Flow behaviour and constitutive modeling for hot deformation of austenitic stainless steel

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Abstract

The flow behavior of 316H austenitic stainless steel is investigated using hot compression tests. The modified Johnson-Cook and Zerilli-Armstrong models are developed, and modified Arrhenius-type model is established using an approach by dividing low and high stress region for determining key material constant and an uncomplicated approach for compensating strain in which the activation energy is determined from peak stress and only other material constants are considered as strain-dependent constants. The performance of all developed constitutive models is comparatively analyzed. It is indicated that the significant sensitivity of flow stress to temperature and strain rate is exhibited, and at 900 and 950 °C, strain rate sensitivity is closely related to temperature and strain rate, which can be explained by low stacking fault energy for 316H austenitic stainless steel. The modified Arrhenius type model has a noticeably higher accuracy in predicting flow behaviour than other two developed models in spite of a good performance of all developed models according to visual examination and statistical analyses.

1. Introduction

316H austenitic stainless steel (ASS) possesses excellent oxidation resistance and high strength at elevated temperatures in comparison with traditional 316 ASSs due to their high carbon content \cite{1}. However, 316H ASS forms readily coarse and mixed grained structures, especially when their thickness is larger than 50 mm, which cannot be significantly improved by controlling solution treatment. Moreover, it has poor intergranular corrosion resistance in comparison with traditional 316 ASSs due to its high carbon content \cite{2}, which can lead to unexpected failures and thus huge losses.

Hot deformation is extensively performed for manufacturing final products of ASSs. Undoubtedly, controlling hot working process and thereby optimizing hot deformation behaviour are of significance to optimize microstructure and improve properties. Nevertheless, the performing successfully of an improved hot working process can be guaranteed by a precise establishment of the relationship between flow behavior and appropriate deformation parameters (such as strain, strain rate, and temperature) encountered during the processing operation \cite{3–6}, and this relationship can often be described by the constitutive models. Therefore, the in-depth understanding of flow behavior and establishing of constitutive models with higher accuracy for hot deformation are of critical importance to controlling hot working process with the purpose of improving microstructure and quality.

So far, many papers have paid attention to understanding flow behavior and establishing constitutive models of the ASSs, such as 304L (0.028%C, 18.6%Cr, 10.3%Ni) \cite{7}, 316L (0.02%C, 18.2%Cr, 11.6%Ni) \cite{7}, 15Cr-
Table 1. Chemical composition of 316H austenitic stainless steel used in this study, wt%.

| C   | N     | Ni    | Mo   | Cr   | Si   | Mn  | P    | S    | Fe  |
|-----|-------|-------|------|------|------|-----|------|------|-----|
| 0.052| 0.005 | 12.2  | 2.0  | 17.5 | 0.4  | 1.5 | 0.015| 0.001| Bal.|

15Ni-2.2Mo-Ti modified ASS [7–9], 304H (0.067%C, 17.65%Cr, 7.91%Ni) [10], 316 (0.018%C, 16.63%Cr, 10.85%Ni) [11], 316LN [12] and 316LN (0.023%C, 16.61%Cr, 11.54%Ni) [13]. In these studies, the constitutive equations for ASSs were always developed based on the Johnson-Cook, Zerilli-Armstrong and Arrhenius type models by determining the material constants in the corresponding model. Amongst these models, the Arrhenius type model is typically employed for the description of peak and steady state behavior. At the peak or steady state, the effect of strain on flow behaviour is in absence or insignificant and this effect has not been considered in this model. In fact, the effect of strain is significant during hot deformation. As a result, the Arrhenius type model was modified by incorporating strain-dependent material constants for compensating the strain and predicting the flow behaviour of ASSs [9, 10]. Nevertheless, during the development of this modified Arrhenius type model, there are two important problems that need to be addressed not only for ASSs but also for most other materials.

