MODELING MICROSTRUCTURAL DEVELOPMENT IN THE FORGING OF WASPALOY TURBINE ENGINE DISKS

DISSERTATION

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By

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To My Wife, Fang Wen
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PUBLICATIONS


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CHAPTER I

INTRODUCTION

1.1 Background

Aeroengine turbine disks operate under severe conditions including high stress caused by combined centrifugal force and thermal gradient during operation. These severe working environments require the disks to be made from superalloys, and to be properly forged and heat treated to possess good microstructure and properties. Grain size control in the forging of superalloy disks is an important issue in pursuing integrity of structure and properties. This is because the grain size is closely related to mechanical properties and performance of the disks.

For satisfactory performance in service two types of microstructures are required in a forged disk: (1) a grain size of ASTM 10-14 for tensile strength, ductility and resistance to crack nucleation in low cycle fatigue; or (2) a grain size of ASTM 4-8 required for creep strength and resistance to crack propagation (Couts, Jr. et al., 1987).

1.2 Thermomechanical Processing of Superalloys

Thermomechanical processing is the practice which was developed in the mid 1960’s for nickel alloys to enhance mechanical properties by exploiting
the interaction of temperature and strain (deformation) (Wilkinson, 1989). In thermomechanical processing, the deformation operation and the heat treatment are combined together to obtain desired microstructure (Howson et al., 1989). A so called "MINIGRAIN" process developed by Pratt & Whitney engineers produced extremely fine grain size for Incoloy 901 and Inconel 718 (Brown et al., 1972). These two gamma matrix alloys are hardened by two precipitation phases \((\eta/\gamma'\text{ for 901 and } \delta/\gamma''\text{ for 718})\). The control of precipitation phases in these alloys serves as the basis for thermomechanical processing in order to obtain desired properties. Phase relationships for alloy 901 and Waspaloy are shown in Fig. 1.1 and 1.2 respectively (Muzyka, 1979).

The Ladish Co. developed a controlled manufacturing process called "ISOCON" in order to overcome the non-uniformity in structure and properties in the Waspaloy forging (Stewart, 1988). The "ISOCON" process is a combination of "isothermal" and "conventional hammer" forging techniques which gives rise to improved, uniform and defect tolerant microstructure. The initial forging sequence is performed isothermally in order to produce a uniform microstructure which is conformed by stringent ultrasonic inspection. The final forging sequence is performed on a hammer for improved tensile strengths and refined contour outline.

Taking Incoloy 901 as an example, in the process presented in Fig. 1.3, a conditioning heat treatment, on the order of eight hours at 1650°F, is applied which precipitates the phase, \(\eta\), in a needle-like Widmanstatten pattern. Working is then done at about 1750°F, which is below the \(\eta\) solvus temperature, with final deformation occurring below the recrystallization
Fig. 1.1 Phase relationships for Incoloy 901 (Muzyka, 1979).
Fig. 1.2 Phase relationships for Waspaloy (Muzyka, 1979).
Fig. 1.3 Proprietary process developed by Pratt & Witney engineers which produces extremely fine grain size for Incoloy 901 and Inconel 718 parts (Bradley et al., 1974).
temperature. A fine grained structure is generated by subsequent recrystallization below the η solvus. The needle-like η becomes spherical and pins the grain boundaries. Thus grain growth is inhibited during recrystallization. Because subsequent aging to precipitate γ' and some additional η does not change the grain size, a structure of ASTM 10 or finer is obtained (Bradley et al., 1974). In the process presented in Fig. 1.3, the conditioning heat treatment is critical.

Jackman (1976) has shown the microstructural differences in conventionally forged and thermomechanically processed Incoloy 901 (Fig. 1.4). In conventionally forged alloy 901, which has had no η-sub-solvus heat treatment, η particles are very few in the matrix before forging (Fig. 1.4c). By contrast, in thermomechanical processing of 901 which has the sub-solvus heat treatment, abundant η is presented in the matrix before forging (Fig. 1.4d) thereby pinning grain boundaries during recrystallization in the forging. Similar practice is also reported by Luo et al. (1991).

Previous researches on grain size control for superalloy disk forging are mainly empirical. In these, qualitative relationship between process variables and microstructures as well as properties are discussed. However, they are limited by their inability to link thermomechanical histories of various material points in the workpiece during the forging process with their structure and properties. Without these relationships, it is difficult to predict correctly the microstructural phenomena for a process, especially a new process.
1.3 The Objectives of This Research

The successful control of grain size in the forging of superalloy disks requires a complete model of the microstructural evolution. The grain size obtained from a superalloy forging is related to the dynamic and metadynamic recrystallization as well as grain growth behavior of the material. Those behaviors are in turn related to the thermomechanical histories of the workpiece. The thermomechanical histories of the workpiece during the forging process include (a) the initial structure, (b) the preheat condition before forging, (c) the temperature, strain, strain rate experienced in the forging, and (d) the post forging dwell and cooling conditions. To control grain size, the relationships between the thermomechanical history, and the recrystallization and grain growth behavior of the material need to be quantified. The thermomechanical histories of the material can be obtained from FEM simulations. The quantified recrystallization and grain growth behavior of the material can be obtained from controlled experiments.

The objective of this research is to combine physical testing and FEM simulations to develop an empirical based physical metallurgy model for the microstructural evolution in superalloy deformation, and to combine this model with FEM to predict the percentage of recrystallization and grain size in the forging of disks.
Fig. 1.4 Microstructures in conventionally forged (a, c) and thermomechanically processed (b, d) Incoloy 901 (Jackman, 1976).
CHAPTER II

REVIEW OF MICROSTRUCTURAL MODELS

In this chapter the major models for prediction of microstructural evolution for steels and other materials are reviewed. Collection and evaluation of these models were helpful for the determination of research approach.

2.1 Complex Model (Miller)

A classical example of a complex model is MATMOD (MATerial MODel) proposed by Miller (1976) and later developed by others. In this unified model a single equation is proposed to calculate all of the nonelastic strain due to composition and processing. The nature of the equations are such that when appropriate boundary conditions are imposed appropriate phenomena are simulated. The main disadvantage of this model is that a large number of constants need to be evaluated for a given material. This prompted Henshall and Miller (1990) to propose changes to reduce the number of constants in their modified model. The modified model MATMOD-BSSOL (Back Stresses from Solute) contains 28 constants as compared to 44 in MATMOD. Table 2.1 shows the values of the constants needed in MATMOD-BSSOL for aluminum and Al-Mg alloys (Henshall, et al., 1990).
Table 2.1 The MATMOD-BSSOL material constants for pure aluminum and Al-Mg alloys.

<table>
<thead>
<tr>
<th>Constant</th>
<th>Pure Al</th>
<th>Al-1.15% Mg*</th>
<th>Al-3.1% Mg*</th>
<th>Al-5.8% Mg*</th>
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<td>B</td>
<td>$1.0 \times 10^4$</td>
<td>$1.0 \times 10^4$</td>
<td>$1.0 \times 10^4$</td>
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<td>d</td>
<td>2.0</td>
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<tr>
<td>$T_1$</td>
<td>461 K</td>
<td>461 K</td>
<td>461 K</td>
<td>461 K</td>
</tr>
<tr>
<td>$Q^*$</td>
<td>35,500 cal/mol</td>
<td>35,500 cal/mol</td>
<td>35,500 cal/mol</td>
<td>35,500 cal/mol</td>
</tr>
<tr>
<td>$E_0$</td>
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<td>$1.16 \times 10^4$</td>
<td>$1.16 \times 10^4$</td>
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</tr>
<tr>
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<td>$-1.55 \times 10^{-3}$</td>
<td>$-1.55 \times 10^{-3}$</td>
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<td>$2.00 \times 10^{6}$</td>
<td>$2.00 \times 10^{6}$</td>
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<td>3.1</td>
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<td>$6.00 \times 10^{-3}$</td>
<td>$3.30 \times 10^{-3}$</td>
<td>$2.00 \times 10^{-3}$</td>
</tr>
<tr>
<td>$A_4$</td>
<td>$5.50 \times 10^{-3}$</td>
<td>$1.22 \times 10^{-4}$</td>
<td>$3.35 \times 10^{-4}$</td>
<td>$1.80 \times 10^{-4}$</td>
</tr>
<tr>
<td>$C_3$</td>
<td>$7.53 \times 10^{-7}$</td>
<td>$7.53 \times 10^{-7}$</td>
<td>$7.53 \times 10^{-7}$</td>
<td>$7.53 \times 10^{-7}$</td>
</tr>
<tr>
<td>$G_3$</td>
<td>$3.00 \times 10^{-5}$</td>
<td>$2.00 \times 10^{-4}$</td>
<td>$2.40 \times 10^{-4}$</td>
<td>$2.87 \times 10^{-4}$</td>
</tr>
<tr>
<td>$g_3$</td>
<td>0.5</td>
<td>0.5</td>
<td>0.5</td>
<td>0.5</td>
</tr>
<tr>
<td>$F_{s, \text{sol, max}}$</td>
<td>1.33</td>
<td>3.75</td>
<td>7.00</td>
<td></td>
</tr>
<tr>
<td>$Z_{\text{max}}$</td>
<td>$9.83 \times 10^{10}$</td>
<td>$9.83 \times 10^{10}$</td>
<td>$9.83 \times 10^{10}$</td>
<td></td>
</tr>
<tr>
<td>$\beta$</td>
<td>25.0</td>
<td>25.0</td>
<td>25.0</td>
<td>25.0</td>
</tr>
<tr>
<td>$Q_{\text{sol}}$</td>
<td>30,000 cal/mol</td>
<td>30,000 cal/mol</td>
<td>30,000 cal/mol</td>
<td></td>
</tr>
</tbody>
</table>

*Constants affecting cyclic deformation are only approximate due to the lack of cyclic data.*
2.2 Monte Carlo Model

A Monte Carlo model for recrystallization and grain growth has been proposed to simulate the phenomenon at the lattice level (Anderson et al., 1983; Rollett et al., 1990).

