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RECENT TRENDS IN METAL FORMING: FROM PROCESS SIMULATION AND MICROSTRUCTURE CONTROL IN CLASSICAL FORMING PROCESSES TO HYBRID COMBINATIONS BETWEEN FORMING AND ADDITIVE MANUFACTURING

This paper describes some recent trends in metal forming such as isothermal forging of titanium aluminides and process combinations between metal forming and additive manufacturing. These trends rely on accurate process and material models for process design. Process and material models must hence be able to track the microstructure evolution in complex materials such as titanium aluminides as well as predict the microstructure evolution along process histories with multiple deformation and/or heat input steps. In models for such processes, JMAK-type kinetics for and phase transformation are still common. For processes involving deformation and heat, the accuracy, consistency and limits of JMAK-type models are discussed. It is shown that the consistency of DRX models as well as the stability of model predictions in multi-stage processes require further attention.

1. INTRODUCTION

The field of metal forming comprises all manufacturing processes in which the shape, surface and material properties of a solid (metallic) body are changed in a well-defined, targeted way, while its mass and topological connectedness remain unchanged. The importance of metal forming processes can be assessed from the share of the annual production quantity processed by metal forming. For steel and aluminium, which are the dominant metallic construction materials, figures of the global flow have been collected by Allwood et al. and Cullen et al. In the year 2008, roughly 90% of the steel produced worldwide annually (about 1.4 billion tons) [1] and more than 2/3 of the world's annual production of wrought aluminum [2] were processed to semi-finished products using forming processes, which highlights the importance of metal forming. Metal forming is currently evolving in different directions. Some current challenges and trends can be summarized as follows:

1. **Forming of alloys with low workability.** Some new products, such as turbine blades made of intermetallic titanium aluminides, require isothermal forging conditions with

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very low forming speeds, setting high demands on tool materials and process control. These processes have not yet reached a state of highest productivity and demand for research into novel alloys and processing concepts, die materials as well as process models.

2. **Scale-independent individualization.** Metal forming processes for finished products often use dedicated tooling which makes these processes only economically viable for mass production. The increasing demand for individualized products or small production series calls for solutions with reduced tooling effort. Flexible processes such as additive manufacturing (AM) are advantageous for these cases but currently still possess very limited build rates. This opens the way to explore process combinations between metal forming and AM processes.
3. **Cyber-Physical Systems.** Trends in information technology and automation, currently summarized under the buzz word 'industry 4.0', possess a great influence on metal forming value chains. Both process chains for the production of semi-finished products (e.g. steel plants with integrated rolling mills) as well as production chains for finished products are more and more perceived as cyber-physical systems which offer new opportunities for operation such as machine-to-machine communication and self-optimization. In rolling mills, the vision of schedule-free rolling could be realized, where the rolling mill would develop the rolling schedule during the first trial passes itself.

The trends discussed above are only a selection of current streams which also comprise flexible forming processes, development of new machine tools and sensors etc. They have been selected here since they set demands for process and materials modeling, i.e. the processes cannot be run based on experience but require substantial efforts in process and materials modeling. Being a field with long-standing tradition, process design and property control in metal forming used to be based on experience in the past. Nowadays, tools for numerical simulation of metal forming processes and process chains are extensively used in the metal forming industry. With the advent of powerful computers, parallel processing and the development of robust numerical methods, it seems that not only individual metal forming processes can be simulated but that the entire fabrication chain of a product could be simulated and designed 'from the cradle to the grave' on the computer. This scenario is currently being pursued in the emerging field of 'integrated computational materials engineering' (ICME), which is an approach for developing products, materials and the corresponding manufacturing processes by coupling of simulations across scales and along the manufacturing process chain [3],[4],[5]. Due to the ever increasing complexity of metal forming processes and process chains, process and material models can be seen as enablers allowing for the prediction and control of final product properties.

If we restrict our attention to processes running at elevated temperature, the prediction of process variables and product properties in the processes described above requires models for microstructure evolution by recrystallization and phase transformations, i.e. processes typically described using transformation kinetics. For these processes, two kinds of models are available: (i) models for the kinetics for processes of nucleation and growth in terms of volume fractions and (ii) models with spatial resolution of the microstructure. For comprehensive ICME model chains and on-line process control, only the first model type seems to offer viable run times. However, the simplified treatment of transformation processes may cause problems such as model inconsistencies and instabilities that are not well researched up to date.