One problem lies in the compensation of strain. For considering compensation of strain, in general, the strain-dependent material constants are incorporated in the Arrhenius type model by the assumption that all material constants $Q$, $\alpha$, $A$ and $n$ (as stated in equation (19) in this study) are polynomial function of strain [3, 9, 10, 14]. This conventional approach for compensating strain is complicated and time consuming due to the large amount of calculation required for determining all material constants at each given strain, especially for $Q$ and $\alpha$. In fact, at a given strain, different deformation conditions correspond to the different microstructures owing to different stages of deformation or dynamic softening. That is to say, in the conventional approach for compensating strain, the material constants such as activation energy $Q$ and constitutive equation for a particular strain will be developed using the flow stress determined from the different microstructures. The other problem lies in the calculation of material constant $\alpha$ according to $\alpha \approx \beta/n'$, where the $n'$ and $\beta$ (as stated in equations (20) and (21) in this study) are determined from the slopes of the lines in the $\ln \varepsilon - \ln \sigma$ plot and $\ln \varepsilon - \sigma$ plot for different temperatures, respectively [3–5, 15, 16]. Note that the determination of material constant $\alpha$ is very important for developing Arrhenius type model with high accuracy. In generally, $\alpha$ is obtained by employing the measured data at all stress levels containing high and low stress level to calculate $n'$ and $\beta$ [9–19]. In fact, the high stress level ($\alpha \sigma > 1.2$) cannot be applicable for calculating $n'$, while the low stress level ($\alpha \sigma < 0.8$) cannot be suitable for calculating $\beta$ (as stated in equations (20) and (21) in this study).

In this study, the flow behavior of a 316H ASS was investigated using isothermal hot compression tests on a thermo-mechanical simulator, and the microstructure was investigated using electron backscattering diffraction (EBSD). The modified constitutive equation was developed based on the Johnson–Cook and Zerilli-Armstrong models, and the Arrhenius-type model was also employed to establish the modified constitutive equations using the method for addressing the above-mentioned problems. The reliability of these modified constitutive models throughout the deformation conditions was comparatively examined and analyzed in detail with an objective to obtain an optimum method to model the flow behaviour of 316H ASS during hot deformation.

2. Experimental section

The 316H used in this study was prepared by vacuum melting and casting in the laboratory. The chemical composition of experimental steel is given in table 1. In order to obtain the specimens with uniform microstructure for hot compression test, the ingot was heated at 1200 °C for 2 h and hot rolled in the temperature range from 1150 °C to 950 °C, and subsequently the hot rolled plate was solution treated at 1100 °C for 30 min and then immediately water quenched.

Isothermal hot compression test was performed using Gleeble-3800 thermo-mechanical simulator according to the hot deformation schedule (figure 1). Cylindrical specimens with the diameter of 8 mm and height of 15 mm were machined from the final plate and the cylinder axes are parallel to the rolling direction. For minimizing the friction coefficient and barreling during hot compression tests, the specimens were coated with tantalum foils with the thickness of 0.1 mm and high-temperature lubricant (MoS$_2$). Microstructure was analyzed using EBSD and the corresponding samples were produced using the standard method in the literature [3]. Note that cross-sections of cylindrical compression specimens halved parallel to the cylinder axis were used for analyzing microstructure.
3. Results and discussion

