallized grain size will also be dependent on the starting grain size and microstructure of the preform. In the study performed by Mataya and Matlock at Colorado School of Mines (1989), a four pass sequence was simulated using wrought Alloy 718 billet with a starting grain size of ASTM 1.0 (254 μm) at 0.25 strain per pass and a forging temperature of 1050 °C. The as-forged grain size in their study showed continuous refinement with each simulated pass. This would indicate that when forging an as-cast ingot that a larger number of passes will be required to obtain a steady-state grain size than for a wrought Alloy 718 preform which has a smaller starting grain size. This result is also consistent with the engineering laws of recrystallization which predict that for a larger starting grain size that additional work will be required to achieve the same recrystallized grain size obtained from a smaller starting grain size.

8.4.2 Effect of Strain per pass

Due to the repeated static recrystallization experienced by Alloy 718 during multiple pass forging at super-solvus temperatures, it is not meaningful to consider the total or accumulated strain. Therefore, it is necessary to consider
the level of strain imposed during each pass. The result of increasing the strain per pass will be to form a greater dislocation density with a greater driving force for recrystallization and additional sites for nucleation. If \( \bar{d} \) is the mean grain diameter, then:

\[
\bar{d} = \left( \frac{G}{N} \right)^{0.25}
\]  (8.2)

where \( G \) is the velocity at which the transforming boundary moves and \( N \) is the number of nuclei formed per unit time. While \( G \) and \( N \) will increase with temperature (Shewmon, 1983), the number of potential sites is likely to be a function of deformation or the strain per pass. Based on equation (8.2), it is apparent that factors which result in an increase in \( N \) will cause a reduction in the recrystallized grain size. This includes a smaller starting grain size and the amount of applied strain.

The effect of increasing the strain per pass was found to result in a finer recrystallized grain size at both temperatures as shown in Figure 8.9. Microstructures for 0.1 and 0.2 strain per pass are shown in Figures 8.10 and 8.11. The result of increasing the strain applied during each simulated pass from 0.1 to 0.2 was found to slightly promote the percent recrystallized but not to a very noticeable extent.
Figure 8.9. Relationship of recrystallized grain size with applied strain per pass at (a) 1010 °C and (b) 1066 °C. All data points were taken after simulating four forging passes.
Figure 8.10. Recrystallized microstructure after simulating three passes at 1066 °C and 0.15 strain per pass with 60 seconds per pass. The microstructure is 100% ASTM 6.5.

Figure 8.11. Recrystallized microstructure after simulating three passes at 1066 °C and 0.30 strain per pass with 60 seconds per pass. The microstructure is 100% ASTM 8.0.
Therefore, it is advisable to forge using maximum strain per pass to obtain the finest grain size possible. This result is also confirmed by the laws of recrystallization.

8.4.2.1 Equivalence of Pass Schedules

Based on a comparison of equivalent total reductions, it is preferable to take a small number of passes at heavier forging reductions rather than a large number of passes at lighter reductions to obtain a finer grain size. The results in Figure 8.12 indicate that a finer recrystallized grain size will be achieved for heavier reductions. This is a consequence of the fact that deformation is the dominant factor in refining the microstructure and increased strain will reduce the recrystallized grain size for a given temperature. This result is also inferred by the ANOVA results which indicated that deformation dominates the as-forged grain size and is not coupled with the number of passes. Therefore, the degree of grain size refinement will be deformation driven.
Figure 8.12. Comparison of as-forged grain size for equivalent total forging reductions at (a) 0.2 total effective strain and (b) 0.4 total effective strain.
8.4.3 Effect of Time per Pass

In view of the dominance of deformation and temperature in controlling the microstructure, it might be expected that the grain size would be less sensitive to the time per pass. For the pass sequences simulated at 1010 °C, little difference in grain size was noted for increased pass times. However, as temperature was increased to 1066 °C there was a slight increase in grain size by about 1 ASTM as the time per pass was increased from 30 to 90 seconds. Therefore, slightly larger grains are expected to result from longer or increased pass times at higher temperatures as a result of the increased thermal energy which is available for grain growth. This interaction is the result of the temperature/time per pass interaction term which was predicted by the ANOVA test and is shown in Figure 8.13.
Additional evidence of static recrystallization may also be seen from the percent recrystallized. Neglecting grain size differences, Block 1 had the most instances of partial recrystallization. Block 2 had less incomplete and partial recrystallization and Block 3 had no partial or incomplete recrystallization after the second pass. The number of passes which were required to achieve full recrystallization was found to increase with decreasing pass time as shown in
Table 8.4. This was seen by the reduction in the percent and number of unrecrystallized grains remaining as the time per pass was increased. The effect of increasing time per pass was found to allow a greater amount of the deformed material to recrystallize. The effect of increasing time per pass was to enable 100% recrystallization after the fourth pass and eliminate the unrecrystallized grains which were observed in some of the Block 2 specimens.

### Table 8.4. Number of forging passes required to obtain 100% recrystallization.

<table>
<thead>
<tr>
<th>Pass Time (seconds)</th>
<th>Strain/Pass</th>
<th>Pass for 100% Recrystallization</th>
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<tr>
<td></td>
<td></td>
<td>1010 °C</td>
</tr>
<tr>
<td>30</td>
<td>0.1</td>
<td>4</td>
</tr>
<tr>
<td></td>
<td>0.2</td>
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<tr>
<td></td>
<td>0.2</td>
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</tr>
<tr>
<td>90</td>
<td>0.1</td>
<td>3</td>
</tr>
<tr>
<td></td>
<td>0.2</td>
<td>2</td>
</tr>
</tbody>
</table>

8.4.4 Effect of Forging Temperature

The effect of temperature was predicted to be one of the dominant factors in the ANOVA. The effect of temperature on the grain size may also be seen by the Zener-Holloman parameter, Z, which predicts that a lower value of Z (increased temperature) will cause a coarser grain size. This trend was also found to hold for the data obtained from the simulation study. A reduction in temperature of
56 °C from 1066 to 1010 °C was found to result in a finer as-forged grain size as shown in Figure 8.14. Microstructures for 0.2 strain per pass at each of the temperatures are shown in Figures 8.15 and 8.16. The average reduction in grain size was found to be on the order of 2.5 ASTM. The effect of increasing the forging temperature was to promote the recrystallization kinetics as evidenced by the percent recrystallized and the number of unrecrystallized grains. By increasing the temperature, percent recrystallized was substantially completed by the second pass simulation.

There was also a moderate effect on the recrystallized grain size where the recrystallized grain size after the first simulated pass was slightly larger than the remaining passes at 1010 °C forging temperature. In comparison, the recrystallized grain size was constant after the first simulated pass for all forging passes carried out at 1066 °C.
Figure 8.14. Grain coarsening resulting from increased forging temperature. All data points were taken after simulating four passes.

Figure 8.15. As-forged microstructure at 1010 °C. The microstructure is 100% ASTM 9.5. The data points were taken after a four pass simulation at 0.20 strain per pass with 90 seconds per pass.
Figure 8.16. As-forged microstructure at 1066 °C. The microstructure is 100% ASTM 7.0. The data points were taken after a four pass simulation at 0.20 strain per pass with 90 seconds per pass.

8.5 Guidelines for Radial Forging Pass Design

Based on the study results, it was found that the laws of recrystallization are valid for qualitative predictions regarding the effect of temperature and deformation in multiple pass radial forging of fine-grain Alloy 718 preforms. On the basis of the simulation study results, additional guidelines were developed which may be applied to the design of multiple pass radial forging sequences for fine grain Alloy 718 billet. However, it should be noted that the guidelines presented do not consider the effect of grain growth during post-forging cooling or non-uniform recrystallization resulting from thermomechanical gradients for large diameter billets.
Based on the percentage of material transformed, it is better to forge at super-solvus temperatures than sub-solvus temperatures as a more uniform grain size and full recrystallization are likely to occur though the resulting grain size will be slightly larger. Forging at super-solvus temperatures should reduce the number of forging passes which are required to refine and obtain a uniform forged microstructure in the billet.