The grain growth model is defined as the following. A certain numbers of lattice sites are chosen and assigned a spin number between 1 and Q, which corresponds to the orientation of the lattice. The lattice sites are then randomly selected and a new spin number is chosen. The change in energy associated with the change in orientation is evaluated. The change in orientation which is energy favorable is kept. The neighboring sites which share the same orientation belong to the same grain. The grain boundary is defined between sites with different orientations. The recrystallization is simulated by adding new grains or nuclei to the model at random locations (Rollett et al., 1990).

The Monte Carlo model for simulation of recrystallization and grain growth though a fundamental metallurgical model, is still far away from the application to the modeling of real materials for engineering applications.

2.3 FEM Based Model

Habraken (1989) presented a finite element model which performed simultaneously thermal, mechanical and microstructural analysis for steel. The intent of the coupled model was to predict the final microstructures which include percentage of each phase (i.e. the austenite, ferrite, pearlite,
bainite and martensite), the residual stresses, and the precise final geometry of the steel piece.

In the heat transfer model used by Habraken, the enthalpy transformation from austenite into various components such as ferrite, pearlite, cementite, bainite, and martensite is included in heat conduction equation. In his phase transformation model, both diffusional transformation and martensitic transformation are considered. In their stress and strain model the total strain rate is divided into five parts: elastic strain rate, thermal strain rate, transformation strain rate, classical plastic strain rate, and plasticity transformation rate. In his results, the only microstructural prediction shown is the percentage of martensite.

The FEM based microstructural constitutive model could be a model eventually used in prediction of microstructural evolution. However, a complete material data base needs to be available before this model can be applied to solve engineering problems.

2.4 Metallurgical Model

Sellars and his co-workers (University of Sheffield) have been working on modeling microstructural evolution for hot rolling since late 70s (Sellars, 1979). Their model covers all possible physical metallurgical phenomenon related to recrystallization and grain growth in hot rolling of steel (Sellars, 1990). In addition to Sellars, Yada and his coworkers at Nippon Steel (1990), Saito et al. at Kawasaki Steel (1985) also published their models. Recently, Devadas et al. (1991) summarized models developed by the above four groups
of researchers. This summary is shown in Table 2.2 (Devadas, 1991). These models established the recrystallization and grain growth relationships for C-Mn steel. Though the detailed model differs from group to group they basically represent a similar approach that is characterizing dynamic structural changes (dynamic recrystallization) and static structural changes (metadynamic recrystallization, static recrystallization and grain growth) with empirical equations based on fundamental physical metallurgy principles.

These models are not simple linear regression models which may not be valid when processing conditions change or the geometry of the part changes. The physical metallurgy based empirical models grasp the essence of the microstructural changes. Thus, they are an ideal candidates to combine with FEM simulation to predict microstructures resulted from different process conditions and part geometries.
Table 2.2 Summary of reported recrystallization and grain growth relationships (Devadas, 1991).

<table>
<thead>
<tr>
<th>University of Sheffield</th>
<th>Nippon Steel</th>
<th>Kawasaki Steel</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sellars and Co-Workers\textsuperscript{11-13}</td>
<td>Yada \textit{et al.}\textsuperscript{16,18}</td>
<td>Suito \textit{et al.}\textsuperscript{19,20}</td>
</tr>
<tr>
<td><strong>Dynamic Recrystallization</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>( \varepsilon_r = 4.9 \cdot 10^{-4} d_0^{1/2} Z^{0.15} )</td>
<td>( \varepsilon_r = 4.76 \cdot 10^{-4} \exp \left( \frac{8000}{T} \right) )</td>
<td>( \varepsilon_r = 3.68 \cdot 10^{-4} d_0^{0.60} )</td>
</tr>
<tr>
<td>( \varepsilon_r = \sigma_0 )</td>
<td>( d_{\alpha r} = 22.600 Z^{-0.17} )</td>
<td>( d_{\alpha r} = 2.82 \cdot 10^2 Z^{-0.14} )</td>
</tr>
<tr>
<td>( Z = \dot{\varepsilon} \exp \frac{Q}{R \cdot T} )</td>
<td>( Z = \dot{\varepsilon} \exp \frac{Q}{R \cdot T} )</td>
<td>( Z = \dot{\varepsilon} \exp \frac{Q}{R \cdot T} )</td>
</tr>
<tr>
<td>( Q = 312 \text{ kJ/mol} ) (metadynamic)</td>
<td>( Q = 257.1 \text{ kJ/mol} )</td>
<td>( Q = 312 \text{ kJ/mol} )</td>
</tr>
<tr>
<td>( X = 1 - \exp \left( -0.693 \frac{t}{t_0} \right) )</td>
<td>( X_{\alpha r} = 1 - \exp \left( -0.693 \frac{t}{t_{0r}} \right) )</td>
<td>( \varepsilon_{0r} = 1.144 \cdot 10^{-7} d_0^{0.60} \exp \left( \frac{6420}{T} \right) )</td>
</tr>
</tbody>
</table>

| **Static Recrystallization** | | |
| \( X = 1 - \exp \left( -0.693 \frac{t}{t_0} \right) \) | \( X = 1 - \exp \left( -0.693 \frac{t}{t_{0r}} \right) \) | \( X = 1 - \exp \left( -0.693 \frac{t}{t_{0r}} \right) \) |
| \( \varepsilon < 0.8 \varepsilon_r \) | \( t_{0r} = 2.2 \cdot 10^3 \) | \( t_{0r} = 2.5 \cdot 10^4 \varepsilon^{-4} d_0^2 \exp \left( \frac{30,000}{T} \right) \) |
| \( t_{0r} = 2.5 \cdot 10^{-7} Z^{-0.6} \exp \left( \frac{Q_r}{R \cdot T} \right) \) | \( Q_r = 300 \text{ kJ/mol} \) | \( d_{\alpha r} = 0.5 \dot{d}_0^{0.6} \varepsilon^{-1} \) |
| \( \varepsilon > 0.8 \varepsilon_r \) | | |
| \( t_{0r} = 1.06 \cdot 10^{-7} Z^{-0.6} \exp \left( \frac{Q_r}{R \cdot T} \right) \) | | |
| \( Q_r = 300 \text{ kJ/mol} \) | | |
| \( d_{\alpha r} = 0.5 \dot{d}_0^{0.6} \varepsilon^{-1} \) | | |
| \( (\varepsilon < \varepsilon^*) \) | | |
| \( d_{\alpha r} = 1.8 \cdot 10^{-8} Z^{-0.15} \) | | |
| \( (\varepsilon > \varepsilon^*) \) | | |
| \( \varepsilon^* = 0.57 \dot{d}_0^{0.17} \varepsilon_r \) | | |

| **Grain Growth** | | |
| \( d^{10} = d_0^{10} + A \dot{t} \exp \left( \frac{-Q_u}{RT} \right) \) | \( d^{10} = d_0^{10} + A \dot{t} \exp \left( \frac{-Q_u}{RT} \right) \) | \( d^{10} = d_0^{10} + A \dot{t} \exp \left( \frac{-Q_u}{RT} \right) \) |
| \( T > 1100 \text{ °C} \) | \( A = 3.87 \cdot 10^{13} \) | \( A = 3.87 \cdot 10^{13} \) |
| \( A = 400 \text{ kJ/mol} \) | \( T < 1100 \text{ °C} \) | \( Q_u = 400 \text{ kJ/mol} \) |
| \( A = 3.87 \cdot 10^{13} \) | | |
| \( Q_u = 914 \text{ kJ/mol} \) | | |
CHAPTER III

CURRENT APPROACH

3.1 The Methodology of Modeling

Our approach to model microstructural evolution is to adapt Sellars's approach, and to develop empirical equations for Waspaloy based on fundamental physical metallurgy principles. The empirical metallurgical model developed will be used in FEM post processing and in the prediction of recrystallization and grain size for Waspaloy disk forgings.

In the development of the model, both experiments and FEM simulations are combined together. The experiments provide microstructural data with specified thermomechanical history. The FEM simulations provide temperature, strain, strain rate and their histories for the experimental samples regardless whether they are lab samples or real forging parts. Thus, the experimental data can be processed more precisely. Details of the experiments and FEM simulation are discussed in the following sections.

3.2 Experimental Investigations

To investigate the evolution of microstructure in the forging of Waspaloy disks, both laboratory experiments and real forging tryout were conducted. The purpose of lab experiments was to investigate in depth the effect of
process variables under controlled conditions, and to develop a prototype of a model for the analysis of generic forging processes. The lab experiments included (a) preheating tests to investigate the as heated grain size; and (b) compression tests with different process parameters and different cooling conditions in a Gleeble testing machine to investigate recrystallization and immediate grain growth. The real forging trials included both pancake and closed-die forgings which were conducted at an aerospace forging company to verify and refine the model developed from the lab experiments.

3.2.1 Material

The samples used in both preheating and compression tests were obtained from hot rolled bar of ø152 mm (ø 6 inch). The composition of the material used in this study is shown in Table 3.1.

Table 3.1 The composition of Waspaloy used in this study.

<table>
<thead>
<tr>
<th>Ni</th>
<th>Cr</th>
<th>Co</th>
<th>Mo</th>
<th>Ti</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>58</td>
<td>19.5</td>
<td>13.5</td>
<td>4.3</td>
<td>3.1</td>
<td>1.4</td>
</tr>
</tbody>
</table>

The lower boundary of the γ' solvus of the Waspaloy is around 1027°C (1880°F). The phase relationships of Waspaloy is shown in Fig. 3.1.

The grain size from the as received bar was examined in three representative locations: the outer layer, the mid radius, and the center.
Fig. 3.1 Phase relationships for Waspaloy (Muzyka, 1979).
Fig. 3.2 (a) shows the grain structure at the outer layer which can be characterized as a necklace structure with grains of ASTM 1.4 surrounded by grains of ASTM 7. This kind of structure is possibly due to partial recrystallization during hot rolling. Fig. 3.2(b) shows the structure in the mid radius; it has a grain size of around ASTM 6 with bands of γ' segregation. Fig. 3.2 (c) shows the structure at the center of the bar where relatively uniformed grains with a size of around ASTM 4-5 are observed.

