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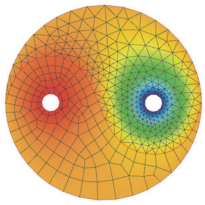

ON THE APPLICABILITY OF JMAK-TYPE MODELS IN PREDICTING IN718 MICROSTRUCTURAL EVOLUTION

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Abstract

Nickel-based superalloys are widely used in the aerospace sector for their mechanical properties, which are directly related to the microstructural and physical properties of these materials. JMAK-type models have been applied to this class of superalloys for the prediction of microstructural evolution phenomena such as recrystallisation. However, these models often lack a clear range of applicability. The majority of the successful applications normally address rather idealised processes (relatively slow forging, simple geometry). However, the industrial production environment generally involves complex strain paths and thermal histories. Thus, there arises the question of whether the JMAK-type models can be applied to such cases. This paper's research focus is to investigate the applicability of JMAK-type models for such processes. To do this, screw press forging of disks was used to validate the in-built JMAK-type model of Inconel 718® available in DEFORM™. In particular, the applicability of the model was examined using a comparison between the results from simulation and from metallographic analysis. At first, the appropriateness of the JMAK outputs in describing the observed microstructures was investigated and then quantitative results were evaluated. The model's outputs were found to be insufficient in describing the observed microstructural states and additional parameters were deemed necessary. The model's predictions ranged from a broadly good match, for which the model could be calibrated with a proposed new methodology, to a qualitative mismatch that highlights the limits of the model's applicability.

Key words: Inconel 718®, Hot Forging, Microstructure Modelling, JMAK

1. INTRODUCTION

Superalloys were developed in the mid-20th century to meet the need of the growing aerospace sector. In particular, the evolution of the gas turbine engine played a crucial role in creating a demand for materials that exhibited high strength at elevated temperatures (Sims, 1984). The developers of the nickel-based superalloys responded to this challenge with the addition of anti-corrosion/oxidation properties, typical of this family of alloys. This fact granted them a key role in the development of materials for high-temperature applications. During this period, the focus shifted from chemical composition to phase constitution and microstructure. Indeed, advanced microscopy techniques highlighted the relation between the special properties of these superalloys and their microstructural state.

The demand for the manufacturing of critical parts with enhanced operational properties led to the question of whether it was possible to tailor the obtained microstructure. A large number of process models and modelling approaches that link processing parameters and microstructure were suggested in the last century, but only a very few were really suitable for industrial application (Humphreys & Hatherly, 2004; Hallberg, 2011).

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The Johnson-Mehl-Avrami-Kolmogorov (JMAK)-type model is one example of this; it is often used for the description of recrystallisation in quasi-single phase alloys. It is a relatively simple and stable model, but its utilisation has to be accompanied with an understanding of its applicability limits, which are almost never clearly specified. The model originated in the 1930's from the work of Kolmogorov, Johnson and Mehl who described the crystallisation of melts, and Avrami who used it to describe the kinetics of phase transformation (Kolmogorov, 1937; Avrami, 1939; Avrami, 1940; Avrami, 1941; Johnson & Mehl, 1939). The original model was based on the concept that phase change takes place through the processes of nucleation and growth. It was formalised with the now-famous simple mathematical relationship demonstrated through the theory of probability, shown in equation 1.

$$V(t) = 1 - e^{-V_E(t)} \tag{1}$$

where $V(t)$ is the fraction of the transformed phase at time t and $V_E(t)$ is the extended volume of the transformed phase (the volume the transformed phase would acquire by omitting impingement of the growing nuclei).

Several derivatives of the basic JMAK model are presently available. Nowadays, modern Finite-Element Analysis (FEA) packages (e.g. DEFORM, QFORM) offer these models as embedded solutions in the finite element code (DEFORM^{TM}, 2014; Micas Simulation LTD, n.d.). These are often based on a few simplifications of the initial theory, through the use of special cases and assumptions. In the literature, there are many successful implementations of these JMAK-type models for modelling recrystallisation in hot forging (Huang et al., 2001; Huber et al., 2008; Reshetov et al., 2016; Medeiros et al., 2000; Na et al., 2003). The majority of these normally address rather idealised processes (relatively slow forging, simple geometry). However, industrial production environment generally involves complex strain paths and thermal histories. Thus, there arises the question of whether the JMAK-type models can be safely applied to such cases. This paper's research focus is to investigate the applicability of JMAK-type model for such processes.

In addition, JMAK-type models as described here are mainly used for the prediction of levels of recrystallisation. However, this may not be the only, nor the dominant, microstructure evolution process. For this reason, an accurate prediction of microstructural evolution cannot uncritically neglect the presence of other phenomena (e.g. dynamic recovery). In particular, this study is based on the hot forging of the nickel superalloy INCONEL 718®.

2. EXPERIMENTAL PROCEDURE

An analysis of the development history of the JMAK model showed that this model was initially developed for isothermal phase change (Kolmogorov, 1937; Johnson & Mehl, 1939; Avrami, 1939). Further model developments and calibrations were often based on laboratory tests with constant temperature and strain rates (Brand et al., 1996; Sellars & Whiteman, 1979; Medeiros et al., 2000; Zhang et al., 1999). Moreover, the majority of successful model implementations (e.g. for extrusion, cylinder upsetting etc.) were characterised by linear material flows and simple thermo-mechanical histories (Huang et al., 2001; Huber et al., 2008; Reshetov et al., 2016; Medeiros et al., 2000; Na et al., 2003). In these cases, the changes in the direction of material flow are limited, which in turn, from a mechanical point of view, results in limited change in the strain tensor principal axes. Meanwhile, processing maps show how different thermo-mechanical states are linked to different microstructure evolution changes as shown in figure 1 (Sui et al., 2011). Since the majority of the applications studied so far have a simple thermo-mechanical history, their microstructure evolution is often limited to one microstructure phenomenon (e.g. only dynamic recrystallisation).