This paper gives an overview of current trends and pending problems in the field of metal forming, such as the operation of process chains, forging of intermetallics and hybrid combinations between forming and additive manufacturing. These processes are described in more detail in section 2. They set demands for process and microstructure modeling. The subsequent section 3 is dedicated to discussing the status and necessary developments that result from the demands of current process developments.

2. TRENDS IN METAL FORMING AND DEMANDS FOR PROCESS MODELING

2.1. FORGING OF INTERMETALLIC TIAL

In recent years β-solidifying TiAl alloys, such as the TNB alloys, have been intensively studied due to their excellent mechanical properties, i.e. low density, high strength at elevated temperatures and oxidation/corrosion resistance [6]. As a consequence of these promising mechanical properties these alloys could replace other materials, for example heavier steels or nickel base alloys, in applications for turbocharger turbine wheels

Fig. 1. (Left) 1700 ton isothermal forging press. (Top right) Lower die with part in vacuum chamber. (Bottom right) TiAl workpiece and forged turbine blade

in automotive engines or for turbine blades in the low pressure turbine of aircraft engines [7]. Due to the low workability of TiAl alloys at room temperature, they are usually processed to components by hot working. The chair of mechanical design and manufacture at BTU Cottbus operates a forging press dedicated to isothermal forging of TiAl alloys, Fig. 1. As die material, special Molybdenum alloys are used which offer sufficient strength at hot forging conditions but are highly susceptible to oxidation. For this reason, the press is equipped with a vacuum chamber to allow for forging in the absence of oxygen.

In order to optimize the hot forging processes for material flow and minimal die load, predicting the flow behaviour of TiAl alloys at hot deformation conditions thus has a great importance for the design of hot working processes. Recently, increasing attention has been paid to the workability of intermetallic titanium aluminides, due to the fact that deformation without defects and cracks is possible only in a very narrow stability region under isothermal forging conditions. As observed by Kim et. al. [8] and Imayev et. al. [9] the intrinsic workability of intermetallic titanium aluminides alloys depends upon the initial microstructure as determined by the alloy chemistry and prior processing history, and also on its response to the temperature, strain rate and strain during processing. Several authors have put forward constitutive equations for TiAl-based alloys to describe the hot deformation behaviour, i.e. the evolution of flow stress and grain size, see e.g. Cheng et al. [10]. However, as will be shown in section 3, modeling of micrustructure evolution in TiAl alloys still poses substantial challenges due to the intricate phase composition and other unsolved problems in the description of dynamic recrystallization phenomena.

2.2. HYBRID PROCESS COMBINATIONS OF METAL FORMING AND AM

Both in sheet and bulk metal forming, mass production processes are common which profit from the so-called economies of scale and cannot easily be scaled down to small-scale and individualized production. The cost structure of hot forging and additive manufacturing processes are fundamentally different. Hot forging processes use expensive dies, whose cost increase rapidly with part size and part complexity. Further downsides are that large forgings often are milled to the final product shape with a large volume of wasted material since the microstructure is only appropriate for the final application in a small portion of the workpiece. In the aerospace industry, hot forging is used to produce both structural parts as well as e.g. turbine blades. Hot forging thus offers various possibilities for the design of value chains in combination with additive manufacturing processes, e.g. the production of variants based on a generic shape made by forging to reduce material waste by milling.

A demonstrator of such a process combination has been realized at BTU in collaboration between the chair for mechanical design and manufacture and the chair for welding and joining, Fig. 2. By the combination of processes, the part can be produced with ~30-40% of the time, cost and energy involved in 3d printing of the entire part. Hence, prototpyes or small series seem feasible.

A challenge in using hot-forged parts for subsequent additive manufacturing is, in particular, that the microstructure, the residual stresses and the surface conditions of the semi-finished product will be influenced by the AM process, such that suitable process

