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Benchmarking mean-field models available in commercial FE software in application to two-blow forging of IN718 alloy

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Abstract. This paper analyses the potential of standard mean-field models available in commercial FE software Deform®, Forge® and QForm® for a microstructural prediction during multistage forging process of Inconel 718 at conditions close to industrial ones. The special set of experimental trials including heating, forging, reheating and final forging were conducted on 5 MN hydraulic press with detailed measurements of temperature distributions, timings and forging parameters. The microstructure distribution was investigated after each stage of the process (optical and EBSD) and compared with the predictions obtained in three softwares. Standard and optional capabilities as well as limitations and challenges of the models were investigated, and some improvement ideas were proposed.

Introduction

Microstructural modelling, as well as appropriate calibration and application of available mean-field microstructure evolution models for industrial hot forging processes, remains a bottleneck [1,2]. It is getting even more complex when these models are applied for the real-life technological processes where material goes through number of forging and reheating stages. The goal of such complicated route is to get a final part with a required microstructure. Due to a large number of technological parameters involved in this process, troubleshooting and optimisation of the manufacturing route requires the capability of microstructure evolution models to predict final microstructure with some acceptable accuracy.

All the most popular commercial FE metal forming softwares, like Deform®, Forge® and QForm® have some in-built capabilities for microstructural modelling. The study presented in this paper carried out using Inconel 718 alloy. For this and similar to it nickel-based superalloys dominating mechanisms during hot forging, which are intended to be simulated, are Dynamic (DRX), Meta-Dynamic (MDRX) and Static (SRX) recrystallisation along with the static Grain Growth (GG). There also few other microstructural mechanisms which are not directly modelled in available in-built models, but which should however be kept in mind as they can affect the results: Recovery (RC), Twinning (TW) and Continues Recrystallisation (CRX). The question of their consideration will be briefly touched in the Results and Discussion section.

The microstructural models in-built into the three above mentioned metal forming softwares for DRX, MDRX, SRX and GG have quite similar mathematical structure (with minor variations in shape and coding realisation) and individual material library of default parameters. Three recrystallisation models belong to the so-called JMAK-type class of the models [3] and the grain-growth model is described with the classical model of P.A. Beck [4].

The limitations of the JMAK-type models are well known and analysed in many papers [1, 5-7]. They are not well fitted for the processes with the sharp changes in temperatures and strain rates. The set of material parameters is normally suitable for a fixed range of temperatures and

strain rates which they are optimised for, etc. However, these models are the simplest, intuitively "physically" clear and available as in-built models in the commercial FE softwares.

The main goal of this benchmarking study was to work out some practical conclusions regarding application of standard models. The set of tests consisting of heating, forging, reheating and second blow forging was specially designed for this purpose. Accurate direct measurements of all possible process parameters was done for the fine tuning of FE models providing robust prediction of the thermo-mechanical histories of all points of the performs. All the data gathered was further used as input for the microstructural modelling. Separate analysis of microstructure at every stage provided an ability to decouple the models of recrystallisation and give additional information on the reason of the "wrong" behavior of the models.

The special attention was addressed to investigation of the areas of the forged billets where microstructure prediction was not accurate. The reasons of inaccurate prediction were analysed to suggest possible improvements of the models.

Methodology

Forging trials.

Two-blows forging route with reheating between the blows was employed to obtain experimental data for benchmarking microstructure evolution models (see Fig. 1). The standard double truncated cone (DTC) geometry with 50.8 mm initial height was used for the forging. The billets were machined from IN718 material with standard chemical composition. Forging trials of IN718 DTC geometry were carried out at the sub-solvus temperature of 990°C on 5 MN hydraulic press. The dies of the press were heated up to 600°C. The billets were reheated in the furnace back to 990°C between the blows. Taking into account non-isothermal conditions of the forging, relatively bulky billets, as well as multi-stage forging route, these trials mimic a real industrial forging conditions for this alloy. To track the microstructure evolution at the various stages of the forging route, three DTC-s were forged using following routes:

- DTC#1: 1st blow air cooling (AC)
- DTC#2: 1st blow reheating in furnace back to 990°C AC
- DTC#3: 1st blow reheating 2nd blow AC (full route)

A number of real-time process parameters was recorded during the forging trials. Such press readings as velocity of top ram, force and displacement were recorded as a function of a time. Along with that, temperature in 3 reference points was measured for each processed billet during all the forging route including initial heating and final air cooling. For this purpose 3 thermocouples were embedded into the each proceed billet: one thermocouple for the centre of the billet and two thermocouples were embedded on the 5mm distance from the top and bottom surface of the billet in the same way as it was shown in [2, 8]. The example of obtained thermocouples readings is shown in Fig. 2. The trial of each billet was video-recorded to define the exact time of each operation of the processing route. The shape of final geometry of forged parts was analysed using a 3D blue light-based metrology scanning system (by GOM ATOS TripleScan III).

