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The effects of thermomechanical history on the microstructure of a nickel-base superalloy during forging

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ABSTRACT

The effect of thermo-mechanical history on hot compression behaviour and resulting microstructures of a nickel base superalloy is presented. Hot compression tests were carried out on HAYNES® 282® specimens to varying strains from 0.1 to 0.8. Both single pass and multi-pass tests were completed. 60 min inter-pass times were utilized to accurately replicate industrial forging practices. The effect of dynamic, metadynamic and static recrystallization during inter-pass times on flow stress was investigated. The resulting microstructures were analysed using scanning electron, optical microscopy and EBSD to relate grain size and homogeneity with flow stress data. The study showed a negligible difference between multi-pass and single pass tests for strain increments above 0.2. Therefore, when modelling similar low strain and strain rate forging processes in HAYNES® 282®, previous forging steps can be ignored.

1. Introduction

Casings are key structural components in a gas turbine engine. They are designed to contain the gas stream, provide housing for the various hot section components, and provide containment in the event of a mechanical failure. Nickel-based superalloys are able to withstand the high temperatures experienced in the combustor and turbine sections. Typically, casings are manufactured by multiple forging operations which include: upsetting, punching, piercing, becking, and ring rolling. Ring rolling is a low strain rate operation typically completed in multiple steps, requiring reheats in-between to bring the material back up to a workable temperature. By altering various forging and heat treatment parameters, the final microstructure and mechanical properties of the component can be tailored to specific requirements. To save materials, costs and time, FE models are typically used to predict microstructural evolution during a single forging processes [1]. However due to the multi-pass forging route employed during the manufacture of casings, it is unknown whether modelling the final stage of a forging route will yield accurate results, or whether the thermomechanical history of the material needs to be taken into account.

Attempts have been made to model the various ring rolling parameters and therefore predict the final microstructure and mechanical properties of a finished component. An early study [2] developed a coupled thermomechanical model incorporating heat transfer. Recently studies have utilized finite element software DEFORM 3D [3] to model the ring rolling process. Previous studies investigated the effect of processing parameters on recrystallization and final grain size, although these studies did not take into account previous thermomechanical history [4,5]. Previous work has investigated the effect of prior deformation steps and final annealing on microstructure evolution though interrupted hot compression studies. These studies looked into varying strain, and different materials such as magnesium [6], steel [7] and nickel [8]. These studies concluded that there was a difference between single-pass and multi-pass microstructures. Orientation has been shown to effect recrystallization rates in multi-hot experiments [9]. However these studies all utilized short inter-pass times, focusing on metadynamic rather than static recrystallization – therefore not representative of the industrial practice for a ring rolling route. Slightly longer inter-pass times were used in an investigation into magnesium, which showed flow hardening occurring during the inter-pass anneal [10]. Multi-pass flow stress behaviour in Aluminium has also been predicted in recent work [11]. Recently a study has shown that the final grain size of HAYNES® 282® is primarily dependent on annealing time and temperature, not forging strain and strain rate. However this study only considered a single-step process [12].

The aim of this experiment is to investigate the effect of thermomechanical history on final microstructure through a series of

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interrupted hot compression experiments. In particular, comparing the differences, if any, between single and multi-pass compression tests and whether this is dependent on strain. The results will determine whether thermomechanical history should be taken into account when replicating forging conditions in both a laboratory environment and within a model.

2. Materials and methods

The material used for this investigation was HAYNES® 282®an age-hardenable gamma-prime-strengthened nickel-based superalloy. The composition for Haynes 282 is shown in Table 1 [13]. The material was supplied in the form of a 100 mm diameter billet that was 80 mm high and approximately 5 kg in weight. 14 cylindrical specimens (8 mm diameter, 12 mm height) were sectioned from the billet using wire electric discharge machining (EDM).

The initial microstructure was typical of an annealed billet product with a grain size of approximately 75 μm, shown in Fig. 1.

Compression testing was completed on a uniaxial hydraulic compression rig with a split furnace capable of 1200 °C. Tests complied with ASTM E209-00(2010) and were performed under stroke control. Specimens underwent a 3 min soak time at 1100 °C prior to the start of the test to equalise temperatures within the specimen and furnace. Boron nitride spray was used as a lubricant between the specimen faces in contact with the loading platens. Post-test specimens were removed immediately and left to air cool; replicating industrial practice. Multistage tests underwent a 1 h dwell at 1100 °C between compression stages. In total 14 tests were conducted at 1100 °C at a strain rate of 0.2/s. Specimens were deformed to compressive strains of 0.6, 0.8 and 0.1. Additionally, multistage tests were deformed by compressive strains of 3 × 0.2 and 0.7+0.1. The test conditions aim to replicate the typical strains, strain rates, and temperatures experienced during a complete ring rolling process. The conditions were extracted from a FE DEFORM-3D V.11 [3] ring rolling model replicating observed typical industrial processing parameters for nickel-based superalloys. Seven specimens underwent a direct age heat treatment, which consisted of 2 h soak at 1010 °C then air cooled, followed by an 8 h soak at 788 °C then air cooled.