For sub-solvus forging, the following guidelines are applicable:

1) For sub-solvus forging, an increased number of forging passes and heavy forging reductions will result in a very fine recrystallized grain size
2) Starting grain size will dictate the feasibility of recrystallization for sub-solvus forging and a small starting grain size will promote recrystallization

For super-solvus forging, the following criteria may be used:

1) For super-solvus forging, once a uniform recrystallized microstructure is formed, additional forging passes will not result in additional grain refinement
2) A finer as-forged grain size will result as forging temperature is decreased
3) A finer grain size will result with increased forging reduction
4) To obtain a finer grain size, it is preferable to minimize the number of forging passes and maximize the reductions taken
5) When forging above the δ solvus, the number of forging passes necessary to
promote full recrystallization will decrease with an increase in forging reduction.

6) Once the microstructure has transformed completely to the recrystallized grain size, additional forging passes will result in the maintenance of the as-forged grain size.

7) As forging temperature is increased above the δ solvus, the effect of grain coarsening will become more significant.
Chapter IX

FUTURE WORK AND RECOMMENDATIONS

The original intent of this work was to develop a foundation for integrating FEM modeling and process-property relations into an integrated computer based module for predicting microstructure in Alloy 718. At the start of the project, little published work existed in either area pertaining to incremental forging processes such as radial forging. While substantial progress has been made towards modeling the thermomechanical history during radial forging and in predicting the as-forged grain size, there are a number of tasks in each area which need to be addressed in future work. In regards to FEM modeling, additional work is needed to improve the model efficiency and accuracy before it can be deployed for practical use in the forging plant. The microstructural modeling study needs to include the effect of post-forging parameters in order to yield predictions which may be directly applied to the billet.

9.1 FEM Modeling Work

Substantial progress has been made in modeling the radial forging process using the Updated Lagrangian code DEFORM. Prior to this study, the simulation of multiple passes was generally impractical due to the time and effort required for preprocessing. The preprocessing procedure has now been automated using
command language routines so that complete multiple pass sequences may be simulated with relatively little user interface effort.

A second significant obstacle which existed at the start of the study was in validating the predicted temperature and strain profiles in the billet. Based on the work performed in this study, a simple methodology has been developed which can be used to determine the effective strain profile as a function of radial position in the billet. The validation efforts showed acceptable correlation between the model and the actual forged billet considering that the process was approximated by an axisymmetric deformation model. While a three-dimensional model will result in improved predictions, the lack of automatic remeshing and the computation time required at present would make it impractical for use in simulating multiple pass sequences at the plant level. Therefore, a dedicated special purpose model which include the material spread between the tools will be more appropriate to concentrate the future efforts on.

Due to the steady-state nature of the process, as confirmed in this study, the possibility exists for using a reduced workpiece length to represent the billet. This would reduce the number of redundant calculations needed during each simulation for a Lagrangian based formulation. However, the selection of the kinematic boundary conditions imposed at the ends will be a determining factor in the model accuracy. Given the significantly reduced computational time required for dedicated, special purpose 2-1/2 D codes such as TASKS (Shivpuri et al., 1990), this formulation should be considered for radial forging as well.
A second possibility is to simulate only a partial pass whereby the first one-third to one-half of each pass is modeled under deformation and thermal conditions. Heat transfer only could be modeled to account for the cumulative dwell time for the remainder of the current pass and the time during the subsequent pass required for the forging tools to return to the specified axial position. Then both deformation and thermal analyses could be resumed to model the thermomechanical behavior of the region being considered. Due to the steady-state nature of the thermomechanical behavior, only a selected section of the workpiece would need to be considered as the remainder of the workpiece would have the same thermomechanical history. Neglecting deformation in the remainder of the workpiece would be feasible due to the steady-state strain fields in the billet. Due to the uni-dimensional heat flow in the workpiece as a result of the large aspect ratio, heat flow was found to be negligible in the axial direction based on DEFORM simulation results. Therefore it may be sufficient to model heat transfer only which has markedly reduced computational requirements in comparison to a combined thermal and deformation analysis.

While the use of optical pyrometers for temperature measurement effort was not as successful, this area warrants additional effort to improve the methodology. The use if optical pyrometers is probably the only feasible method at present to measure the thermal profile as a result of safety considerations. An additional item which needs to be characterized is the emissivity of the billet surface to determine the effect of deformation on the rate of heat loss. By accurately characterizing the total emissivity, the surface temperature correlation can be further improved.
9.2 Process-Property Relations

When this study was initiated, very little information was available which described the effect of process parameters or relations for predicting the microstructure for incremental forging processes such as radial forging. Based on the Gleeble simulation data, a quantitative relation was proposed which related the starting grain size and the strain per pass to the as-forged grain size under isothermal deformation conditions.

Because of the relatively light strains which are applied and the long dwell times between successive deformations, static recovery and recrystallization are the dominant transformation mechanisms involved in billet conversion. In predicting the microstructure, the grain size has been found to follow a relation which is consistent with static recrystallization behavior. The relation was found to give good correlation for Alloy 718 billet preform material by simulating radial forging using high temperature compression testing. When compared with forged billet, the simulation methodology was also found to yield excellent agreement with the grain size obtained from actual billet forging. Therefore, the simulation methodology and the predictive relation are valid for the processing window studied and represent substantial progress in this area. Because the bulk of Alloy 718 forging is done in the super-solvus region, this temperature regime is where future work should be focused on.
The same methodology which was employed to simulate the radial forging process using the Gleeble in the study should be maintained but the technique may need to be modified to improve the efficiency in the number of samples needed to do future studies as well as to expand its utility. Because static recrystallization behavior predominates, the study may be set-up to consider additional hold times as well as cooling rate while neglecting the number of passes on the grain size. This would reduce the number of factors to be considered and keep the study within a manageable level.

While the simulation methodology was found to yield results which were in excellent agreement when the cooling rate was simulated, this parameter will also need to be considered in future work. By including the effect of post forging cooling, it will be possible to directly apply the relation to actual pass design. Due to the significant time at which interior points in the billet remain above the δ solvus after forging, grain growth will continue and will affect the final grain size. Two possible approaches may be considered for incorporating the effect of cooling rate. Grain coarsening may be considered from the standpoint of adding an additional term to the relation developed in Chapter 7 or it may prove necessary to include it in the relation directly. The cooling curves could be obtained by FEM simulation and should be performed for a set of different billet diameters. This cooling rate information could then be directly incorporated into the Gleeble simulations.

The effect of adiabatic heating may be a factor for heavy reductions and this may need to be included in the Gleeble simulations as well. However, this will also
need to found through FEM simulations as the amount of deformation heating can not be directly calculated for each pass.

The use of experimental design techniques and methods in metallurgical studies was also found to be very useful in this study. Experimental design provides a quantitative means for identifying significant factors and interactions, provided that the methods are properly understood and applied. Regression techniques are useful for a first approximation of the process-property relations but such relations are better developed using metallurgical based relations. All future work should be approached using experimental design context.

9.3 Closure

Once progress has been made in terms of completing the above items, it will then become appropriate and worthwhile to begin the task of linking the FEM and process-property relations into a seamless, computer integrated module for application in pass design. At the present time, additional understanding is required in terms of both thermomechanical processing and microstructural evolution before process and microstructural modeling can effectively be combined into one predictive model for radial forging of nickel-based superalloys.
LIST OF REFERENCES


Evans, R.W., Metallurgical Modeling of Superalloy Disc Isothermal Forgings, Aerospace Materials Process Modeling, AGARD Conference Proceedings No. CP-26, Cesme, Turkey, October 1987, pp. 4-1 to 4-17.


Tzseng, T.C., Ph.D. Dissertation, Department of Mechanical Engineering, University of California, Berkeley, 1987.


Appendix A

Description of Labels Used to Identify Alloy 718 Samples Simulated Using the Gleeble 1500 System
Each of the Alloy 718 specimens simulated on the Gleeble 1500 is identified by a label of the form, $BP_S1S2S3$ to identify the experimental block and the thermomechanical testing sequence used. The prefix $BP$ refers to the block of material from which the sample was taken as well as the assigned time for the completion of each pass as listed in Table A.1.

The suffix $S1S2S3$ refers to the particular thermomechanical sequence applied to the specimen. $S1$ refers to the number of passes applied which is an integer value ranging from one to four. $S2$ refers to the applied strain per pass which is denoted by $L$ ($= 0.0 \text{ mm/mm}$), $M$ ($= 0.1 \text{ mm/mm}$), and $H$ ($= 0.2 \text{ mm/mm}$). $S3$ is the simulation temperature used and is also denoted by $L$ ($= 954 \text{ degrees C}$), $M$ ($= 1010 \text{ degrees C}$), and $H$ ($= 1066 \text{ degrees C}$).