3.2.2 Preheating Experiments

To obtain quantitative information on the status of grain size prior to forging, samples were heated under different conditions using a small furnace. This study was useful for the estimation of the grain size developed right after preheating before the forging started. The dimensions of the Waspaloy sample used in both the preheating and compression tests were φ6.4 x 11.4 mm (φ 0.25x0.45 inch). The sample was obtained by EDM machining of the as received hot rolled bar.

The conditions selected for preheating study were:

(1) Preheat temperatures: 1010°C (1850°F), 1066 (1950°F), and 1121°C (2050°F);

(2) Initial grain sizes: ASTM 2, 4 and 6;

(3) Soaking times: 10, 30, 50 minutes.

The samples with grain sizes of ASTM 4 and 6 were obtained from mid radius and the center of the as received bar, respectively. The samples with the
Fig. 3.2 As received grain structures of hot rolled Waspaloy bar: (a) outer layer, (b) mid radius, and (c) center.
Fig. 3.2 Continued.
ASTM 2 grain size were obtained by heating the samples obtained from the outside layer at 1080°C (1976°F) for one hour.

In each of the preheating tests, the sample was put in a furnace which was set to the selected temperature and filled with argon. After a predetermined soaking time was reached the sample was quenched immediately to "freeze" the structure for metallographic examination.

3.2.3 Gleeble Compression Tests

Compression tests in a Gleeble testing machine were designed to simulate press and hammer forgings: Three sets of compression tests were conducted (1) single stroke compression tests, (2) multiblow compression tests and (3) single stroke compression tests with different post forging holding times.

3.2.3.1 Single Stroke Compression Tests

Single stroke compression tests were conducted to study (1) the effects of different process parameters on the final grain size obtained after forging and (2) the behavior of dynamic recrystallization.

Variables investigated in the compression tests were:

(1) Forging temperatures: 1010°C (1850°F), 1066 (1950°F), and 1121°C (2050°F);

(2) Initial grain sizes: ASTM 2, 4 and 6;

(3) Strain rates: 0.02/s, 0.2/s, 3.5/s;
(4) Strains: 0.4 (33% deformation), 0.8 (55% deformation), 1.2 (70% deformation).

To efficiently design the experiments, the "Taguchi Method" was used for the single stroke compression tests. The "Taguchi Method" makes it possible to study the relationship between a large number of variables with a relatively small number of experiments (Phadke, 1989). The matrix of the experiments conducted to simulate press forging is shown in Table 3.2. To obtain more data for the largest deformation (70%), tests to simulate press forging with a strain of 1.2 were combined with all three temperatures. Therefore, there are two sets of orthogonal matrices of experiments in Table 3.2. Experiments 1-9 belong to the first one and experiments 3, 6, 9, 10 - 15 belong to the other.

3.2.3.1 Multiblow Compression Tests

Multiblow compression tests were conducted to simulate the hammer forging conditions with dwell time.

Variables investigated in multiblow tests were:

(1) Initial grain sizes, GS: ASTM 2, 4 and 6;
(2) γ sub-solvus heat treatment before forging, HT;
(3) Forging temperatures, T₀: 1010°C (1850°F), 1066°C (1950°F), and 1121°C (2050°F);
(4) Strain rates, \( \dot{\varepsilon} \): 0.02/s, 0.2/s, 3.5/s;
(5) Strains, \( \varepsilon \): 0.4 (33% deformation), 0.8 (55% deformation), 1.2 (70% deformation);
(6) Dwell times between blows, \( t_d \): 1s, 3s, 5s;
Table 3.2 Matrix for Single Stroke Compression Tests.

<table>
<thead>
<tr>
<th>Ex #</th>
<th>Temperature °C (°F)</th>
<th>Strain</th>
<th>Strain Rate 1/s</th>
<th>Initial ASTM Grain Size</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>1010 (1850)</td>
<td>0.4</td>
<td>0.02</td>
<td>2</td>
</tr>
<tr>
<td>2</td>
<td>1010 (1850)</td>
<td>0.8</td>
<td>0.2</td>
<td>4</td>
</tr>
<tr>
<td>3</td>
<td>1010 (1850)</td>
<td>1.2</td>
<td>3.5</td>
<td>6</td>
</tr>
<tr>
<td>4</td>
<td>1066 (1950)</td>
<td>0.4</td>
<td>0.2</td>
<td>6</td>
</tr>
<tr>
<td>5</td>
<td>1066 (1950)</td>
<td>0.8</td>
<td>3.5</td>
<td>2</td>
</tr>
<tr>
<td>6</td>
<td>1066 (1950)</td>
<td>1.2</td>
<td>0.02</td>
<td>4</td>
</tr>
<tr>
<td>7</td>
<td>1121 (2050)</td>
<td>0.4</td>
<td>3.5</td>
<td>4</td>
</tr>
<tr>
<td>8</td>
<td>1121 (2050)</td>
<td>0.8</td>
<td>0.02</td>
<td>6</td>
</tr>
<tr>
<td>9</td>
<td>1121 (2050)</td>
<td>1.2</td>
<td>0.2</td>
<td>2</td>
</tr>
<tr>
<td>10</td>
<td>1010 (1850)</td>
<td>1.2</td>
<td>0.02</td>
<td>2</td>
</tr>
<tr>
<td>11</td>
<td>1010 (1850)</td>
<td>1.2</td>
<td>0.2</td>
<td>4</td>
</tr>
<tr>
<td>12</td>
<td>1066 (1950)</td>
<td>1.2</td>
<td>0.2</td>
<td>6</td>
</tr>
<tr>
<td>13</td>
<td>1066 (1950)</td>
<td>1.2</td>
<td>3.5</td>
<td>2</td>
</tr>
<tr>
<td>14</td>
<td>1121 (2050)</td>
<td>1.2</td>
<td>0.02</td>
<td>6</td>
</tr>
<tr>
<td>15</td>
<td>1121 (2050)</td>
<td>1.2</td>
<td>3.5</td>
<td>4</td>
</tr>
</tbody>
</table>
(7) Strain increments in each blow, $\Delta \bar{\varepsilon}$: 0.2 (for simulation of blocker), 0.1 (for simulation of finisher).

The "Taguchi Method" was also used for the multiblow compression tests. The matrix of the experiments conducted is shown in Table 3.3.

3.2.3.3 Single Stroke Compression Tests with Different Post Forging Holding Times

To model metadynamic/static recrystallization and grain growth mechanisms single stroke compression tests with different post forging holding times were also conducted.

Variables investigated in multiblow tests were:

(1) Forging temperatures: 1010°C (1850°F), 1066 (1950°F), and 1121°C (2050°F);
(2) Initial grain sizes: ASTM 2, 4 and 6;
(3) Strain rates: 0.02/s, 3.5/s;
(4) Strains: 0.14 (13% deformation);
(5) Post forging holding times before fast cooling: 0 s, 1s, 3s, 5s, 10s, 20s, 60s;

A randomized block design was used for single stroke compression tests with different post forging holding times. A part of the matrix of the experiments conducted to simulate press forging is shown in Table 3.4.

Additional post forging holding tests were also conducted for some selected single stroke tests in Table 3.2 and multiblow tests in Table 3.3. The conditions for these tests are shown in Tables 3.5 to 3.6.
Table 3.3 Matrix for Multiblow Compression Tests.

<table>
<thead>
<tr>
<th>Ex #</th>
<th>GS</th>
<th>HT</th>
<th>T₀</th>
<th>( \ddot{\varepsilon} )</th>
<th>( \bar{\varepsilon} )</th>
<th>t₀</th>
<th>( \Delta \bar{\varepsilon} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>2</td>
<td>yes</td>
<td>1850</td>
<td>3.5</td>
<td>0.4</td>
<td>1</td>
<td>0.2</td>
</tr>
<tr>
<td>2</td>
<td>4</td>
<td>yes</td>
<td>1950</td>
<td>0.2</td>
<td>0.8</td>
<td>3</td>
<td>0.1</td>
</tr>
<tr>
<td>3</td>
<td>6</td>
<td>yes</td>
<td>2050</td>
<td>0.02</td>
<td>1.2</td>
<td>5</td>
<td>0.2</td>
</tr>
<tr>
<td>4</td>
<td>2</td>
<td>yes</td>
<td>1850</td>
<td>0.2</td>
<td>0.8</td>
<td>5</td>
<td>0.2</td>
</tr>
<tr>
<td>5</td>
<td>4</td>
<td>yes</td>
<td>1950</td>
<td>0.02</td>
<td>1.2</td>
<td>1</td>
<td>0.2</td>
</tr>
<tr>
<td>6</td>
<td>6</td>
<td>yes</td>
<td>2050</td>
<td>3.5</td>
<td>0.4</td>
<td>3</td>
<td>0.1</td>
</tr>
<tr>
<td>7</td>
<td>4</td>
<td>yes</td>
<td>1850</td>
<td>0.02</td>
<td>0.4</td>
<td>5</td>
<td>0.1</td>
</tr>
<tr>
<td>8</td>
<td>6</td>
<td>yes</td>
<td>1950</td>
<td>3.5</td>
<td>0.8</td>
<td>1</td>
<td>0.2</td>
</tr>
<tr>
<td>9</td>
<td>2</td>
<td>yes</td>
<td>2050</td>
<td>0.2</td>
<td>1.2</td>
<td>3</td>
<td>0.2</td>
</tr>
<tr>
<td>10</td>
<td>6</td>
<td>no</td>
<td>1850</td>
<td>0.2</td>
<td>1.2</td>
<td>1</td>
<td>0.1</td>
</tr>
<tr>
<td>11</td>
<td>2</td>
<td>no</td>
<td>1950</td>
<td>0.02</td>
<td>0.4</td>
<td>3</td>
<td>0.2</td>
</tr>
<tr>
<td>12</td>
<td>4</td>
<td>no</td>
<td>2050</td>
<td>3.5</td>
<td>0.8</td>
<td>5</td>
<td>0.2</td>
</tr>
<tr>
<td>13</td>
<td>4</td>
<td>no</td>
<td>1850</td>
<td>3.5</td>
<td>1.2</td>
<td>3</td>
<td>0.2</td>
</tr>
<tr>
<td>14</td>
<td>6</td>
<td>no</td>
<td>1950</td>
<td>0.2</td>
<td>0.4</td>
<td>5</td>
<td>0.2</td>
</tr>
<tr>
<td>15</td>
<td>2</td>
<td>no</td>
<td>2050</td>
<td>0.02</td>
<td>0.8</td>
<td>1</td>
<td>0.1</td>
</tr>
<tr>
<td>16</td>
<td>6</td>
<td>no</td>
<td>1850</td>
<td>0.02</td>
<td>0.8</td>
<td>3</td>
<td>0.2</td>
</tr>
<tr>
<td>17</td>
<td>2</td>
<td>no</td>
<td>1950</td>
<td>3.5</td>
<td>1.2</td>
<td>5</td>
<td>0.1</td>
</tr>
<tr>
<td>18</td>
<td>4</td>
<td>no</td>
<td>2050</td>
<td>0.02</td>
<td>0.4</td>
<td>1</td>
<td>0.2</td>
</tr>
<tr>
<td>19</td>
<td>4</td>
<td>yes</td>
<td>1850</td>
<td>0.02</td>
<td>1.2</td>
<td>4</td>
<td>0.2</td>
</tr>
<tr>
<td>20</td>
<td>4</td>
<td>yes</td>
<td>1850</td>
<td>0.02</td>
<td>1.2</td>
<td>3</td>
<td>0.1</td>
</tr>
</tbody>
</table>
Table 3.4 Matrix Used for Single Stroke Compression Tests with Different Post Forging Holding Times.

