Control of Grain Size Via Forging Strain Rate Limits for R'88DT

Eric Huron
General Electric Engine Services

Shesh Srivatsa

General Electric Aircraft Engines

Ed Raymond
General Electric Aircraft Engines

Abstract

R'88DT is a powder metallurgy nickel-based superalloy used for rotating compressor and turbine disks. It was developed using a supersolvus heat treatment to achieve good creep and fatigue crack growth resistance. It was observed during development that the supersolvus heat treatment could result in large grains occuring in bands in the microstructure. Forging window studies were carried out using a specimen designed to produce a range of strain rates and strains for a given nominal specimen upset. These specimens were deformed over a range of nominal strain rates and the regions of large grains were correlated with strain rate. The presence of large grains was found to be related to a critical strain rate range depending on temperature. The large grains were termed Critical Grain Growth (CGG) because of this relationship. Deformation finite element analysis with commercial forging codes was used to establish a model to predict when CGG would occur and to design safe forging designs and process routes. This model has allowed the successful production of R'88DT forgings.

Superalloys 2000
Edited by T.M. Pollock, R.D. Kissinger, R.R. Bowman,
K.A. Green, M. McLean, S. Olson, and J.J. Schirra
TMS (The Minerals, Metals & Materials Society), 2000

Introduction:

Rene'88DT is an advanced high strength Ni-base powder metallurgy alloy used for rotating disks in gas turbines (References 1 and 2). It was developed with a nominal composition of 13Co-16Cr-4Mo-4W-2.1Al-3.7Ti-0.7Nb-0.03C (Reference 2).

This alloy is processed to a relatively coarse grain size (typically ASTM 6-10, or nominally 10 to 40 microns) to achieve good creep strength and damage tolerance. The processing route involves several major processes. First, argon atomized powder is produced and screened to desired mesh size (-270 mesh for R'88DT). The powder is extruded into billet, typically 20-30 cm in diameter. Cylindrical sections ("mults") are cut from the billet at target weights. The mults are isothermally forged in specially designed, controlled-environment forging presses. To achieve a consistent grain size, the forging condition must be carefully controlled. This paper describes the definition of the forging window using experiments designed to produce a controlled strain rate distribution in test coupons. The test coupons were forged, heat treated, and evaluated for grain size. Finite Element Modeling analysis was used to map grain size response vs. local strain rate to develop a model for the proper strain rate conditions. This model allows reliable forging of R'88DT hardware. Some of the other factors affecting the model are discussed in light of requirements for technology in forging of these alloys.

Procedure:

Test coupons were machined from extruded P/M R'88DT billet according to Figure 1. The development and composition of R'88DT are described in detail elsewhere (Reference 2). The coupons were tested on a servohydraulic testing machine equipped with a resistance furnace to control temperature and programmed to provide a nominal constant strain rate. Coupons were tested over a matrix of strain rates and temperatures from 954C to 1066C at strain rates from 0.0032/sec to 0.32/sec. Strain rates were programmed on a nominal value based on the overall specimen height to simplify the testing. Later, computer modeling was used to correlate the deformation conditions to actual local strain rate contours within the samples. After testing the coupons were heat treated with a supersolvus exposure of 1149C for 1 hour. The coupons were sectioned on radial planes and etched in Kalling's reagent to identify regions of grain size variation.

Some portions of the double cone matrix were replicated using standard right circular cylinder compression tests. The flow stress data from these specimens were used to calculate values of "m", the strain rate sensitivity factor, as a function of temperature and strain rate

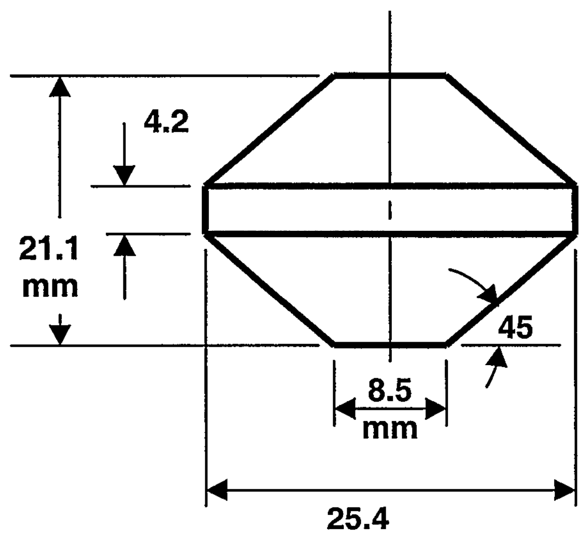

Figure 1: Double Cone Specimen Geometry. This specimen produces gradients of strain rate which simulate strain rate gradients caused by die radii and flow constriction in isothermal forging.

