Optimization of the hot forging parameters for 4340 steel by processing maps

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Abstract

The study presents an investigation of hot deformation behaviour of 4340 steel and optimization of its processing parameters. The isothermal compression tests were performed at the temperatures ranging from 800 to 1200°C and at the strain rates in the range of 0.01–100 s⁻¹. The changes in the microstructure of 4340 steel, resulting from processing, have also been studied. The workability parameters and processing windows were determined, and the processing maps were correlated with microstructural changes. The optimal parameters for the forging process were also determined and then the numerical modelling and also the industrial tests of closed-die forging with flash were performed. Thermomechanical processing parameters as well as their influence on the investigated steel microstructure evolution, mechanisms of plastic deformation and its formability were discussed.

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1. Introduction

Understanding the material flow behaviour and relationship between microstructure evolution and workability is essential for proper thermomechanical processing. One of the important tools enabling the control and choice of the parameters of processing and, in consequence, ensuring appropriate mechanical properties and microstructure of the products, are processing maps elaborated on the basis of dynamic material model [1–3]. Many scientists have been developing the processing maps in order to control and optimize the parameters of hot working [4–7]. It has been confirmed that processing maps are quite accurate and effective thanks to a number of experimental tests for steel [8,9], aluminium alloys [10,11], titanium alloys [12] and nickel alloys [13]. The investigations conducted on high-strength and ultra-high-strength steels are more and more important for designing structural parts widely used in automotive and railway industry. The advanced modern ultra-high-strength steels require very strictly controlled microstructure and mechanical properties, especially when they are used for forgings. There are some research works concerning ultra-high-strength steels flow behavior, processing maps, the influence of different tempering temperatures on the susceptibility of the material to delayed fracture [14], determination of the energy spectra of hydrogen [15], or on theoretical and experimental investigations of the effect of static shear stresses on the cyclic fatigue behavior of those steels [16]. Sajadifar et al. have analysed the hot deformation behavior, the dynamic recovery and the dynamic recrystallization of 4340 steel in the strain rate range of 0.01–1 s⁻¹ [16]. Gong et al. [18] has investigated the as-forged 34CrNiMo6 steel by isothermal hot compression tests at various temperatures and
strain rates, and have established a physically based constitutive model and processing maps for hot working of the investigated material. The comparison between the present processing maps based on Prasad’s theory, coupling the material grain size and processing conditions for 300M steel, was performed by Sun et al. [19]. It was validated, that the present processing maps can be more accurately predicted, and the processing parameters could more accurately control the microstructure in the isothermally compressed 300M steel samples. Based on the theory of thermal activation dislocation and damage evolution during impact loading, the Zerilli-Armstrong constitutive model has been improved by Li et al. [20] to determine the impact compression mechanical properties of 25CrMo4 steel.

4340 steel, representing high-strength steels, finds many applications as forged parts used in the commercial and military aircrafts, and ground support systems, automotive systems, race cars, hydraulic tools and other machine tools. A comprehensive approach to the design of 4340 steel forging was presented in this paper. The hot deformation behaviour of the investigated steel at different temperature and strain rate ranges was analysed through the compression tests performed on thermomechanical simulator Gleeble 3800. Basing on the flow stress curves, the processing maps for the intermediate stages of deformation were elaborated and correlated with microstructures observed in the compressed samples. On that basis, the optimal hot deformation parameters were obtained and verified during numerical and physical modelling of closed-die forging process.

2. Experimental method

2.1. Experimental procedures

The chemical composition of 4340 steel was as follows: 0.40 C, 0.77 Mn, 1.72 Ni, 0.79 Cr, 0.22 Si, 0.20 Cu, 0.21 Mo, 0.008 P, 0.006 S, 0.004 V, 0.01 Sn, 0.027 Al, 0.0084 wt% N and the balance was Fe. The isothermal compression tests were conducted on Gleeble thermo-mechanical simulator. Specimens of 10 mm in diameter and 12 mm in height, machined from as-received 4340 steel rod, were heated at 2.5 °C/s up to the specified temperature. The tests were conducted at the temperatures of 1000, 1050, 1100, 1150, and 1200 °C, and at strain rates of 0.01, 0.1; 1; 10 and 100 s⁻¹. Samples were deformed in compression to a total true strain of 1. Microstructural investigations were performed on Axiovert 200MAT optical microscope, and on scanning microscope FEI VERSA 3D to analyse the microstructure of the obtained forging with BSE technique.