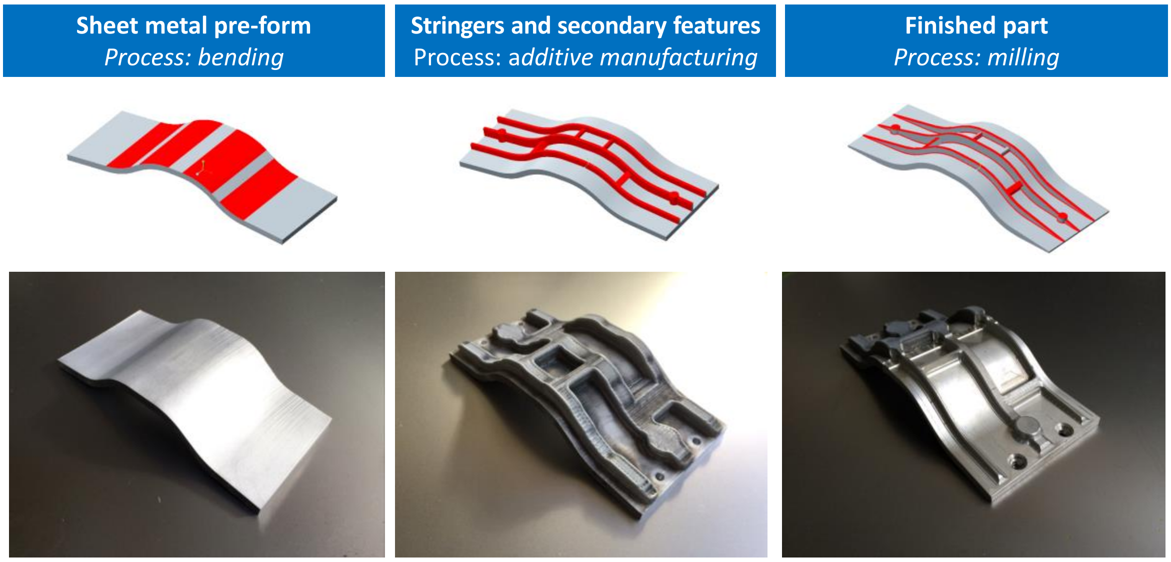

Fig. 2. Process combination of pre-form manufacturing and laser powder decomposition welding

parameters (laser power, feed rate,…) need to be identified. The powder deposition process imposes a cyclic thermal load onto the preformed base material, which triggers both the creation of residual stresses as well as a complex microstructure evolution. In order to predict the microstructure evolution in AM and process combinations of AM and forming, models of substantial complexity need to be built.

2.3. MULTI-STAGE HOT ROLLING PROCESSES

The process chains of strip and plate rolling have been optimized over decades and work at highest productivity. However, their resource-saving, energy-efficient and cost-minimizing design and operation is a continuous challenge for the steel and aluminum industry. For the design and control of rolling processes, various types of models have been developed. A main challenge, however, lies in the fact that rolling processes are multi-stage processes. For the computation of the process chain in terms of 'schedule-free rolling' and property prediction in the context of 'industry 4.0', stability of the model predictions poses a major problem.

3. PROCESS AND MATERIALS MODELING

3.1. ICME AND THE VIABILITY OF MICROSTRUCTURE SIMULATION ALONG PROCESS CHAINS

The process examples in section 2 have in common that microstructure evolution must be understood under highly demanding conditions, either in alloys with a complex microstructure (TiAl) or in processes with a complex thermomechanical history (cyclic heat input, multiple deformation or multi-physics). Recent success stories [11],[12],[13] suggest that ICME provides a versatile, ready-to-use solution for the development of new materials, products and the design of the corresponding manufacturing process chains. It seems that by integrating models across length scales and along the process chain, product properties could be predicted and optimized. However, there are several factors that jeopardize the viability of ICME simulations in general and also for specific cases in which only a part of a process chain is considered. In this section, the viability of simulations of individual processes and process chains that involve multi-stage hot working operations are analysed, with a focus on modeling of recrystallization events.

3.2. RECRYSTALLIZATION AND HOT WORKING

In metal forming and heat treatment operations, three types of recrystallization processes are distinguished, static, dynamic and meta-dynamic recrystallization. Static recrystallization (SRX) refers to recrystallization occurring in interpass times after a previous hot working operation. Unlike primary recrystallization of a cold worked material that is heated above recrystallization temperature, the material has undergone deformation at elevated temperature already. Recrystallization occurring during plastic deformation is called dynamic recrystallization (DRX) and leads to the formation of new, relatively defect-free grains, which cover the former high angle grain boundaries of the material. DRX nuclei that are created during hot working may grow very rapidly even if the input of deformation energy stalls. This kind of recrystallization shows no incubation period and is referred to as post- or meta-dynamic recrystallization (MDRX). Phase transformations may also occur during or after hot working, as well as during heat input and cooling. They will not be considered further here.