<table>
<thead>
<tr>
<th>Ex #</th>
<th>T₀, °F</th>
<th>Gₐ₀ (ASTM #)</th>
<th>˙ε</th>
<th>ε</th>
<th>Holding time (s)</th>
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<td>0.14</td>
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<td>0.14</td>
<td>20</td>
</tr>
<tr>
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<td>0.14</td>
<td>60</td>
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<tr>
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<td>2/4/6</td>
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</tr>
<tr>
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<td>2/4/6</td>
<td>3.5</td>
<td>0.14</td>
<td>20</td>
</tr>
<tr>
<td>6</td>
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<td>2/4/6</td>
<td>3.5</td>
<td>0.14</td>
<td>60</td>
</tr>
<tr>
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<td>60</td>
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<tr>
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<td>2050</td>
<td>2/6</td>
<td>3.5</td>
<td>0.14</td>
<td>60</td>
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</table>
Table 3.5 Matrix used for Additional Single Stroke Compression Tests with Different Post Forging Holding Times.

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<tr>
<th>Ex #</th>
<th>$T_0, , ^\circ\text{F}$</th>
<th>$G_{S_0}$</th>
<th>$\dot{\varepsilon}$</th>
<th>$\bar{\varepsilon}$</th>
<th>Holding time (s)</th>
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</thead>
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<td>0.2/s</td>
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<td>0.2/s</td>
<td>0.4</td>
<td>3</td>
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<td>3.5/s</td>
<td>0.8</td>
<td>0</td>
</tr>
<tr>
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<td>1950</td>
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<td>3.5/s</td>
<td>0.8</td>
<td>3</td>
</tr>
<tr>
<td>8F</td>
<td>2050</td>
<td>6</td>
<td>0.02/s</td>
<td>0.4</td>
<td>0</td>
</tr>
<tr>
<td>8H</td>
<td>2050</td>
<td>6</td>
<td>0.02/s</td>
<td>0.4</td>
<td>3</td>
</tr>
<tr>
<td>7F</td>
<td>2050</td>
<td>4</td>
<td>3.5/s</td>
<td>0.4</td>
<td>0</td>
</tr>
<tr>
<td>7H</td>
<td>2050</td>
<td>4</td>
<td>3.5/s</td>
<td>0.4</td>
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</tbody>
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Table 3.6 Matrix used for Additional Multiblow Compression Tests with Fast Post Forging Cooling.

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<th>$G_{S_0}$</th>
<th>$\dot{\varepsilon}$</th>
<th>$\bar{\varepsilon}$</th>
<th>Holding time (s)</th>
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<td>0.2/s</td>
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<td>sh18F</td>
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<td>4</td>
<td>0.02/s</td>
<td>0.4</td>
<td>0</td>
</tr>
<tr>
<td>sh15F</td>
<td>2050</td>
<td>2</td>
<td>0.02/s</td>
<td>0.8</td>
<td>0</td>
</tr>
</tbody>
</table>
3.2.3.4 Conditions in the Gleeble Tests

In each of the Gleeble tests, the sample was heated to the forging temperature for one minute and soaked at that temperature for 9 minutes. This heating condition is very close to the condition used with 10 minute soaking in preheating tests. This is because those samples which were soaked for 10 minutes in the preheating test also needed certain time to reach the preheated temperature. Compression was conducted in a vacuum chamber. Graphite foil was placed between the sample and platens to provide interface lubrication. A pair of thermocouples was welded at the middle cylindrical surface of the sample to monitor its temperature. The temperature at the flat end of the sample where it contacted the platen was also measured in the test. The measured temperatures at the ends are shown in Fig. 3.3. It can be seen from this figure that the temperature is not uniform for the dimensions of the sample used in the Gleeble test. After each compression, the sample was either quickly or slowly cooled below the solvus of \( \gamma \) which is around 1027°C (1880°F). The cooling rate for fast cooling and slow cooling were around 60°C/s and 6°C/s respectively. The cooling rate used for all the tests in Tables 3.4 to 3.6 were fast cooling. The measured cooling rates from each individual tests were between 60 - 80°C/s.

Two tests were conducted for each set of experimental conditions. To reduce the influence of unknown variables on the experimental results, the sequence of experiments was randomized by a random number table. Each of the two tests was randomized independently. The sample used in each test was sectioned, mounted, ground, polished, and etched, after
which micrographs of the grain structure were taken. Grain size
distributions were calculated from the micrographs by the intercept
method (Vander Voort, 1984). Due to the small size of the samples, four
micrographs at 50 X were found to cover an entire quarter of the cross
section.

3.2.4 Pancake and Closed Die Forgings

Besides laboratory tests, fourteen pancake and two closed die forging tests
were conducted in an aerospace forging company for the purpose of
verification and refinement of the model in a manufacturing
environment. These tests include six isothermal forgings, six hammer
forgings and four a combination of the two. The forging covers a wide
temperature range from subsolvus to supersolvus and a wide strain rate
range from the low strain rate used in a isothermal forging to the high
strain rate present in a hammer forging.

3.3 Process Simulations

Process simulations were carried out using the finite element program
DEFORM (Oh, 1982; Wu and Oh, 1985) to simulate each of the
compression tests. Flow stress data used in the simulations came from
three sources: Kattus (1975), Guimaraes and Jonas (1981), and unpublished
research from Battelle Columbus Laboratory. A heat transfer simulation
was first run to obtain an initial temperature distribution of the sample
which matched the measured temperature at the end of the sample prior
to compression (Fig. 3.3). This initial temperature distribution was used as
the starting temperature for the FEM simulation of the corresponding compression test. The temperature of the platens used in FEM simulation was also selected according to the measured temperature mentioned above. The FEM simulation results were verified by checking whether predicted bulge shapes agreed with the bulge shapes obtained from the corresponding tests. Fig. 3.4 (a) shows the bulge shape of a sample obtained from a Gleeble compression test. Fig. 3.4 (b) shows the temperature distribution obtained from the corresponding FEM simulation. These figures show that not only does the bulge shape predicted from the FEM simulation agree with the experiment, but the predicted location of the temperature of 1880°F (Fig. 3.4b) which corresponds to the lower boundary of the γ' solvus temperature shown in Fig. 3.1 also agrees with the location where the bands of γ' segregation disappear (Fig. 3.4a).

The local grain sizes in the samples obtained by metallographic examination were correlated to the corresponding local strains and strain rates obtained from FEM simulations, as well as the measured temperatures in each test. Typical micrographs cover a quarter of the cross section showing the grain size distribution are as presented in Fig. 3.5a. A corresponding FEM simulation result of the effective strain distribution for this test is shown in Fig. 3.5b.

The advantage of using FEM simulation to process the experimental data is that the local strain, strain rate, and temperature can be determined for correlations. Thus, more than one data point can be obtained from one test. A group of FEM simulation results showing the locations of "target"
points and the corresponding histories of temperature, strain and strain rate are presented in Figs. 3.6 (a)-(d).
Fig. 3.3 Temperatures measured at the ends of Gleeble test samples.
Fig. 3.4 (a) Macrograph showing the bulge shape and the γ segregation in a Gleeble test sample and (b) the predicted temperature distribution obtained from the FEM simulation for this test.
Fig. 3.5 (a) Micrographs showing a quarter of the sample compressed at 1121°C (2050°F) with a starting strain rate of 3.5/s (fast cooling, due to die chilling ts=0). (b) Strain distribution obtained from the FEM simulation for this test.
Fig. 3.6 FEM simulation results for compression tests at 1066°C (1950°F) with a starting strain rate of 3.5/s, showing (a) location of tracing points, and the histories of (b) temperature, (c) effective strain, and (d) effective strain rate.
(c) Effective Strain vs Time (Second)

(d) Effective Strain Rate (1/s) vs Time (Second)

Fig. 3.6 Continued.
CHAPTER IV

DEVELOPMENT OF AN EMPIRICAL MODEL

In this chapter the procedure is discussed in the development of the empirical model for the prediction of the microstructural development in the forging of Waspaloy disks. There are four sections in this chapter. Section 4.1 discusses the development of the model for dynamic recrystallization. Section 4.2 discusses the development of the model for metadynamic recrystallization. Section 4.3 discusses the development of the model for grain growth. The last section, 4.4, summarized the equations developed in the model.

4.1 Dynamic Recrystallization

Dynamic recrystallization happens instantaneously during high temperature deformation. The fraction of dynamic recrystallization can be obtained by examining micrographs obtained from fast cooled samples after their deformation. Under production conditions pure dynamic recrystallization is difficult to achieve. However, a study of dynamic recrystallization is helpful for understanding the microstructural development in the forging.