Digital twin of the experiment. Digital twins of thermo-mechanical processing were created for each processed billet. By the term "digital twin" here is implied validated FE simulations which reproduce particular forging conditions, heating and cooling profiles, timings of operations, as well as peculiarities of metal flow for each forged billet. The FE models were created using several commercially available FE packages: DEFORM, QFORM and FORGE. Each of these software developers has its own mean-field microstructure evolution model for IN718 material, which was benchmarked in this work.

This phase of study is critically important because microstructure evolution models employ macroscopic process parameters, namely strain, temperature and strain rate obtained from FE

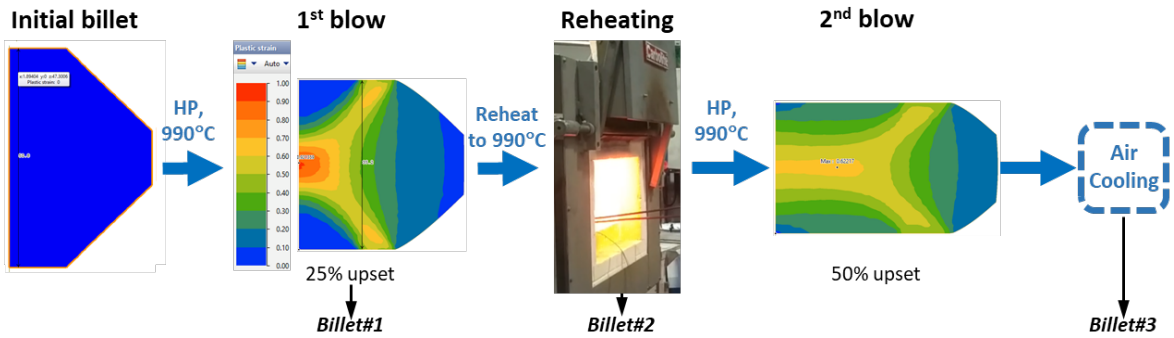


Fig. 1. Forging route used for two-blow forging trials of IN718 on hydraulic press.

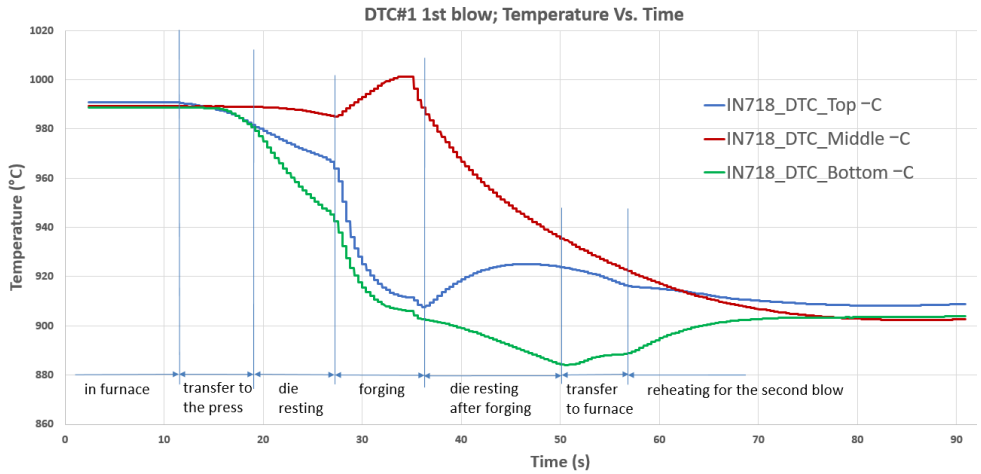


Fig. 2. Thermocouples readings for the DTC#2 (just the 1st blow and transfer back to furnace).