Results:

Most regions of the samples displayed the expected typical supersolvus microstructure of ASTM 6-10 grain size. However, regions of very large grains, ASTM 00, exceeding 500 microns in diameter, were observed. Figure 2 shows a typical uniform microstructure, and Figure 3 shows a region of the enlarged grains. The location of the enlarged grains was observed to change within the sample depending on the overall strain rate and temperature. Because it showed a definite relationship with a critical strain rate, this type of grain growth was termed Critical Grain Growth (CGG). For a given temperature, the location of CGG moved outward as nominal strain rate increased. Figure 4 shows the patterns of large grain regions observed in the samples. Generally the CGG occurred in bands within the samples. The grain size in the regions without CGG was generally ASTM 6-10. Within the CGG grains up to ASTM 00 were present. Maintaining the uniform ASTM 6-10 grain size is desirable to provide uniform mechanical properties. Although repetitive specimens did not always show CGG in exactly the same location, the same general patterns and locations of CGG were observed to occur. The emergent patterns observed over the repetitive sampling are shown schematically in Figure 5.

The main issue was then to transfer the results of the experiment into limits for production forging of R'88DT. This was accomplished by Finite Element Modeling (FEM) of the local strain rates in the coupons. Modeling was done using the commercial software package DEFORM. Flow stress data of R'88DT generated from 9" fine grain billet material was used. The models considered non-isothermal deformation with effects of adiabatic heating included. A shear friction factor of 0.2 corresponding to well lubricated specimens was used, and this value was independently verified by measurements in ring tests. The modeling results were verified by comparing measured and predicted load-displacement as well as the deformed specimen shapes.

The strain rates and the adiabatic heating experienced by all material points (finite elements) were tracked through the process and the maximum strain rate for each point was noted. The locations of CGG within the samples were matched visually to the modeled output to establish predictive maximum strain rate and adiabatic heating regions where CGG occurred. Using this approach, a strain rate window to predict CGG was established. If the maximum strain at a location occurs between