An example of how to use the labeling system, consider the specimen designation $B2_4HL$. This specimen was taken from the Teledyne Allvac Alloy 718 billet slice where the assumed time to complete a single pass is 60 seconds ($B2$). Four passes were simulated (designated by 4) with 0.2 strain per pass (designated by $H$) at 954 degrees C (designated by $L$).

Table A.1. Description of the notation used to represent the experimental blocks used in the metallurgical simulations.

<table>
<thead>
<tr>
<th>Block</th>
<th>Description</th>
<th>Pass Time</th>
</tr>
</thead>
<tbody>
<tr>
<td>B1</td>
<td>Block 1 Carpenter Billet Slice, Pass Time = 30 seconds</td>
<td></td>
</tr>
<tr>
<td>B2</td>
<td>Block 2 Teledyne Allvac Billet Slice, Pass Time = 60 seconds</td>
<td></td>
</tr>
<tr>
<td>B3</td>
<td>Block 3 Teledyne Allvac Billet Slice, Pass Time = 90 seconds</td>
<td></td>
</tr>
</tbody>
</table>
Appendix B

Metallurgical Data for Alloy 718 Samples Simulated
Using the Gleeble 1500 System
Table B.1. Block 1 grain sizes after simulating multiple pass radial forging (an * denotes unrecrystallized grain after Gleeble testing). The time per pass was 30 seconds. Samples were taken from the Carpenter Technology billet slice.

<table>
<thead>
<tr>
<th>Low Reduction (0.0 strain per pass)</th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td>954 °C</td>
<td>6</td>
<td>6</td>
<td>6</td>
<td>6</td>
</tr>
<tr>
<td>1010 °C</td>
<td>6</td>
<td>6</td>
<td>6</td>
<td>6</td>
</tr>
<tr>
<td>1066 °C</td>
<td>6</td>
<td>5</td>
<td>5 ALA 3</td>
<td>5 ALA 3</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Mid Reduction (0.1 strain per pass)</th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td>954 °C</td>
<td>6</td>
<td>6</td>
<td>6</td>
<td>6</td>
</tr>
<tr>
<td>1010 °C</td>
<td>6.5</td>
<td>6, 4% 8.5</td>
<td>8.5, 5% 6*</td>
<td>8.5 ALA 6*</td>
</tr>
<tr>
<td>1066 °C</td>
<td>7, 1% 5 ALA 2</td>
<td>7, 1% 5</td>
<td>6.5</td>
<td>7</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>High Reduction (0.2 strain per pass)</th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td>954 °C</td>
<td>6</td>
<td>6</td>
<td>9, 40% 6</td>
<td>9, 20% 6</td>
</tr>
<tr>
<td>1010 °C</td>
<td>7, 30% 8.5</td>
<td>9, 10% 6</td>
<td>9, 5% 6</td>
<td>9 ALA 6</td>
</tr>
<tr>
<td>1066 °C</td>
<td>4, 30% 7</td>
<td>7</td>
<td>7</td>
<td>7</td>
</tr>
</tbody>
</table>
Table B.2. Block 2 grain sizes after simulating multiple pass radial forging (an * denotes unrecrystallized grain after Gleeble testing). The time per pass was 60 seconds. Samples were taken from the Teledyne Allvac billet slice.

<table>
<thead>
<tr>
<th></th>
<th>Low Reduction (0.0 strain per pass)</th>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Pass 1</td>
<td>Pass 2</td>
<td>Pass 3</td>
<td>Pass 4</td>
</tr>
<tr>
<td>954 °C</td>
<td>6,50% 8.5</td>
<td>6,50% 8.5</td>
<td>6,50% 8.5</td>
<td>6,50% 8.5</td>
</tr>
<tr>
<td>1010 °C</td>
<td>6,3% 8</td>
<td>6</td>
<td>6</td>
<td>6</td>
</tr>
<tr>
<td>1066 °C</td>
<td>4</td>
<td>4</td>
<td>4</td>
<td>4</td>
</tr>
</tbody>
</table>

|                   | Mid Reduction (0.1 strain per pass) |                      |                      |                      |
|                   | Pass 1 | Pass 2 | Pass 3 | Pass 4 |                      |
| 954 °C            | 6,50% 8.5 | 6,50% 8.5 | 6,50% 8.5 | 6,50% 8.5 |                      |
| 1010 °C           | 6,50% 8.5 | 6,55% 8.5 | 8.5, 1% 6 | 8.5 ALA 6* |                      |
| 1066 °C           | 6.51% 5  | 6.5 ALA 5 | 6.5 | 6.5 |                      |

|                   | High Reduction (0.2 strain per pass) |                      |                      |                      |
|                   | Pass 1 | Pass 2 | Pass 3 | Pass 4 |                      |
| 954 °C            | 6,50% 8.5 | 6,50% 8.5 | 6,50% 8.5 | 6,50% 8.5 |                      |
| 1010 °C           | 8.3% 6*  | 9.5, 3% 6* | 9.5 ALA 6* | 8.5 |                      |
| 1066 °C           | 4.30% 7  | 7      | 7      | 7      |                      |
Table B.3. Block 3 grain sizes after simulating multiple pass radial forging (an * denotes unrecrystallized grain after Gleeble testing). The time per pass was 90 seconds. Samples were taken from the Teledyne Allvac billet slice.

<table>
<thead>
<tr>
<th>Low Reduction (0.0 strain per pass)</th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td>954 °C</td>
<td>6, 50% 8.5</td>
<td>8.5 ALA 6</td>
<td>5, 5% 9</td>
<td>8.5, 30% 6 ALA 4</td>
</tr>
<tr>
<td>1010 °C</td>
<td>6, 3% 8</td>
<td>6.5</td>
<td>6.5</td>
<td>6</td>
</tr>
<tr>
<td>1066 °C</td>
<td>4</td>
<td>4</td>
<td>4</td>
<td>4</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Mid Reduction (0.1 strain per pass)</th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td>954 °C</td>
<td>6</td>
<td>8.5, 40% 6</td>
<td>6, 2% 8.5</td>
<td>6, 10% 8.5</td>
</tr>
<tr>
<td>1010 °C</td>
<td>6, 2% 8.5</td>
<td>6, 50% 8.5</td>
<td>8.5</td>
<td>8.5</td>
</tr>
<tr>
<td>1066 °C</td>
<td>6, 3% 4</td>
<td>6, 2% 4</td>
<td>6</td>
<td>6.5</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>High Reduction (0.2 strain per pass)</th>
<th>Pass 1</th>
<th>Pass 2</th>
<th>Pass 3</th>
<th>Pass 4</th>
</tr>
</thead>
<tbody>
<tr>
<td>954 °C</td>
<td>6, 15% 8.5</td>
<td>6.5</td>
<td>6, 10% 8</td>
<td>6, 40% 10</td>
</tr>
<tr>
<td>1010 °C</td>
<td>8.5, 15% 6*</td>
<td>9</td>
<td>9</td>
<td>9</td>
</tr>
<tr>
<td>1066 °C</td>
<td>7, 30% 4</td>
<td>7</td>
<td>7</td>
<td>7</td>
</tr>
</tbody>
</table>
Appendix C

Slab Analysis Equations for Predicting Forging Load in Radial Forging
C.1 Introduction to RAFSTRAN

Based on the slab analysis method, the computer code RAFSTRAN (RAdial Forging STResS ANalysis) was developed. The program is intended for use on IBM personal computers and compatibles with graphics capabilities, though the program can be modified to run without graphics if desired. The only system requirement for using RAFSTRAN is that a BASIC interpreter written for MS-DOS 2.2 or GWBASIC 3.11, or higher, be used.

The software was written using the BASIC computer language. This is due to the simplicity and the minimal hardware and software requirements for running most BASIC programs. The program file structure is shown in Figure C.1.

The program inputs are die geometry, starting and finished workpiece diameter/height, and average material flow stress as a function of strain rate and temperature. Both axisymmetric and plane strain analyses may be performed. In its present configuration, RAFSTRAN can analyze tools having up to three inlet angles. Input data can be manually entered by the user from a keyboard or called from an existing database. Either SI or English system of units may be used to input data to RAFSTRAN as long as the units used are consistent. After performing the analysis algorithm the program outputs forging load (total and contribution from each zone), estimated neutral plane location, and the "h/L" ratio (where "h" is the average workpiece of the diameter/height
and "L" is the die contact length). The distribution of the normal stresses along the die-material interface for each hammer may also be displayed graphically.