Unfortunately, this may not be the case in an industrial scenario, and it is of interest to investigate the behaviour of material at boundary regions of maps such as that shown in figure 1. To do this, it is necessary to study processes with more complicated thermo-mechanical histories and flow directions. Herein, we performed hot-forging of disks with a screw press (with a maximum ram speed of 700 *mm/s*) so as to observe the effect of different strain trajectories in a single disk as shown in figure 2. This allows us to take into account the possible effects of concurrent microstructure evolution phenomena.

The screw press energy was set to 80 *kJ* resulting in a ram speed of 400 *mm/s* at impact. Samples were forged at 970°*C* and were quenched to isolate dynamic recrystallisation and prevent other post-deformation microstructural evolution phenomena (e.g. static recrystallisation, grain growth). The pre-

forging heat treatment time was estimated through FE modelling and the surface temperature of the billet in the furnace was monitored with thermo-couples. The initial material was examined through a scanning electron microscope (SEM) as shown in figure 2b. The average grain size (120 μm in our case) was calculated using planimetric method (ASTM E112-13, 2014). A qualitative evaluation showed the absence of δ phase, the implications of which, however, will not be discussed in this paper.

3. MICROSTRUCTURE MODELLING

As the focus of the JMAK model was to describe phase change through nucleation and growth, the main model output was the percentage of the new phase at a given moment in time, as shown in equation 1. JMAK-type models for dynamic recrystallisation retained the format, and the principal model output is represented by the recrystallised volume fraction. This is calculated as a function of strain,

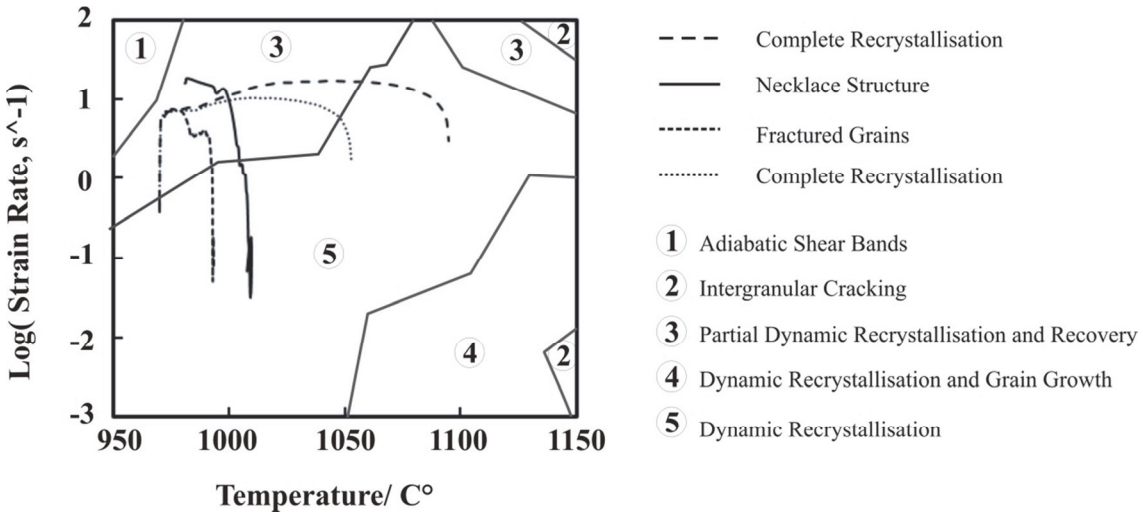


Fig. 1. *Processing map of IN718 for a plastic strain of 0.5 (Sui et al., 2011) and thermo-mechanical history of 4 points taken from a section of a forged disk with reference to their final microstructure*

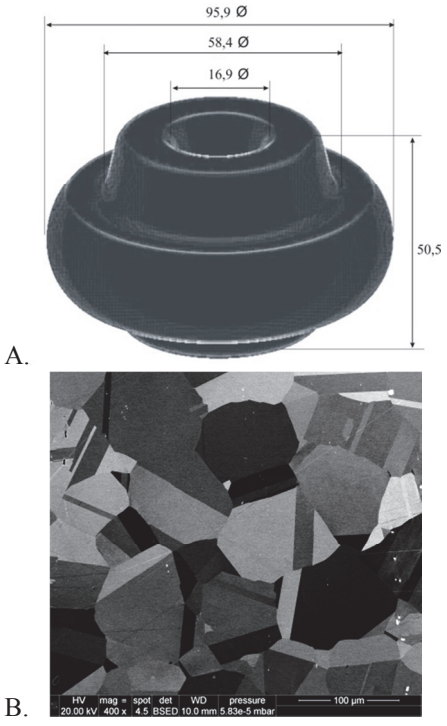
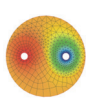


Fig. 2. *A - Disk Geometry (dimensions shown in mm), B - As-received Material Microstructure (BSED image)*

instead of time, owing to the short duration of the deformation process. In addition to this, modern variations of the model offer outputs related to the grain size of the new phase. These are usually the grain size of newly recrystallized grains and the average grain size (DEFORM^{TM}, 2014).

There are two main modes in which such micro-structural models are implemented: coupled models generated during the simulation itself and those using a post-processor. In the former, the microstructure model and the metal forming simulation run at the same time, allowing the microstructure model to influence the deformation process. In the post-processor approach, the metal forming simulation is carried out first, and then the sequence of the thermo-mechanical states is used as an input for the microstructure evolution model.