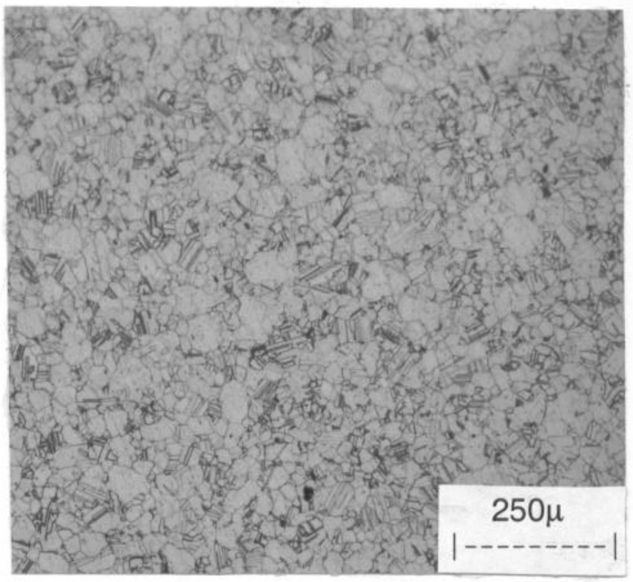

Figure 2: Typical Microstructure of R'88DT Away from CGG

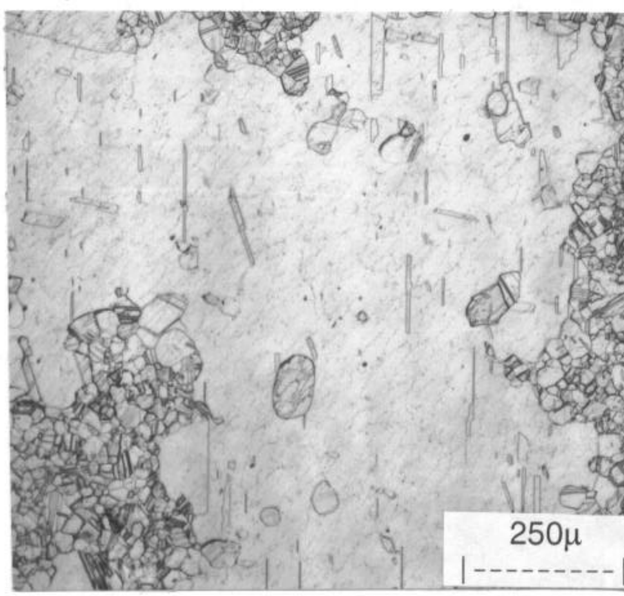

Figure 3: Typical Microstructure of R'88DT In Large Grain Band

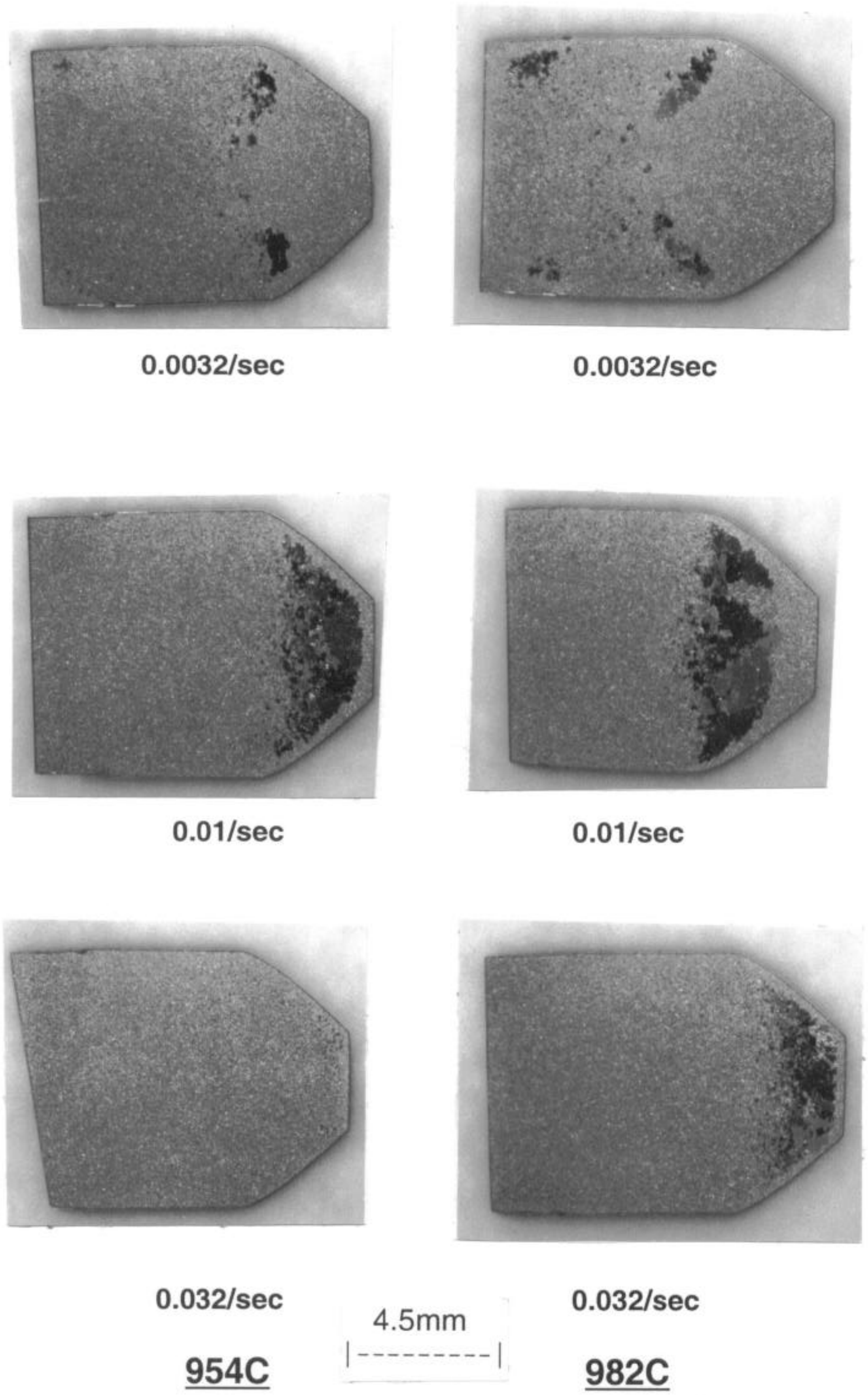