3.1. Flow behavior

Figure 2 shows the flow stress curves at different deformation conditions in this study. It can be seen that all flow curves typically show remarkable work hardening at the initial stage followed by prominent dynamic softening such as DRX and dynamic recovery (DRV). Under deformation conditions with high temperature and low strain rate, the flow curves undergo a single peak stress followed by a gradual fall toward a steady state stress, indicating the occurrence of pronounced DRX. But at the rest of the deformation conditions, the flow curves show a little change in stress with an increase in strain after initial work hardening. This indicates that no or very limited DRX is taking place and the operative softening mechanism is DRV [3, 20]. The occurrence of this softening behavior
can be closely related to high Mo content and relatively low temperature. Naghizadeh and Mirzadeh [21] suggested that the Mo has significant effect on the retardation of grain growth for ASSs due to an interaction energy between Mo and grain boundaries. On the other hand, relatively low temperature can decrease the driving force for recrystallization during the hot deformation. In order to in depth analyze the difference in the flow curves showing different softening mechanism, the work hardening rate ($\theta = \partial \sigma / \partial \varepsilon$) is calculated and the relationship between work hardening rate and true stress (work hardening curve) is established at strain rate of 0.1 s$^{-1}$ and deformation temperatures of 900 to 1100 °C, as shown in figure 3. The change of work hardening rate for the flow curves having characteristic of DRX is faster than that of DRV, indicating the different softening rates of DRX and DRV.

It also can be seen form figure 2 that the flow stress decreases with increasing deformation temperature and decreasing strain rate. That is to say, the experimental steel displays thermal softening and strain rate hardening. This is due to the fact that higher temperatures and lower strain rates offer higher mobility of dislocation and grain boundary and longer time, respectively, which can benefit the annihilation and rearrangement of dislocations, the formation and coalescence of subgrains and/or the nucleation and growth of new grains. Moreover, the sensitivity of flow stress to strain rate is closely connected with deformation temperature and strain rate on deformation at relatively low temperatures of 900 and 950 °C. With decreasing deformation temperature from 950 to 900 °C the change of strain rate has less effect on flow stress, and at deformation temperature of 900 °C the change of strain rate show little impact on flow stress as the strain rate increases.

In order to further analyze the sensitivity of flow stress to strain rate, the strain rate sensitivity should be evaluated. Taylor and Hodgson proposed that during hot deformation, the higher slope of the line in the $\ln \sigma - \ln \dot{\varepsilon}$ plot for certain strain rate region reflects the higher strain rate sensitivity [22]. Take an example of the strains of 0.1 and 0.3, the $\ln \sigma - \ln \dot{\varepsilon}$ plot is established for experimental steel after deformation at 900 and 950 °C, as shown in figure 4. It can be seen that the strain rate sensitivity decreases with decreasing deformation temperature, and at deformation temperature of 900 °C, the strain rate sensitivity are significantly different in low ($<1$ s$^{-1}$) and high ($>1$ s$^{-1}$) strain rate regions and decreases with increasing strain rate. A similar phenomenon can be observed at other strains. The results and analyses mentioned above indicate the experimental steel exhibits strain rate insensitive with the decrease of deformation temperature from 950 to 900 °C and/or the increase of strain rate at deformation temperature of 900 °C.

During deformation at temperatures of 900 and 950 °C, the effect of deformation temperature and strain rate on strain rate sensitivity can be closely connected with the low stacking fault energy (SFE) of experimental steel. According to a formula for estimation of the SFE of ASSs employed in the literature [23], the SFE of experimental steel was determined as 58.3 mJ m$^{-2}$, which is relatively low. For experimental steel during deformation at 900 and 950 °C, no or very limited DRX occur and DRV is main restoration mechanism according to the flow stress curve analysis (as shown in figures 2 and 3). The higher SFE the shorter distance between the partial dislocations is obtained [3]. In high SFE materials, the recombination of partial dislocations is prone to occur due to short distance between the partial dislocations, which can promote the cross-slip of dislocations. This dislocation migration prompts the annihilation and rearrangement of dislocations, which can decrease the dislocation density and lower the work hardening rate. When the strain rate is lower the partial

![Figure 3. Relationship between work hardening rate and true stress for the experimental steel after deformation at strain rate of 0.1 s$^{-1}$ and temperatures of 900 to 1100 °C.](image-url)
dislocations can easily recombine, but as the strain rate increases the recombination of partial dislocations becomes difficult, which can heighten the work hardening rate. A similar sensitivity of work hardening rate to strain rate has been proved by other researchers [24]. In low SFE materials, the recombination of partial dislocations cannot occur, which can retard the activation of dislocation cross-slip and eventually suppress the process of recovery even at low strain rates. Hence, the work hardening rate has nothing to do with the strain rate. That is to say, the strain rate sensitivity is independent of the strain rate.