The investigation described here is based on a post-processor microstructure model. Thus, the first step is the simulation of the forging process, which was performed with DEFORM^{TM} as shown in figure 3. Material data on IN718 was taken from the hot working guide published by Prasad et al. (1997). For strain rates above the 10^2 s^{-1} threshold, data was



extrapolated. With regard to boundary conditions, friction was calculated as pure shear, with a friction factor of 0.3, based on the inbuilt settings available in DEFORM™. A simple heat transfer was chosen between the dies (T = 250°C) and workpiece (T = 970°C) with an assumed heat transfer coefficient of 22500 $W/m^2{\cdot}K$. Since the forging simulation parameters (e.g. T, strain rate) serve as the input for the microstructure evolution model, the simulation results were validated by comparison with experimental data (geometry of final part and load).

The in-built JMAK-type model and material-related coefficients for IN718® (see table 1) were taken from the DEFORM™ library (DEFORM™, 2014), while the average grain size was measured as noted in the experimental section. Four main aspects of dynamic recrystallisation were calculated: recrystallised volume fraction (equation 2-3), activation conditions (equation 4-5) size of the recrystallised grains (equation 6), and the average grain size (equation 7).

$$X_{drx} = \begin{cases} 1-\exp\left[-\beta_d\left(\dfrac{\varepsilon-\varepsilon_c}{\varepsilon_{0.5}}\right)^{k_d}\right], & \varepsilon \geq \varepsilon_c \\ 0, & \varepsilon \leq \varepsilon_c \end{cases} \quad (2)$$

$$\varepsilon_{0.5} = a_5 d_0^{h_5} \varepsilon^{n_5} \dot{\varepsilon}^{m_1} \exp\left(\frac{Q_5}{RT}\right) + c_5 \quad (3)$$

where: X_{drx} – recrystallised volume, ε_c – critical strain, $\varepsilon_{0.5}$ – strain required to achieve 50% recrystallisation, d_0 – initial grain size, Q_5 – activation energy, ε – strain, $\dot{\varepsilon}$ – strain rate, T – temperature, R – gas constant, and $\beta_d, k_d, a_5, h_5, n_5, m_5, c_5$ – material constants.

$$\varepsilon_c = a_2 \varepsilon_p \quad (4)$$

$$\varepsilon_p = a_1 d_0^{n_1} \dot{\varepsilon}^{m_1} \exp\left(\frac{Q_1}{RT}\right) + c_1 \quad (5)$$

where: ε_c – critical strain, ε_p – peak strain, d_0 – initial grain size, Q_1 – activation energy, $\dot{\varepsilon}$ – strain rate, T – temperature, R – gas constant, and $a_1/a_2/n_1/m_1/c_1$-material constants.

$$d_{drx} = a_8 d_0^{h_8} \varepsilon^{n_8} \dot{\varepsilon}^{m_{81}} \exp\left(\frac{Q_8}{RT}\right) + c_8 \quad (6)$$

where: d_{drx} – dynamically recrystallised grain size, d_0 – initial grain size, Q_8 – activation energy, ε – strain, $\dot{\varepsilon}$ – strain rate, T – temperature, R – gas constant, and $a_8/h_8/n_8/m_8/c_8$ – material constants.

$$d_{avg} = X_{drx} \cdot d_{drx} + (1 - X_{drx}) \cdot d_0 \quad (7)$$

where: X_{drx} – recrystallised volume, d_{drx} – dynamically recrystallised grain size, and d_0 – initial grain size.

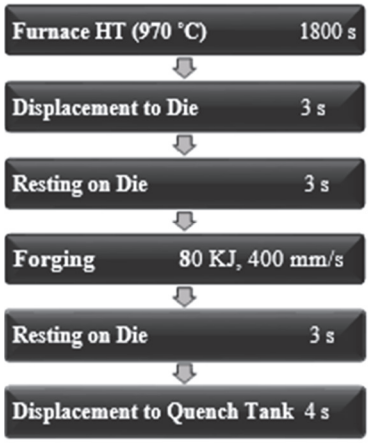


Fig. 3. *Metal Forming Simulation Steps*

Table 1. *Material Coefficients for IN718 (DEFORM™, 2014)*

a_1	$4{,}659{\cdot}10^{-3}$	n_5	-1,42
a_2	0,83	m_5	-0,48
n_1	0	c_5	0
m_1	0,1238	Q_5	$1{,}96{\cdot}10^{5}$
c_1	0	a_8	$4{,}85{\cdot}10^{10}$
Q_1	$4{,}952{\cdot}10^{4}$	h_8	0
β_d	0,693	n_8	-0,41
k_d	1	m_8	-0,028
a_5	$6{,}887{\cdot}10^{-4}$	c_8	0
h_5	0	Q_8	$-2{,}4{\cdot}10^{5}$

4. RESULTS AND DISCUSSION

As mentioned, the aim of this paper is to investigate the applicability of JMAK-type models in industrial manufacturing processes. These are characterised by complicated thermo-mechanical history and flow trajectories, which may result in a variety of microstructural evolution phenomena (e.g. Recovery, Recrystallisation, etc.). However, JMAK-type models are mainly used only for the description of recrystallisation. Therefore, when evaluating the applicability of these models it cannot be implied that they are wrongly predicting microstructure, because their use may be simply inappropriate. To give an example, it could be the case in which additional models and/or parameters may be needed in conjunction with the JMAK-type model used. Hence, there arises the first important question of when and where it is appropriate to use these mod-

els. Only then, can the applicability be investigated in terms of numerical mismatch between the microstructure model outputs and experimental results.

