Prediction of recrystallization in investment cast single-crystal superalloys

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Received 13 June 2012; received in revised form 30 August 2012; accepted 5 September 2012
Available online 24 October 2012

Abstract

Modelling and targeted experimentation are used to quantify the processing conditions which cause recrystallization in a single-crystal superalloy. The plasticity needed is traced to the differential thermal contractions of the metal and its ceramic mould during processing. For typical cooling rates, the plasticity causing recrystallization is induced above 1000 °C, thus over a temperature interval of approximately 300 °C after solidification is complete. The total accumulated plastic strain needed for recrystallization is estimated to be in the range of 2–3%. Modelling is used to rationalize the influence of mould thickness, stress concentration factor and geometry on the induced plasticity. Negligible plastic strains were predicted in a solid casting with no stress concentration features, as found experimentally. However, recrystallization occurred in thin-walled sections, particularly beneath shroud-like features due to the plasticity induced there. The model provides the foundation for a systems-based approach which enables recrystallization to be predicted and thus avoided in new designs of turbine blade aerofoil.

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Keywords: Mechanical deformation; Solidification; Investment casting; Superalloys

1. Introduction

Turbine blades for gas turbine applications are investment cast, often into single-crystal form. During processing, deformation is induced in the nickel-based superalloy during solidification and subsequent cooling; this is due to differential thermal contraction of the metal, mould and core, arising from their differing thermal expansion coefficients [1]. In practice, this effect has a number of ramifications which need to be recognized. First, account needs to be taken of the shrinkages which occur – the final casting will not exhibit the same dimensions as the wax model to which it is related. Second, and perhaps of greater practical importance, plastic strains are produced which can be large enough to induce recrystallization during subsequent heat treatment [2–5]. Particularly for components cast in single-crystal form, the occurrence of recrystallization cannot be tolerated – the associated high-angle grain boundaries degrade the creep [6,7] and fatigue [8–10] properties significantly. Work has been done to study the recrystallization behaviour of single-crystal superalloys under the influence of different annealing conditions [11–13] and microstructural features [14–16]. However, very little attention has been given to developing a systematic approach for design purposes.

This paper is concerned with the mathematical modelling of the investment casting of single-crystal superalloys, with particular emphasis on thermal–mechanical effects so that processing-induced plasticity can be predicted and rationalized. One aim is to elucidate the amount of plastic strain needed for recrystallization to occur, and to infer the
temperature range over which this is induced in the material. The overarching goal is to build a physics-based tool for the prediction of recrystallization during the processing of single-crystal castings. Traditionally, the avoidance of recrystallization has been dealt with in a rather empirical way, with reliance placed on existing casting practice, experience or else rules of thumb. Mathematical modelling allows the physical effects causing recrystallization to be anticipated, which is obviously advantageous. Such modelling might be used for the optimization of processing conditions, so that the likelihood of recrystallization can be reduced. If sufficiently robust, it might also be used during the early stages of design process to influence the geometry chosen for the turbine blade – as part of a systems-based approach to component design.

2. Methods

2.1. Modelling approach

A simplified geometry was designed to represent an analogue of a turbine blade aerofoil, of comparable size and containing flanges to simulate the mechanical constraints provided by platforms and shrouds. Fig. 1 illustrates the details of the test piece analysed, which has been christened the “bobbin” geometry. Fillet radii of 2 mm were introduced between the three platforms/shroud flanges of thickness 5 mm, consistent with a stress concentration factor of 1.80. The three gauge lengths of diameter 15 mm are identified according to the notation \( L_1, L_2 \) and \( L_3 \), with \( L_1 \) and \( L_3 \) being the first and last to solidify, respectively, during processing. In some cases – in order to introduce greater complexity – further stress concentration features of known factor \( K \), were introduced into the gauge lengths, and in other cases a ceramic core was used to facilitate the casting of hollow test pieces of wall thickness as little as 1.5 mm. The details of the various geometries considered are summarized in Table 1. The nominal thickness of the shell used was \( \sim 5 \) mm.