2.2. Processing maps

To investigate the influence of flow stress on the processing parameters, it is necessary to calculate the strain rate sensitivity parameter. Based on the flow stress-strain data obtained from hot compression tests, the parameter of the strain rate sensitivity of flow stress m at given strain of 1 and given deformation temperature could be calculated using formula (1) [21,22].

\[ m = \left( \frac{\partial \log \sigma}{\partial \log \varepsilon} \right)_{T, \varepsilon} \]  

(1)

where \( \sigma \) – the flow stress, MPa, \( \dot{\varepsilon} \) – the strain rate, s⁻¹, \( \varepsilon \) – constant true strain value, T – temperature, °C.

The strain rate sensitivity exponent (m) is a key parameter defining the relative (rather than absolute) partitioning of power between heat generation and microstructural changes [23].

The processing map is a clear representation of the response of a material, in terms of the microstructural mechanisms, to the imposed process parameters [24].

In the Dynamic Material Model (DMM), the workpiece is considered to be the dissipator of power. The unit power (P) absorbed by the material during plastic working is expressed in the following way [25,26]:

\[ P = \sigma \cdot \dot{\varepsilon} = G + \int_{0}^{\varepsilon} \sigma \cdot \dot{\varepsilon} \, d\varepsilon + \int_{0}^{\varepsilon} \dot{\varepsilon} \, d\sigma. \]  

(2)

where \( \sigma \) – the flow stress, MPa, \( \dot{\varepsilon} \) – the strain rate, s⁻¹, G – dissipator content, J – dissipator co-content.

J co-content (related to metallurgical processes or mechanisms proceeding dynamically with power dissipation, which are responsible for the intrinsic workability of the material) as the function of temperature and strain rate at constant strains is evaluated as [25,26]:

\[ J = \int_{0}^{\dot{\varepsilon}} \dot{\varepsilon} \, d\sigma = \frac{\sigma \cdot \dot{\varepsilon} \cdot m}{m + 1}. \]  

(3)

The efficiency of power dissipation – \( \eta \), as a measure of the material workability, is determined by the following formula [25]:

\[ \eta = \frac{J}{J_{\text{max}}} = \frac{2m}{m + 1}. \]  

(4)

This parameter could be plotted as a contour map in a frame of temperature and \log(strain rate) to obtain the power dissipation map [1]. The domains with the high value of the efficiency of the power dissipation parameter \( \eta \) correspond to the preferable processing condition [27–29], provided that those processing windows lie within the “safe” regions. However, the most efficient power dissipation mechanism is not necessarily the best for hot workability since it could be cracking or damage mechanism [25].

The criterion, which makes possible identifying the flow instability of the alloy during hot working, can be given as [25,29]:

\[ \xi_1 = \frac{\alpha \log(m/(m + 1))}{\alpha \log \dot{\varepsilon}} + m \leq 0. \]  

(5)

The variations of the instability parameter \( \xi_1 \) as the function of strain rate and temperature make it possible to plot instability maps. The regions on the map, where \( \xi_1 < 0 \), correspond to the microstructural instabilities (unstable flow) in the
material. The flow instability regions are generally manifested in the form of void formation, wedge cracking, intercrystalline cracking and other types of cracking processes and adiabatic shear bands, Lüder’s bands, kink bands and mechanical twinning [30–32]. The parameter $\xi$, is considered to be an important warning when designing the technology of plastic working for a specific material. Information from the processing map is useful for the designing and optimizing of bulk metal working processes like forging. The processing windows describe the most favourable forging parameters (the recommended optimum range of temperature and strain rate). Moreover, on the basis of strain rate range, it is possible to conduct the initial selection of a machine for closed-die forging process. The most advantageous parameters of the forging process determined by processing windows were applied for the numerical simulation and verification during forging trials performed in industrial conditions. The geometry of the model forging was selected in a way providing complex material flow within the volume of the processed material.