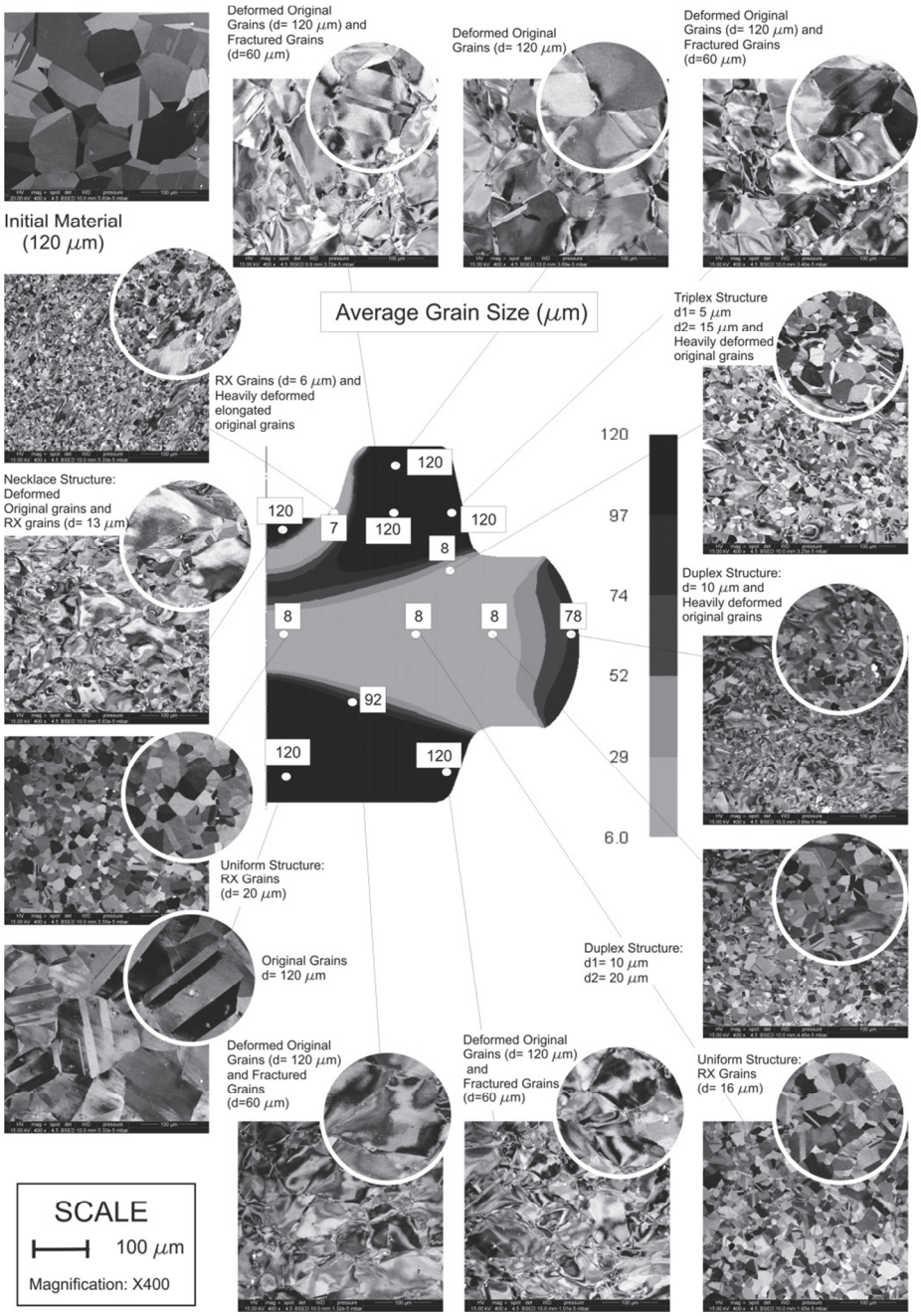


Fig. 4. *Grain-size comparison between metallographic analysis and microstructure simulation. In the central section are shown the results of the Average Grain Size output from microstructure simulation, and beside each point of interest is shown the specific value predicted by the model*

Figure 4 shows the results from simulation and SEM analysis, and the presence of a few different structural states (e.g., duplex and triplex grain structures, quasi-original and heavily deformed grains) can be observed. In figure 5, based on our observation, we have classified the available types of microstructure into three zones: non-recrystallised (A), partially recrystallised (B) and completely recrystallised (C). Each zone was further clustered into subgroups as detailed below:

A. Non-Recrystallised:

A1. Original Grains with Small Deformation - where grains have maintained the original shape and size with the addition of slight deformation features (little intragranular deformation).

A2. Heavily Deformed Grains with Fragmentation - where heavy deformation features can be

seen (significant intragranular deformation) and a portion of grains seem fractured.

B. Partially Recrystallised:

B1. Duplex Necklace - where two types of grains – larger, deformed grains with smaller , recrystallized, grains on the boundaries – are observed.
B2. Triplex - where 3 different types of grains – larger, deformed; small, recrystallised (necklace); and an intermediate size with a clean surface (possibly recrystallised) – can be observed. Each grain type has a clear prevailing size with no smooth transition between the types (no intermediate grain sizes).

C. Completely Recrystallised:

C1. Homogeneous - where all the grains have approximately the same grain size and a deformation free surface.
C2. Duplex - where two distinctive homogenously distributed grain sizes (no intermediate grain sizes) can be observed.