Figure C.1. RAFSTRAN program files and structure.
C.2 Development and Formulation of the Slab Method for Radial Forging
Using Compound Angle Dies

Some of the important process variables which need to be considered in a slab analysis of the radial forging process are:

1) Applied axial front and back-pull forces
2) Friction at the die-workpiece interface (friction factor assumed)
3) Die geometry
4) Flow stress as a function of strain rate and temperature

In performing the slab analysis, a determination must be made between employing a plane-strain or axisymmetric analysis. It may be argued that both methods can provide a reasonable description of the radial forging process under idealized conditions. In developing the differential equations for the analysis, both axisymmetric and plane-strain analyses were carried out and implemented in the program.

The following discussion details the development of the equations used to solve for the die stresses under axisymmetric conditions. Polar coordinate notation is used to facilitate the discussion. Due to the axisymmetric conditions, the dies are assumed to be in contact with the entire workpiece circumference along the length of the dies. Thus the small clearances between the dies are neglected in the analysis. It is further assumed that the Von-Mises yield criterion adequately describes the flow stress of Inconel 718 under hot-working conditions. The stress
field is also assumed to be non-zero under the die and zero elsewhere. The effect of front and back-pull are neglected in the present analysis as this data was not available at the time of the analysis.

In order to perform the analysis it is first necessary to divide the die into sections or zones corresponding to each inlet angle. This enables a more simplified analysis and makes it easier to handle the discontinuities arising from the change in angle. Each section is then analyzed separately with suitable boundary conditions imposed to maintain equilibrium conditions. The results from the analysis of each section can then be summed together to obtain an estimate of the total forging load.

The deforming material in each zone is first divided into a series of elemental slabs as shown in Figure C.2. During the analysis, the coordinate axes are oriented in the directions of material flow and are assumed to coincide with the principal directions. Frictional stresses are assumed to have a negligible effect on the orientation of the principal stress axes.

C.2.1 Analysis of Stresses in the Forging Zone

Employing the nomenclature given in Figures C.3 and C.4 for the forging stresses and element geometry, the governing differential equation for axial equilibrium acting on an annular element may be written as:

$$\sigma_z A = (\sigma_z + d\sigma_z)(A + dA) - mk \cos \theta (2\pi r dS) + p \sin \theta (2\pi r dS)$$  \hspace{1cm} (C.3)
Figure C.2. Division of deforming material into elemental slabs.

Figure C.3. Equilibrium of an elemental slab in the axial direction.
The +/- sign is used to denote a change in direction of the frictional stress at the neutral plane.

Expanding and simplifying equation (C.3) by neglecting higher order terms results in:

\[-d\sigma_A = \sigma\:dA \pm mk \cos \theta (2\pi rdS) + p\sin \theta (2\pi rdS)\]  \hspace{1cm} (C.4)

Equation (C.4) can then be rewritten by making use of the geometric relation shown in Figure C.5.

\[dS = \frac{dr}{\sin \theta}\]  \hspace{1cm} (C.5)
Substituting equation (C.5) into equation (C.4) results in:

\[-d\sigma_z A = \sigma_z dA \pm mk \frac{2\pi rd}{\tan \theta} + p 2\pi rd\]  

(C.6)

Further simplification may be obtained by noting that \( dA = 2\pi rd \):

\[-d\sigma_z = \frac{2dr}{r} (\sigma_z \pm \frac{mk}{\tan \theta} + p)\]  

(C.7)

In order to solve equation (C.7) directly it is necessary to first write the die pressure, \( p \), in terms of \( \sigma_z \) and \( \bar{\sigma} \). This may be done by taking a force balance in the \( r \)-direction as shown in Figure C.6. It is assumed in the analysis that \( \sigma_r \) is constant over the height of the element.
Taking equilibrium in the $z$-direction, the following equation is obtained:

$$p \cos \theta (2\pi r dS) + \sigma_r dz 2\pi r = \pm m k \sin \theta (2\pi r dS)$$ \hspace{1cm} (C.8)

Simplifying equation (C.8) by dividing by $2\pi r$ and noting that $dS = \frac{dz}{\cos \theta}$,

$$pdz + \sigma_r dz = \pm m k \sin \theta \left[ \frac{dz}{\cos \theta} \right]$$ \hspace{1cm} (C.9)

Rearranging and simplifying equation (C.9) results in:

$$\sigma_r = \pm mk \tan \theta - p$$ \hspace{1cm} (C.10)
From the Von-Mises Flow Rule for Axisymmetry the yield condition in terms of principal stresses may be written as:

\[ \sigma_1 - \sigma_3 = |\bar{\sigma}| \]  

(C.11)

Based on the assumption that co-ordinate axes coincide with principal directions equation (C.11) becomes:

\[ \sigma_z - \sigma_r = \bar{\sigma} \]  

(C.12)

Inserting equation (C.12) into equation (C.10) results in:

\[ \sigma_z \mp mk \tan \theta + p = \bar{\sigma} \]  

(C.13)

The die pressure, P, may then be found by rearranging equation (C.13):

\[ p = \bar{\sigma} \pm mk \tan \theta - \sigma_z \]  

(C.14)

The above analysis is then repeated for each of the inlet angles and noting that the die stress at each zone inlet is bounded by the die pressure at the exit of the preceding zone.
C.2.2 Analysis of Stresses in the Sizing Zone

In the sizing zone the material is not expected to undergo a significant amount of plastic deformation. Therefore, elastic deformation is a significant portion of the total deformation. In employing a rigid-plastic analysis, elastic deformation is neglected and for material which is in the elastic regime, the stress is therefore indeterminate. In the actual radial forging process, the contribution to the total load from the sizing zone is still expected to be significant due to both frictional and elastic stresses.

In the following discussion, the material in the sizing zone is assumed to be elastically deformed to the yield limit, hence satisfying the Von-Mises yield criterion. This assumption was first used by Lahoti and Altan in their analysis of tube forging (Lahoti and Altan, 1974).

The material in the sizing zone may be modeled as a right circular cylinder with stresses acting as shown in Figure C.7. The governing differential equation for axial equilibrium may be written as:

\[ \sigma_z A = (\sigma_z + d\sigma_z)A \pm mkdz2\pi r \]  

(C.15)

Expanding and simplifying equation (C.15) results in:

\[-d\sigma_z A = \pm \frac{2mk}{r} dz\]  

(C.16)
Noting that \( A = 2\pi r^2 \), equation (C.16) may be integrated to obtain:

\[
\sigma_z = \pm \frac{2mk}{r} z + C
\]

(C.17)

where \( C \) is a constant of integration.

Figure C.7. Axial equilibrium of an elemental slab in the sizing zone.

C.2.3 Neutral Plane Location

To find the neutral plane it is necessary to first determine the peak die pressure or, equivalently, the point at which the friction force changes direction. This can be done by analyzing the die pressure first from right to left and then from left to right and then finding the point at which the two curves intersect, as is
commonly done in rolling analysis. In performing each analysis it is assumed that the friction stress is acting in the direction opposite to the direction of metal flow. The neutral point is located by searching each of the deformation zones and finding the point at which the die pressures are equal.

The limiting positions for the neutral plane lie at the die inlet and exit. If the neutral plane lies outside the deformation zone, plastic deformation is not feasible as the billet would simply be "kicked-out" of the forging box by the approaching hammers (Lahoti and Altan, 1974).

C.3 Adjustment of Axisymmetric Load to Actual Case

In the axisymmetric analysis, it is assumed that the die lands are cylindrical and form a conical surface which fully encircles the workpiece surface. This neglects the clearances between the dies at the end of the stroke and the angle of tangency on the die land of flat tools. Due to the axisymmetric assumptions, the predicted load will be significantly higher than the actual load. To adjust the axisymmetric load to account for the fact that the dies make partial contact with the workpiece surface, it was assumed that in a single stroke the dies contacted 1/6 th of the workpiece circumference along the length of the die land. This value was based on a previous analysis performed to calculate the number of blows that each tool makes during a 90 degree rotation of the workpiece (Domblesky et al., 1992).
C.4. Validation of RAFSTRAN Predictions

To verify the accuracy of the program formulation of RAFSTRAN, two in-plant forging experiments were performed using Inconel 718 billets. Large and small diameter billets were forged using proprietary forging sequences and standard GFM forging tools. The forging loads during multiple pass radial forging were measured by engineering personnel at Teledyne Allvac in Monroe, North Carolina using a GFM SX-55 machine (Jackman and Ramesh, 1992).