The amount of dynamic recrystallization is related to as heated grain size, strain, temperature and strain rate in a hot deformation condition. In this section four important issues related to dynamic recrystallization are
discussed: (1) the peak strain, (2) the strain for 50% dynamic recrystallization, (3) the fraction of dynamic recrystallization, and (4) the size of dynamically recrystallized grain.

4.1.1 Peak Strain

Strain corresponding to the peak of the flow curve is an important measure for the onset of dynamic recrystallization. The occurrence of dynamic recrystallization modifies the appearance of flow curves. At the strain rates typical for forging of Waspaloy single peak flow curves are most common. A characteristic of this kind of flow curve is that the flow stress rises to a maximum at the peak strain $\varepsilon_p$. Thereupon, as a result of dynamic recrystallization, it diminishes to a value intermediate between the yield stress and the peak stress. The reason for this steady state curve following a single peak is that under the condition of high $Z$, the dislocation density can be built up very fast. Before recrystallization is complete, the dislocation densities at the center of recrystallizing grain have increased sufficiently that another cycle of nucleation occurs and new grain begin to grow again. Thus, average flow stress intermediate between the yield stress and the peak stress is maintained.

Typical flow curves for Waspaloy are shown in Fig. 4.1. The equations developed for the peak strain are:

\[
\bar{\varepsilon}_p = 5.375 \times 10^{-4} d_0^{0.54} Z^{0.106} \quad (T < 1850^\circ F) \tag{4.1}
\]

\[
\bar{\varepsilon}_p = 1.685 \times 10^{-4} d_0^{0.54} Z^{0.106} \quad (T > 1850^\circ F) \tag{4.2}
\]
Fig. 4.1 Flow curves show the peak of flow stress (Guimaraes and Jonas, 1981).
The form of the equation (4.1) and (4.2) is based on the equations developed by Sellars (1990) for steel. The constants were obtained from the Gleeble tests and isothermal pancake forgings.

The "d_0" denotes the "as heated grain size" before compression. The "Z" denotes the "Zener-Hollomon parameter", or the temperature compensated strain rate:

\[ Z = \dot{\varepsilon} \exp(Q/RT) \]  \hspace{1cm} (4.3)

where Q used for calculating Z was 468 kJ/mole which is obtained from Battelle.

4.1.2 Strain for 50% Dynamic Recrystallization

Micrographs taken from fast cooled samples obtained from compression tests conducted at a given temperature, strain rate, and as heated grain size with varying strains show that the dynamic recrystallization progresses in a sigmoidal manner with respect to strain. The Avrami equation can be used to describe a sigmoidal curve for the fraction of dynamic recrystallization vs. strain:

\[ X = 1 - \exp\left[-\ln2\left[\frac{\varepsilon}{\varepsilon_{0.5}}\right]^n\right] \]  \hspace{1cm} (4.4)

When the constants \( \varepsilon_{0.5} \) and n are determined, the relation for the fraction of dynamic recrystallization is determined.

The strain for 50% recrystallization, \( \varepsilon_{0.5} \) can be obtained from compression tests with different magnitudes of strain for a given condition of temperature,
strain rate, and as heated grain size as shown in Fig. 4.2. The \( n \) can be obtained by taking logarithm for equation (4.4) as shown in Fig. 4.3. The \( \bar{\varepsilon}_{0.5} \) is related to as heated grain size and \( Z \). The \( \bar{\varepsilon}_{0.5} \) and the \( n \) are determined to follow the relationship:

\[
\bar{\varepsilon}_{0.5} = 0.1449 \ d_0^{0.32} \ Z^{0.03}, \quad n=3.0 \quad (T < 1850^\circ F) \tag{4.5}
\]

\[
\bar{\varepsilon}_{0.5} = 0.056 \ d_0^{0.32} \ Z^{0.03}, \quad n=2.0 \quad (1850^\circ F < T < 1880^\circ F) \tag{4.6}
\]

\[
\bar{\varepsilon}_{0.5} = 0.035 \ d_0^{0.29} \ Z^{0.04}, \quad n=1.8 \quad (T > 1880^\circ F) \tag{4.7}
\]

When the temperature is below 1850°F the \( \gamma' \) precipitation present in the material impedes dynamic recrystallization. When the temperature is between 1850°F and 1880°F the impeding effects reduced a great amount. When the temperature is above 1880°F, which is the lower boundary of \( \gamma' \) precipitation, the impeding effects are even less due to the dissolve of the \( \gamma' \) precipitation into the matrix.

4.1.3 Fraction of Dynamic Recrystallization

After the strain for 50% dynamic recrystallization and the exponent \( n \) for equation (4.4) are determined the fraction of dynamic recrystallized grain is expressed as follows:

\[
X_{\text{dyn}} = 1 - \exp\{-\ln2[\bar{\varepsilon} / \bar{\varepsilon}_{0.5}]^{3.0}\} \quad (T < 1850^\circ F) \tag{4.8}
\]

\[
X_{\text{dyn}} = 1 - \exp\{-\ln2[\bar{\varepsilon} / \bar{\varepsilon}_{0.5}]^{2.0}\} \quad (1850^\circ F < T < 1880^\circ F) \tag{4.9}
\]

\[
X_{\text{dyn}} = 1 - \exp\{-\ln2[\bar{\varepsilon} / \bar{\varepsilon}_{0.5}]^{1.8}\} \quad (T > 1880^\circ F) \tag{4.10}
\]

Figs. 4.4 to 4.7 summarizes the experimental data and the fitted model for dynamic recrystallization at 1830°F, 1850°F, 1950°F, and 2050°F respectively. 1830°F and 1850°F are below the \( \gamma' \) solvus temperature and 1950°F and 2050°F
Fig. 4.2 Schematic of strain corresponding to 50% dynamic recrystallization.
Fig. 4.3 Determination of $n$ for dynamic recrystallization. "n" is the slope of the line fitted line.
Fig. 4.4 Fraction of dynamic recrystallization at 1830°F.
Fig. 4.5 Fraction of dynamic recrystallization at 1850°F.
Fig. 4.6 Fraction of dynamic recrystallization at 1950°F.
Fig. 4.7 Fraction of dynamic recrystallization at 2050°F.
are above it. The model curves in Fig. 4.7 shift to the right is caused by the reason that when the sample is compressed at 2050°F, with the fastest cooling rate possible in the Gleeble testing machine, it still takes at least two seconds for the sample to cool down to a temperature below the γ solvus. This is especially true for the sample with less deformation. To prove this, Fig. 4.8 shows a few tests with large strain and instantaneous cooling due to die chilling. The prediction is observed to be well within the range of experimental data.

It should be noted that the model developed here uses the form of equation (4.4) instead of some other form such as:

\[ X = 1 - \exp\{-\ln2[(\bar{\varepsilon} - \bar{\varepsilon}_c)/\bar{\varepsilon}_{0.5}]^n\} \]  

(4.11)

This is based on the observation that the critical strain for the start of dynamic recrystallization, which is usually given by:

\[ \bar{\varepsilon}_c = 0.8 \bar{\varepsilon}_p \]  

(4.12)

has been considered automatically in the sigmoidal equation (4.4). When \( \bar{\varepsilon}_{0.5} \) and \( n \) is determined there is a corresponding threshold. Fig. 4.9 shows that the dynamic recrystallization does not start when strain is less than 0.1. The calculated \( \bar{\varepsilon}_c \) for this case based on equation (4.2) and (4.12) is around 0.08. On the other hand when the form of equation (4.11) is used the prediction is way off the experimental data as shown in Fig. 4.10.
Fig. 4.8 Fraction of dynamic recrystallization at temperatures between 1996°F - 2034°F.
Fig. 4.9 Fraction of dynamic recrystallization at a temperature of 1850°F and low strain.
Fig. 4.10 Fraction of dynamic recrystallization predicted with the equation with the format of (4.11).
4.1.4 The Size of Dynamically Recrystallized Grain

The size of dynamically recrystallized grain is the function of Zener-Hollomon parameter, $Z$, only. This is because that when dynamic recrystallization reaches steady state the new grain occupies the entire space of the old grain. Thus, the effect of as heated grain size disappears. Though the dynamically recrystallized grain size is not related to strain, the strain has to reach the value of steady state strain to result in fully dynamic recrystallization. The relationship between dynamically recrystallized grain, $d_{\text{dyn}}$, and $Z$, is shown as follows:

$$d_{\text{dyn}} = 8103 \ Z^{-0.16} \ \text{(\mu m)} \quad (1850^\circ F < T < 1900^\circ F) \quad (4.13)$$

$$d_{\text{dyn}} = 108.85 \ Z^{-0.0456} \ \text{(\mu m)} \quad (T > 1900^\circ F) \quad (4.14)$$

Fig. 4.11 schematically represents Equations (4.13) and (4.14). It is seen that there is a difference between subsolvus forging and supersolvus forging in terms of dynamically recrystallized grain. The subsolvus forging gives fine grain size. The supersolvus forging gives coarse grain size. However, the subsolvus forging needs large strain to finish dynamic recrystallization as shown in equation (4.5) to (4.10).

4.2 Metadynamic Recrystallization

Metadynamic recrystallization is important in the determination of the grain size obtained under practical forging conditions. Metadynamic
Fig. 4.11 Logarithm of dynamic recrystallized grain size in μm vs. ln (Z) obtained from Gleeble compression tests \((Z=\dot{\varepsilon}\exp(Q/RT))\). Q used for calculating Z was 468 kJ/mole.
recrystallization occurs when a deformation stops at a strain which passes the peak strain but does not reach the steady state strain for dynamic recrystallization (McQueen and Jonas, 1975), which is the case for most regions in a forged part. Under metadynamic recrystallization condition, the partially recrystallized grain structure (Fig. 4.12a), which is observed right after deformation, changes to a fully recrystallized grain structure (Fig. 4.12b and c) by continuous growth of these dynamically recrystallized nuclei at a high temperature. The metadynamically recrystallized grains are coarser than the fully dynamically recrystallized grains. However, understanding this phenomenon is the key issue in the control of the size and the uniformity of the grain produced under a production condition. The amount of metadynamic recrystallization is related to the as heated grain size, the strain, the temperature, the strain rate and the holding time in a hot deformation. Important issues related to metadynamic recrystallization are: (1) the time for 50% metadynamic recrystallization, (2) the fraction of metadynamic recrystallization, and (3) the size of metadynamically recrystallized grain. They are discussed in the following sections.