Figure 4 : Photographs of Critical Grain Growth (CGG) bands within deformed and supersolvus heat treated R'88DT specimens

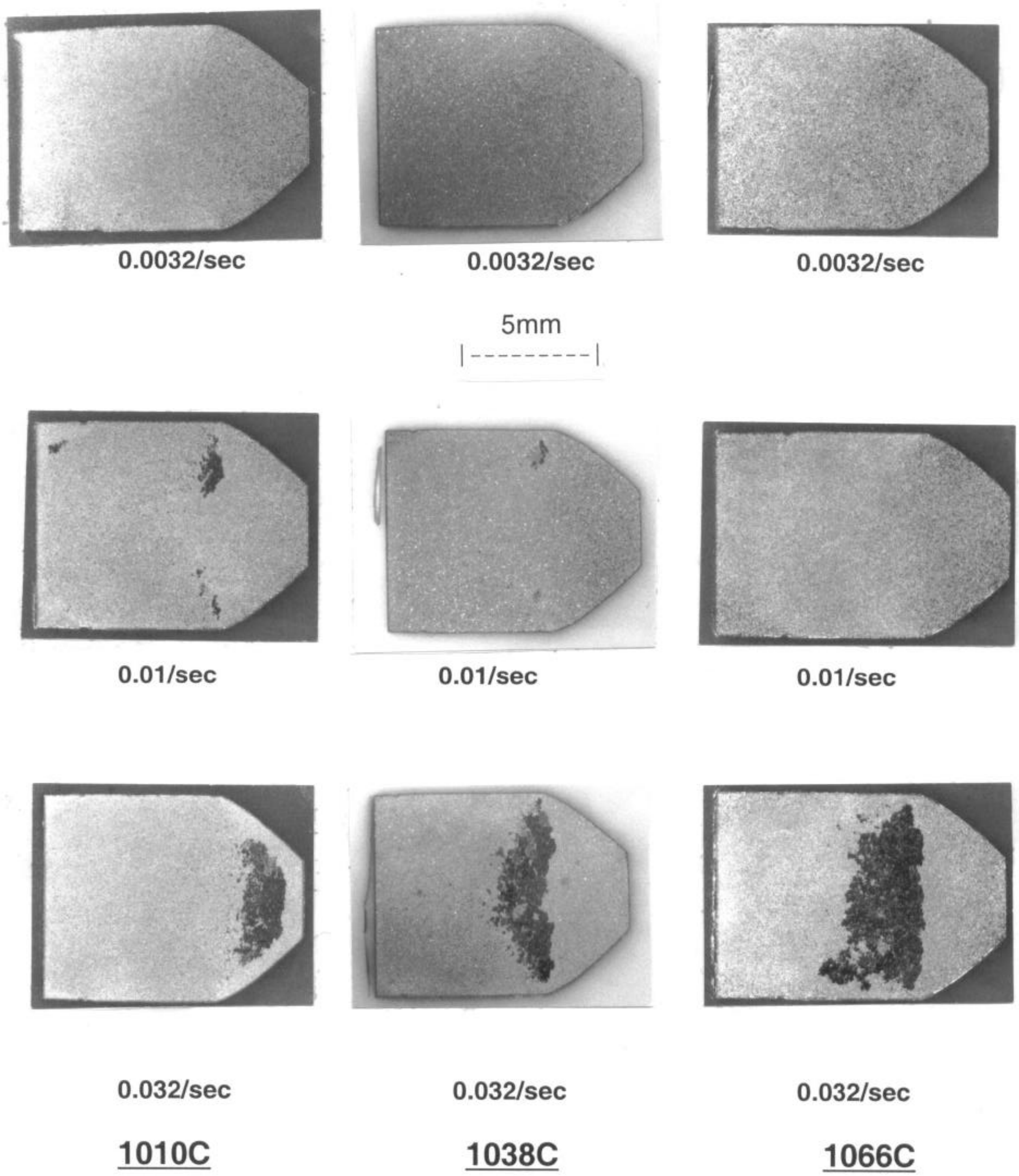

Figure 4 (continued): Photographs of Critical Grain Growth (CGG) bands within deformed and supersolvus heat treated R'88DT specimens