C.4.1 Analysis Results

Results of the comparison between actual and predicted forging loads are shown in Figure C.8. For the analysis an assumed friction factor of $m = 0.3$ was used. The ratio of predicted loads to experimental loads has been plotted for each pass. For each experiment, both axisymmetric and plane-strain analyses were performed. In both experiments the analysis predicts an upper-bound load. The axisymmetric analysis appears to provide the closest estimate of forging load of the two methods. From Figure C.8 it can also be seen that the ratio of predicted and actual load is sensitive to the ratio of $h/L$. For $h/L$ less than 1.0 the assumption that $1/6$ th of the total load in an axisymmetric analysis would approximate the actual load was found to provide reasonable estimates. However for $h/L$ greater than or equal to 1.0, the predicted results were found to consistently be 2.0 to 2.25 times higher than the actual. Therefore it was
necessary to further adjust the predicted load by a factor of 2.0 for cases where h/L was greater than or equal to 1.0.

For each of the cases analyzed, the neutral plane was found to lie in the sizing zone or the forging zone adjacent to the sizing zone. The location of the neutral plane was found to be sensitive to the friction factor used. As the friction factor, m, was decreased, the neutral plane was found to move towards the die exit accompanied by a decrease in forging load. This is in agreement with the results of Lahoti and Altan’s analysis for tube forging (Lahoti and Altan, 1974).

![Graph showing forging loads predicted by RAFSTRAN normalized to experimental forging loads.](image)

Figure C.8. Forging loads predicted by RAFSTRAN normalized to experimental forging loads.
C.4.2 Discussion

A possible explanation for the sensitivity of load to the h/L ratio is given by considering a narrow punch indenting an overhanging part. The classic analysis performed by Hill (1959) showed that for a rectangular workpiece, when the h/L ratio was less than 1.0, the two deformation zones fully co-operate and result in homogeneous deformation. Additionally the effect of friction is important. This situation most closely approximates the conditions of the slab analysis. For values of h/L greater than 1.0, the two deformation zones do not fully co-operate and at high h/L ratios, the deformation zones are independent of each other (Schey, 1987). The latter case corresponds to two punches indenting a semi-infinite body where the die pressure required to initiate plastic deformation is three times that for simple compression.

In the present slab analysis it is thought that for h/L ratios less than 1.0, the slab analysis will correspond most closely to the actual case and the load will need only to be adjusted for the limited die contact. For h/L ratios greater than 1.0, the stress distribution will not be uniform as assumed in the slab analysis. Therefore, based on the analysis by Hill, it is likely that additional adjustment will be required to account for the stress gradients.

While the above discussion was based on Hill's analysis of plane-strain conditions, it is to be envisaged that a similar relationship is valid for axisymmetric workpieces.
The reason for the upper bound load predictions is due to the fact that front and back-pull were not considered. If the axial force exerted by the chuckheads were considered, the imposed axial stress would act to reduce the die pressure required to initiate and maintain plastic deformation using the rigid-plastic theory. The reduced die pressure would most likely have decreased the load predictions shown in Figure C.8 and provided a lower bound estimate of the actual value. It is also expected that this would have caused the position of the neutral plane to shift away from the sizing zone and towards the forging zone.
Appendix D

Application of the Finite Element Method to Large Diameter Tube Forging
Previous investigations of large diameter tube forging using the radial forging process have been based on highly simplified slab and upper bound process models. In order to make accurate predictions of thermal and mechanical history of the workpiece, more detailed models are required. In this paper, a model of the tube forging process based on the rigid-viscoplastic finite element method is presented. Results presented include effective strain, strain rates, and temperature distributions in the forged tube. Deformation was found to be uniform but sensitive to the axial feed rate. Strain rates were found to be sensitive to the axial feed rate and were within the range typical for a mechanical press. Large temperature gradients across the wall thickness were predicted in the tube. These temperature gradients were generated during forging due to contact with the mandrel and are thought to give rise to variation in mechanical properties in the forged tube.

D.1. Introduction

Hot radial forging is an important near-net shape manufacturing process used in the production of large diameter tubes having thick walls. The main products of the tube forging process are large caliber gun tubes, automotive axles, and oil field equipment. Deformation in radial forging of tubes results from a large number of short-stroke side-pressing operations by four forging tools arranged radially around the workpiece. A typical hammer arrangement is shown in Figure D.1.
The forging action of the radial forge takes place within a vertically arranged forging box which houses the four hammers and two drives which are located at right angles to one another. Due to the opposing motion of the hammers, no forces are transmitted to the machine base (GFM Precision Forging Machines, 1976). A schematic of the radial forging process is shown in Figure D.2. The primary features of the machine are the forging box and two synchronous
chuckheads which act to center and index the workpiece. During forging, the motion of the chuckheads is intermittent and synchronized with that of the dies to prevent twisting of the workpiece. During the time that the dies are in contact with the workpiece, the chuckheads remain stationary. When the dies are not in contact, the workpiece is rotated and axially indexed in a simultaneous motion as depicted in Figure D.3. After each blow of the tools, the workpiece is axially fed towards the die inlet and rotated by 15 degrees to obtain a good surface finish (Lahoti and Altan, 1976-1). The tube is generally forged down in a single pass with no intermediate reheats.
Figure D.2. A schematic of the radial forging process.
Figure D.3. Kinematics of the radial forging process.
During hot forging, the mandrel is water cooled due to heat radiation and conduction from the workpiece. The purpose of the mandrel during forging is to prevent the inner wall of the tube from collapsing and to form the inner wall diameter. Because the inner wall of the forged tube takes on the impression of the mandrel, steps and other shapes may be formed with a suitable mandrel design. Depending on the internal profile to be produced, there are two basic types of mandrels which are employed in radial forging. These are "short" and "long" mandrels. Short mandrels are used in forging long tubes where the inner diameter of the forged tube is constant along the length of the workpiece. Once the mandrel has been positioned between the forging tools, it remains axially stationary during the forging process while the tube is indexed through the forging box as depicted in Figure D.4. Long mandrels are used in forging short tubes where the inner diameter of the forged tube is stepped along the length of the workpiece. For stepped tubes, the mandrel moves simultaneously with the workpiece through the forging box as shown in Figure D.5.
Figure D.4. Short mandrel used in forging tubes with constant inner diameter.
Figure D.5. Long mandrel used in forging tubes with stepped inner diameter.
Radial forging of tubes has received very little attention due to its highly specialized applications and complex deformation behavior. Due to the rather specialized nature of the process, only a limited number of studies (Lahoti and Altan, 1976; Lahoti et al, 1980; Raghpathi et al., 1980; Tszeng, 1987) have considered the process and tried to model thermomechanical behavior during tube forging (Lahoti and Altan, 1976; Lahoti et al, 1980; Raghpathi et al., 1980). Lahoti et al (Lahoti and Altan, 1976; Lahoti et al, 1980) were the first to extensively study the process and develop mathematical models of the process for forging large caliber gun barrels. Their analysis used a simplified deformation model based on the slab and upper bound methods. Tszeng and Kobayashi (1986) were the first to model tube forging using the finite element method (FEM). However their analysis was based on an isothermal analysis where the front and back pull were ignored and it was assumed that the bore diameter of the preform was equal to the mandrel diameter.

The objective of this study was to develop an efficient FEM model which could be used to predict the thermomechanical behavior of the tube during radial forging. Due to the limited number of studies of tube forging, it is also necessary to study the relations between process parameters and product quality. In the following, the model development and selected simulation results are presented. Due to the lack of available experimental and calculated data, only a limited comparison of the predicted results could be made based on data from previously published literature.
D.2. Mathematical Modeling

The radial forging process is characterized by cyclical and transient loading conditions. To analyze the material flow/deformation, the model needs to have the capability of analyzing complex deformation and thermal conditions. Due to the limited state of development of 3-D FEM codes at the present time, such an analysis is not yet possible. However, a reasonable approximation may be made based on an axisymmetric deformation model. In doing so, the following assumptions need to be made:

1) The flat surfaces on the forging tools may be approximated by semi-circular arcs
2) Small clearances between the dies are neglected
3) Rotational feed is neglected, only the axial feed component is considered

Because of the usefulness in solving a wide variety of metal forming problems, the forging industry has made increasing use of the finite element method as a process modeling and analysis tool. To model the tube forging process, the commercially available FEM code, DEFORM (formerly ALPID) from Scientific Forming Technologies Corporation was used. DEFORM is based on the Updated Lagrangian formulation with rigid-viscoplastic material behavior.