2.3. Numerical modelling

The numerical calculations of hot forging were performed using QForm 3D commercial software based on FEM. The shape and size of model forging selected for the purpose of numerical simulation is presented in Fig. 1. The gear wheel is a representative example of forgings having a complex shape with a circular contour in the die parting plane. The forging has a compact structure – a large cylindrical surface, a small height, and also gear teeth. It should have a surface free of laps and the assumed high level of the mechanical properties. It is forged with the narrow ranges of tolerance. One of the most important problems in forging process is large pressure per unit area (the area of a web, cavities and teeth), and also the areas of difficulty flow – gear teeth (the tooth point and the tooth root).

The User’s Defined Function (UDF) was programmed with the application of the LUA language and was applied for calculating the power dissipation efficiency parameter.

The forging process was conducted on a hydraulic screw press having the maximum capacity of 10 MN. Forging process was conducted in two operations: upset forging and forging in a finishing impression. The temperature of forging the billet was 1100°C. An initial temperature of the tools was assumed as 250°C. The time of the billet transportation to the die cavity was 2 s, and the time of billet cooling in a die cavity before forging was 1 s. In the calculations, a graphite-water emulsion was assumed as a lubricant, with a friction factor of 0.4. The billet dimensions were 50 mm in diameter and 62 mm in height.

2.4. Physical modelling – forging test

The billet was heated up in an electrical resistance furnace until the temperature of 1150°C. The drop forging dies were made of X37CrMoV5-1 tool steel. They were heated up to the temperature of 250°C. The obtained forging was cooled down very slowly in air (approximately 30° C/s from the temperature of 880°C down to the temperature of 320°C, and, subsequently from the temperature of 320°C to room temperature with cooling rate of 17° C/s).

3. Results and discussion

The investigated material was supplied in the form of a 950 bar in a heat treated condition. The microstructures of the investigated steels in as-delivered condition are presented in Fig. 2.
The microstructure is composed of ferrite, and also partly coagulated pearlite. It can be considered as homogeneous. The microstructure of steel reflects the history of deformation, after which, as the result of cooling, diffusion-free structures or partly diffusive ones were obtained. Dispersive precipitates of iron carbides in a ferritic matrix can be observed in the microstructure. Microstructural banding can be noticed in longitudinal cross-section, which is particularly clearly visible in the vicinity of the bar axis.

The true stress–true strain curves for 4340 steel deformed at different strain rates and temperatures to a total true strain of 1 are shown in Fig. 3. An increase in stress with increasing strain till a maximum value can be noticed. Then, stress uniformly decreases till almost constant value. Chen et al. [33] has found that in the case of processing of nickel alloy the flow stress is composed of three stages: work hardening, softening and steady state stage. Similar material flow behaviour can be observed in the case of the investigated 4340 steel. During the initial stage of deformation more and more dislocations are generated, what causes the increase of the stress value. In the works [33–35] it was noted, that dynamic recovery caused by climbing and cross-slip of dislocations is insufficient to counterbalance work hardening. The investigated material flow response is often connected with on-going dynamic recrystallization or dynamic recovery processes. It can be seen, that the effects of the temperature and strain rate on the flow stress are significant for all the tested conditions, what was also pointed out by some investigators [5,17]. Dynamic recrystallization occurs when critical deformation value is exceeded, what is connected
with high density of accumulated dislocations. After reaching the stress peak, DRX becomes the main mechanism of softening (softening effect caused by DRV is very small) [35].

The maps of changes to the strain rate sensitivity parameter of flow stress for the 4340 steel are presented in Fig. 4.