the limits of the CGG window, CGG is likely to occur. The strain rate window was found to be a strong function of temperature. At 954C, the maximum strain rate to avoid CGG was only 0.00244/sec, while at 1066C, a strain rate up to 0.0256/sec could be used. In addition, a strain rate above which CGG could be avoided was noted. For 954C, this strain rate was 0.01/sec, while at 1066C, it was much higher, 0.07/sec. The overall strain rate limits where CGG was observed are summarized in Table 1.

Table 1: Regions of Critical Grain Growth

Temperature (degrees C)	Strain Rates for Critical Grain Growth (/sec)
954	0.00244-0.01
982	0.00288-0.0128
1010	0.008-0.04
1038	0.010-0.05
1066	0.0256-0.07

The modeling results are summarized in Figure 6. The CGG regions are indicated by black color. The darker the black color, the closer is the max strain rate to the center of the CGG window. The predicted CGG locations agree well with the observed CGG (Figures 4 and 5) in all the specimens. This shows that the max strain rate is one of the variables controlling the occurrence of CGG. Further experimental tests and modeling are needed to establish the effects of gradients of strain, strain rate and temperature.

The strain rate regions where CGG was observed in the double cone specimens generally corresponded to regions of low "m" (strain rate sensitivity). Although the full analysis is beyond the scope of this paper, generally deformation under conditions of low "m" correspond to relatively low superplasticity. Instead of deforming superplastically (in theory, with little buildup of deformation debris and stored energy), the material deformed in this region is not highly superplastic and thus deformation debris and storage of energy could in fact occur.

Discussion:

The application of the testing to production forging has been fairly straightforward. The strain rate limits achieved by modeling analysis are used to guide forging design. Die design and press speeds are controlled so that no regions of the forging during isothermal forging are predicted to lie within the strain rate window. Exact details of the design are left to the forging supplier, but in general slower speeds or more generous die radii can be used to address any forging regions predicted to lie within the window. The model data have been provided to the commercial isothermal forging suppliers and have resulted in successful production of several thousand R'88DT isothermal forgings.

The model has been practically applied with great success even though the theoretical microstructural causes of CGG have not been fully determined. In extensive study at GE, the exact microstructural event leading to CGG has not been determined. If classical superplasticity is the cause, or rather, forging in regions of limited superplasticity, some microstructural evidence of stored energy should be expected. Superplasticity is commonly discussed in terms of the relative position along the classical stress vs. strain rate sensitivity curve, which is divided into three regions (References 3, 4). Region I, at low strain rates, is dominated by diffusion-accomodated grain boundary sliding with little stored energy. Region III, at high strain rates, is characterized by dislocation climb in addition to grain boundary sliding and extensive stored energy. Region II is an intermediate region with several possible contributing mechanisms. The present work for R'88DT appears to be associated with the upper end of Region II and the lower end of Region III. For IN100, as deformation conditions moved into this reason an increased dislocation density was noted and "m" decreased (Reference 5). This would correlate to increased tendency for CGG if it could be shown that the dislocation structure helps provide the driving force for CGG to occur. However, extensive Transmission Electron Microscope studies attempting to correlate measured dislocation densities in as-forged samples, with observed CGG in samples forged