4.2.1 Time for 50% Metadynamic Recrystallization

The metadynamic recrystallization is time dependent. For a given strain, strain rate, and as heated initial grain size, the metadynamic recrystallization progresses in a sigmoidal manner with respect to time:

\[ X_{\text{meta-dyn}} = 1 - \exp \left[ -\ln(2/t_0.5)^n \right] \]  

(4.15)
Fig. 4.12 Micrographs obtained from samples with different cooling histories after forging: (a) immediately fast cooling, (b) fast cooling after 3 seconds of holding, and (c) slow cooling. Forging temperature = 1066°C (1950°F), strain rate = 0.2/s, amount of deformation = 33%.
Fig. 4.12 Continued.
The $t_{0.5}$ can be obtained from compression tests with different holding times for a given temperature, strain, strain rate, and as heated grain size as shown in Fig. 4.13. The empirical $t_{0.5}$ for metadynamic recrystallization follows the following relationship:

$$t_{0.5-\text{meta-dyn}} = 4.54 \times 10^{-5} \bar{e}^{-1.2815} d_0^{0.5062} \bar{\varepsilon}^{-0.0729} \exp(9703/T)$$ (4.16)

The experimentally obtained $t_{0.5-\text{meta-dyn}}$ is shown in Fig. 4.14. It is seen that when the strain is larger than 0.5 the $t_{0.5-\text{meta-dyn}}$ is below 2 second for the process condition used here.

The $n$ for metadynamic recrystallization is found to be 1 as shown in Fig. 4.15. This number is typical for metadynamic recrystallization (Jonas, 1976, Devadas, 1991).

### 4.2.2 Fraction of Metadynamic Recrystallization

The fraction of metadynamic recrystallization progresses in a sigmoidal manner according to the following equation:

$$X_{\text{meta-dyn}} = 1 - \exp\left[-\ln(2)\left(t/t_{0.5}\right)^{1.0}\right]$$ (4.17)

Figs. 4.16 to 4.21 shows the fraction of metadynamic recrystallization vs. time obtained from the experiments and the predictions at 1950°F and 2050°F with different as heated initial grain size, strain, and strain rate conditions. The metadynamic recrystallization finishes sooner in the case with larger strain, finer as heated grain size, larger strain rate and higher holding temperature.
Fig. 4.13 Schematic for time corresponding to 50% metadynamic recrystallization.
Fig. 4.14 Experimentally obtained time corresponding to 50% metadynamic recrystallization.
Fig. 4.15 Determination of n for metadynamic recrystallization. "n" is the slope of the fitted line.
Fig. 4.16 Fraction of metadynamic recrystallization obtained at 1950°F with a strain of 0.22 and different holding time.
Fig. 4.17 Fraction of metadynamic recrystallization obtained at 1950°F with a strain of 0.33 and different holding time.
Fig. 4.18 Fraction of metadynamic recrystallization obtained at 1950°F with strain between 0.6-1.3 and different holding time.
Fig. 4.19 Fraction of metadynamic recrystallization obtained at 2050°F with a strain of 0.2 and different holding time.
Fig. 4.20 Fraction of metadynamic recrystallization obtained at 2050°F with a strain of 0.26 and different holding time.
Fig. 4.21 Fraction of metadynamic recrystallization obtained at 2050°F with strain between 0.5 - 1.8 and different holding time.
4.2.3 Metadynamic Recrystallized Grain Size

The grain size obtained at the end of metadynamic recrystallization is found to have the following relationship with the strain, the as heated grain size $d_0$, and Zener-Hollomon parameter, $Z$,

$$d_{\text{meta-dyn}} = 14.56 \varepsilon^{-0.44} d_0^{0.325} Z^{-0.0258} \tag{4.18}$$

A finer metadynamically recrystallized grain size is associated with cases with larger strain, finer as heated grain size, and higher $Z$.

The metadynamic recrystallized grain size in ASTM number vs. strain under conditions with different as heated grain sizes, temperatures and strain rates is shown in Fig. 4.22.

4.3 Grain Growth

Under high temperature deformation conditions grain growth happens rapidly after the completion of metadynamic recrystallization. Grain boundary energy is the driving force causing grain boundary motion at high temperature. Grain boundary energy is like the surface energy - it tends to minimize itself whenever possible by decreasing the grain boundary area. In general, grain growth will continue to occur at elevated temperatures until the balance between the grain boundary energy and the size and pacing of precipitation is reached.

Grain growth are studied under two conditions in the present research: (1) short time grain growth which is the grain growth right after metadynamic
Fig. 4.22 Metadynamic recrystallized grain size vs. strain under conditions with different as heated grain size, temperature and strain rate.
recrystallization, and (2) long time grain growth which is the grain growth during preheating period before the forging.

4.3.1 Short Time Grain Growth

The short time grain growth is characterized by compression tests with different post-forging holding times. From the micrographs obtained from the above mentioned compression tests, the microstructural evolution from partially dynamic recrystallization to fully metadynamic recrystallization, and to grain growth were observed. The change in grain size vs. time after the completion of metadynamic recrystallization is found to follow relationship:

\[ d^3 = d_0^3 + 2 \times 10^{26} \exp(-595000/RT)t \]  \hspace{1cm} (d: \mu m) \hspace{1cm} (4.19)

where \( d_0 \) is the grain size after complete metadynamic recrystallization, \( T \) is the temperature at which the sample is held in Kelvin, and the \( t \) is the holding time in second. The form of the equation (4.19) is well known for the characterization of grain growth. The reason for emphasizing that the \( d_0 \) is the grain size after complete metadynamic recrystallization is that after the completion of metadynamic recrystallization the dislocations have basically disappeared and the driving force for grain size change is the grain boundary energy only.

The experimentally obtained data and the model prediction for the short time grain growth are shown in Fig. 4.23. It is seen from Fig. 4.23 that the grain growth at a temperature around 2050°F is very fast. There was not much
Fig. 4.23 Grain growth vs. time after the completion of metadynamic recrystallization in Waspaloy.
difference in grain size after the completion of metadynamic recrystallization between the two samples obtained from compression tests at 1950°F and 2050°F (in Fig. 4.23). However, the grain growth makes a big difference in the final grain size for the two sets of test conducted at 1950°F and 2050°F.

4.3.2 Long Time Grain Growth

Long time grain growth occurs during the preheating period. As seen from the model developed for dynamic and metadynamic recrystallization that the as heated grain size plays an significant role there. Therefore modeling the as heated grain size is important. Grain size vs. time data for three different preheating temperatures, namely 1850°F, 1950°F, and 2050°F, are shown in Figs. 4.24 (a) - (c). It is seen from these figures that when heating was carried out at 1010°C (1850°F), there was no obvious grain growth up to a 50 minute holding time for the three group starting grain sizes (ASTM 2, 4 and 6). This was due to the temperature being below the solvus temperature of γ which is around 1038°C (1900°F). However, for heating performed at 1066°C (1950°F) and 1121°C (2050°F), which is above the solvus temperature of γ, noticeable grain growth occurred in a short time and steady grain sizes are approached after 30 minutes of holding time. This steady state grain size is around ASTM 3.5 at 1066°C (1950°F) and around ASTM 1 at 1121°C (2050°F). Longer heating time up to four hours was also carried out. It showed that the steady state grain size kept the same. Thus the grain size for long time grain growth is defined to be a function of only temperature:

\[ d = 4.85 \times 10^{-63} T^{20.58} \quad (d: \mu m) \]  

(4.20)
Fig. 4.24 ASTM grain size vs. time at preheating temperatures of (a) 1010°C (1850°F), (b) 1066°C (1950°F), and (c) 1121°C (2050°F).
Fig. 4.24 Continued.
4.4 Summary of Equations Developed

The empirical equations developed in this chapter are summarized in Table 4.1. It is seen from these equations that the major factors in the control of the grain size in the forging of Waspaloy are (1) strain; (2) temperature compensated strain rate \( Z \), and (3) as heated grain size, \( d_0 \).

Strains create localized high densities of dislocations. To reduce their energy dislocations rearrange into subgrains. When the subgrains reach certain size the nuclei of new grains form. The higher the strain, the more dense the nuclei and more the cycles of recrystallization. This results in a finer recrystallized grain size.

The reason for deformation under high \( Z \) condition giving finer grain size is that an increase in \( Z \) results in the increase in the subgrain density which gives a higher density of nuclei. There are also more cycles of recrystallization present under high values of \( Z \). Thus, the size of recrystallized grain decreases (McQueen and Jonas, 1975).