To model the process, it is first necessary to develop a suitable representation of the forging dies, chuckheads, and the mandrel. Each of these will be considered
separately below. The modeling strategy used in the present study was to model the process on a "stroke-by-stroke" basis assuming a 2-D axisymmetric deformation model. The underlying principle used was to model each tool stroke as a separate simulation with the accumulated non-contact time between each loading cycle included at the end of the pass simulation. The workpiece is moved by an amount equal to the axial feed rate between each simulation. The FEM representation of the tube forging process is shown in Figure D.6. The figure is shown with the longitudinal axis (axis of symmetry) positioned vertically for simulation purposes.
Figure D.6. Finite element representation of the tube forging process.
Although the forging dies are driven in pairs, the motion of each die may be considered to follow that of a harmonic or sinusoidal function. Each die may then be considered to act as a short-stroke mechanical press where the tool velocity may be written as a function of the stroke using the relation (after Altan et al., 1983):

$$V = \frac{\pi nh}{30} \sqrt{\frac{s}{h} - 1}$$  \hspace{1cm} (D.1)

where:  
- \text{n = Number of strokes per minute}  
- \text{h = Distance from Bottom Dead Center (BDC) of the Press}  
- \text{s = Total stroke of the Forging Tools}

From equation (D.1) it is apparent that the die velocity will be continuously decreasing during the stroke. While the die velocities are high due to the high stroking rates, most deformation occurs near the end of the stroke where the die velocity will be much lower. The strain rate in the deformation zone is generally thought to be within the range typical of those found in mechanical presses.

The bases of the forging tools are fabricated from a high chrome die steel with a welded layer of Ni-based alloy for better temperature resistance. The tool geometry is typically composed of compound or multiple inlet angles and a flat sizing zone. Deformation of the workpiece almost entirely takes place under the die inlet in the sinking and forging zones as shown in Figure D.7. The sole
purpose of the sizing zone is to provide a good surface finish (Lahoti et al., 1977). In developing the model, the actual die geometry was used. In practice the dies are nominally pre-heated to 200 °C (Nagpal and Lahoti, 1980).

In most radial forging analyses the effect of die chilling has been assumed to be negligible due to the relatively short time over which the die is in contact with the workpiece (Lahoti and Altan, 1974; Lahoti et al., 1980; Isogawa et al., 1990). It has been estimated that the tools are in contact with the workpiece for less than 10-15% of each forging cycle (Nagpal and Lahoti, 1980). For a machine operating at 180-200 strokes per minute, the non-contact time will be on the order of 0.3 seconds. While the amount of heat loss occurring between each stroke is insignificant, the accumulated effect can be important. To account for the non-contact time of the tools between each stroke, in an earlier study an implicit method was used to introduce a period of heat transfer equal to the dwell time of the forging tools between each forging cycle during each simulation (Domblesky et al., 1994). However, in the present case the implicit scheme could not be used to model heat transfer losses between forging cycles. This was due to the fact that when the dies were not in contact with the workpiece, an applied moment was generated causing the workpiece nodes to separate from the mandrel and chuckhead allowing rigid body motion to occur.
Figure D.7. Schematic representation of the three deformation zones in radial forging of tubes.
The effect of the workpiece restraint imposed by the chuckheads greatly influences the metal flow. Therefore it is necessary to accurately model the kinematic boundary conditions at the workpiece ends. If insufficient restraint is provided, rigid body translation results and a singularity occurs in the FEM solution. Previous simulations (Tszeng and Kobyashi, 1986; Isogawa et al., 1990) of the radial forging process have assumed that the nodes at the trailing end of the workpiece are fixed in the axial direction. In actuality there is some give in the chuckhead which allows limited movement of the workpiece in the axial direction. These intermittent motions result in compression of springs located in the chuckhead (Lahoti et al., 1976-2). In the present study the chuckhead was initially modeled as an elastic object with an elastic modulus one order of magnitude less than the workpiece material. However, the computational time required for each simulation was prohibitive. Because the chuckheads securely grip the workpiece ends at four points, the chuckheads may be modeled as being a rigid object with a constant friction factor of \( m = 1.0 \) (sticking friction) where the frictional stress is defined as:

\[
\tau_i = m \frac{\sigma}{\sqrt{3}}
\]  

(D.2)

and \( m \) is defined as a constant expressing friction conditions at the tool/workpiece interface.
In the actual process, the chuckheads exert a front and back tension on the workpiece ends. For the FEM model the front and back tension were modeled by imposing distributed nodal forces along the leading edge of the workpiece. The mandrel used in tube forging is a hollow cylinder of H-10 or H-13 hot-work chromium die steel with an internal cavity for water coolant. Due to intense heat radiation and conduction it is necessary to cool the mandrel to prevent excessive heat buildup and adhesion to the inner surface of the tube. There is some heating of the water during forging, but due to the water being continuously circulated in the mandrel, the temperature will ultimately reach a constant or steady-state temperature (Raghupathi et al., 1980). By assuming that the water temperature stays at a constant temperature, the mandrel may be modeled as a hollow cylinder with an internal diameter equal to that of the coolant channel. The mandrel may then be considered as a rigid object with the nodes on the inner surface being prescribed at a temperature equal to the outlet temperature of the coolant.

D.3. Simulation of the Tube Forging Process

For the study, large diameter AISI 4337 tubes were modeled using three different axial feed rates. The conditions used are shown in Table D.1.
Table D.1. Process variables used in the simulation study of tube forging.

<table>
<thead>
<tr>
<th>Process Variables Held Constant:</th>
</tr>
</thead>
<tbody>
<tr>
<td>Preform Material: AISI 4337</td>
</tr>
<tr>
<td>Initial Preform Temperature: 1000.0 °C</td>
</tr>
<tr>
<td>Inside Diameter of the Preform: 180 mm</td>
</tr>
<tr>
<td>Mandrel Outside Diameter: 160 mm</td>
</tr>
<tr>
<td>Cooling Chamber Diameter: 50 mm</td>
</tr>
<tr>
<td>Machine Stroking Rate: 200 Strokes/minute</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Process Variables Changed:</th>
</tr>
</thead>
<tbody>
<tr>
<td>Axial Feed Rate: 10, 20, and 30 mm/stroke (feed per minute)</td>
</tr>
</tbody>
</table>

Due to the large number of simulations which were required to simulate each pass, an automated routine was developed to minimize engineering time required. Each simulation required approximately 40-52 minutes of CPU time to perform on a VAX Station 4100 computer system.