simulations. The settings in the FE models were adjusted in a way to reflect actual forging conditions. The calibration procedure consisted of following steps:

- Defining exact time of each operation using video-recordings and thermocouples' readings. The full list of simulated operations was: pre-heating, transfer to furnace, die resting before the 1st blow, 1st blow forging, die resting after the 1st blow, transfer back to the furnace, re-heating, transfer for the 2nd blow, die resting, 2nd blow forging, die resting, transfer for air cooling, air cooling.
- Adjusting heat-exchange coefficients in simulations in a way to repeat experimentally obtained profiles of temperature in reference points.
- Friction conditions were validated by superimposing images from GOM scan and those obtained in the simulations.
- Validating metal flow in the simulations using available experimental data. The final shapes and positions of the holes drilled for the thermocouples were used to validate the metal flow and strain values locally in the vicinity of the billets' dead-zones and central locations. Also force readings from the press were used to check the force prediction in simulations.

More detailed description of the procedure of creating digital twins is shown in earlier studies [2, 8].

The main calibration criteria for the simulations were temperature and strain fields, which have had to be similar to those obtained in the experiment and to be equal between FE simulations created in various FE packages. Fig. 3 demonstrates the comparison between the temperature fields and strain fields in all utilised FE packages for the billet after the first blow (Fig. 3*a*) and the second blow (Fig. 3*b*).