Recrystallization and phase transformations are nucleation and growth processes, whose kinetics are typically described using the so-called JMAK equation due to Kolmogorov [14], Johnson and Mehl [15] and Avrami [16] who studied the kinetics of phase transformations. $X(t)$ is the recrystallized volume fraction, k is the Avrami constant and q the Avrami exponent. JMAK kinetics are based on four main assumptions [17]:

1. The transformation proceeds by nucleation and growth.
2. The nucleation rate   and the growth rate v are constant in time.
3. The nucleation sites are randomly distributed.
4. The nuclei grow isotropically until impingement.

The Avrami exponent typically assumes a value of the order of 1 for MDRX, 2 for SRX (cf. [17]) and 3 for DRX (see Jonas et al. [18]). JMAK kinetics have been used over centuries and modified in various ways. However, the behaviour of JMAK kinetics in cases where multiple hot deformation or heat input stages occur have not been analysed in detail. Also, the validity of JMAK kinetics in cases where deformation and recrystallization overlap need further attention. In the subsequent sections, current challenges posed by using JMAK kinetics are discussed.

3.3. CHALLENGES IN DRX MODELING BY JMAK KINETICS

Various models have been derived for DRX in the past (see e.g. Bariani et al. [19] for a review), most of which rely on JMAK kinetics, some use generalized transformation kinetics. Poliak and Jonas [20] showed that the onset of DRX corresponds to an inflection point in the course of the strain hardening rate θ as a function of stress σ, i.e. the critical stress is a root of the second derivative of the strain hardening rate with respect to stress:

$$\frac{\partial}{\partial\sigma}\left(-\frac{\partial\theta}{\partial\sigma}\right)=0 \tag{1}$$

Using JMAK kinetics for DRX causes two pending problems, (i) inconsistency of JMAK kinetics with the Poliak-Jonas-criterion and (ii) absense of inflection point in practical experiments despite occurrence of DRX.

Model inconsistency

Since the criterion by Poliak and Jonas was derived from principles of irreversible thermodynamics and has been confirmed with experimental data, a flow stress model that takes DRX into account should adhere to this criterion, not only to be consistent with experimental data but also to make sure that the model is sound in a thermodynamical sense. Hence, the flow stress σ_y should be three times continuously differentiable with respect to ε. It was recently shown by Bambach [21] that the Poliak-Jonas criterion requires that the Avrami exponent exceeds a value of 3 in a DRX model for the model to be consistent with the criterion. Hence, generalized transformation kinetics were considered to find the root