The research and analyses proved that an increase in the value of the $m$ parameter is simultaneously accompanied by improvement in the steel ductility, which should ensure good workability of the 4340 steel. Furthermore, the $m$ parameter increases together with the temperature of the process for the stable metal flow. The 4340 steel is significantly sensitive to thermal changes of the process conditions. Its deformation resistance is determined by the deformation mechanisms and is also depends on the amount of deformation, the influence exerted by the deformation history on the degree of the microstructure disintegration, without clearly observable correlations with the strain rate. Chen et al. [33] noted that the growth of DRX grains is associated with greater mobility of grain boundaries, increasing with increasing temperature. Wen et al. [34] found that the long deformation time associated with low strain rates is sufficient for grain boundaries migration and facilitates dynamic recrystallization, because the energy accumulated during deformation becomes the main driving force for the recrystallization process.

3.1. Processing maps

Processing maps were plotted for the constant value of true strain of $\varepsilon=0.4, 0.6, 0.8, 1$ and varying temperatures (Fig. 5). The contour numbers represent percent efficiency of power dissipation $\eta$, which could be interpreted by the microstructural evolution mechanism. The variations of contours value show

![Processing maps](image)

Fig. 5 – Processing maps for the 4340 steel based on the Prasad criterion and contour maps for instability parameter $\xi$ at the true strain of: 0.4, 0.6, 0.8, 1. Contour numbers represent percent efficiency of power dissipation $\eta$; shaded regions correspond to flow instability.
the process of microstructural evolution, such as DRV, DRX or grain coarsening [36,37]. The intermediary true strains are significantly important for the qualitative designing of the thermomechanical processing of the material. They provide information about the extension of, or decrease in, the areas of the material flow instability, and the possibility of the structural defects occurrence. Moreover, the flow instability areas indicated the inhomogeneous deformation with bands oriented according to the compression direction. These areas should be avoided during designing of hot working conditions for the studied steel.

On the processing map (Fig. 5) for true strain of 1, the processing windows and four areas of instability were shown. The high power dissipation efficiency indicates, that the material dissipates more energy for the microstructural changes, what determines the best process parameters, recommended for designing and conducting hot thermomechanical forging processes (processing windows) [38,39]. The peak value is found in the temperature range of 1050–1200 °C and the strain rate range of 3–57 s⁻¹, and it is higher than other domains (processing windows). It was found, that this domain shows the best process parameters: Processing window 1 (η = 32 – 36%), and these conditions can lead to the occurrence of dynamic recrystallization (Fig. 6) [18]. The maximum efficiency of power dissipation for DRX is approximately 0.3–0.4 for the materials with a low stacking fault energy [21]. High temperature
increases mobility of grain boundaries (providing sufficient driving force for grain boundaries migration), what intensifies DRX (facilitates nucleation of DRX grains) [35]. It is recommended for the purpose of conducting the conventional processes of die impression forging for a wide gamut of forgings with the application of horizontal forging machines and hydraulic screw presses, screw presses and double-action air-steam hammers. The shape of the $\eta$ isoclines is regular. The processing window 2 has the following parameters: $\eta = 24 - 36\%$, $i = 1 - 6 s^{-1}$ and $T = 875 - 1050\, \text{C}$. That range of parameters indicates, that the microstructure is entirely recrystallized above the temperature of 900 $\text{C}$, with austenitic matrix and a clearly-observable microstructural banding, and below that temperature the microstructure is only partially recrystallized. That fact is also confirmed by the observations of the microstructures. Gong et al. have observed, that the instability areas occur at a high strain rate and low temperature and vary with increasing strain on the processing maps for as-forged 34CrNiMo6 alloy steel, which was investigated at the temperature in a range of 1173–1473 $\text{K}$ and at strain rate in a range of $0.005–1 s^{-1}$ [18]. For intermediary true strain of 0.8, the processing windows undergo size modification, and assume the following parameters: processing window 1 $1 - \eta = 32 – 38\%$, $i = 2.5 – 32 s^{-1}$, $T = 1000 – 1200\, \text{C}$. Processing window 2 $- \eta = 26 – 38\%$, $i = 1 – 16 s^{-1}$, $T = 850 – 1000\, \text{C}$. In those windows, the peak of efficiency is found, which is higher than that in processing window for true strain of 1. In the case of true strains of 0.6 and 0.4, the processing windows undergo size limitation. The processing window 2 for true strain of 0.4 was eliminated and excluded from the recommendation for forging process because of small value of the parameter of power efficiency, which may exert a negative influence on the material workability at the temperature range of 850–1050 $\text{C}$ and strain rate range of 1–16 $s^{-1}$. For that reason, a forging may fail to fulfill qualitative requirements. The closeness of contours and decreases in power efficiency parameter could also be observed on the processing maps. Probably this situation is caused by rapid growth of the recrystallized grains and due to that, consuming the energy [18].