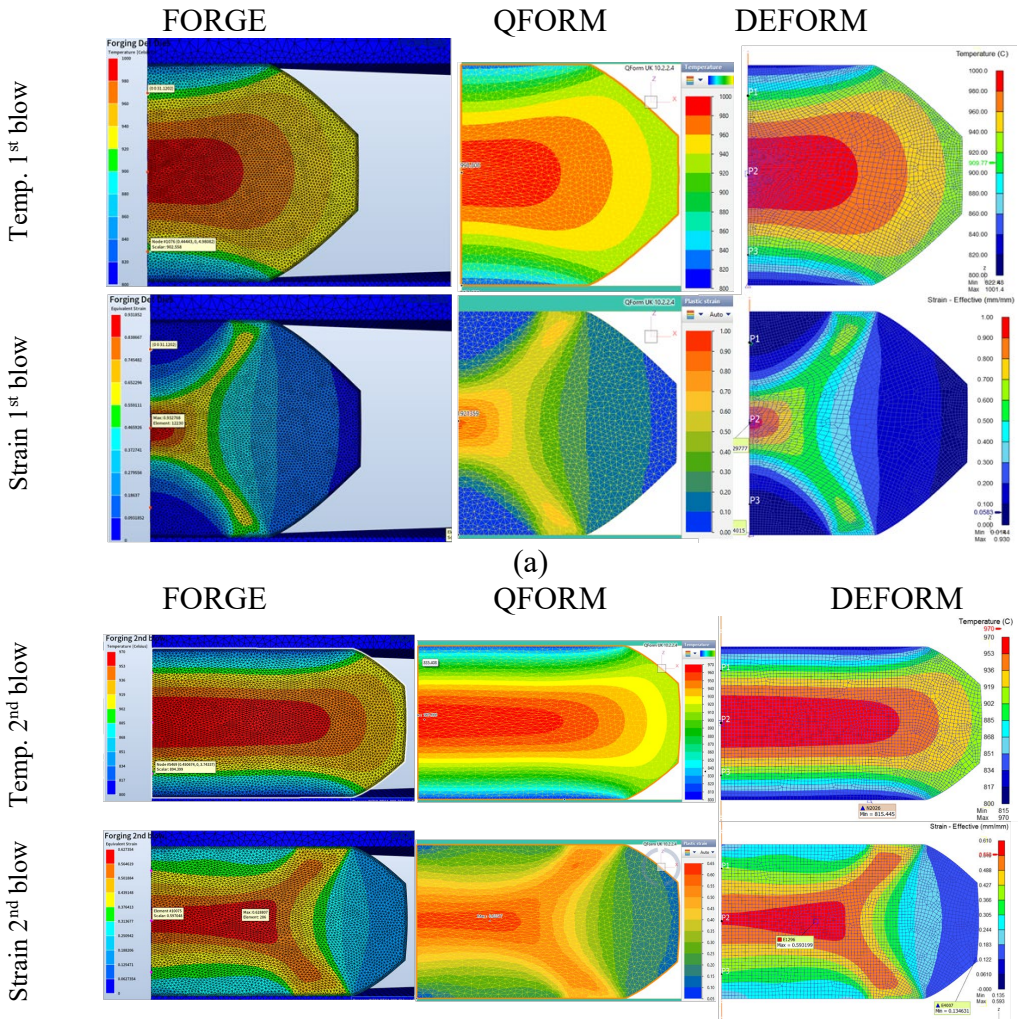


Fig. 3. Comparison between the temperature fields and strain fields in all utilised FE packages: (a) temperature and strain fields straight after the first blow; (b) temperature and strain fields straight after the second blow.

Methodology of microstructure study.

Processed double truncated cones were cut in half and prepared for optic microscopy and EBSD analysis. The boundaries of recristallised area in the billet's cross-sections were defined using optical microscopy with the accuracy 5%. EBSD maps were obtained in reference points of each forged billet, namely central area and area where recrystallisation was initiated.

Benchmarked mean-field microstructure evolution models.

Standard mean-field models available in commercial FEA software Deform®, Forge® and QForm® were used for microstructural prediction during multistage forging process. All the models were used "as is" with the default set of parameters for IN718 available in the material libraries of each software. For the detailed description of the models and values of the model parameters please refer to description of the models [9-11]. The initial grain size of the material was set to 17μm and temperature limit of 950°C was used for the initiation of microstructure evolution in all the models. All the microstructure evolution mechanisms available in the models were set as active: DRX, MDRX, SRX and static GG. It was also assumed that material was fully recovered (no retained strain) after the re-heating operation before the second blow.

Results and Discussion

The outputs of microstructure evolution models for total recrystallized volume fraction are compared with the experimental observations in Table 1 (for the scale from 0% to 100%) and Table 2 (for the scale from 0% to 5%). The scale used in Table 2 allows to compare the boundaries of the recrystallised area more precisely. Note that total RX volume fraction presented in the tables below shown as a summary of contribution from all three various types of recrystallisation: DRX, MDRX and SRX. According to [12, 13] all three mechanisms were in play during the selected processing route.

As it follows from the Table 1 and Table 2, none of the standard JMAK-type models embedded in the commercially available FEA software and used with the default settings have demonstrated accurate predictions. Though the boundaries of recrystallised area for DTC#1 was predicted by FORGE and QFORM models quite well (see the first row of Table 2), the volume fraction of the recrystallised material was predicted wrongly. It is assumed that this type of models should predict microstructure evolution for one single forging operation correctly on the condition that model is calibrated in proper way. Such a big disagreement with the experiment indicates that default settings of the models are not suitable for this case and model should be re-calibrated. In this particular case, large value of total RX vol% predicted by models is caused by overestimated MDRX vol%, while prediction of DRX is much closer to the experimental value of 26%.

Table 1. Comparison of microstructure evolution models outputs for RX vol% versus experiment in scale 0 to 100%.