To analyse the applicability of the JMAK-type model, we studied the significance of these three outputs in describing the different zones and subgroups:

- X_{drx} is an important parameter that has to be implemented to allow the main distinction between zone A, B and C.
- d_{drx} may be used as is to describe homogeneous recrystallisation (C1). For duplex necklace structures (B1), it can be used for calculating the size of the smaller recrystallised grains; however the remaining (unrecrystallised) grains need to be described by an additional parameter. For duplex and triplex structures (B2, C2), d_{drx} may not be sufficient because, as we have observed, there may be different sizes of recrystallised grains.
- d_{avg} can be used only for original and homogeneous grain sizes (A1 and C1). In contrast, for morphologies like C1, C2 and B2, this parameter may become misleading as it may specify

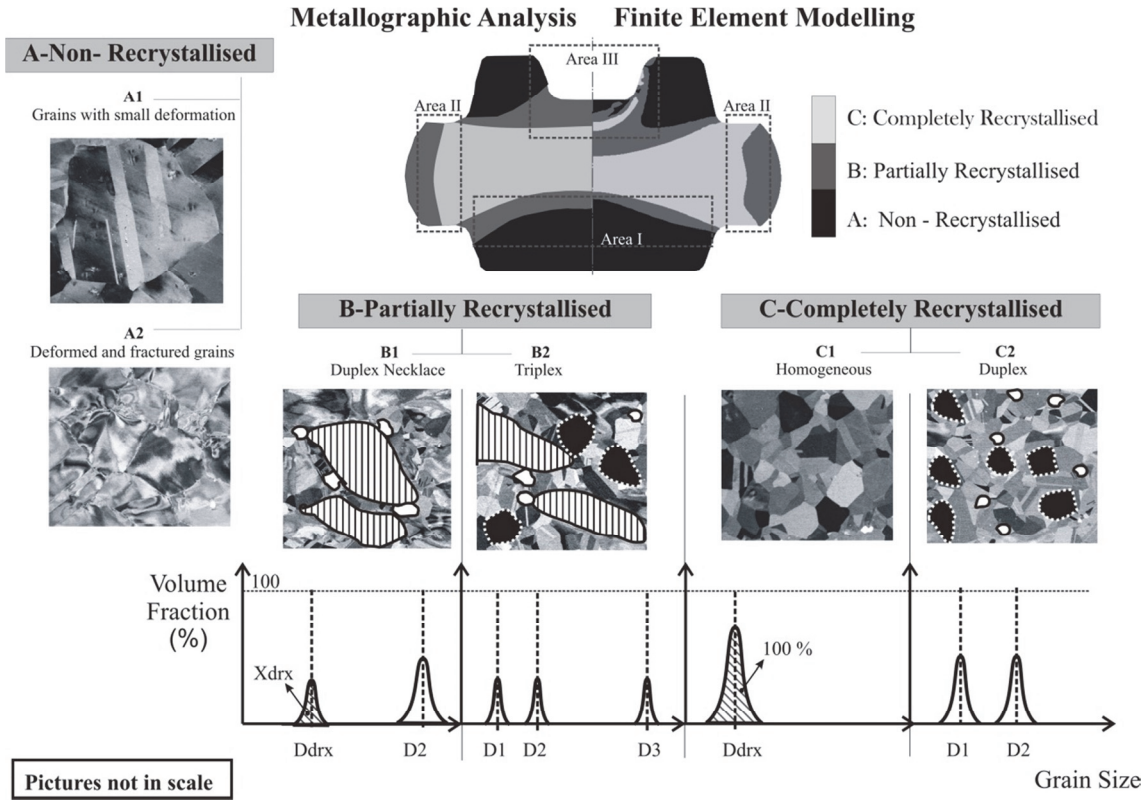


Fig. 5. *(Top) Metallographic analysis and microstructure simulation comparison for recrystallised volume fraction; (Periphery and bottom) microstructure groups and related grain-size distributions*

However, as implied in the previous section, we have only three main outputs from the model, namely, X_{drx} - Recrystallised volume fraction, d_{drx} - Dimensions of recrystallised grains and d_{avg} - Average grain size.

a size that does not actually exist in the material.

Evidently, additional parameters are necessary to characterise the microstructural states identified. However, it must be remembered that JMAK-type models are mainly used in industrial applications

because of their simplicity. Therefore, it is important not to overcomplicate these models and to add only sufficient parameters to be able to more adequately describe the microstructure in a more thorough way.

As observed, the average grain size is insufficient for description in zones where different grain sizes are present. In these cases, the easiest solution can be the use of a grain-size distribution. However, if this is taken as a vector, its implementation in the model may be difficult. Fortunately, there was an absence of intermediate sizes between each grain-type in the subgroups, resulting in distribution functions characterised by a few main peaks (one for each grain-type). Consequently, it may be possible to use the mean values of the peaks. To do this, it is necessary to find the volume fraction of each peak. While this may be easy for A1, B1 and C1 (only X_{drx} required), more parameters are required for duplex and triplex structures (B2, C2).

Now that we have assessed where the model could in principle be used, it is possible to examine how well it performed. The two most relevant aspects to be addressed for industrial applications are the initial and final stages of recrystallisation (the 0% and 100% boundary). The initial stage or 0% boundary (Zone A in figure 5) corresponds to the "dead-zone" boundary, wherein are found the biggest gradients in material properties. On the other hand the final stage or 100% boundary (Zone C in figure 5) represents industry's desired microstructure outcome (i.e. complete homogeneous recrystallisation). Intermediate stages of recrystallisation (e.g. Zone B in figure 5) are still important, but are only secondary to these, and as observed, only relevant to limited areas. Similarly, grain size related outputs (d_{drx}, d_{avg}) are likely to be of use in only a few cases. Hence, these and partial recrystallisation will be taken on in future research and as such not included in this study.

Analysing '0%' and '100%' boundaries, two situations can be observed. In some areas (e.g. Area I and Area II, figure 5), the matching between the prediction and experimental data is qualitatively reasonable. At the same time, in Area III, shown in figure 5, the results from the simulation were found to give a poor match to experimental results.