For the modelling, calculations were carried out using the finite element method and commercially available software. Thermal–elastic–plastic analysis was carried out, with the high-temperature plasticity assumed to be rate independent. This assumption is critiqued later in Section 3.6. Temperature-dependent material parameters for the CMSX-4 superalloy were assumed. Thus, the following thermal and mechanical properties of CMSX-4 were used: density, specific heat capacity, thermal conductivity, yield stress, ultimate yield strength, hardening exponent, thermal expansion coefficient, Young’s modulus and Poisson’s ratio, taken in the (001) solidification direction. The stresses and strains, especially the plastic strains, are of primary interest here; since radial symmetry was present in all cases, it proved sufficient to model a 30 degree section of the components for the mechanical part of the calculation. For the superalloy, an elastic–plastic formulation was employed, together with the von Mises criterion; elastic isotropy was assumed. In what follows, consideration of the effective (accumulated) plastic strain is made and this is denoted \( \varepsilon_{pl} \). For the definition of \( \varepsilon_{pl} \), the flow function for an isotropically hardening material is introduced in the form

\[
f\{\sigma, \kappa(\varepsilon_{pl})\} = \sigma(\varepsilon) - \sigma_f(\varepsilon_{pl})
\]

where the effective stress \( \sigma(\varepsilon) \) is defined as

\[
\sigma(\varepsilon) = \sqrt{3} \frac{\sigma : \varepsilon}{2}
\]

and the plastic flow stress due to isotropic hardening, \( \sigma_f(\varepsilon_{pl}) \), is defined by

\[
\sigma_f(\varepsilon_{pl}) = \sigma(\varepsilon) - \sigma_f(\varepsilon_{pl}) \exp(-H \varepsilon_{pl})
\]

where \( \sigma(\varepsilon) \), \( \sigma_f(\varepsilon_{pl}) \) and \( H \) are the ultimate yield stress, the yield strength and the hardening exponent, respectively. Therefore, from the isotropic hardening criterion, one obtains

\[
\sigma(\varepsilon) = \sigma_f(\varepsilon_{pl}).
\]

This gives rise to the definition of effective (accumulated) plastic strain as

\[
\varepsilon_{pl} = \int_0^{\varepsilon_{pl}} d\varepsilon_{pl} = \int_0^{\varepsilon_{pl}} \frac{2}{3} \frac{\sigma_f(\varepsilon_{pl})}{\sigma : \varepsilon_{pl}}
\]

Note that the definition of \( d\varepsilon_{pl} \) is such that stored energy is conserved:

\[
dW = \sigma d\varepsilon_{pl} = \sigma d\varepsilon_{pl}
\]

For the superalloy, model parameters were adopted from Refs. [17–20]. For the shell and core, representative values for alumina-and silica-based shell materials adapted from Refs. [21–23] were used. The variation in the flow stress with temperature for CMSX-4 is plotted in Fig. 2; the data are taken from Ref. [1].
2.2. Validation studies – investment casting

For the purposes of model validation, some of the geometries identified in Table 1 were fabricated from the CMSX-4 single-crystal superalloy. A semi-industrial-scale single-shot investment casting facility at the University of Birmingham was employed. Moulds were prepared using wax assembly, ceramic processing and steam-autoclave dewaxing methods, following standard procedures. The casting cycle followed is shown in Fig. 3. The molten metal was poured under vacuum conditions at 1500 °C into the mould, and withdrawn thereafter at 0.06 mm s\(^{-1}\); when all of the mould had passed the baffle, cooling was allowed to proceed in air. Some of the castings were instrumented using Rh/Pt Type B thermocouples, which have measurement capability up to 1750 °C, with use being made of alumina sheaths of 4 mm outside diameter, 2 mm inside diameter and 15 mm length. All castings received a standard solution heat treatment, consisting of a final solutioning step of 1315 °C/6 h [1]. An electrolytic etching process involving HCl/H\(_2\)O\(_2\) etch was used to reveal whether recrystallization had occurred.