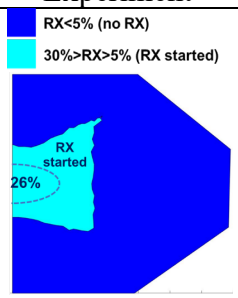
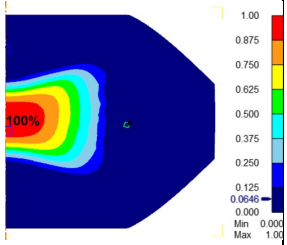
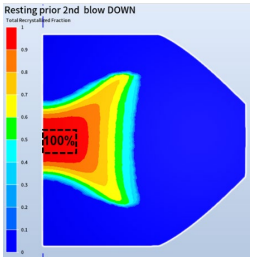
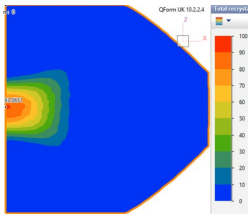
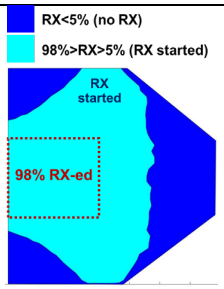
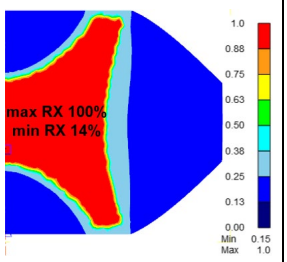
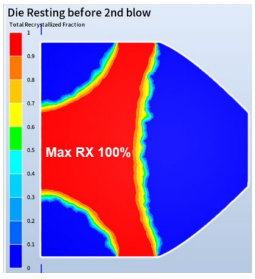
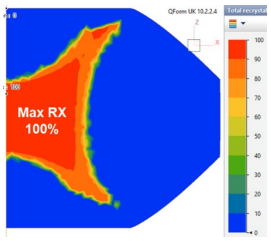
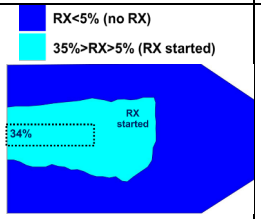
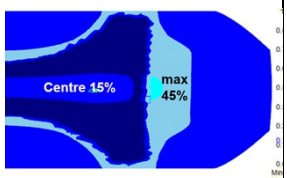
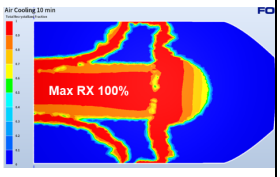
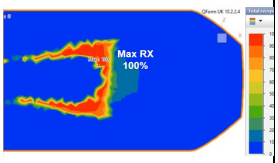
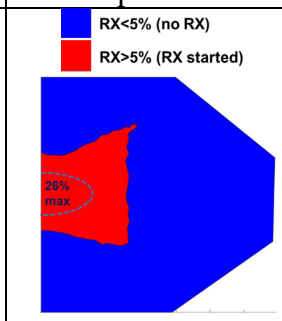
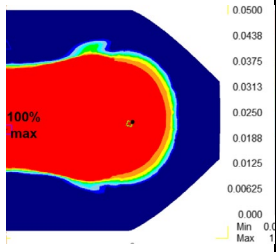
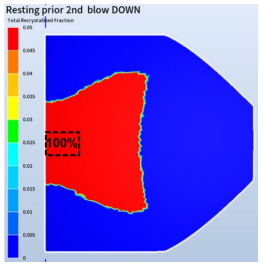
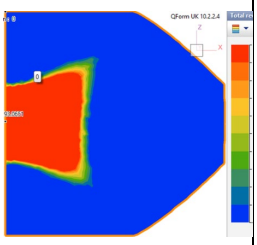
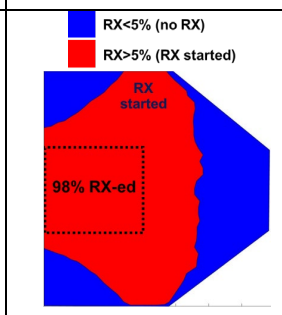
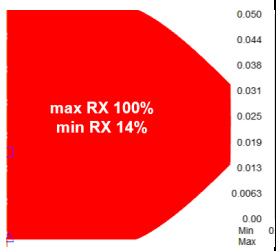
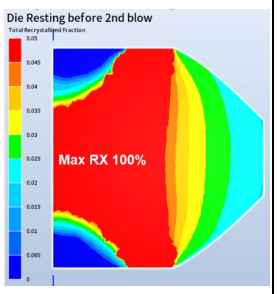
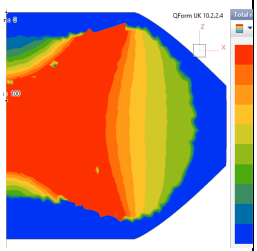
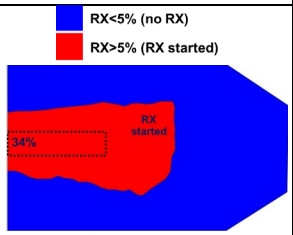
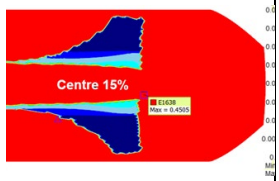
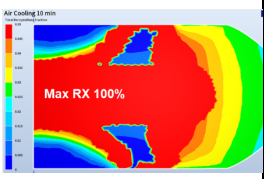
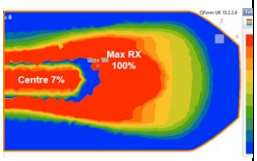
Route	Total RX volume fraction in scale 0 to 100%			
	Experiment	DEFORM	FORGE	QFORM
DTC#1: 1st blow + AC				
DTC#2: 1st blow + reheating + AC				
DTC#3: 1st blow + reheating + 2nd blow + AC				

Table 2. Comparison of microstructure evolution models outputs for RX vol% versus experiment in scale 0 to 5%.