D.4. Simulation Results

The effective strain distributions in the forged tube at mid-pass for each case simulated are shown in Figures D.8-D.10. Based on the predicted deformation, it is evident that three deformation zones exist during forging. These are the sinking, forging, and sizing zones. In the sinking zone, the diameter of the tube is reduced with only a small change in tube wall thickness. In the forging zone, the tube is forged onto the mandrel with most deformation taking place in this zone. The tube material under the sizing zone is assumed to undergo elastic or elasto-plastic deformation (Lahoti and Altan, 1974; Lahoti et al, 1980). However,
because DEFORM is based on rigid-viscoplastic material behavior where the elastic response is neglected, the effect of the sizing zone could not be considered in the analysis. For the three axial feed rates simulated, it is evident from the iso-contours that only a small amount of straining occurs in the sinking zone. Due to the lack of frictional restraint offered by the mandrel in the sinking zone, deformation at the tube inner diameter tends to lead or extend ahead of deformation at the tube outer diameter. In the forging zone, deformation is observed to increase to the maximum value of strain predicted. For all feed rates simulated, deformation is observed to penetrate through the tube thickness. It is also evident that the process maintains essentially steady-state deformation along the length of the forged tube with only minor variations.
Figure D.8. Effective strain in the forged tube for an axial feed rate of 10 mm/stroke.
Figure D.9. Effective strain in the forged tube for an axial feed rate of 20 mm/stroke.
Figure D.10. Effective strain in the forged tube for an axial feed rate of 30 mm/stroke.
The effect of increasing axial feed rate from 10 to 20 mm/stroke was found to result in marginally higher effective strains in the axial and radial directions as shown in Figures D.11-D.13. A further increase in feed rate was not observed to have a discernible impact on the strain distribution. An interesting result of the simulation also shows that deformation is marginally higher at the tube inner diameter and outer diameter than the sub-surface. These results are different from those found in billet conversion where the strain is predicted to decrease with increasing distance from the surface of the billet (Boyko et al., 1991). The difference may be the result of larger height of the deforming material (radial thickness) in billet forging.
Figure D.11. Effective strain distribution in the radial direction after forging for an axial feed rate of 10 mm/stroke.
Figure D.12. Effective strain distribution in the radial direction after forging for an axial feed rate of 20 mm/stroke.
Figure D.13. Effective strain distribution in the radial direction after forging for an axial feed rate of 30 mm/stroke.
For all three axial feed rates simulated (10, 20, and 30 mm/stroke), the maximum strain rate was typically found to occur at the start of deformation when the tool contacted the workpiece. The effect of axial feed rate on strain rate during deformation is shown in Figures D.14-D.16. As expected, the strain rate in the tube was maximum at the beginning of deformation and approached zero at the end of the tool stroke. The maximum strain rates in all three cases were found to be on the order of 4.0 and 6.0/sec. which are typical deformation rates for mechanical presses. The maximum strain rates for all feed rates were observed under the tool radius. Deformation occurred most rapidly in the forging zone with lower strain rates generally occurring in the sizing zone. The strain rate appears to be influenced by the axial feed rate to a limited extent. When feed rate was doubled from 10 to 20 mm/stroke, the maximum strain rate was also doubled from 3.0 to 6.0/sec. A further increase of feed rate to 30 mm/stroke resulted in little additional change to the strain rate; indicating the existence of a threshold in strain rate values. While the effect of changing feed rate altered the maximum strain rate, it did not appear to significantly affect the geometry of the deforming zone for the cases simulated. The presence of a neutral plane is clearly discernible in Figures D.14-D.16.
Figure D.14. Maximum effective strain rate in the deformation zone for an axial feed rate of 10 mm/stroke.
Figure D.15. Effective strain rate in the forged tube for an axial feed rate of 20 mm/stroke.
Figure D.16. Effective strain rate in the forged tube for an axial feed rate of 30 mm/stroke.
The temperature history of the billet for the three radial locations is shown in Figures D.17-D.19. In order to facilitate the comparison between the three feed rates, the temperature is plotted as a function of the axial position of the die along the workpiece rather than time. The effect of increasing axial feed rate does not appear to significantly influence temperatures in the workpiece for the reduction simulated (Figure D.20). The inner and outer surfaces of the forged tube tend to experience a significant decrease in temperature during forging due to contact with the cooler mandrel and forging tools. After forging, temperatures appear to be maintained at a relatively constant level (Figures D.17-D.19). Sub-surface locations experienced a slight temperature rise due to deformation heating. Temperature gradients which developed at the outer surface of the workpiece are presumed to be the result of contact with the cooler forging tools which were prescribed as having a nominal pre-heat temperature of 200 °C. Even though the effect of heat losses between subsequent tool strokes were neglected, the predicted surface temperatures were found to be in reasonable agreement with those measured during actual forging. This indicates that the assumption made in previous studies (Lahoti et al., 1980; Raghupathi et al., 1980) that heat losses to the tools are negligible may not always be applicable when simulating tube forging.

Based on the temperature gradients observed, it is likely that discernible differences in product properties may result after forging in the radial direction. Temperature losses at the outer diameter of the forged tube could be minimized by using a higher pre-heat temperature for the dies. Minimizing thermal
gradients in the forged product would have an important impact on the properties of the forged workpiece since microstructural development is mainly influenced by the strain and thermal history imposed.

Figure D.17. Radial temperature distribution during forging for axial tube location = 500 mm for 10 mm/stroke feed rate.
Figure D.18. Radial temperature distribution during forging for axial tube location = 500 mm for 20 mm/stroke feed rate.
Figure D.19. Radial temperature distribution during forging for axial tube location = 500 mm for 30 mm/stroke feed rate.
Figure D.20. Temperature distributions at axial location = 500 mm for nodes at inner, mid, and outer radius locations for the forged tube.
D.5. Conclusions/Summary

In the study presented, it was found feasible to model radial forging of large diameter tubes using the finite element method. Such a model would be useful for predicting the thermomechanical history of the forged tube. While the effect of heat loss between subsequent strokes was ignored, some predictions may be made regarding metal flow and heat loss trends during the forging process. Based on the results of this simulation study, the following conclusions may be made for radial forging of large diameter tubes.

1) Very little deformation occurs in the sinking zone. Most deformation is predicted to occur in the forging zone.

2) The effect of increasing axial feed rate was found to result in marginally higher effective strains in the axial and radial directions in the forged tube.

3) The strain rate appears to be influenced by the axial feed rate to a limited extent. The most significant effect was observed at small feeds for the reduction simulated.

4) The effect of axial feed rate does not appear to significantly influence temperatures in the forged tube for the reduction simulated.
5) Temperature gradients developed at the outer surface of the tube are most likely to be the result of contact with the cooler forging tools.
Appendix E