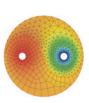
In principle, some quantitative mismatch in Areas I and II can be written off as caused by the poor calibration of the model. Unfortunately, microstructural models often lack a clear methodology for calibration, resulting in a great limitation on their capabilities. Indeed, mathematical equations are just one part of a model, and cannot be definitive without a defined calibration methodology and applicability limits. The model used in this paper utilizes the standard values from the DEFORM^{TM} database. These values were obtained on the basis of a method in which the coefficients of equation 3 were calculated using the peak stress and strain values taken from the mechanical stress-strain curve (Mirzadeh et al., 2009). However, it is pretty clear that this method is very rough. The accuracy of experimental location of the peak in the stress-strain curve is quite limited and its relation to the beginning of DRX is not proved (due to the simultaneous presence of other microstructural mechanisms that can be also responsible for the softening). Besides this, uniaxial tests with constant temperature and strain rate do not give the full picture of the phenomena involved in complex forging processes.

A combination of forging trials, modelling and microstructural analysis can provide more room for the improvement of the calibration methodology. As was noted earlier industrial forging processes give rise to very different sequences of thermomechanical conditions in different points in the body. The microstructure formed at the end of the process is the accumulated result of the total history (changes in temperature, strain rates, triaxiality, etc.). Hence, it is not surprising that any instantaneous values of accumulated plastic strain, temperature or strain rate sometimes give a confusing picture and poor calibration results. To overcome this problem, the following approach is suggested. The threshold for the beginning of recrystallisation corresponds to the condition where the left-hand side of equation 2 equals zero. One of the most likely conditions for this to happen is when the accumulated plastic strain (ε) becomes equal to the critical strain (ε_c), which finally will lead to the equation:

$$\varepsilon = k_1 d_0^{n_1} \dot{\varepsilon}^{m_1} \exp\left(\frac{Q_1}{RT}\right) + k_2 \qquad (8)$$

where $\dot{\varepsilon}$ (strain rate) and T (temperature) are the process variables, while k_1, n_1, m_1, Q_1 and k_2 are the material constants, which have to be found for the model calibration (R is the gas constant).

Being plotted in the space of temperature, strain rate and accumulated plastic strain, this equation will form a surface, splitting the space into two parts – where recrystallisation should or should not take place (figure 6b). At the same time, for every point in the forged part, the history of the process can be depicted as a trajectory in the same space. Then, if



the threshold part of the model is calibrated correctly, we should get a picture where the trajectories of all the points in the A-Zones (figure 5) – zones with no recrystallisation – are located inside the surface

– if the mathematical form of equation 8 does not permit the description of the surface obtained with a single set of constants, either local approximations can be used, or a different spline

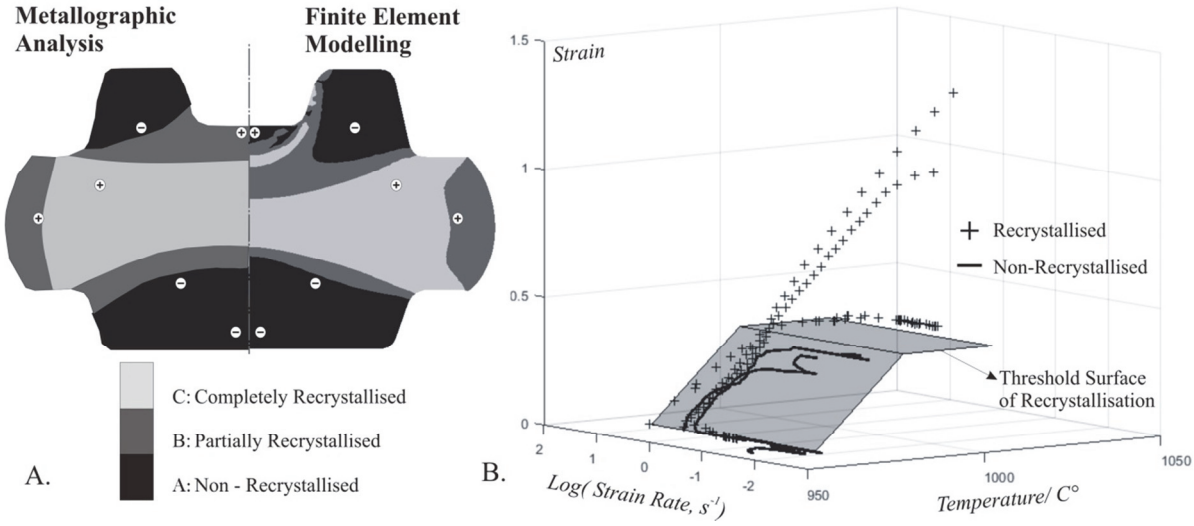


Fig. 6. *The 3D visualisation of the thermomechanical history of the deformation process (B) in the different points of the forged body (A) and the principle scheme of the threshold surface of recrystallisation (B)*