Route	Total RX volume fraction in scale 0 to 5% (for indicating onset of RX)			
	Experiment	DEFORM	FORGE	QFORM
DTC#1: 1st blow + AC				
DTC#2: 1st blow + reheating + AC				
DTC#3: 1st blow + reheating + 2nd blow + AC				

For the more complex processing routes used for DTC#2 and DTC#3 the prediction of the models became worse. For these more complicated cases the problem of transferring material and its properties between in-line manufacturing operations prevails.

It is important to note that all used standard JMAK-type models were de-coupled models. This fact limits their application for multiple-blows simulations as changes in the microstructural state are not connected with the model of the material (i.e. flow stress data) directly. To compensate this essential drawbacks and transfer key material parameters between the operations, the software developers use number of assumptions and patches [5]. However, even in their current de-coupled form these models can be improved significantly to make their prediction more accurate. Several reasons causing wrong predictions, as well as suggestions regarding model improvements are given below.

Temperature limits for activation various microstructure evolution mechanisms.

The value of temperature limit is important as it largely defines (together with accumulated plastic strain) recrystallised area and RX volume fraction. JMAK-type models implemented into FE softwares have single temperature limit for all microstructure evolution mechanisms and do not allow to set various temperature limit for DRX, SRX and MDRX separately. Probably it is not correct approach, because according to the experimental observations grain growth activation requires higher temperature threshold than nucleation process. Ideally temperature limits should

be set separately for each microstructure evolution mechanism as a function of temperature and some internal state variables which characterise microstructure state, e.g. dislocation density, grain size, etc. Also, it could be beneficial do not use "hard cut-off" by temperature with some particular temperature value, but have temperature limit as function of some process parameters instead, e.g. strain and strain rate. It will better fit "physics" of the recrystallisation, as bigger strain and strain rate results in bigger internal energy and, correspondently, requires lower temperature limit for the activation of RX mechanism.

The way of representing multiple rounds of recrystallisation by the model.

As it can be seen from the third row of Table 1 and Table 2, the prediction for the RX vol% after the second blow looking strange. The reason is that in all the benchmarked models the RX vol% during the second blow is overlapped with the RX vol% fields remained after the first blow. There is an assumption in the models that if RX vol% getting close to 100% during the recrystallisation, then all recrystallised fractions are reset back to 0% considering a fully non-recrystallised new microstructure obtained. This solution is inconvenient for representation of multiple rounds of RX, because it is not possible to distinguish between the fully recrystallised material during the previous operations and material which was never recrystallised, for example, dead-zones. It is also hard to distinguish RX during the second blow and compare it with experimentally defined RX-ed area after the second blow. It would be beneficial to reconsider the way of representation of RX vol% in such a way to make the previous rounds of recrystallisation visible to the user.

Transferring non-uniform microstructure to the re-heating operation.

As can be seen from Fig. 4 (a, b, c) DTC#1 has various types of non-uniform microstructure after the first blow. The microstructures differ by grain size, grain shapes and dislocation densities. These three types of microstructure later re-heated in the furnace for the second blow. During the re-heating SRX occurs. It is assumed that SRX will have different kinetics for all these 3 cases. The problem is that SRX models in their current form cannot take into account such peculiarities of microstructure. Current equations for SRX volume fraction and grain size have average grain size parameter in some power (material constant) [3-5], but it cannot be used to describe bimodal microstructure or partially recrystallised microstructure adequately. The possible solution of this problem within current architecture of the SRX model is utilising both average recrystallised and non-recrystallised grain sizes in the SRX model equations. Such modifications of the model was already tested by developers of FORGE FEA software [5].

Transferring accumulated plastic strain and microstructure to the second blow.