### Table 1

Details of the bobbin castings which were modelled (those chosen for experimental casting are labelled by 1)

<table>
<thead>
<tr>
<th>No.</th>
<th>Label</th>
<th>Gauge diameter ((\phi)) (mm)</th>
<th>Notch position</th>
<th>(K_t)</th>
<th>Wall thickness ((W)) (mm)</th>
<th>Core diameter (mm) ((\phi_{core}))</th>
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<tr>
<td>1</td>
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<td>15</td>
<td>–</td>
<td>–</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>2</td>
<td>SB9</td>
<td>9</td>
<td>–</td>
<td>–</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>3</td>
<td>MSB15(^1)</td>
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<td>2.6</td>
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<td>2.0</td>
<td>2.6</td>
</tr>
<tr>
<td>5</td>
<td>CB3.0(^\circ)</td>
<td>15</td>
<td>–</td>
<td>–</td>
<td>–</td>
<td>3.0</td>
</tr>
<tr>
<td>6</td>
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<td>9</td>
<td>Middle</td>
<td>1.6</td>
<td>2.0</td>
<td>2.6</td>
</tr>
<tr>
<td>7</td>
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<td>–</td>
<td>–</td>
<td>–</td>
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<td>9</td>
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</tr>
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</tr>
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<td>2.6</td>
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<td>1.6</td>
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</tr>
</tbody>
</table>

Fig. 2. Yield strength of CMSX-4 material used in the simulation. Data are from Ref. [1].

2.2. Validation studies – investment casting

For the purposes of model validation, some of the geometries identified in Table 1 were fabricated from the CMSX-4 single-crystal superalloy. A semi-industrial-scale single-shot investment casting facility at the University of Birmingham was employed. Moulds were prepared using wax assembly, ceramic processing and steam-autoclave dewaxing methods, following standard procedures. The casting cycle followed is shown in Fig. 3. The molten metal was poured under vacuum conditions at 1500 °C into the mould, and withdrawn thereafter at 0.06 mm s\(^{-1}\); when all of the mould had passed the baffle, cooling was allowed to proceed in air. Some of the castings were instrumented using Rh/Pt Type B thermocouples, which have measurement capability up to 1750 °C, with use being made of alumina sheaths of 4 mm outside diameter, 2 mm inside diameter and 15 mm length. All castings received a standard solution heat treatment, consisting of a final solutioning step of 1315 °C/6 h [1]. An electrolytic etching process involving HCl/H\(_2\)O\(_2\) etch was used to reveal whether recrystallization had occurred.

2.3. Transmission electron microscopy

Transmission electron microscopy (TEM) was used to study the deformation induced during the investment casting process. CMSX-4 superalloy in the as-cast condition...
was used for this purpose. For sample preparation, discs of ~3 mm diameter and 0.2–0.3 mm thickness were subjected to twin-jet electropolishing with 10% perchloric acid solution in methanol; 20.5 V and −5 °C were the working conditions. A JEOL 200CX transmission electron microscope was used at an accelerating voltage of 200 kV for bright-field imaging of the deformed microstructure.

2.4. Mechanical testing – critical plastic strain for recrystallization

For later parts of the study, as-cast tensile test pieces of CMSX-4 of diameter 5.85 mm and gauge length 29 mm were cast. To ensure that the surface finish of the test pieces was representative of that arising from the investment casting furnace, tensile test pieces were cast in such a way that the gauge length remained in the as-cast condition; only the shoulders and grips of the test pieces were subjected to electrical discharge machining (see Fig. 4). Care was taken to ensure that only test pieces cast within 20 ° of ⟨001⟩ were employed for the tensile testing. To evaluate the critical straining conditions necessary for recrystallization, mechanical testing was subsequently carried out in tension at different temperatures at a strain rate of 0.2% min⁻¹, to induce various levels of plastic strain into the material. The yield stress data were extracted from the mechanical testing results for modelling purposes. Typical stress–strain curves are given in Fig. 5 and the variation in the flow stress with temperature is given in Fig. 5. Subsequently, samples pulled in tension to various strain levels were subjected to the standard solution heat treatment for CMSX-4. The samples were then sectioned longitudinally along ⟨001⟩ for examination. The electrolytic etching process was used as before, to reveal any recrystallization.
3. Results

In this section our results are presented as follows. First, some basic considerations relevant to the problem are discussed; results from a basic one-dimensional model are given. Next, results from (i) the thermal and (ii) thermal–mechanical finite element modelling are presented. There follows a sensitivity analysis, which explores the effects of the important processing parameters. Finally, the predictions are compared with the results from our experimental studies.