The SFE can be closely related to some metallurgical factors such as chemical composition and temperature [25, 26]. The increase of deformation temperature can increase the SFE, which can decrease the distance between partial dislocations and promote the activation of dislocation recombination and cross slip. This implies that for experimental steel with low SFE, the partial dislocations will recombine easily, causing that strain rate has an influence on work hardening rate, especially at low strain rate region during deformation at elevated temperatures of 900 and 950 °C. As the deformation temperature decreases from 950 to 900 °C, the SFE reduces which can increase the distance between partial dislocations, hinder the activation of recombination and cross slip, and eventually weaken the influence of strain rate on work hardening rate even at low strain rate region. Therefore, the strain rate sensitivity decreases with decreasing the deformation temperature from 950 to 900 °C. Furthermore, at deformation temperature of 900 °C, as the strain rate increase the probability of recombination of partial dislocations continues to decreases, which suppress further the occurrence of dislocation cross-slip. This means that at high strain rate region, the strain rate has no or lower influence on work hardening rate. Hence, the strain rate sensitivity decreases with increasing the strain rate at deformation temperature of 900 °C.

3.2. Modified Johnson-Cook model

The Johnson-Cook model was firstly presented by Johnson and Cook [27] for describing the flow behaviour of metallic materials under a wide ranges of strain, strain rate and deformation temperature. This model can be described by the following equation:

\[
\sigma = (A_1 + A_2\varepsilon^{n_2})(1 + A_3 \ln \dot{\varepsilon}^*)(1 - T^* a_2)
\]  

where \(\sigma\) is flow stress (MPa); \(\varepsilon\) is strain; \(\dot{\varepsilon}^* = \dot{\varepsilon}/\dot{\varepsilon}_0\) and here, \(\dot{\varepsilon}\) and \(\dot{\varepsilon}_0\), are strain rate and reference strain rate (s\(^{-1}\)), respectively; \(T^* = (T - T_i)/(T_m - T_i)\) and here \(T\), \(T_i\), and \(T_m\), are deformation temperature, reference temperature and melting temperature of the metallic material (K), respectively; \(A_1\) is yield stress at reference temperature and reference strain rate (MPa); \(A_2, n_1, A_3, n_2\) are material constants. The Johnson-Cook model has been widely used due to simple expression form and less material constants that can be determined by a few measured data.

Clearly, in the Johnson-Cook model the flow behaviour can be described by the product of three mutually independent factors of strain hardening, strain rate hardening and thermal softening. As a matter of fact, the temperature, strain rate and strain have a coupled effect on the flow behaviour according to the section 3.1, as shown in figures 2 and 4. For addressing this problem which will be undoubtedly harmful to the accuracy of original model, the modified Johnson-Cook model was proposed by Liu et al [28] which can be described by the following equation:

![Figure 4](image-url)
\[ \sigma = (B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3)(1 + B_5 \ln \varepsilon^*) \exp[\{(\lambda_1 + \lambda_2 \ln \varepsilon^*)(T - T_f)\}] \] (2)

where \(B_1, B_2, B_3, B_4, B_5, \lambda_1\) and \(\lambda_2\) are material constants, and the meanings of the other parameters are identical to those in the original Johnson-Cook model. This model has been successfully utilized to predict flow behaviour for carbon steels [28, 29], nickel-based alloy [30] and FeCr alloy [31]. So far, however, almost no effort based on this modified model has been devoted to model flow behaviour of ASSs, especially 316H.