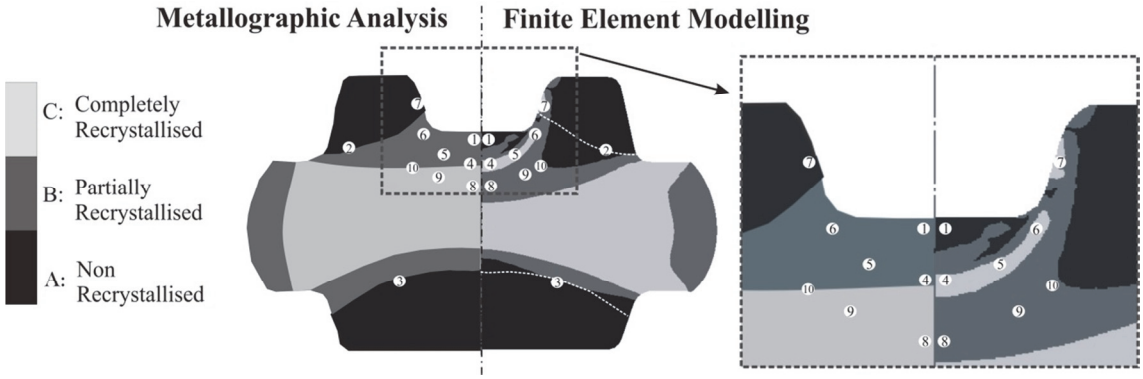


Fig. 7. *The area of most significant mismatch between simulation results and experimentally observed microstructure*

described with equation 8. In summary, the proposed calibration process will involve the following steps:

– part is forged;
– the forging process is FE-simulated;
– microstructural analysis of the section of the forged part is done;
– zones without recrystallisation are detected;
– for the points in material belonging to these zones, the history of the temperature, strain rate and plastic strain are taken from the results of the FE-simulation;
– the trajectories of the corresponding histories are plotted in the space of strain rate-temperature-plastic strain;
– the points of the enveloping surface are detected;
– the coefficients of equation 8 are found;

chosen.

However, while there is a promising chance to rectify the prediction in Areas I and II (figure 5) by improving the calibration of the model, the situation with Area III is more difficult. This area is shown in more detail in figure 7. It can be seen that, for example, point 1, which is expected to be a non-recrystallized dead zone, has in reality a necklace structure (see figure 4). Points 4, 5, 6 and 7 are predicted to have a completely recrystallized microstructure. In reality, in point 7, recrystallisation did not even start, while at the other points, a partly recrystallized microstructure is observed. In contrary, points 8, 9 and 10 should be partially recrystallized according to the model prediction but are actually completely recrystallized.

Can these observations witness that JMAK models give qualitatively wrong predictions and cannot be used for this type of forging process? Most probably yes, but it is definitely too early to make this conclusion without a proper understanding of the nature of this mismatch. It is first required to recalibrate the model by following the steps listed above. Following this, the trajectories of the points of mismatch have to be carefully analysed to understand how the history actually affected recrystallisation.

5. SUMMARY

Simple models based on the JMAK assumptions can be effective for through-process modelling, but they require clear specifications with regard to their range of applicability. Utilisation of JMAK models for recrystallisation prediction in quasi-steady processes (e.g. simple geometry of metal flow, low-speed forging, extrusion, etc.) have shown good results (Huang et al., 2001; Medeiros et al., 2000; Huber et al., 2008; Reshetov et al., 2016; Na et al., 2003). However, industrial manufacturing processes are often more complex. Therefore, experiments were designed to achieve different deformation trajectories and thermo-mechanical histories in a single disk section. The results of the simulation were compared to those of the metallographic analysis. In many areas, the examined microstructure did not match the results from simulation. In particular:

- Six main microstructure groups were identified and the applicability of the model outputs for their description was evaluated. It was observed that in regions where recrystallisation was not present, there were microstructural changes that have to be taken into account (e.g. fractured grains, deformed grains). Moreover, in certain areas multiple grain types were found (with different characteristic sizes), resulting in duplex and triplex grain structures. The analysis on the applicability of the model outputs revealed that:
  - X_{drx} - Recrystallised volume fraction is an important parameter that has to be implemented.
  - d_{drx} - Size of recrystallised grains may be used but it is not sufficient because, as we have observed, there may be different sizes of recrystallised grains.
  - d_{avg} –The applicability of this parameter is very limited and in the majority of cases it specifies a size that does not actually exist in the material.
  - More parameters are necessary for the description of the observed microstructural groups.
- Once the applicability of the outputs was evaluated, predictions were compared where appropriate. Due to its industrial relevance and its availability in the workpiece examined, the zero (0%) and complete (100%) recrystallisation boundaries were taken for an in-depth analysis. It was observed that:
  - In certain parts of the workpiece, the boundary prediction was qualitatively acceptable but still not satisfactory (the model seems broadly applicable). This may be simply due to the model being wrongly calibrated. Therefore, a possible methodology for calibration was proposed, however until this is performed the applicability cannot be fully evaluated.
  - In other areas, the results from the simulation were found to give a poor match to experimental results. These most probably cannot be corrected by simple recalibration of the model, although further investigations are required for final conclusions on the applicability of JMAK model.

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REFERENCES

ASTM E112-13, 2014, Standard Test Methods for Determining Average Grain Size, *ASTM International*, 1-28.

Avrami, M., 1941, Granulation, Phase Change, and Microstructure Kinetics of Phase Change. III, *The Journal of Chemical Physics*, 9(2), 177.

Avrami, M., 1940, Kinetics of Phase Change. II - Transformation-Time Relations for Random Distribution of Nuclei, *Journal of Chemical Physics*, 8(1940), 212-224.

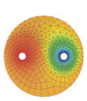
Avrami, M., 1939, Kinetics of phase change. I general theory, *The Journal of Chemical Physics*, 7(12), 1103.

Brand, A. J., Karhausen, K., Kopp, R., 1996, Microstructural simulation of nickel base alloy Incone* 718 in production of turbine discs, *Materials Science and Technology*, 12(11), 963-969.

DEFORM™, 2014, DEFORM™ V11.0.1 System Documentation, *Reed Road Columbus, Ohio: Scientific Forming Technologies Corporation*.

Hallberg, H., 2011, Approaches to Modeling of Recrystallization, *Metals*, 1(1), 16-48.