3.1. Preliminary considerations

Consider the situation of a solid cylinder of a single-crystal superalloy held within a mould assumed to be totally rigid; this simple test case can be modelled in one dimension using an elasto-plastic material law under the assumption of isotropic elasticity. The cooling rate is assumed to be uniform at 0.06 mm s\(^{-1}\), consistent with the experimental work. The plastic strain \(\varepsilon_{pl}[T]\) at each temperature is determined using

\[
\varepsilon_{pl}(T) = \int_{T_{solidus}}^{T} \frac{\partial \varepsilon_{pl}}{\partial t} \frac{\partial T}{\partial t} dT
\]  

(6)

The results (see Fig. 6) indicate that about 0.1% plastic strain is induced at temperatures in excess of 1200 °C (see the zone A marked in the figure). On further cooling, further thermal strain can be accommodated elastically in zone B down to a temperature of approximately 600 °C.
Further plastic strain is induced only when the temperature falls below 600 °C, building up substantially thereafter (zone C). These calculations suggest the possibility of the plasticity causing recrystallization being introduced at temperatures in excess of 1000 °C, soon after solidification is completed.

3.2. Thermal model

Thermocouple measurements were made at the locations $L_1$ and $L_2$ during the casting of the bobbin test piece SB15 of Table 1. In Fig. 7, the thermocouple readings are compared with the predictions from the model. Consistent with the location of the thermocouples, the readings from $L_2$ are higher than those from $L_1$ at any given time; a maximum difference of up to 150 °C was measured. The model predictions are in good agreement with the experimental results between approximately 1515 and 1000 °C.

The important temperature range for this study is between 1300 and 1000 °C, since strain accumulation occurs only below the solidus temperature of the alloy, which is about 1330 °C; this is consistent with the predictions in Fig. 6. Below 1000 °C, the model underestimates the cooling rates somewhat, due to the rather complicated
heat transfer effects, which are not accounted for perfectly. However, as will be confirmed later, only temperatures above 1000 °C are significant for recrystallization, so the results in Fig. 7 demonstrate that the thermal characteristics of the furnace used are captured sufficiently well for the present purposes.
3.3 Thermal–mechanical model

Thermal–mechanical modelling was carried out to quantify the effects of geometry, stress concentration features, shell thickness, shell type, core diameter and core type. In summarizing the results, the notation of Table 1 is used. Preliminary work was carried out first to confirm the necessary scale of mesh discretization needed for the finite element analysis. The geometry SB15 was used for this purpose. Fig. 8 illustrates the variation in the computed plastic strain for this geometry at three distinct locations. No significant benefit is gained from using a mesh size less than 1 mm; this was used for the work reported in this rest of the paper.

Fig. 9 illustrates the influences of shell thickness and stress concentration factor, which have been found for the solid test pieces MSB-1 to TSB-3 of Table 1. One identifies first that the shell thickness has an important influence, particularly at larger stress concentration factors. The effect is greatest when the stress concentration factor $K_t$ is as high as 2.6. The plastic strains produced by $K_t$ values of 1.6 and 2.0 were found to be approximately the same and not strongly influenced by shell thickness.

The plastic strains predicted by the model were found to be consistently higher for the $L_3$ location than for $L_1$, and in general were in the order $L_3 > L_2 > L_1$. This can be rationalized in the following way. For illustrative purposes, consider results for the CB1.5 casting (see Fig. 10). The lower part of the casting solidifies first (Fig. 10a) and therefore, during its cooling towards 1000°C, the material above it (which has yet to solidify) remains in the liquid state, so that little stress builds up. Thus the stress at location $L_2$ does not rise markedly until location $L_1$ solidifies. Location $L_3$ passes through the critical temperature range towards 1000°C with the material below it already solidified, so that greater stresses can be set up on it. Also important are the rates of cooling, which then relate directly to the rate of plastic straining. The cooling rates are higher at location $L_1$ than at $L_3$, due to its greater proximity to the water-cooled chill plate on which the casting sits. The definition of plastic strain in Eq. (6) then helps to rationalize the greater plastic strain at lower cooling rates, consistent with $L_3$ – for which

![Fig. 10](image_url)
the cooling rate is lowest – experiencing the highest plastic strain. Note the greater rate of increase of the plastic strain with time for location \( L_3 \) (see Fig. 10c). The maximum plastic strain predicted is also affected by the location of the notch, e.g. whether it is placed at the very middle of the gauge length or alternatively directly under the flange (see Fig. 1). Higher plastic strains were found in the latter cases, due to the superposition of stresses arising from the fillet radii (under the flanges) with those due to the machined notches introduced for their stress concentration effect.