In this study, the modified Johnson-Cook model is developed for working out an optimum model to predict the flow behaviour of 316H ASS. Obviously, different material constants of modified Johnson-Cook model can be obtained by selecting the different reference temperatures and strain rates. In this case, the average absolute relative error AARE and a constrained optimization procedure is employed for the determination of suitable reference condition and material constants and the more accurate prediction of flow behaviour by using this modified model. The AARE can be expressed as follows:

\[ \text{AARE} \% = \frac{1}{N} \sum_{i=1}^{N} \left| \frac{M_i - P_i}{M_i} \right| \times 100 \] (3)

where \(M\) is the measured value; \(P\) is the predicted value obtained from the modified constitutive equation; \(N\) is the total number of data points. In a constrained optimization procedure, the different AAREs are firstly obtained by substituting the different material constants acquired by selecting the different reference conditions into the modified Johnson-Cook model. Subsequently, the most suitable reference condition and material constants which corresponds to minimum AARE between the measured and predicted flow stresses can be determined on the basis of the comparison between different AAREs, and eventually the modified Johnson-Cook model with high accuracy are obtained for experimental steel. As an example, the determination of material constants at reference temperature of 900 °C and reference strain rate of 0.05 s\(^{-1}\) is as follows.

According to equation (2), it can be obtained at reference temperature and reference strain rate:

\[ \sigma = B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3 \] (4)

It is clear that the material constants \(B_1, B_2, B_3, B_4, B_5, \lambda_1\) are obtained to be 162.0146, 515.883 19, –957.122 22 and 570.17434 MPa from the coefficient of the three-order polynomial generated by fitting in the \(\sigma - \varepsilon\) plot at reference temperature, reference strain rate and constant strains (0.1–0.7 at an interval of 0.05) (figure 5(a)), respectively.

According to equation (2), it can be obtained at reference temperature:

\[ \frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3} = 1 + B_5 \ln \varepsilon^* \] (5)

It is clear that the material constant \(B_5\) can be acquired to be 0.0579 from the slope of the line in the \(\frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3} = \ln \varepsilon^*\) plot at reference temperature, constant strains (0.1–0.7 at an interval of 0.05) (figure 5(b)) by consideration of the intercept of 1.

According to equation (2), it can be obtained:

\[ \ln \left\{ \frac{\sigma}{(B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3)(1 + B_5 \ln \varepsilon^*)} \right\} = (\lambda_1 + \lambda_2 \ln \varepsilon^*)(T - T_f) \] (6)

It is clear that the \(\lambda_1 + \lambda_2 \ln \varepsilon^*\) can be acquired to be \(-0.004\ 28, -0.003\ 85, -0.003\ 02\) and \(-0.002\ 32\) from the slope of the line in the \(\ln \left\{ \frac{\sigma}{(B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3)(1 + B_5 \ln \varepsilon^*)} \right\} = (T - T_f)\) plot at constant strains (0.1–0.7 at an interval of 0.05) and the strain rates 0.05 s\(^{-1}\) (figure 5(c)), 0.1 s\(^{-1}\) (figure 5(d)), 1 s\(^{-1}\) (figure 5(e)) and 10 s\(^{-1}\) (figure 5(f)) by consideration of the intercept of 0, respectively. These slopes are denote as S. So,

\[ S = \lambda_1 + \lambda_2 \ln \varepsilon^* \] (7)

It is clear that the \(S\) is only a function of strain rate, and the \(\lambda_1\) and \(\lambda_2\) can be acquired to be \(-0.004\ 18\) and 0.000 360 929 from the intercept and slope of the line in the \(S - \varepsilon\) plot (figure 5(g)), respectively.

Subsequently, the material constants \(B_1, B_2, B_3, B_4, B_5, \lambda_1\) and \(\lambda_2\) are determined at different reference conditions. Substituting different set of material constants at different reference conditions into the modified Johnson-Cook model, the different AAREs are obtained according to the equation (3), as shown in figure 6. It is clear that the AARE is minimum at 3.058% when the reference temperature and strain rate are 1100 °C and 10 s\(^{-1}\), respectively. The optimized material constants of modified Johnson-Cook model for experimental steel are given in table 2.