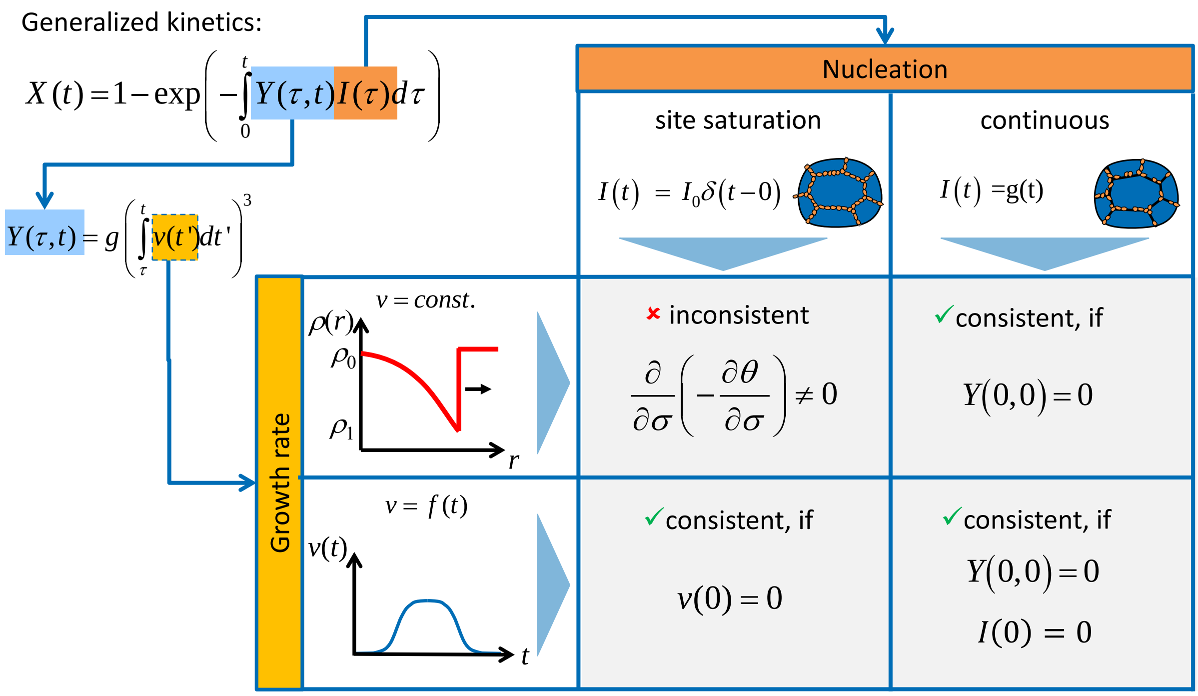

Fig. 3. Conditions on nucleation and growth rate of generalized transformation kinetics stemming from Eq. (1)

cause of the inconsistency and to alleviate them. Let $I(t)$ denote the rate of nucleation of recrystallized grains and the function $Y(\tau,t)$ represent the volume at time t of particles nucleated at time τ. Then, the Poliak-Jonas criterion leads to the conditions summarized in Fig. 3 on the functions I and Y.

For site saturation in the sense that all nucleation sites are occupied at the onset of DRX, consistency with the second derivative criterion requires a growth exponent > 3 (i.e. growth in more than 3 dimensions) or a nucleus size of zero together with a vanishing grain boundary velocity at the onset of DRX, i.e., a model for the acceleration of an initially motionless grain boundary. By introducing a time-dependent growth velocity with $v(0)=0$ (where $t=0$ marks the critical point), a mathematically consistent formulation may be obtained for site saturation. From a physical point of view, however, it does not seem consistent to assume that nuclei with zero size and an initially motionless grain boundary are present at the start of DRX, since these could not be considered real 'nuclei'. For continuous nucleation and growth, less restrictive conditions need to be fulfilled. One possibility is to assume that the nucleation rate and nucleus size are zero at the start of DRX. In this case, only 1D growth requires the grain boundary velocity to be zero at the onset of DRX. This leaves enough freedom for a consistent formulation. Another possibility is to assume that nuclei of zero size are present, which imposes the condition of a zero growth velocity in all dimensions ≤2. Again, the assumption of a nucleus size of zero contradicts the notion of a nucleus but follows mathematically. From a physical point of view it thus seems expedient to include nucleation into the DRX model, i.e. to account for the generation of nuclei with mobile high-angle grain boundaries after the critical conditions have been reached. This will automatically lead to a zero growth velocity at the critical point and avoid the inconsistency with the second derivative criterion altogether.

Flow curves an DRX of TiAl

Fig. 4 shows the true stress-true strain curves of a Ti–44.5Al–6.25Nb–0.8Mo–0.1B alloy at the deformation temperature of 1250°C and different strain rates. All curves exhibit a sharp flow stress peak at a relatively low strain and a drop towards steady-state conditions.

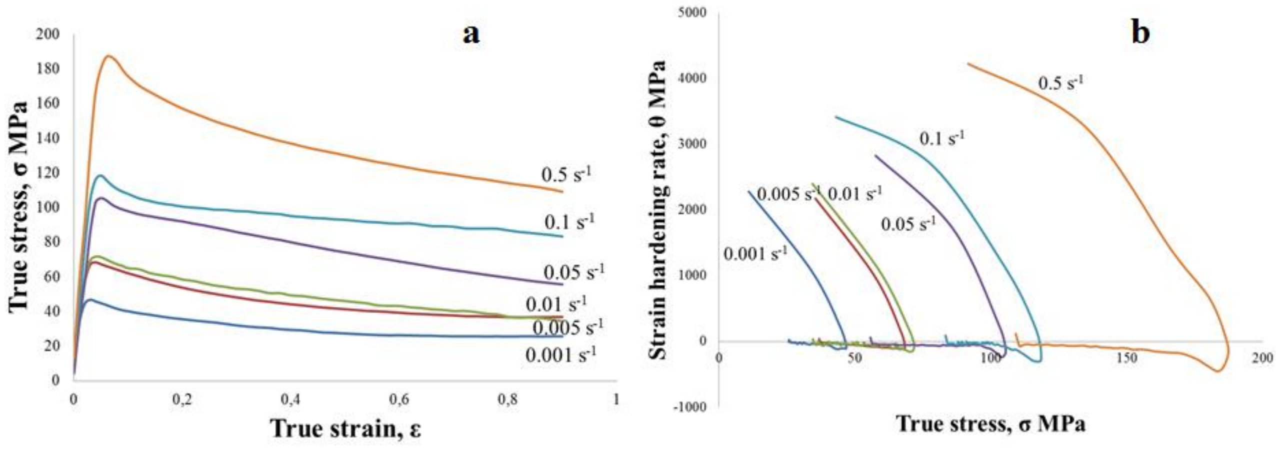

Fig. 4. Measured flow curves (a) and Mecking plots (b) for temperature of 1250°C applying different strain rates