In addition to the above, a sensitivity analysis for plastic strain accumulation was carried out to explore the effects of shell material. Two distinct shell systems were studied: an alumina-based one (referred to here as Shell1) and a silica-based one (Shell2). Thermal expansion coefficients and Young’s moduli were assumed to be linear functions of temperature, and thus, given the materials used, Shell1 is appreciably stiffer than Shell2; consequently, more plastic strain might be anticipated within the metal cast in Shell1 than in Shell2. Calculations were carried out for the SB15 and MSB15 geometries (see Fig. 11). The predicted plastic strain induced in the unnotched solid test piece SB15 is small and not appreciably influenced by shell type; moreover, the plastic strain was distributed uniformly along the casting. For the notched test piece MSB15, the strain was concentrated strongly around the notch, and for Shell2 in particular it showed again the tendency \( L_3 > L_2 > L_1 \). For Shell1 the plastic strain did not vary strongly from gauge length to gauge length. Slower withdrawal rates caused slower cooling rates, hence lower strain rates for deformation and thus lower plastic strains. Fig. 12 illustrates the quantitative data associated with this cooling
rate effect. More deformation was found around the notches of MSB15 than under the flange features. These findings emphasize the importance of shell properties.

Finally, the influence of the material used for the core was studied. The larger values are typical of alumina-based shells and cores, and the smaller ones representative of silica-based ones. The test piece CB1.5 is emphasized here, since it is this one which was found to exhibit extensive recrystallization when an alumina core was employed. Fig. 13 summarizes the results of the analysis. Castings produced with an alumina-based core are predicted to have an induced plastic strain of about 2%, which varies slightly with position along the casting. When a silica-based core is used, the induced plasticity is reduced by a factor of approximately half. Hence one advantage of using a silica-based core – in addition to the ease by which it can be leached out of the casting – is a reduced propensity for recrystallization to occur.

To summarize, the castings listed in Table 1 were modelled to examine the effects of casting cross-section, the stress concentration factor, its location and the presence of a core. Fig. 14 plots the maximum plastic strain generated in the different castings and confirms that there is a strong influence of casting geometry. In the solid castings, the induced strain is low and approximately constant irrespective of the gauge diameter and, for the thicker castings, the stress concentration features. However, as the gauge cross-sections were reduced and when a ceramic core is
introduced, the plastic strains increase significantly. The introduction of stress concentration features enhances the accumulated strains. Notches in the top positions of the gauge generated the highest strains; for the TCB test piece, a total accumulated plastic strain of 13% plastic strain was predicted.

3.4. Comparison of predictions with experimental casting trials

In order to assess the accuracy of the modelling, targeted model-driven experimentation was carried out by casting some of the geometries modelled above experimentally.
The castings fabricated are identified in Fig. 15, being plotted in order of increasing predicted plastic strain. After subsequent solution heat treatment, the solid castings of 15 mm gauge diameter were never found to recrystallize; however, when an alumina core was used to reduce the casting wall thickness to 1.5 mm, recrystallization was produced consistently over the entire length of the casting (see Fig. 16). In Fig. 17, the evolution of the plastic strain in the gauge sections of the simple solid casting and the cored casting are compared. During cooling, a plastic strain of only 0.35–0.4% is predicted in the former, and ~2.5% in the latter. The sense of the stress field is also important. Fig. 18 compares the distribution of effective plastic strain and average normal stresses on a longitudinal section and the surface, for both the solid and cored castings. The normal stresses within the gauge regions are primarily tensile, consistent with the lower thermal expansion coefficient of the ceramic mould. However, the intensity and distribution of normal stresses increases with solidification height, as the cooling of the lower regions imparts contraction stresses on the regions above which remain at higher temperatures; this is consistent with the results presented in Fig. 9. This, consequently, is reflected in the strain distribution. Fig. 17 shows that during cooling of the cored casting position L3 is strained first and L1 last. To summarize, the results confirm conclusively that the critical level of accumulated plastic strain for recrystallization to occur is about ~2.5%, and that this strain – consistent with the one-dimensional analysis presented in Section 3.1 – is induced before the temperature drops to 1000 °C.