This problem is partially overlapped with the previous one as microstructures after the reheating operation can be also quite different, see Fig. (d, e, f), so it is assumed that it should influence kinetics of DRX during the second blow. This part of the problem could be partially sorted out in the similar manner as suggested above.

The second part of the problem is figuring out what ratio of the accumulated plastic strain should go for the second blow after the re-heating operation. Obviously material which was not SRX-ed in full and have remained strain will be DRX-ed in different way and its rheology also will be different. There is a patch to solve this problem in the FE softwares, namely, "retaining strain" which is calculated as inverse ratio to "RX total vol%":

$$\varepsilon_i = (1 - X_{rex})\varepsilon_{i-1} \quad (1)$$

Alternatively "retaining strain" can be calculated as a function of parameters which define recovery, e.g. time and temperature. This way is better as it covers the cases when SRX did not happen, but deformed grains were recovered during the re-heating operation.

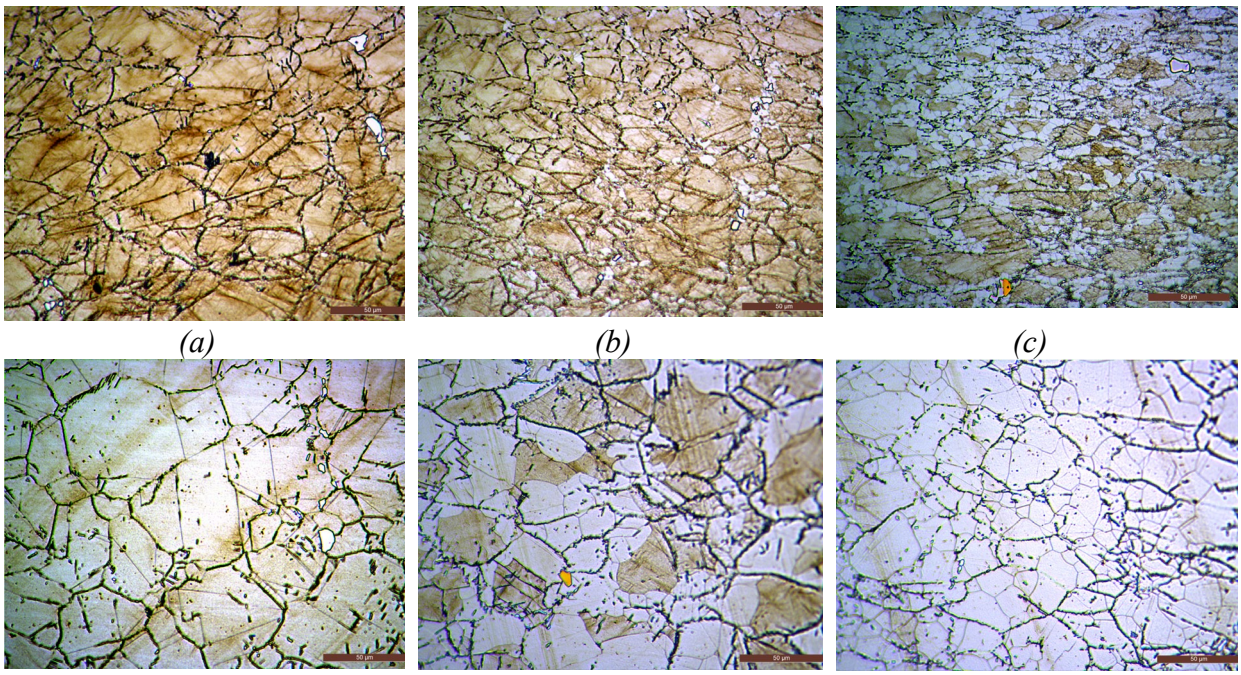


Fig. 4. Optic microscopy images obtained at some reference points of the processed billets: (a) DTC#1: non-recrystallised location with heavily-deformed initial grains; (b) DTC#1: necklace microstructure - location where DRX initiated; (c) DTC#1: heavily DRX-ed location in the centre of the billet; (d) DTC#2: location at the vicinity of bottom dead-zone where SRX was not started; (e) DTC#2: partially SRX-ed location; (f) DTC#2: fully SRX-ed location in the centre of the billet.