Finally, the modified Johnson-Cook model with relatively higher accuracy in predicting the flow behaviour of experimental steel can be expressed as follows:
Figure 5. Plots used to determine the material constants of modified Johnson-Cook model at reference temperature of 900 °C and reference strain rate of 0.05 s⁻¹ for the experimental steel. (a) $\sigma - \varepsilon$ plot, (b) $\ln \left( \frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2} \right)$ plot, (c) $\ln \left( \frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3} \right)$ plot, (d) $\ln \left( \frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3 + B_5 \varepsilon^4} \right)$ plot, (e) $\ln \left( \frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3 + B_5 \varepsilon^4 + B_6 \varepsilon^5} \right)$ plot, (f) $\ln \left( \frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3 + B_5 \varepsilon^4 + B_6 \varepsilon^5 + B_7 \varepsilon^6} \right)$ plot, (g) $\ln \left( \frac{\sigma}{B_1 + B_2 \varepsilon + B_3 \varepsilon^2 + B_4 \varepsilon^3 + B_5 \varepsilon^4 + B_6 \varepsilon^5 + B_7 \varepsilon^6 + B_8 \varepsilon^7} \right)$ plot.
The Zerilli-Armstrong model was originally suggested by Zerilli and Armstrong based on the mechanics of dislocation motion which play an essential role in the plastic deformation of metallic materials [32, 33], and can describe the flow behaviour of bcc and fcc metallic materials under different strain rates at temperatures between room temperature and \(0.6T_m\) [11]. For the materials with different crystalline structures, this model display different expressions due to the difference in the nature of dislocation interactions. When the materials possess fcc structure, the Zerilli-Armstrong model can be described by the following equation:

\[
\sigma = (142.05561 + 396.10523 \varepsilon - 897.51908 \varepsilon^2 + 655.50451 \varepsilon^3) \exp\left((-0.00264 + 2.97974 \times 10^{-4} \ln 0.1 \varepsilon)(T - 1373.15)\right) \text{ MPa}
\]

where \(C_1, C_2, C_3, C_4\) and \(k_1\) are material constants; the meanings of other parameters are identical to those in the original Johnson-Cook model. Although the Zerilli-Armstrong model has complex form in comparison with Johnson-Cook model, it has also been widely used in much finite element software [10].

Obviously, the Zerilli-Armstrong model considers the coupled effect of strain rate and temperature on flow stress. However, this model does not fully consider the coupled effect of temperature, strain rate and strain on flow behaviour, such as the coupled effect of temperature and strain, and it is not applicable for modeling the flow behaviour of materials under hot working temperature domain (above \(0.6T_m\)) [10, 11]. For overcoming this insufficiency, the modified Zerilli-Armstrong model was proposed by Samantaray et al [11] which will more accurately predict the flow behaviour of materials over a wider strain rate-temperature domain and can be described by the following equation:

\[
\sigma = (\varepsilon + D_1 \varepsilon^2) \exp\left(-D_3 + D_4 \varepsilon^2(T - T_1) + D_5 + D_6 (T - T_2) \ln \varepsilon^k \right) \text{ MPa}
\]

where \(D_1\) is the yield stress at reference temperature and reference strain rate (MPa); \(D_2, D_3, D_4, D_5, D_6\) and \(k_2\) are material constants; the meanings of the other parameters are identical to those in the original Johnson-Cook model. This model has been successfully utilized to predict flow behaviour for austenitic stainless steel such as 304L, 316L and Ti-modified ASS [10, 11] and carbon steels [34]. Up to now, however, no work about this modified model has been focused on modeling flow behaviour of 316H ASS.
In this study, the modified Zerilli-Armstrong model is also developed for working out an optimum model to predict the flow behaviour of 316H ASS. Like modified Johnson-Cook model, the different material constants can be obtained by selecting the different reference conditions for the modified Zerilli-Armstrong model. In order to derive the modified Zerilli-Armstrong model to more accurately predict the flow behaviour, the most suitable reference condition and material constants are found using a constrained optimization procedure, as described in the section 3.2. As an example, the determination of material constants at reference temperature of 900 °C and reference strain rate of 0.05 s⁻¹ is as follows.