At a given strain the reason for as heated grain size playing an important role in fraction of recrystallization and recrystallized grain size is that when polycrystalline metal is deformed, the grain boundary interrupt the slip processes. Thus, the lattice adjacent to grain boundary distorts more than the center of the grain. The smaller the as heated grain, the larger the grain boundary area and the volume of distorted metal. As a consequence, the number of possible sites of nucleation increases. The rate of nucleation increases. The size of the recrystallized grain decreases. Moreover, the
uniformity of distortion increases with the decrease in as heated grain size. Therefore, having a fine as heated initial grain size is very important for obtaining a fine recrystallized final grain size.
Table 4.1 Model for Microstructure Development in Forging of Waspaloy

<table>
<thead>
<tr>
<th>Dynamic Recrystallization</th>
</tr>
</thead>
<tbody>
<tr>
<td>$Z = \dot{\varepsilon} \exp(Q/RT)$</td>
</tr>
<tr>
<td>$\bar{\varepsilon}_p = 5.375 \times 10^{-4} d_0^{0.54} Z^{0.106}$ (T &lt; 1850°F)</td>
</tr>
<tr>
<td>$\bar{\varepsilon}_p = 1.685 \times 10^{-4} d_0^{0.54} Z^{0.106}$ (T &gt; 1850°F)</td>
</tr>
<tr>
<td>$\bar{\varepsilon}_c = 0.8 \bar{\varepsilon}_p$</td>
</tr>
</tbody>
</table>

**sub-solvus forging**

| $\bar{\varepsilon}_{0.5} = 0.1449 d_0^{0.32} Z^{0.03}$ (T < 1850°F) |
| $X_{d_{\text{dyn}}} = 1 - \exp[-\ln2(\bar{\varepsilon}/\bar{\varepsilon}_{0.5})^{3.0}]$ |
| $\bar{\varepsilon}_{0.5} = 0.056 d_0^{0.32} Z^{0.03}$ (1850°F < T < 1880°F) |
| $X_{d_{\text{dyn}}} = 1 - \exp[-\ln2(\bar{\varepsilon}/\bar{\varepsilon}_{0.5})^{2.0}]$ |
| $d_{d_{\text{dyn}}} = 8103 Z^{-0.16}$ (μm) |

**super-solvus forging**

| $\bar{\varepsilon}_{0.5} = 0.035 d_0^{0.29} Z^{0.04}$ (T > 1880°F) |
| $X_{d_{\text{dyn}}} = 1 - \exp[-\ln2(\bar{\varepsilon}/\bar{\varepsilon}_{0.5})^{1.8}]$ |
| $d_{d_{\text{dyn}}} = 108.85 Z^{-0.0456}$ (μm) |

<table>
<thead>
<tr>
<th>Metadynamic Recrystallization (T &gt; 1880°F)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$t_{0.5-\text{meta-dyn}} = 4.54 \times 10^{-5} \bar{\varepsilon}^{-1.2815} d_0^{0.5062} \dot{\varepsilon}^{-0.0729} \exp(9703/T)$</td>
</tr>
<tr>
<td>$X_{\text{meta-dyn}} = 1 - \exp[\ln2(t/t_{0.5})^{1.0}]$</td>
</tr>
<tr>
<td>$d_{\text{meta-dyn}} = 14.56 \bar{\varepsilon}^{-0.44} d_0^{0.325} Z^{-0.0258}$ (μm)</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Grain Growth</th>
</tr>
</thead>
<tbody>
<tr>
<td>time &lt; 30 min.: $d^3 = d_0^3 + 2 \times 10^{26} \exp(-595000/RT)t$ (d: μm)</td>
</tr>
<tr>
<td>time &gt; 30 min.: $d = 4.85 \times 10^{-63} T^{20.58}$ (d: μm)</td>
</tr>
</tbody>
</table>
CHAPTER V

APPLICATION OF MODEL TO MICROSTRUCTURE PREDICTION IN ISOTHERMAL AND HAMMER FORGINGS

The microstructural model developed in Chapter IV was tested in the prediction of microstructure under real production conditions. Two types of forging considered were isothermal and hammer forgings. The forging temperatures chosen for each type of forging covered a wide range from subsolvus to supersolvus.

5.1. FEM Process Simulation

Before the model can be applied to the prediction of microstructure a complete thermomechanical history should be available. FEM simulation can provide the thermomechanical histories of various material points during deformation processing. The method used in the simulation of isothermal forging and hammer forging is discussed below.

5.1.1 Simulation of Isothermal Forging

To ensure a correct FEM output the whole thermomechanical history in the isothermal forging should be correctly reproduced in FEM simulation. The preheating temperature, the transfer and equalizing times, the corresponding environment temperature and die temperature, the post forging dwell and
cooling condition etc., all need to be considered. Besides, the correct material flow stress should be used. Without the correct flow stress the correct flow pattern cannot be reproduced. Thirdly, the correct interface heat transfer and friction coefficients need to be selected. In the selection of interface heat transfer coefficient the research results from Semiatin (1987) and Burte (1989) among others can be used as a reference. In the selection of interface friction coefficient, the comparison of the flow line obtained from the macro etched sample and that obtained from FEM prediction can be used to make the judgment.

In the simulation of isothermal forging process several iterations were carried out before a pair of interface friction coefficient was chosen. Fig. 5.1a shows a macro etched sample on which the flow lines are observable. Fig. 5.1b shows the flow line predicted by FEM simulation after selecting a pair of correct interface friction coefficient. It is seen that the flow pattern produced by FEM simulation agrees well with the flow pattern shown on the experimental sample. This method is a feasible way for the selection of the friction coefficient under practical forging conditions in general.

5.1.2 Simulation of Hammer Forging

The whole thermomechanical history is considered again in the simulation of hammer forging. This includes the transfer of the billet to the die, the dwell of the billet on the lower die, the blow and the dwell between blows, the post forging dwell and cooling. Heat transfer simulations are carried out for the transfer, dwell, and post forging cooling periods. Coupled deformation and heat transfer is carried out for each blow. A special treatment should be
Fig. 5.1 (a) Macro etched sample shows the flow line; (b) FEM predicted flow line for the forging case.
mentioned is the simulation of heat transfer during the dwell between the blows.

To ensure a correct temperature field predicted by FEM simulation the boundary marks caused by the precipitation of γ' are used to check the interface heat transfer coefficient and the temperature prediction. For example we have a couple forgings which were forged below the γ' solvus. Due to the deformation heating, the temperature of the center of the workpiece rises to above the solvus. From the etched sample, a boundary between the region of dissolved and undissolved γ' can be seen. This boundary should correspond to the temperature for lower boundary of the γ which is around 1880°F according to the composition of the material used here.

Using this method, it is found is that the traditional way of simulation of hammer forging over predicts the temperature of the workpiece in the region in contact with the upper die. This is mainly due to the under prediction of the contact time of the upper die and the workpiece in a hammer forging. It is traditionally assumed (Shen et al., 1993) that when the energy of the hammer is used up in a blow the upper die separates from the workpiece immediately. However, this is true only for perfectly rigid or rigid contact. This is not true when plastic deformation is involved as in the forging. Therefore, there should be extra contact time for the upper die and the workpiece in the dwell period. Though this contact time is short for one blow, the accumulation of the time is significant for a hammer forging which involved quite a number of blows.
For engineering application, the exact contact time is not significant as long as there is a combination of the contact time and heat transfer coefficient which can yield the correct prediction of the temperature for the forging. The approach used here is that the dwell time in a hammer forging is divided into two time periods. In the first time period which was short, the heat transfer between workpiece and both the upper and lower dies was performed. In the second time period which was long, the heat transfer between the workpiece and the lower die was performed. After a few iteration for the selection of the interface heat transfer coefficient, the temperature predictions give very good agreement to real forging when the boundary of γ' precipitation is compared to the predicted temperature for the boundary which is around 1880°F.

Fig. 5.2a shows the macrograph which reveals the boundary between the dissolved and undissolved γ' regions. Fig. 5.2b shows the FEM predicted temperature field at the end of the hammer forging for reference. However, the average temperature of the initial, the intermediate and the final stage is preferred for this comparison.

5.2 Method to Combine the Model with FEM Analysis

Equations developed in Chapter IV are put in a code which use FEM predicted thermomechanical histories as input to predict microstructure development in Waspaloy forging.

5.2.1 As Heated Grain Size

It is seen from Table 4.1 that the as heated grain size is an important variable in the calculation of the microstructural development. The as heated grain
Fig. 5.2 (a) Macrograph which reveals the boundary between the dissolved and undissolved \( \gamma' \) regions. (b) FEM predicted temperature field at the end of the hammer forging.
size is decided according to preheating temperature in the program. If the preheating temperature is below the solvus temperature the as received grain size is used as the as heated grain size. This is because there is little change in grain size in subsolvus heating as shown in Fig. 4.24a. If the preheating temperature is above the solvus temperature the as heated grain size is calculated using equation (4.20).

5.2.2 Element Temperature

The recrystallization and grain size is calculated element by element. To do so the temperatures (1) in preheating, (2) at the beginning of deformation which is after transfer and dwell, (3) at the end of deformation, (4) in post-forging dwell, and (5) in post-forging cooling must be available for each element. However, the available temperature in DEFORM is the nodal temperature. Thus, the element temperature is calculated based on the average of the temperature at the four nodal points. This element temperature is then used for the microstructure prediction.

5.2.3 Critical Strain For Dynamic Recrystallization

Whether an element is recrystallized or not is judged by its critical strain for dynamic recrystallization. The critical strain is equal to 80% of the peak strain as shown in Table 4.1. If the strain of that element is below its critical strain for dynamic recrystallization the as heated grain size is assigned to that element without letting that element go through the calculation of dynamic and metadynamic recrystallization. If the strain of that element is larger than its critical strain for dynamic recrystallization the calculation for dynamic or
metadynamic recrystallization is performed depending on the average temperature at the beginning, the intermediate and the end of the deformation. The equations for the calculation of the critical strain for dynamic recrystallization are (4.1), (4.2), and (4.12).

5.2.4 Dynamic Recrystallization

The fraction of dynamic recrystallization and dynamically recrystallized grain size is calculated for those elements which have temperatures below the solvus. This is because when the temperature is low the effect of metadynamic recrystallization and grain growth is negligible. However, the dynamic recrystallization is not calculated for those elements whose average temperature are above the solvus temperature. This is because metadynamic recrystallization and grain growth occur. Therefore, the calculation of dynamic recrystallization is redundant for those elements. The equations used for the calculation of dynamic recrystallization for subsolvus elements are (4.5), (4.6), (4.8), (4.9), and (4.13).

5.2.5 Metadynamic Recrystallization and Grain Growth in Post-Forging Dwell

As mentioned above, for the elements whose average temperature are above the solvus temperature metadynamic recrystallization and the grain growth following it are calculated. Equations (4.16) to (4.18) are used for the calculation of the fraction of metadynamic recrystallization. Equation (4.19) is used for the calculation of grain growth right after the metadynamic recrystallization is finished. It is assumed that the metadynamic
recrystallization is finished when the fraction of metadynamic recrystallization is more than 0.95.

The temperature and time in the post-forging dwell are important for the metadynamic recrystallization and the following grain growth. The temperature used for the calculation of time for 50% metadynamic recrystallization in Equation (4.16) is the temperature at the end of forging for both isothermal and hammer forgings. The post-forging dwell time for the isothermal forging is the direct use of the time involved in the post-forging dwell. The post-forging dwell time used for the hammer forging is selected as two times the average dwell time for a blow. The reason for not using the total accumulated dwell time as the time for post-forging dwell is explained below.