The analysis revealed further effects which should be noted. When a core is present, the predicted plastic strain is higher on the inner surface than on the outer one; on the other hand, the normal stresses are lower on the inner surface than on the outer one. Furthermore, in the cored castings, the fillet radii between the flanges and the gauges did not cause any substantial stress concentration; consequently, no plastic strain was observed in these regions. However, stress concentration was observed in the fillet regions of the solid casting; it was primarily tensile in nature, but the fillet region below L1 also showed compressive stresses. In the cored casting, the ratio of the gauge cross-section to the flange thickness is much smaller than in the solid
casting. Hence, in the cored casting the gauge regions are the most susceptible to developing the very highest plastic strains (~2.5%). On the other hand, in the solid casting, strains concentrated on features such as fillet regions, and the gauge regions develop only half the maximum strain at most (~0.15%). In the solid casting, strain concentration was observed only in the fillet regions below the flanges, due to the effect of axial contraction during cooling. In all castings, the gauge regions closest to the flanges displayed higher normal stresses, and consequently higher plastic strains. In cored castings, the middle two flanges showed local compressive stresses at the interfaces with the gauge regions. These developed plastic strains of 0.5–1%; such regions might be additional sites for recrystallization.

One of the hollow castings (CB1.5) which exhibited recrystallization after solution heat treatment was examined in detail. Material from location $L_1$ was electrical discharge machining wire cut longitudinally (see Fig. 19a) and transversely (see Fig. 20a), and characterized in the as-cast condition. The horizontal section was examined using scanning electron microscopy, and showed that recrystallization was evident right across the wall thickness (see Fig. 19b and Fig. 20b).

3.5. Transmission electron microscopy

As a further test of the modelling predictions, the microstructure of the as-cast CMSX-4 was examined using TEM. Solid bars were used for this purpose. The dendrite cores – which are characterized by cubic $\gamma'$ of ~100–400 nm – are more or less dislocation-free (see Fig. 21a). However, as the interdendritic regions were approached – comprising

![Fig. 19. Longitudinal cross-section of cored casting at location $L_1$: (a) as-cast condition, (b) after solution heat treatment.](image-url)
~200–1000 nm irregular $\gamma'$ and large $\gamma/\gamma'$ eutectic pools – an appreciable increase in the dislocation density was observed (see Fig. 21b). The dislocations in the as-cast microstructure were primarily at the $\gamma/\gamma'$ interface, forming loops around the precipitates.

To place these observations into context, deformation in as-cast samples strained at various temperatures were studied. Between 20 and 550 °C, shearing by dislocations along the {111}{110} slip systems was the primary deformation mechanism, with dislocation dipoles, loops and pairs in the $\gamma'$ being found (see Fig. 21c). However, with deformation at 750 °C, stacking faults were also observed within the $\gamma'$ (see Fig. 21d); these form due to the activation of {111}{111,2} slip systems. At the higher temperatures of 1050 and 1200 °C, dislocation networks formed at the $\gamma/\gamma'$ interface (see Fig. 21e and f), leaving both $\gamma$ and $\gamma'$ relatively dislocation free. These observations are consistent with those reported in the literature [1,24-25].

The dislocation structure within an as-cast microstructure is rarely reported, and it is significant that a high dislocation density is observed. The conventional solution heat treatment would dissolve the $\gamma'$, allowing the loops to annihilate to give the more familiar dislocation-free microstructure. The simple bar geometry examined experiences the lowest possible stresses during casting, yet clearly undergoes deformation at the yield stress appropriate to the deformation temperature; according to Fig. 18, the induced plastic strain in the bar is estimated to be $\leq 0.15\%$. Comparison of the microstructures in Fig. 21 shows that the as-cast microstructures of Fig. 21a and b most nearly resemble the higher-temperature deformation of Fig. 21e and f, evidenced by the absence of stacking faults, dislocation dipoles and loops within the $\gamma'$. It should be noted that the coarser $\gamma'$ in the strained samples is due to the longer soak time at temperature during the straining tests.