Although the flow curves show differences to, e.g., austenitic steels, at first glance they show all common characteristics of single peak flow curves evoked by dynamic recrystallization. In fact, for γ-TiAl-based alloys with low stacking fault energy, it is commonly believed that DRX is responsible for the flow softening behaviour [22]. However, when the flow curves are converted into a so-called Mecking plot which consists of the strain hardening rate as a function of flow stress, peculiarities are observed. In the Kocks-Mecking plots shown in Fig. 4b the strain hardening rate is plotted as a function of stress. The Kocks-Mecking plots decrease monotonically initially ('concave down' curvature) but do not show a linear decrease of θ (stage III hardening) and exhibit neither a plateau (stage IV) nor an inflexion point for all strain rates. This is in strong contrast to ordinary single phase alloys such as austenitic steels, which show hardening stages III-V in Kocks-Mecking plots and inflection points corresponding to the onset of DRX.

3.4. CHALLENGES FOR MICROSTRUCTURE MODELING OVER MULTIPLE PROCESS STEPS

A typical model for multi-stage hot rolling was proposed by Beynon and Sellars [23]. To track the energy stored in the material, the concept of accumulated strain is used. Here, ε_i is the accumulated strain present after the *i*-th roll pass which can be seen as a (rough) measure of the stored energy of deformation that will trigger SRX. $\Delta\varepsilon_i$ is the strain increment imposed in pass *i*. The kinetics of static recrystallization are described by a JMAK equation. X_i is the fraction recrystallized after the time Δt_i, which is the interpass time between the passes *i* and *i*+1. T_i is the absolute temperature and d_i the grain size in μm in the *i*-th pass. The grain size after complete recrystallization is modeled by a typical rule of mixture. Also the expression for the evolution of grain size when partial recrystallization takes place is taken from Beynon and Sellars [23], Eq. (7). When recrystallization is complete, further grain growth can take place even in relatively short times between the passes according to Eq. (8) (Table 1).

Table 1. summarized the most important model equations

Quantity	equation	Quantity	equation
Accumulated strain	$\varepsilon_i = \Delta\varepsilon_i + \varepsilon_{r,i-1}$ (2) $\varepsilon_{r,i} = \varepsilon_i\left(1 - X_i\right)$ (3)	JMAK kinetics	$X_i = 1 - \exp\left(\ln\frac{1}{2}\left(\frac{\Delta t_i}{t_{50,i}}\right)^k\right)$ (5)
Grain size after SRX	$d_i^{SRX} = Cd_i^{C_1}\varepsilon_i^{C_2}$ (4)	Time for 50% SRX:	$t_{50,i} = Bd_i^{B_1}\varepsilon_i^{B_2}\exp\left(\frac{Q_{RX}}{RT_i}\right)$ (6)
Grain size for partial SRX	$d_{i+1} = X_i^{4/3}d_i^{SRX} + \left(1 - X_i\right)^2 d_i$ (7)		
Grain growth	$d_i = \left[\left(d_i^{SRX}\right)^{D1} + D\left(\Delta t_i - t_{95,i}\right)\exp\left(-\frac{Q_{GG}}{RT_i}\right)\right]^{1/D_1}$ (8)		

In order to analyse the stability of models such as the Beynon-Sellars model, it is worth mentioning that grain size evolution defined by these models is represented by a non-linear recurrence equation or fixed-point iteration if these models are given as closed-form equations and used over multiple passes. Equations of this type occur as models for discrete dynamical systems. In mathematics, a dynamical system is defined by a fixed rule which describes the evolution of a system state in time. If discrete times are considered, a discrete dynamical system described by an iteration scheme is obtained. This is exactly the case for the models for microstructure evolution in hot working, in which the grain size d_{i+1} in a deformation pass i+1 is obtained from a function that depends on the grain size di in pass i:

$$d_{i+1} = f(d_i) \tag{9}$$

In general, a scalar fixed-point iteration is defined by some rule

$$x_{i+1} = f(x_i) \tag{10}$$

Stability of such an iteration scheme with a vector-valued function $\boldsymbol{f}(\boldsymbol{x})$ with $\boldsymbol{x} = (x_{1,} x_2, ..., x_n)^T$ depends on the Jacobian matrix

$$\boldsymbol{J} = \frac{\partial(f_1, \ldots, f_n)}{\partial(x_1, \ldots, x_n)} \tag{11}$$