3.2. The microstructure observation

The analysis of the microstructures of the samples deformed in compression on Gleeeble 3800 simulator was performed. The map of microstructural changes in 4340 steel and the sample microstructures were presented in Fig. 6.

In the case of 4340 steel, a complete recrystallization of austenite occurs at the temperature exceeding Ac3 ($775\, \text{C}$). Apart from deformation bands resulting from compression, no further microstructural changes occurred in the processed samples. After the intensive cooling of deformed samples with compressed air, fine needles of martensite, and partly lower bainite were observed in the microstructure of the samples. The temperature of 900 $\text{C}$ was sufficient to austenitize the matrix of the investigated steel (Fig. 6b). A banded structure of austenitic matrix (typical deformation bands resulting from compression test), and also partial recrystallization, are visible in the microstructure of the sample deformed at the temperature of 800 $\text{C}$ and at the strain rate of $100 s^{-1}$ (Fig. 6c). It is possible to observe (Fig. 6e), that an increase in the temperature causes the growth of the austenite grains, and weakening of banding. The austenitic matrix underwent complete recrystallization, not causing further microstructural changes. However, the analysis of the microstructure indicates, that dynamic recrystallization of the austenite did not undergo in the whole volume of the material processed at the temperature of 800 $\text{C}$ and at strain rate of $0.1 s^{-1}$ and higher (Fig. 6d). As the strain rate increases, the proportion of ferrite precaptations is greater. These separations are no longer visible in the case of samples deformed at 900 $\text{C}$ (the range of existence of austenite). This proves that the temperature of 900 $\text{C}$ was sufficient to austenitize the matrix of the investigated steel. The analysis of the microstructure indicates that at deformation temperature of 900 $\text{C}$ only at the strain rate of $10 s^{-1}$ recrystallization did not take place throughout its volume during deformation. In the case of the material deformed at the temperature of 800 $\text{C}$ and at a strain rate of $0.01 s^{-1}$, the complete recrystallization of austenite should be associated with a long time of exposure of the material to a high temperature. Moreover, it can be concluded that the mechanisms of recovery and recrystallization are favoured by low strain rate [40]. Reducing the deformation time (increasing the strain rate) does not result in the complete recrystallization of austenite during deformation. However, at the deformation temperature of 900 $\text{C}$, the reduction of the strain rate below $10 s^{-1}$ and the associated increase in the time of exposure of the material to this temperature enables the full recrystallization of the austenitic structure. On the other hand, increasing the strain rate to $100 s^{-1}$ increases the intensity of strain, generating probably a stronger exothermic effect, which causes full dynamic recrystallization of austenite. An increase in strain rate on one hand may ensure a high dislocation density favourable to nucleation of DRX nuclei, on the other hand it causes a shorter deformation time which may cause an inhibition of the growth of DRX nuclei, and consequently may result in heterogeneity of grains [35].

Intensive cooling of the samples in forced air resulted in a bainite-martensite transformation in the austenitic areas. The microstructure of the investigated steel matrix consisted thus of martensite and lower bainite. It can be observed, that with the increase of deformation temperature, the grain size of the former austenite increases. This dependence is not so simple regarding the influence of the strain rate on the size of the former austenite grains. It seems, that for a given deformation temperature, the smallest austenite grain occurs at a strain rate of $10 s^{-1}$. This is most likely associated with a shorter time of exposure to high temperatures as compared to the samples deformed at a lower strain rate, what limits grain growth. However, in the case of a higher strain rate of $100 s^{-1}$ due to such a high intensity of strain, the exothermic effect causes a locally high temperature increase, which is a factor intensifying the grain growth. Microstructural banding was clearly visible in the case of the samples deformed at all analysed temperatures and strain rates. However, the higher the strain rate and the lower the deformation temperature, the banding was more visible. The bending of the aforementioned bands, which were originally parallel to the traverse movement during compression tests, indicates the material flow direction during compression.
Fig. 7 – Numerical distribution of: (a) temperature, (b) effective strain, (c) the efficiency of power dissipation parameter η%.