In a hammer forging, multiple blows are used with a dwell time of a couple seconds between the blows. The metadynamic recrystallization and grain growth (if metadynamic recrystallization can finish during the dwell) is interrupted by the next blow. There is never time interval as long as the total accumulate dwell time available for metadynamic recrystallization and grain growth in a hammer forging. A good assumption for the dwell time for metadynamic recrystallization and grain growth is between the dwell time for one blow and the dwell time for a couple blows. There is not much difference in the microstructure prediction among them. Thus, in the prediction presented in the next section the dwell time of two blows are used as the post forging dwell time for all the hammer forging cases.
5.2.6 Treatment for Post-Forging Cooling

In the post-forging cooling period the temperature of the workpiece drops from the as forged temperature to room temperature at a very high cooling rate. However, if there is still sufficient time for which the workpiece stays above the solvus temperature, which is between 1850°F to 1950°F, the metadynamic recrystallization and grain growth can be in progress. This is specially true for the forgings with larger dimensions. Thus, the metadynamic recrystallization and grain growth in post-forging cooling should be also considered in the model. A piece-wise constant temperature scheme is used to handle the continuous cooling condition in post-forging.

Figs. 5.3 (a) & (b) show selected material points and their temperature-time curves in post forging cooling in a pan cake isothermal forging. Fig. 5.4 shows the piece-wise constant temperature scheme used. Still, the exact time for each temperature region needs to be calculated. Another assumption is used to facilitate the calculation which treats the actual cooling curve as a straight line. Thus, from the as forged temperature, the temperature at the end of cooling and the total cooling time any time corresponding to a selected temperature can be easily found out. Treating the cooling curve as a straight line is not a bad assumption when looking at the cooling curve shown in Fig. 5.3 (b).

5.2.7 Percentage of Recrystallization and the Grain Size

The percentage of recrystallization and grain size are calculated according to the thermomechanical history of the element. If the element is subject to a
Fig. 5.3 (a) Selected material points and (b) their temperature-time curves in post forging cooling in a pan cake forging.
Fig. 5.4 A piece-wise constant temperature scheme which is used to handle the continual cooling condition in post-forging cooling.
subsolvus forging condition, the percentage of dynamic recrystallization and its grain size is calculated once for all without considering the post-forging dwell and post-forging cooling conditions. Whether a recrystallized grain size or as heated grain size is assigned to the element depends on whether the percentage of dynamic recrystallization exceeds 50% or not. If it exceeds 50% recrystallization the recrystallized grain size is assigned. If not, the as heated grain size is assigned.

If the element is subject to a supersolvus forging condition, the percentage of metadynamic recrystallization and corresponding grain size is calculated considering both post-forging dwell and post-forging cooling. The percentage of metadynamic recrystallization is calculated accumulatively for the dwell and the cooling if it does not exceed 95%. When 95% metadynamic recrystallization is reached the calculation for grain growth starts with the metadynamically recrystallized grain size as the initial grain size. The grain growth is continuously calculated until the temperature drops to blow the solvus of \( \gamma' \). Again, a new grain size is assigned to the element if the percentage of metadynamic recrystallization exceeds 50% and the as heated grain size is assigned to the element if it does not.

5.3 Results

The experimentally obtained grain size and percentage of recrystallization is presented in this section according to isothermal forging and hammer forging. The exact process conditions are taken out to preserve the proprietary information of the supporting company.
5.3.1 **Isothermal Forging**

Figs. 5.5 to 5.10 show grain size or fraction of recrystallization for six different isothermal pan cake forging cases obtained from (1) the experiment and (2) the prediction from the program which implemented the model developed in Chapter IV. The results indicate that the prediction agrees very well with the experiments. Considering the different temperatures and different reduction were used for these pancakes, it can conclude that the model is proved applicable for isothermal forging.

5.3.2 **Hammer Forging**

Figs. 5.11 to 5.19 show grain size or fraction of recrystallization for eight different pan cake and one closed-die hammer forging cases obtained from (a) the experiment and (b) the prediction from a program which implemented the model developed in Chapter IV. The results also indicate that the prediction agrees very well with the experiments. There are a couple cases, for example P13, the prediction under-estimate the recrystallization in the region in contact with the upper and lower die. Thus, it gives larger region of grain size of ASTM 1 than that shown from the experiment. P14 is another example which over-estimate the grain size at the center of the workpiece. These are individual problems which should be handled after possessing more data from production, which belongs to the category of the fine turning of the program for industrial application. Fig. 5.19 (a) - (c) compared the grain size and fraction of recrystallization obtained from the experiment and model prediction for a most complex case, closed-die hammer forging. Though there are also some local regions show discrepancy the prediction does grasp the
structure of the most region. Thus, it can also concludes that the model is also applicable for hammer forging.
Fig. 5.5 (a) Experimentally obtained grain size for isothermal forging case P1.
P1 - Prediction

Fig. 5.5 (b) Predicted grain size for isothermal forging case P1.
Fig. 5.6 (a) Experimentally obtained fraction of recrystallization for isothermal forging case P2.
Fig. 5.6 (b) Predicted fraction of recrystallization for isothermal forging case P2.
Fig. 5.7 (a) Experimentally obtained grain size for isothermal forging case P3.
P3 - Prediction

Fig. 5.7 (b) Predicted grain size for isothermal forging case P3.
Fig. 5.8 (a) Experimentally obtained grain size for isothermal forging case P5.
P5 - Prediction

ASTM Grain Size
Object # 1
A = 5.4000
B = 5.8000
C = 5.9000
D = 5.7000
E = 5.8000
F = 5.9000
G = 6.0000
H = 6.1000
I = 6.2000
J = 6.3000
K = 6.4000
L = 6.4762
\Delta = 4.3171

Fig. 5.8 (b) Predicted grain size for isothermal forging case P5.
Fig. 5.9 (a) Experimentally obtained grain size for isothermal forging case P6.
Fig. 5.9 (b) Predicted grain size for isothermal forging case P6.
Fig. 5.10 (a) Experimentally obtained grain size for isothermal forging case P8.
Fig. 5.10 (b) Predicted grain size for isothermal forging case P8.
Fig. 5.11 (a) Experimentally obtained grain size for hammer forging case P9.
P9 - Prediction

Fig. 5.11 (b) Predicted grain size for hammer forging case P9.
Fig. 5.12 (a) Experimentally obtained grain size for hammer forging case P10.
Fig. 5.12 (b) Predicted grain size for hammer forging case P10.
Fig. 5.13 (a) Experimentally obtained grain size for hammer forging case P11.
P11 - Prediction

Fig. 5.13 (b) Predicted grain size for hammer forging case P11.
P12 - Experiment

Fig. 5.14 (a) Experimentally obtained grain size for hammer forging case P12.
Fig. 5.14 (b) Predicted grain size for hammer forging case P12.
P4 - Experiment

Fig. 5.15 (a) Experimentally obtained grain size for hammer forging case P4.
P4 - Prediction

Fig. 5.15 (b) Predicted grain size for hammer forging case P4.
Fig. 5.16 (a) Experimentally obtained grain size for hammer forging case P13.
Fig. 5.16 (b) Predicted grain size for hammer forging case P13.
Fig. 5.17 (a) Experimentally obtained grain size for hammer forging case P14.
Fig. 5.17 (b) Predicted grain size for hammer forging case P14.
Fig. 5.18 (a) Experimentally obtained grain size for hammer forging case P7.
Fig. 5.18 (b) Predicted grain size for hammer forging case P7.
Fig. 5.19 (a) Experimentally obtained grain size and fraction of recrystallization for a hammer closed-die forging case.
Closed Die Forging - Prediction

Fig. 5.19 (b) Predicted grain size for the hammer closed-die forging case.
Closed Die Forging - Prediction

Fig. 5.19 (c) Predicted fraction of recrystallization for the hammer closed-die forging case.
CHAPTER VI

CONCLUSIONS AND FUTURE WORK

This research developed a model for the prediction of microstructural development in the forging of Waspaloy turbine engine disks. Three sets of experiments were conducted in the development of the model.

(1) The preheating tests with different temperature and holding time were carried out to model the as-heated grain size (or long time grain growth) before a forging process.

(2) The compression tests in a Gleeble testing machine with different deformation and cooling conditions were performed to model the dynamic recrystallization in the forging, and the metadynamic recrystallization and the immediate grain growth in the post-forging dwell and cooling.

(3) Pancake and closed die forgings were conducted in an aerospace forging company for the purpose of verification and refinement of the model in a manufacturing environment.

6.1 Conclusions

The conclusions drawn from the current research are:
• The model developed is successful and it predicts very well the microstructure obtained from a wide temperature, strain, and strain rate ranges in isothermal and hammer forgings.

• The Zener-Hollomon parameter $Z$ controls the dynamically recrystallized grain size.

• The as heated grain size, the strain, and the Zener-Hollomon parameter $Z$ are major factors in the control of the metadynamically recrystallized grain size.

• Preheating temperature determines the as heated grain size before forging. At 1850°F (1010°C), there was no obvious grain growth up to a 50 minute holding time for the three groups of starting grain sizes: ASTM 2, 4 and 6. This was due to the temperature being below the solvus temperature of $\gamma'$ (around 1038°C or 1900°F). At 1066°C (1950°F) and 1121°C (2050°F), which was above the solvus temperature of $\gamma'$, noticeable grain growth occurred in a short time and a steady state is approached after 30 minutes of holding time. The steady state grain size is around ASTM 3.5 at 1066°C (1950°F) and around ASTM 1 at 1121°C (2050°F).

• Grain growth curves show that the grain size coarsening occurs much faster in the post-forging holding period in the compression conducted at 2050°F, than in the compression conducted at 1950°F.

• Use of the FEM simulations to process microstructural results obtained from compression tests made it possible to relate process data to microstructural data obtained from the experiments.
6.2 Future Work

- Apply the model to more industrial cases and fine tune the model.
- Reduce the tests in the development of the model to a minimum level.
- Apply the methodology to more materials.
REFERENCES


Battelle, unpublished flow stress data