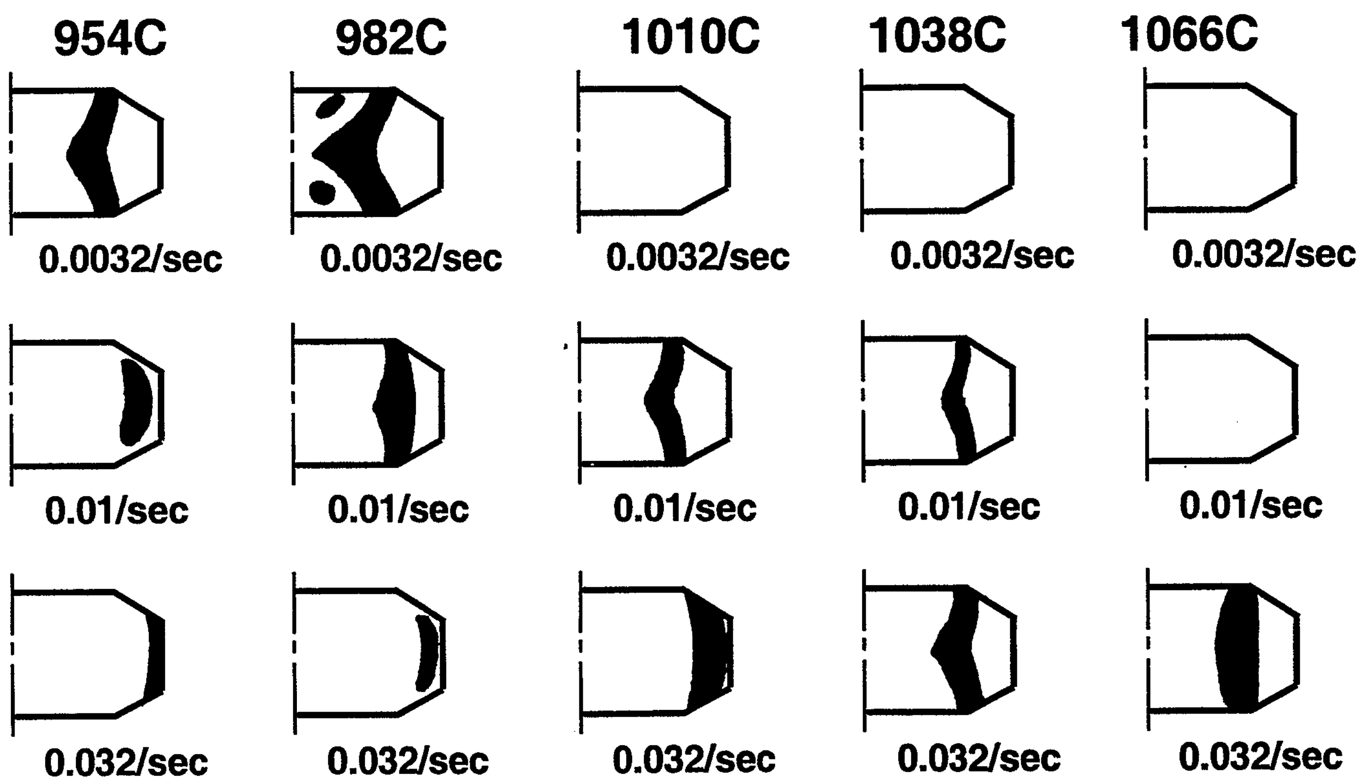

Figure 5: Schematic representation of Critical Grain Growth (CGG) patterns within deformed and supersolvus heat treated R'88DT specimens.