If the norm of the Jacobian matrix fulfils $\|\boldsymbol{J}(\boldsymbol{x}^*)\| < 1$, the fixed point x* is asymptotically stable. Typically, the spectral radius of $\rho(\boldsymbol{J})$, i.e. its largest eigenvalue, is considered. It is then sufficient to check whether $\boldsymbol{J}$ is a convergent matrix, i.e. whether $\rho(\boldsymbol{J}) < 1$. Since the fixed point is typically not known a priori, it can be checked whether the spectral radius of $\boldsymbol{J}$ is bounded by some value L<1 on the entire domain on which the iteration scheme is defined. The evolution of grain size under hot working conditions is described by a recurrence equation which changes from pass to pass, depending on whether no, partial or full recrystallization occurs. To analyse the stability of the model equations put forward by Beynon and Sellars [23], the model equations have to be re-written as an iteration scheme to analyse the evolution of grain size when multiple cycles of deformation and interpass time are employed. The full iteration scheme or system of recurrence equations for the grain size evolution can be written as

$$\begin{pmatrix} \varepsilon_{i+1} \\ d_{i+1} \end{pmatrix} = \mathbf{f}\left(\varepsilon_i, d_i, \Delta\varepsilon_i, \Delta t_i, T, \boldsymbol{\beta}\right) \tag{12}$$

The vector $\boldsymbol{\beta}$=[k, B, B_1, B_2, C, C_1, C_2, D, D_1, Q_{RX}, Q_{GG}]^T comprises all input parameters of the closed-form model. As shown in [24] for a specific case of multi-stage hot rolling and model parameters, the iteration scheme has a stability map with instability regions, Fig. 5. The stability of the evolution of grain size is analysed using the Jacobian of the map between strain and grain size at stages i to i+1. It shows that the algebraic, semi-empirical equations for grain size evolution proposed by Beynon and Sellars [23] define an iteration scheme with an instability band: -Stable evolution of the predicted grain size is obtained for loading conditions that lead to full SRX after each deformation stage (or roll pass), Fig. 5.

This requires sufficiently large amounts of deformation per pass, long enough interpass times and slow grain growth following SRX. Partial recrystallization was found to