Program Listing for AUTODEFORM 2.1
**File: AutoDeform2.1**

```csh
#! /usr/bin/csh
#
===============================================================================
#
HEADER FOR AUTODEFORM
#
===============================================================================
#
To run Deform automatically, these two files must exist in
the directory you wish to run from:
AutoDeform
AutoCommands
#
================== QUICK DIRECTIONS TO BEGIN AUTODEFORM ===========
#
1. Enter the directory you wish to run Deform in.
2. Type 'AutoDeform' to start the interactive program.
#
#
================== LONG DIRECTIONS TO BEGIN AUTODEFORM =====
#
1. Enter the directory you wish to run Deform in as usual.
2. Type 'AutoDeform' to start the interactive portion of the process.
   AutoDeform will ask you for the Problem Name, # of simulations,
   etc. After you answer the questions, the program creates a file
   in the directory containing the problem information. AUTODEFORM
   THEN EXECUTES AUTOCOMMANDS AS A BATCH PROCESS.
3. When the prompt returns, you are free to continue other work
   or LOG OFF. The simulations will run in the background. An
   OUTPUT file will be created indicating results and the CRON
   daemon will e-mail a message to you when the batch is finished.
#
#
===============================================================================
#
REstrictions on AUTODEFORM
#
===============================================================================
#
1. At the moment, only one set of simulations is allowed to run
   per user. This is true even if the simulations are run from
   different directories, so run ONLY one set of simulations at
   a time. The reason for this is that AUTOCOMMANDS constantly
# checks 'getbatch' for a list of processes running and counts
# the number of DEF.EXE processes, without regard to what
# directory it was started from. If AUTOCOMMANDS sees one or
# more DEF.EXE processes, it waits until there are no such
# processes.
#
# 2. Right now, AutoDeform only works in my account (UID = 312).
# This will be changed very soon.
#
# 3. AutoDeform performs NO error checking. If you enter the wrong
# starting step for Deform, it will not tell you, but instead will
# execute the instructions anyway (generally ending up with a
# huge message file when it finishes. There is no guarantee it
# won't hurt any data, so be sure the starting step is correct.
# In fact, make a backup of the database file before attempting
# to run AutoDeform.
#
# 4. AutoDeform is set up to ONLY perform CONTINUING runs, so at least
# one positive step must exist in the database file.
#
# ================ AUTODEFORM: INTERACTIVE PORTION ================
#
# ================== GET PROBLEM NAME & OTHER DATA ======
set DATAFILE = 'ADtmpfile' #Filename of the AutoDeform storage
set PASSFILE = 'DEF_PRE.INI' #File that is required for pre-processor
set INPUTFILE = 'AutoID'

set ENDER = 7
set SELECTION = 1
set PROBLEMNAME = DEFORM
set NUMSTEPS = 38       #Number of Deform steps in any one simulation
set SAVENAME = DEFORM.OUT
set CURRENTSTEP = 0
set NUMSIMS = 1
set DELFILES = y

if (-e $INPUTFILE) then
    tail -6 AutoID | head -1 | set PROBLEMNAME = 'line'
    tail -5 AutoID | head -1 | set SAVENAME = 'line'

tail -4 AutoID | head -1 | set CURRENTSTEP = 'line'
tail -3 AutoID | head -1 | set NUMSIMS = 'line'
tail -2 AutoID | head -1 | set DELFILES = 'line'
tail -1 AutoID | head -1 | set NUMSTEPS = 'line'
endif

while ($SELECTION < $ENDER)
clear
echo "
echo ' AutoDeform'
echo "
echo ' 1. Problem Name' $PROBLEMNAME
echo ' 2. Results File' $SAVENAME
echo ' 3. Current Step' $CURRENTSTEP
echo ' 4. # of Simulations' $NUMSIMS
echo ' 5. Delete Scratch File' $DELFILES
echo ' 6. Steps per Simulation' $NUMSTEPS
echo "
echo ' 7. Begin Simulations'
echo ' 8. Exit'
echo "
echo 'Enter Selection:'
set SELECTION = 'line'
echo "
echo "
switch ($SELECTION)
case 1:
  # ASK FOR PROBLEM NAME
  echo Enter Problem Name
  set PROBLEMNAME = 'line'  #Same as PROBLEM ID in Deform
  break

case 2:
  # ASK FOR MESSAGE FILE
  echo Enter File to Save Results to
  set SAVENAME = 'line'  #Filename that AD will save all output to
  break

case 3:
  # ASK FOR CURRENT STEP NUMBER
  echo Enter Current Step Number
  set CURRENTSTEP = 'line'  #Last positive step Deform has recorded
  break

case 4:
ASK FOR NUMBER OF SIMULATIONS
    echo Enter Total Number of Simulations to Perform
    set NUMSIMS = 'line'
    breaksw
    case 5:
    ASK WHETHER TO DELETE SCRATCH FILES
    echo 'Delete Scratch Files (y/n)'
    set DELFILES = 'line'
    breaksw
    case 6:
    ASK FOR NUMBER OF STEPS PER SIM
    echo Steps per Simulation
    set NUMSTEPS = 'line'
    breaksw
    case 7:
    DO NOTHING HERE
    echo "
    breaksw
    case 8:
    EXIT PROGRAM
    stop
    breaksw
default:
    echo Invalid Selection
    echo 'Press <RETURN>'
    set SELECTION = 'line'
    set SELECTION = 1
    breaksw
endsw
end

BEGIN PROCESSING
if ($SELECTION == $ENDER) then
    RESTATE THE INPUT
    set NSTEP = 'expr $CURRENTSTEP + 1'  #Equals the next negative step
    set NSTEP = 'expr $NSTEP * -1'
    set FINALSTEP = 'expr $NUMSIMS * $NUMSTEPS'
    set FINALSTEP = 'expr $FINALSTEP + $CURRENTSTEP'

    Preparing to write scripts to perform CONTINUING RUN for
    problem '$PROBLEMNAME' beginning at step '$NSTEP'.
    $NUMSIMS' simulations will be performed, each with
    $NUMSTEPS' steps. The last step will be '$FINALSTEP'.

# =========== WRITE THE FILE TRANSFERRED TO AUTOCOMMANDS = 
# (This is read in by AutoCommands in a background process)

echo $PROBLEMNAME > $DATAFILE
echo $NUMSTEPS >> $DATAFILE
echo $SAVENAME >> $DATAFILE
echo $CURRENTSTEP >> $DATAFILE
echo $NSTEP >> $DATAFILE
echo $NUMSIMS >> $DATAFILE
echo $FINALSTEP >> $DATAFILE
echo $DELFILES >> $DATAFILE

echo $PROBLEMNAME > $INPUTFILE
echo $SAVENAME >> $INPUTFILE
echo $CURRENTSTEP >> $INPUTFILE
echo $NUMSIMS >> $INPUTFILE
echo $DELFILES >> $INPUTFILE
echo $NUMSTEPS >> $INPUTFILE

# ============== START THE BATCH PROCESS 'AutoCommands' =====
echo "
echo 'Press <RETURN> to start the job'
set WAITER = 'line'
clear

echo AutoCommands2.1 | at now
echo "
echo "
echo '**** The simulation IS running'

else
  echo '**** Program Exiting ****'
  echo '**** No Simulations Run ****'
endif
FILE: AutoCommands2.1

#!/usr/bin/csh
#
# GET PROBLEM NAME & OTHER DATA
#
set PASSFILE = 'DEF_PRE.INI'
set DATAFILE = 'ADtmpfile'
set INPUTFILE = 'AutoID'
set MENUFILE = 'MenuSelections'
pwd | set WHERE = 'line'
clear

if (-e ADtmpfile) then
    tail -8 $DATAFILE | head -1 | set PROBLEMNAME = 'line'
tail -7 $DATAFILE | head -1 | set NUMSTEPS = 'line'
tail -6 $DATAFILE | head -1 | set SAVENAME = 'line'
tail -5 $DATAFILE | head -1 | set CURRENTSTEP = 'line'
tail -4 $DATAFILE | head -1 | set NSTEP = 'line'
tail -3 $DATAFILE | head -1 | set NUMSIMS = 'line'
tail -2 $DATAFILE | head -1 | set FINALSTEP = 'line'
tail -1 $DATAFILE | head -1 | set DELFILES = 'line'

# PERFORM THE SIMULATIONS IN A LOOP
#
set NS = $NSTEP
set ES = 'expr $CURRENTSTEP + $NUMSTEPS'
set REPEAT = 1
while ($REPEAT <= $NUMSIMS)
    echo Repetition No. $REPEAT » $SAVENAME
    echo Begin Step No. $NS » $SAVENAME
    echo End Step No. $ES » $SAVENAME
    echo MESSAGES FROM SIMULATION RUN > $SAVENAME

# WRITE THE SIMULATION SCRIPT
#
set PROBLEMNAME.DB > AutoSimScript
echo $PROBLEMNAME.MSG >> AutoSimScript
echo b >> AutoSimScript

#
===================================
# WRITE THE TRANSFER FILE BETWEEN MAIN & PRE
#
===================================

echo PROBID > $PASSFILE
echo $PROBLEMNAME >> $PASSFILE
echo RUNTYP 2 >> $PASSFILE
echo DBREAD $CURRENTSTEP >> $PASSFILE
echo $PROBLEMNAME.DB >> $PASSFILE
echo NSTART $NS >> $PASSFILE
echo INIGES 0 >> $PASSFILE

#
===================================
# EXECUTE THE PRE-PROCESSOR AND THE SIMULATOR
#
===================================

echo '**** Simulations being executed'
DEF_PRE < AutoPreScript > $MENUFILE
DEF_SIM < AutoSimScript

# ============== WAIT FOR THE CURRENT SIM TO END =========
if ($REPEAT <$NUMSIMS) then
  set ADZ = 1
  set rr = 1
  set limit = 0
  pwd | set DIREC = `line`
  while ($limit != $ADZ)
    getbatch | grep $DIREC | wc -l | set ADZ = `line`
  end
  echo 'Simulation finished' >> $SAVENAME
  set COPIER = $PROBLEMNAME.MSG
  set COPY = $PROBLEMNAME$CURRENTSTEP.MSG
  if (-e $COPIER) mv $COPIER $COPY
  echo 'Message file saved to ' "$COPY >> $SAVENAME
  echo " >> $SAVENAME
endif
set CURRENTSTEP = `expr $CURRENTSTEP + $NUMSTEPS`
set NS = `expr $NS - $NUMSTEPS`
set ES = `expr $ES + $NUMSTEPS`
set REPEAT = `expr $REPEAT + 1`
end

#
# SIMULATIONS ARE FINISHED, CLEAN UP
#
#
if (-e ADtmpfile) rm ADtmpfile
if (-e AutoSimScript) rm AutoSimScript
if (-e $PASSFILE) rm $PASSFILE

   echo 
   echo "
   echo "**** SCRIPT FINISHED ****"
   echo "**** The last simulation is still running"
   echo "**** Simulation Results are in file '$SAVENAME'
   echo "**** A list of Pre-processor menu choices is in file '$MENUFILE'
   echo "
   echo "$PROBLEMNAME > $INPUTFILE
   echo $SAVENAME >> $INPUTFILE
   echo $CURRENTSTEP >> $INPUTFILE
   echo $NUMSIMS >> $INPUTFILE
   echo $DELFILIES >> $INPUTFILE
   echo $NUMSTEPS >> $INPUTFILE

else
   echo "**** NO INPUT FILE ****"
   echo "**** AutoDeform must be executed first ****"
endif
Example File: AutoPreScript

6

1
Pass_1
5
1
1
6
10
i

e
1
2
4
5
-.20
/
6
3
2
2
3
.10 1.5
4

0
4
2
/
3
2
5
20
12
.005
/
5
1
1
9
1
1
3
1
2

31 1118
7
/
3
3
2
1
/
7
1

e
y