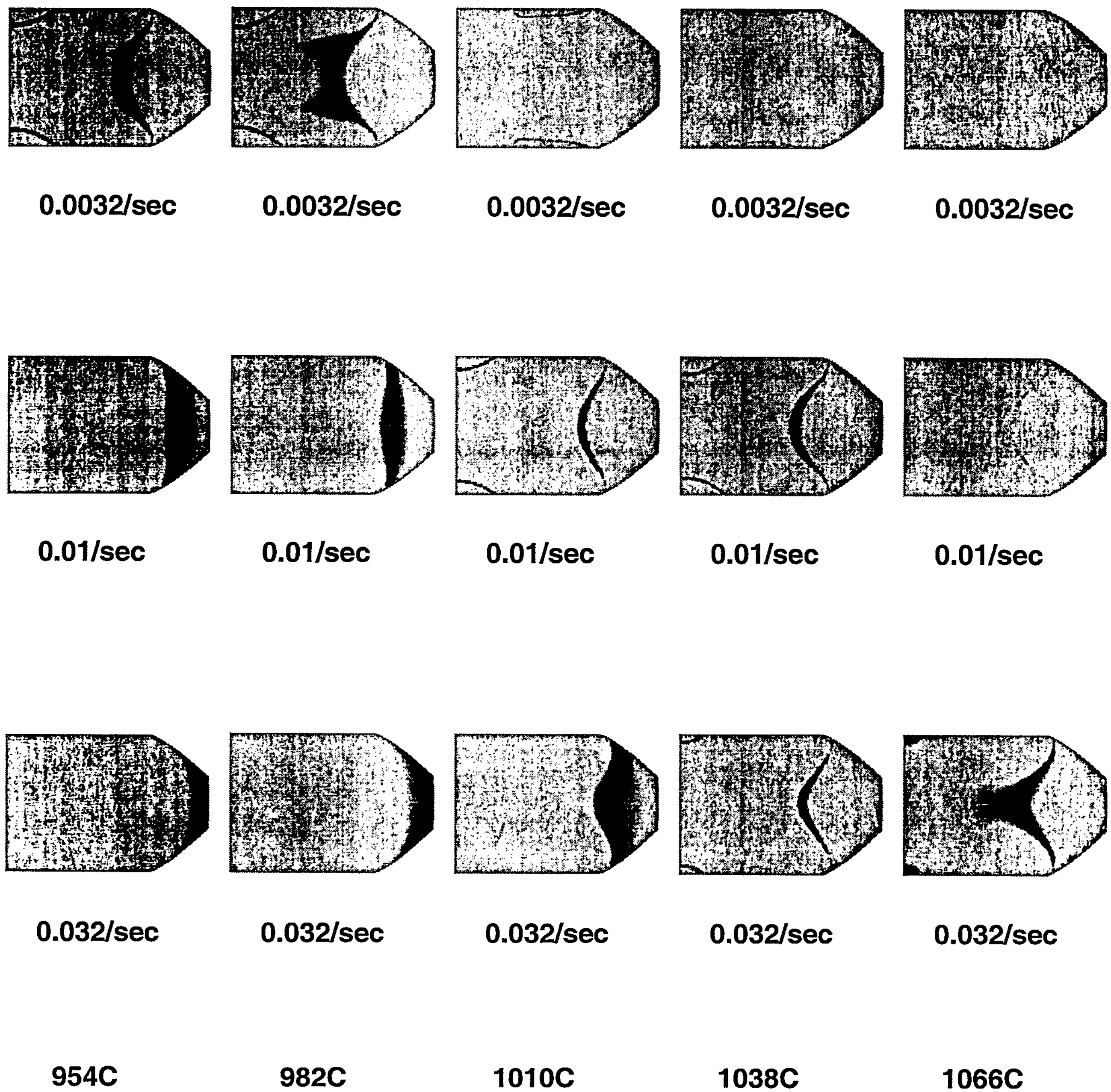

Figure 6 : Critical Grain Growth (CGG) Patterns predicted by modeling of the deformation process. These patterns agree well with the observed patterns shown in Figures 4 and 5.

at those conditions and then supersolvus heat treated, have failed to show a consistent correlation.

The results of CGG in R'88DT agree well with results reported for N18, another P/M Ni-base superalloy (Reference 6). This study also found that high temperature deformation, at a critical strain rate, produced large grains after annealing. The study also showed that the critical strain rate was a function of temperature. For N18, the critical strain rate was 0.0035/sec at 1100C, 0.008/sec at 1120C, and 0.01/sec at 1140C. Although the temperatures were higher, the trend of decreasing critical strain rate with decreasing temperature is consistent with the R'88DT results. The γ' content of N18 is higher than R'88DT (~55% vs. ~40%) and the γ' solvus is correspondingly higher (~1195C vs. ~1100C). The critical strain rate / temperature relationship may depend on the relative γ' amounts.