It is clear that the $D_1$ can be acquired to be 100.31 MPa from the yield stress of flow stress curve at the reference temperature of 900 °C and the reference strain rate of 0.05 s⁻¹ (figure 2).

According to equation (10), it can be obtained at reference strain rate:

$$\sigma = (D_1 + D_2 \varepsilon^{a_2}) \exp\left[-(D_3 + D_4 \varepsilon)(T - T_r)\right]$$

(11)

According to equation (11), it can be obtained:

$$\ln \sigma = \ln(D_1 + D_2 \varepsilon^{a_2}) - (D_3 + D_4 \varepsilon)(T - T_r)$$

(12)

It is clear that the $\ln(D_1 + D_2 \varepsilon^{a_2})$ and $-(D_3 + D_4 \varepsilon)$ can be acquired from the intercepts and slopes of the lines in the $\ln \sigma - (T - T_r)$ plot at reference strain rates and different strains (0.1–0.7 at an interval of 0.05) (figures 7(a), (b)), respectively. These intercepts and slopes are denote as $I_1$ and $S_1$. So,

$$I_1 = \ln(D_1 + D_2 \varepsilon^{a_2})$$

(13)

$$S_1 = -(D_3 + D_4 \varepsilon)$$

(14)

The values of $I_1$ and $S_1$ are only a function of strain.

According to equation (13), it can be obtained:

$$\ln(\exp I_1 - D_1) = \ln D_2 + k_2 \ln \varepsilon$$

(15)

It is clear that the values of $D_2$ and $k_2$ can be acquired to be 177.757 45 and 0.218 97 from the intercept and slope of the line in the $\ln(\exp I_1 - D_1) - \ln \varepsilon$ plot (figure 7(c)), respectively.

According to equation (14), it is clear that the values of $D_3$ and $D_4$ can be acquired to be 0.003 49 and 0.001 93 from the intercept and slope of the line in the $S_1 - \varepsilon$ plot (figure 7(d)), respectively.

From equation (10), it also can be obtained:

$$\ln \sigma = \ln(D_1 + D_2 \varepsilon^{a_2}) - (D_3 + D_4 \varepsilon)(T - T_r) + [D_5 + D_6(T - T_r)] \ln \varepsilon^*$$

(16)

It is clear that the $D_5 + D_6(T - T_r)$ can be acquired from the slopes of the lines in the $\ln \sigma - \ln \varepsilon^*$ plot at different temperatures and different strains (0.1–0.7 at an interval of 0.05). These slopes are denote as $S_2$. So,

$$S_2 = D_5 + D_6(T - T_r)$$

(17)

It is clear that at a particular strain, the $S_2$ is only a function of temperature and the material constants $D_5$ and $D_6$ can be acquired from the intercept and slope of the line in the $S_2 - (T - T_r)$ plot (figures 7(e), (f)). At different strains, different set of material constants $D_5$ and $D_6$ are acquired. In order to acquire the most suitable $D_5$ and $D_6$, a constrained optimization procedure is used and the AAREs under different set of material constants $D_5$ and $D_6$ at different strains are acquired, as shown in figure 7(g). It is clear that the AARE is minimum at 3.583% when $D_5$ is 0.058 64 and $D_6$ is 0.000 334 183.

Subsequently, the material constants $k_2$, $D_1$, $D_2$, $D_3$, $D_4$, $D_5$ and $D_6$ are acquired at different reference conditions. Substituting different set of material constants into the modified Zerilli-Armstrong model, the different AAREs are obtained according to the equation (3), as shown in figure 8. It is clear that the AARE is minimum at 3.482% when the reference temperature and strain rate are 900 °C and 0.1 s⁻¹, respectively. The optimized material constants of modified Zerilli-Armstrong model for experimental steel are given in table 3.