Huang, D., Wu, W.T., Lambert, D., S. L. Semiatin, 2001, Computer simulation of microstructure evolution during



hot forging of Waspaloy and nickel alloy 718, *Proc. Microstruct. Model. Predict. Dur. thermomechanical Process. TMS*, eds, R. Srinivasan, S.L. Semiatin, A. Beaudoin, S. Fox, Z. Jin, Indianapolis (Indiana), 137-146.

Huber, D., Stotter, C., Sommitsch, C., Mitsche, S., Poelt, P., Buchmayr, B., Stockinger, M., 2008, Microstructure modeling of the dynamic recrystallization kinetics during turbine disc forging of the nickel based superalloy Allvac 718Plus™, *Superalloys 2008*, 855-861.

Humphreys, F.J., Hatherly, M., 2004, *Recrystallization and related annealing phenomena*, Elsevier, Amsterdam.

Johnson, W., Mehl, R., 1939, Reaction kinetics in processes of nucleation and growth. *Transactions of the American Institute of Mining and Metallurgical Engineers*, (135), 416-458.

Kolmogorov, A.N., 1937, On the Statistical Theory of the Crystallization of Metals, *Bulletin of the Academy of Sciences of the USSR*, 1(Mathematics Series), 355-359.

Medeiros, S.C., Prasad, Y.V.R.K., Frazies, W.G., Srinivasan, R., 2000, Microstructural modeling of metadynamic recrystallization in hot working of IN 718 superalloy, *Materials Science and Engineering A*, 293(1), 198-207.

Micas Simulation LTD, QFORM v8 FE package documentation: Description of Microstructure evolution model, Available at: http://www.qform3d.co.uk/db_files/17/123.pdf [Accessed September 1, 2016].

Mirzadeh, H., Najafizadeh, A., Moazeny, M., 2009, Flow curve analysis of 17-4 PH stainless steel under hot compression test, *Metallurgical and Materials Transactions A: Physical Metallurgy and Materials Science*, 40(12), 2950-2958.

Na, Y.S., Teom, J.T., Park, N.K., Lee, J.Y., 2003. Simulation of microstructures for Alloy 718 blade forging using 3D FEM simulator, *Journal of Materials Processing Technology*, 141(3), 337-342.

Prasad, Y.V.R.K., Rao, K.P., Sasidhara, S., 1997, *Hot Working Guide: A Compendium of Processing Maps* Second Edi., ASM International, Materials Park (Ohio).

Reshetov, A., Bylya, O., Stefani, N., Rosochowska, M., Blackwell, P., 2016, An Approach To Microstructure Modelling In Nickel Based Superalloys, *TMS 13th International Symposium on Superalloys*, eds., M. Hardy, E. Huron, U. Glatzel, B. Griffin, B. Lewis, C. Rae, V. Seetharaman, S. Ti, Seven Springs (Pennsylvania), 531-538.

Sellars, C.M., Whiteman, J. A., 1979, Recrystallization and grain growth in hot rolling, *Metal Science*, 13(April), 187–194.

Sims, C.T., 1984, A History of Superalloy Metallurgy for Superalloy Metallurgists, *Superalloys 1984 (Fifth International Symposium)*, ed., M. Gel, 399-419.

Sui, F.L., Xu, L.X., Chen, L.Q., Liu, X.H., 2011, Processing map for hot working of Inconel 718 alloy, *Journal of Materials Processing Technology*, 211(3), 433-440.

Zhang, J.M., Gao, Z.Y., Zhuang, J.Y., Zhong, Z.Y., 1999, Mathematical modeling of the hot-deformation behavior of superalloy IN718, *Metallurgical and Materials Transactions A*, 30(10), 2701-1712.

PRZYDATNOŚĆ MODELU TYPU JMAK DO PRZEWIDYWANIA ROZWOJU MIKROSTRUKTURY STOPU IN718

Streszczenie

Ze względu na swoje wysokie własności mechaniczne, ściśle związane z mikrostrukturą i własnościami fizycznymi, stopy na bazie niklu są szeroko stosowane w przemyśle lotniczym. Modele typu JMAK są powszechnie wykorzystywane do opisu rozwoju mikrostruktury w tych stopach. Należy jednak zdawać sobie sprawę z faktu, że te modele mają czasem ograniczone zastosowanie. Większość udanych przykładów wykorzystania modelu JMAK dotyczy wyidealizowanych procesów charakteryzujących się prostą geometrią i małymi prędkościami odkształcania. Z drugiej strony, w przemysłowych procesach kucia mamy często do czynienia ze złożonymi drogami odkształcenia i złożonymi cieplnymi historiami odkształcenia. Powstaje zatem pytanie na ile modele JMAK mogą być zastosowane do analizowanego w pracy przypadku kucia stopu IN718. Aby odpowiedzieć na to pytanie przeprowadzono badania zmierzające do oceny przydatności modeli JMAK. Ten cel osiągnięto przeprowadzając symulacje procesu prasowania dysku ze stopu Inconel 718®, stosując model JMAK wbudowany w program z metody elementów skończonych. Parametry modelu pobrano z bazy danych programu DEFORMTM. Przydatność modelu oceniano na podstawie porównania wyników symulacji numerycznej z analizą metalograficzną próbek po symulacji fizycznej. W pierwszej kolejności oceniono poprawność jakościową modelu w opisie rozwoju mikrostruktury, a następnie porównano wyniki ilościowe. Analiza wszystkich wyników wykazała, że przewidywania modelu nie były satysfakcjonujące i konieczne było uwzględnienie dodatkowych parametrów procesu. Uzyskiwane wyniki wahały się od bardzo dobrej zgodności z doświadczeniem, i dla tych warunków możliwa była kalibracja modelu, do dużych rozbieżności które wyznaczały zakres stosowalności modelu.

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