Fig. 8 – The gear wheel forged on hydraulic screw press: (a) corounded drop forged part, (b) hot drop forged part, (c) the areas of metallographic investigations.

Fig. 9 – The microstructure of the 4340 steel gear wheel: (a) light microscope, (b) scanning microscope; A, B, C – pre-selected points of the microstructure observations.
3.3. Simulations of the impression die forging process

One of the most important problems in forging process are large pressures per unit area (in the case of the analysed model forging the area of a web, cavities and teeth), and also the areas of difficult flow–gear teeth (the tooth point and the tooth root). The complex geometry of the forged part results in the complex material flow. The analysis of the material flow in upsetting operations guarantees the appropriate material flow kinematics in further forging stages. The obtained results of the numerical modelling of gear wheel forging are presented in Fig. 7.

The analysis of the distribution of effective strain on the cross-section of a gear wheel (Fig. 7b) indicates the fairly uniform deformation in the volume of the forging and stable material flow. The forging is free of internal and surface defects. It should have appropriate microstructure and properties, especially in the area of the tooth root, the small zone of the higher value of effective strain. A certain non-uniformity of effective strain occurs in the area of the web being formed. It results from the complex character of the material flow. The maximum values of the $g$ parameter (Fig. 7c) are located in the area of the teeth.

3.4. Industrial drop forging test and microstructure analysis

The temperature of the billet, measured with thermocouples, equalled to 1130 °C. Whereas the temperature on the forged part surface, measured with a pyrometer equalled to approximately 1100 °C. The forgings were corunded in order to check the quality of their surfaces. A gear wheel forged on hydraulic screw press was presented in Fig. 8. The samples for the metallographic investigations were taken from the cross-section of the forged part (Fig. 8c).

Lower bainite was observed in the microstructure of forged gear wheel made of the 4340 steel (Fig. 9). A fine, dispersive precipitates of carbides inside the lattices of ferrite having a lamellar shape were also noticed (Fig. 9 – points A and B). The differentiation of the lattices of ferrite and dispersion of carbides occurring in the forged part are the result of the various amount of deformation in the material during forging process. Two stage forging process caused the deformation of the initial austenitic structure of material. The differences in the amount of deformation of austenite grains exert influence on the differences in diffusion mechanisms and intermediary transformations.

4. Conclusions

The hot deformation behaviour of 4340 steel was analysed in the temperature range of 800–1100 °C, and in a strain rate range of 0.01–100 s$^{-1}$. The main conclusions of this study can be summarized as follows:

1. The distributions of the values of strain rate sensitivity parameter $m$ for a true strain range of 0.4–1 were analysed. 4340 steel is significantly more sensitive to thermal changes than to the changes in strain rate.

2. Based on the analysis of processing maps and microstructure observations, the optimum processing parameters for forging 4340 steel at different strains were obtained. The map at true strain of 1 reveals two stable domains and four flow instability areas. It was found, that the processing window 1 with the peak value (the temperature range of 1050–1200 °C and strain rate range of 3–57 s$^{-1}$) shows the best parameters of processing, and additionally these conditions can lead to the occurrence of dynamic recrystallization.

3. Comparison of the numerical modelling (performed using commercial programme based on FEM – QForm VX) and forging tests performed in industrial conditions was performed. Forging process in industrial conditions was conducted applying the processing parameters being on the border of the processing window 1. Two stage forging process caused deformation of the initial austenitic structure of the investigated steel. Lower bainite as well as dispersive lamellar precipitates of carbides inside the lattices of ferrite were observed in the microstructure of forged gear wheel.

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