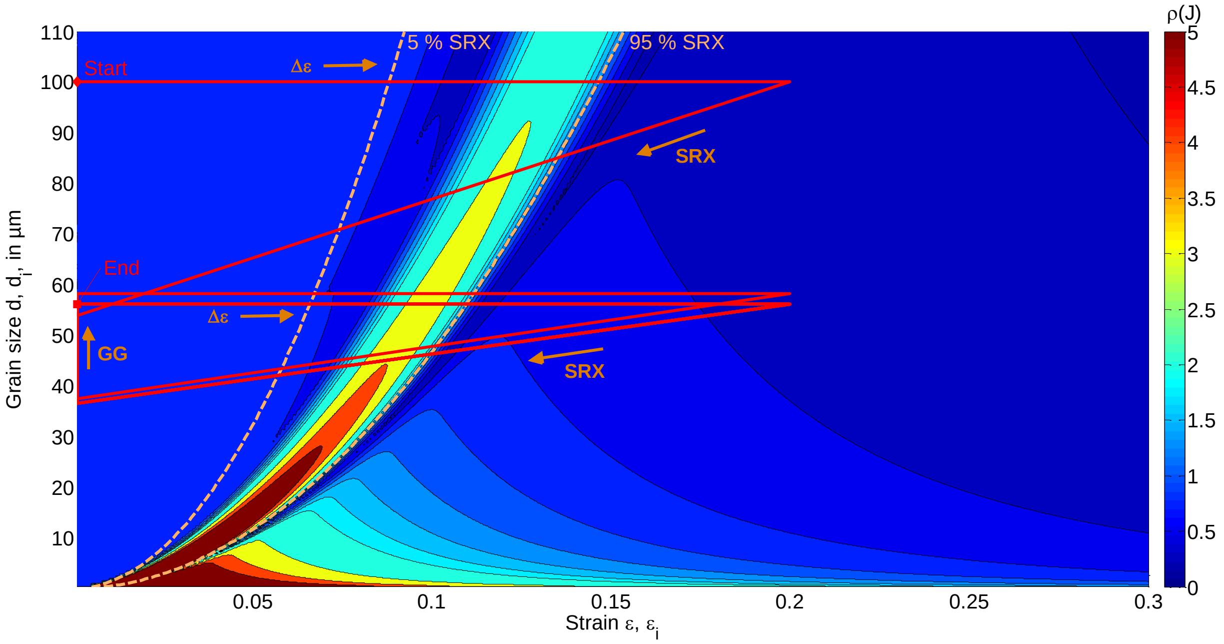

Fig. 5. Stable evolution due to full SRX and GG after each pass

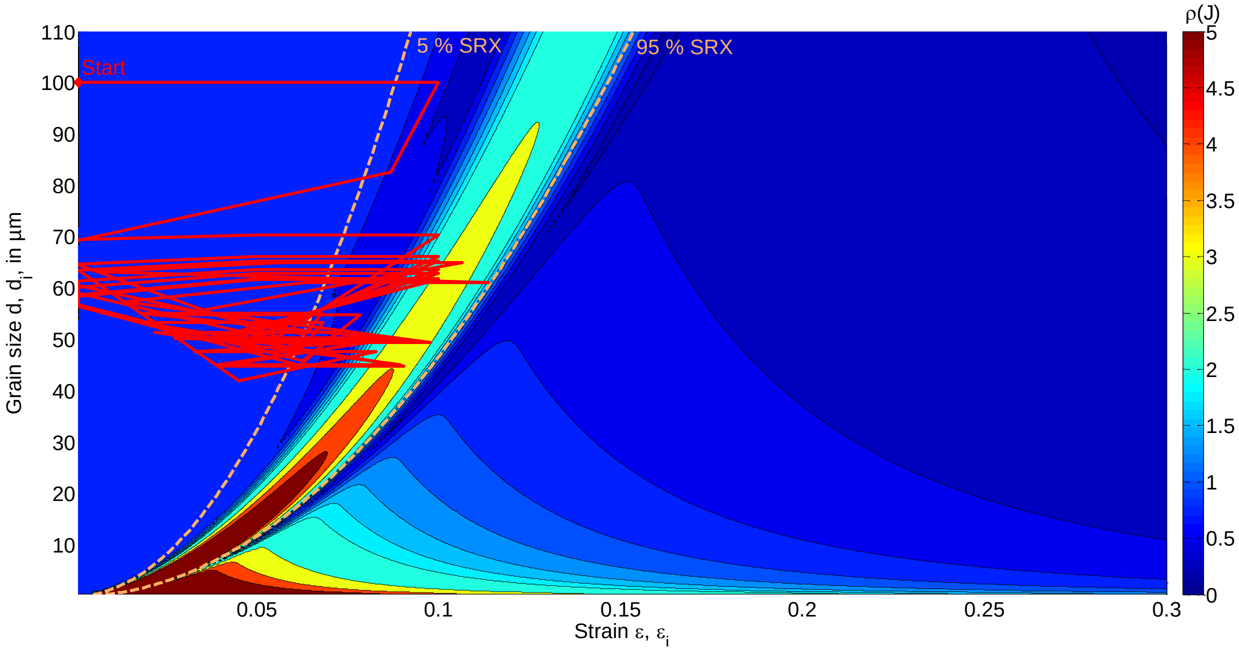

Fig. 6. Chaotic evolution of grain size and accumulated strain. Points in the unstable region are reached repeatedly. The trajectories of the process do not stabilize

be the cause for unstable grain size evolution under repeated thermo-mechanical loading, Fig. 6. Using models such as the Beynon-Sellars model, instabilities may occur in the model predictions when partial SRX occurs. An improvement is proposed in [24] which is based on tracking subsets of the microstructure.

4. SUMMARY AND CONCLUSIONS

This contribution reviews some current trends in metal forming such as forging of titanium aluminies and process combinations between forming and additive manufacturing. These processes pose high demands on microstructure models. The capabilities of 'standard' models for the simulation of transformation kinetics have been analysed. Inconsistencies in DRX models based on JMAK kinetics as well as instabilities in the prediction of multi-pass rolling processes were discussed. Future efforts for model development beyond JMAK kinetics are needed.

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