However, due to de-coupled nature of the investigated models, Eq. 1 influences only the critical strain needed for the following round of RX, but it is not coupled with the model of the material, i.e. flow stress data, so the metal flow on the second blow does not take into account changes in microstructure. The only solution here is employing coupled model where some microstructure parameters are coupled with the material model and flow stress data is recalculated on each step of simulation taking into account microstructure changes. To this end microstructure evolution model should be re-written in incremental form.

Taking into account contribution of other mechanisms of microstructure evolution.

Apart from DRX, MDRX, SRX and static GG described in the investigated models, there are also some microstructure evolution mechanisms which were not taken into account. Among them are recovery (RC), twinning (TW) and continues dynamic recrystallisation (CDRX). All listed mechanisms were also in play during the thermo-mechanical processing used for this study and they influenced average grain size of non-recrystallised grains as well as level of internal energy locked in the material. These mechanisms cannot be taken into account within the decoupled approach and can be described only by coupled models.

To sum up, there are number of ways for improving standard JMAK-type models in their current de-coupled state without changing their mathematical formulation. In brief, the proposed improvements include: setting temperature limits for each microstructure evolution mechanism; calculating retained strain (strain transferred to the next blow); better way of representing multiple rounds of RX; taking into account both average recrystallised and non-recrystallised grain sizes.

Each FEA package tested in this study has strong points in its own mean-field model which are useful for microstructure prediction during multi-pass forging processes. For instance, DEFORM Avrami model has inbuilt capability to use multiple sets of model parameters for various

temperatures and strain rates, which potentially allows to cover wide range of process parameters; it also uses retaining strain concept which allows to some extent compensate drawbacks of the de-coupled model. FORGE JMAK-model has possibility to use tabulated parameters, also software developers suggested the way of modelling heterogeneous microstructures during multi-pass processes [5]. QFORM JMAK-model also has inbuilt equation for calculating retaining strain and option for cyclic recrystallisation inbuilt in GUI, also due to convenient Lua programming the model can be easily modified by user. All these features listed above give user good possibilities to adjust the model in each software for the particular multi-stage forging processes.

Mean-field models in itself operate with some representative volume of the material and averaged characteristics of the process. This approach can work well if the representative volume is uniform. However, the problems arise when material is non-uniform, for instance in the case of partially recrystallised microstructure, or in the case when there are grains with big difference in properties in the same representative volume. Although it is challenging, these more complicated cases could also be described by mean-field approach providing that some new alternative equivalent parameters introduced into the model for representing current state of the microstructure.

Full-field modelling at the meso-scale level, e.g. by using DIGIMU (Transvalor) software, could be useful tool for getting better understanding of microstructure evolution in the cases when prediction of mean-filed models is not accurate, e.g. microstructures with non-uniform distribution of grains by size; or with non-uniform distribution of dislocation density (stored energy) among the grains. It can help to find out what new equivalent parameters could be introduced into the mean-field models to get robust prediction for these cases. Such alternative equivalent parameters potentially could be equivalent dislocation density, equivalent work of plastic deformation, equivalent grain sizes, aspect ratio of grains, etc.

Summary

1) All JMAK-type models considered in the paper have a big potential for predicting microstructure evolution during multiple-blow forging process. The peculiarities of this processes should be taken into account and models should be adjusted to it correspondingly. Developers of FEA packages improving their models continuously and microstructure evolution models embedded into DEFORME, FORGE and QFORM in their current state have plenty of opportunities for the adjustments and improving their predictions.
2) For the cases of bimodal microstructure or necklace microstructure where average grain size cannot represent the microstructure correctly, outputs of the model should be considered critically. For instance, to estimate the scatter in the results, upper and lower-bound methods can be used with two various average grain sizes.
3) Mean-field models should be calibrated for the particular (narrow) range of process parameters (strain rate and temperature range) as they cannot predict microstructure evolution within the wide range of process parameters correctly. In addition to this, model of the material, i.e. flow stress data, should be calibrated for the particular range of process parameters as well.
4) Coupled mean-field models are required for predicting microstructure evolution during multistage thermo-mechanical processing in more accurate way. It is implied that such type of models should have in the model of the material some parameter which characterises current state of the microstructure, for instance, dislocation density or work of plastic deformation. Full-field modelling using DIGIMU Transvalor software can be effective tool for getting better understanding of active microstructure evolution mechanisms in play and identification of parameters for coupled mean-field models.

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