The inhomogeneous dislocation density in Fig. 21a and b supports the deduced temperature range over which casting deformation is induced. This is consistent with results from the process modelling of the bobbin test pieces (Fig. 17). From the TEM, there is no evidence of plastic deformation below 600 °C, and the necessary stresses for yield at 600 °C are probably above the fracture stress of the ceramic shell.

![Fig. 20. Transverse cross-section of cored casting at location L1: (a) as-cast condition, (b) after solution heat treatment.](image-url)
3.6. Critical plastic strain for recrystallization

As a final check of the conclusions drawn from this work, as-cast bars of CMSX-4 were cast and uniaxially strained, at temperatures between 1000 and 1200 °C. They were then subjected to the standard solutioning heat treatment to determine whether recrystallization did indeed occur. Note that the test pieces were prepared with gauge length in the as-cast condition; consistent with the sense of stress expected in practice, testing was carried out in tension. Fig. 22 summarizes the conditions of temperature and induced plastic strain which were found to induce recrystallization. Our results indicate that the critical plastic strain needed to induce recrystallization in CMSX-4 is in the range 1.5–2.0% when it is introduced at temperatures of 1000 °C or higher. Further results of ours, which will be published elsewhere, indicate that substantially greater plastic strains are needed to cause recrystallization when the deformation is...
introduced in the range 550–750 °C and at ambient temperature. Thus, induced strain is appreciably more damaging when introduced at elevated temperatures.

One final point relates to the thermal–elastic–plastic law that is assumed for the modelling. One can see from the high-temperature mechanical testing data of Fig. 5a that it is not easy to define the flow stress of the material precisely at such high temperatures. Some scatter exists in our data – as can be seen from the tests at each temperature which have been duplicated. Moreover, it is probable that at least some of the decrease in apparent Young’s modulus with increase in temperature is due to creep plasticity during tensile straining. Thus, our findings suggest that it would be appropriate to check whether there is a substantial rate-dependence to the plasticity at high temperatures. Substantial creep testing would be required; no data for as-cast CMSX-4 have been reported as yet to the present authors. This is the topic of future research efforts of ours, which will be reported elsewhere.

4. Summary and conclusions

A model for the processing-induced plasticity during investment casting has been presented. It is capable of predicting the sites of localized plasticity needed for recrystallization to occur during heat treatment of single-crystal superalloy castings. The specific conclusions are:

- For typical casting conditions and materials, the analysis suggests that the plasticity needed for recrystallization is induced during the early stages of cooling, when the temperature has not yet fallen below 1000 °C.
- Consistent with the modelling results, TEM microscopy examination indicates that casting deformation is induced at high temperatures close to the γ’ solvus, which is between 1250 and 1310 °C for as-cast CMSX-4.
- The modelling and supporting experimentation indicate that recrystallization is exacerbated by stiffer and thicker shells, by reduced metallic cross-sections and in particular by the presence of stress concentration features. For many of the processing conditions considered – for example, thicker plain cylindrical geometries – the induced plastic strain was insufficient to cause recrystallization.
- Recrystallization was reproduced successfully in laboratory-scale cored castings of CMSX-4 of wall thickness 1.5 mm, for which the modelling indicated that the total plastic strain accumulated during cooling to ambient temperature was ~2.5%. Castings of greater wall thickness were not prone to recrystallization.
- As-cast tensile test pieces were subjected to deformation at elevated temperatures; a plastic strain of only 1.5% induced at temperatures between 1000 and 1300 °C was sufficient to cause recrystallization.
- Such modelling will aid in the assessment of casting strains induced around intricate features, such as blade cooling holes and fillets, and thus to set design criteria to avoid recrystallization in investment castings.

Acknowledgments

C.P. and H.M. acknowledge support from the Engineering & Physical Sciences Research Council (EPSRC) and Rolls-Royce plc, in the form of Dorothy Hodgkin Postgraduate Awards (DHPAs). Helpful discussions with Dr. Hon Tong Pang, University of Cambridge, and Paul Brown and Dr. Neil Jones, Rolls-Royce plc, are acknowledged. The provision of the ProCAST software by the ESI Group, and the help of Dr. Rajab Said in organizing this, is acknowledged as part of the PRISM² collaboration at the University of Birmingham.

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