Specimens were mounted in conductive bakelite and polished to a 1 μm diamond lubricant finish before being etched with Kallings II [14] to reveal the grain boundaries. Images were captured using an optical light microscope and analysed using the software ImageJ [15]. DEFORM models were used to generate a strain map of the specimen cross section, to accurately predict local strains and strain rates; correlating microstructures to local deformation conditions. Energy-dispersive X-ray spectroscopy (EDX) was conducted using a Phillips XL-30 scanning electron microscope (SEM). Electron back scatter diffraction (EBSD) analysis was conducted on the same instrument operated at 20 kV integrated to a Nordlys EBSD detector using HKL Technology Channel 5 software.

3. Results

3.1. Dynamic recrystallization rate

Table 1.
Composition of Haynes 282 wt%.

Ni	Cr	Co	Mo	Ti	Al	Fe	Mn	Si	C	B
57[a]	20	10	8.5	2.1	1.5	1.5[b]	0.3	0.15	0.06	0.005

[a] Nickel as balance.
[b] Maximum.

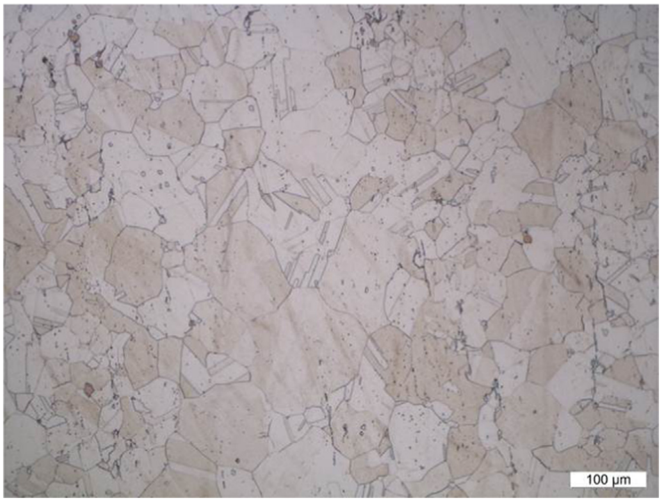

Fig. 1. The initial microstructure of the as received Haynes 282 billet showing a grain size of 75 μm.

In order to establish the amount of dynamic recrystallization (DRX) which occurred during each deformation pass, it was necessary to predict the relationship between fraction of dynamically recrystallized grains and strain. The flow stress data was analysed from a single pass compression test to 0.6 strain. Using a model proposed by Estrin and Mecking [16] and adapted by Chen et al. [17] it is possible to predict the volume fraction of DRX grains (Eq. (1))

$$X = \frac{\sigma_{DRV} - \sigma}{\sigma_{sat} - \sigma_{ss}} \tag{1}$$

where X=fraction recrystallized, σ_{DRV}=theoretical flow stress in the absence of DRX, σ=observed flow stress, σ_{sat}=saturation flow stress, σ_{ss}=steady state flow stress. The theoretical flow stress in the absence of DRX is driven purely by work hardening and dynamic recovery (DRV), and can be calculated using Eq. (2).

$$\sigma_{DRV} = \sqrt{\sigma_{sat}^2 - (\sigma_{sat}^2 - \sigma_{crit}^2)\exp(-\Omega\varepsilon)} \tag{2}$$

σ_{sat}, σ_{ss} and critical strain (σ_{crit}) can be calculated by plotting work hardening rate ($\theta = d\sigma/d\varepsilon$) vs flow stress, as shown in Fig. 2. The critical stress for DRX under these conditions is 178 MPa, which corresponds to a critical true strain of 0.09.

The coefficient of dynamic recovery Ω is obtained using the linear regression method by plotting Eq. (3) as shown in Fig. 3.

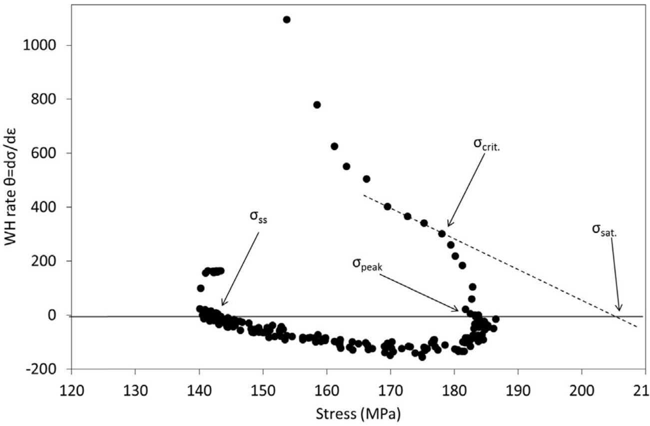

Fig. 2. Work hardening rate vs stress plot used to obtain key stresses.