All of the results reported in this work are for a given billet microstructure and overall chemistry, and with controlled isothermal deformation. Most P/M billet is extruded to a fine grain superplastic structure. Billet microstructure influences the superplasticity and hence the energy storage and resultant grain size response. The results are beyond the scope of the present study, but billet microstructure must be maintained through consistent process controls. Recently new studies have explored alternatives to the most commonly used fine grained billet structure. Hot Isostatic Pressing (HIP) of P/M Udimet 720 was found to allow grain sizes up to ASTM 3, which could be hot worked to ASTM 10 (Reference 7). This work illustrates that minor phases as well as γ' control the grain size. Another study reports that hot die forging under non-isothermal conditions and at high strain rates promoted a different microstructural response (Reference 8). In this study forging was carried out to be deliberately above the critical strain rate, so that on subsequent annealing critical grain growth was intentional. Essentially a uniform high nucleation density promoted uniform grain size instead of a bi-modal distribution. The results did show that high strain rates, above the CGG window, produced uniform grain sizes within the samples. This suggests that forging throughput can be increased by using these higher strain rates. Unfortunately it is not possible to avoid local regions of sticking or low strain rate for areas of the forging in contact with the dies, so it is not practical to forge entirely above the CGG strain rate window.

Although much remains to be understood about the cause of CGG at the microstructural level, the present results and other studies suggest several practical approaches to forging design and practice. For uniform ASTM 6-10 microstructures, controlled isothermal forging, at strain rates below the critical strain rate, can be used to produce a uniform microstructure after supersolvus final heat treatment. Current commercial deformation models can easily be used to design forgings that have local strain rates within safe limits and within die loading and press capacity. Forging geometries and overall forging ram speed are modified iteratively until the models predict that all local regions are within the strain rate limits established by the coupon testing. This may require compromise on geometry radii or slower ram speeds, or both. This places a practical limit on throughput in isothermal forging, but by optimizing the models, good forging practice balancing speed, die design, and grain size control is achieved. Other grain sizes may be achievable by alternate processes, in particular the approach of forging entirely above the critical strain rate, but to date limitations on die capability have limited large-scale commercial use.

Acknowledgements:

All sample deformation testing for this study was done by Dr. Noshir Bhathena of Wyman-Gordon Forging Company, Houston, Texas. Metallographic specimen preparation and heat treatment were provided by Ron Tolbert and Bill Brausch of GE Aircraft Engines. Valuable technical consultation was provided by Drs. Micheal Henry and Charles P. Blankenship, Jr. of GE Corporate Research and Development.

References:

1. D. D. Krueger, R. D. Kissinger, R. G. Menzies, and C. S. Wukusick, U. S. Patent No. 4,957,567, General Electric Company.

2. D. D. Krueger, R. D. Kissinger, and R. G. Menzies, "Development and Introduction of a Damage Tolerant High Temperature Nickel Base Disk Alloy, Rene'88DT", in Superalloys 1992, ed. by S. D. Antolovich et. al., TMS-AIME, Warrendale, PA, 1992, 277-286

3. M. F. Ashby and R. A. Verrall, "Diffusion-Accomodated Flow and Superplasticity", Acta Met. 21, February 1973, pp. 149-173.

4. J. W. Edington, K. N. Melton, and C. P. Culter, "Superplasticity", Progress in Materials Science, 1976, 21, 1976, pp. 61-158.

5. R. G. Menzies, G. J. Davies, and J. W. Edington, "Microstructural Changes During Superplastic Deformation of Powder-Consolidated Nickel-Base Superalloy IN-100", Metal Science, 16, October 1982, pp. 483-494.

6. M. Soucail, M. Harty, and H. Octor, "The Effect of High Temperature Deformation on Grain Growth in a P/M Nickel-Base Superalloy", in Superalloys 1996, ed. by R. D. Kissinger et. al., TMS-AIME, Warrendale, PA, 1996, pp. 663-666.

7. G. Maurer, W. Castledine, F. A. Schweizer, and S. Mancuso, "Development of HIP Consolidated P/M Superalloys for Conventional Forging to Gas Turbine Engine Components, in Superalloys 1996, ed. by R. D. Kissinger et. al., TMS-AIME, Warrendale, PA, 1996, pp. 645-652.

8. C. P. Blankenship, Jr., M. F. Henry, J. M. Hyzak, R. B. Rohling, and E. L. Hall, "Hot Die Forging of P/M Ni-Base Superalloys", in Superalloys 1996, ed. by R. D. Kissinger et. al., TMS-AIME, Warrendale, PA, 1996, pp. 653-662.