Finally, the modified Zerilli-Armstrong model with relatively higher accuracy in predicting the flow behaviour of experimental steel can be expressed as follows:

$$\sigma = (96.9 + 198.56569 \cdot e^{0.20002}) \exp\left[-(0.00329 + 0.0017\varepsilon)(T - 1173.15)\right]$$

$$+ (0.06256 + 2.81057 \times 10^{-4}(T - 1173.15)) \ln 10\varepsilon$$

(18)

3.4. Modified Arrhenius type model with partial strain-dependent constants and key constant determined by division of stress region

The hyperbolic sine type model, firstly employed by Garofalo [35] for describing steady state creep behaviour, was modified subsequently by Sellars and Tegart [36] for describing the flow behaviour of hot deformation over a wider strain rate-temperature domain, which is denoted as Sellars-Tegart-Garofalo model and can be described by the following equation:
Figure 7. Plots used to determine the material constants of modified Zerilli-Armstrong model at reference temperature of 900 °C and reference strain rate of 0.05 s⁻¹ for the experimental steel. (a), (b) ln σ – (T – T₀) plot, (c) ln(exp(1 – Dₐ)) – ln ε plot, (d) S₁ – ε plot, (e), (f) S₁ – (T – T₀) plot, (g) relationship between AARE and strain.
where $A$ and $n$ are material constants; $\alpha$ is stress multiplier (MPa$^{-1}$); $Q$ is activation energy (kJmol$^{-1}$); $R$ is gas constant ($8.3145$ Jmol$^{-1}$ K$^{-1}$); the meanings of the other parameters are identical to those in the original Johnson-Cook model. For low stress level ($\alpha \sigma < 0.8$), the equation (19) can be approximately expressed as follows:

$$\dot{\varepsilon} = A' \sigma^n' \cdot \exp\left( -\frac{Q}{RT} \right)$$

(20)

where $A'$ and $n'$ are the material constants. For high stress level ($\alpha \sigma > 1.2$), the equation (19) can be approximately expressed as follows:

$$\dot{\varepsilon} = A'' \exp(\beta \sigma) \cdot \exp\left( -\frac{Q}{RT} \right)$$

(21)

where $A''$ and $\beta$ are the material constants. The Sellars-Tegart-Garofalo model is also referred to as the Arrhenius type model because Zener-Holloman parameter $Z$ is coupled in this model [37]. The $Z$ can be expressed as follows:

$$Z = \dot{\varepsilon} \cdot \exp\left( -\frac{Q}{RT} \right)$$

(22)

Usually, this model can be employed for establishing the relationship between the peak or steady state stress, deformation temperature and strain rate. Recently, it was modified by considering compensation of strain for predicting the flow behaviour of materials [3, 12, 13, 17–21]. However, two important problems that need to be addressed are usually neglected on the development of Arrhenius type model considering compensation of strain. One problem is that in the conventional approach for compensating strain, the material constants such as activation energy $Q$ and constitutive equation for a particular strain will be developed using the flow stress determined from the different microstructures owing to different stages of deformation or dynamic softening at a given strain for different deformation conditions. For overcoming this shortcoming, the material constants $Q$ and $\alpha$ should be determined by using the peak stress that reflect the similar stage of deformation or dynamic softening for all flow curves, and only $A$ and $n$ should be considered as strain-dependent constants. This modified approach also possesses a relatively uncomplicated calculation for determining the material constants.
3.4.1. Determination of material constants $\alpha$ and $Q$

The stress multiplier $\alpha$ and activation energy $Q$ are determined using the peak stress at different deformation conditions according to equations (19)–(21) in this study. That is to say, the peak stress $\sigma_p$ is taken for the term $\sigma$ in the equations (19)–(21) for determining $\alpha$ and $Q$. Note that the peak stress is determined through the work hardening rate reaching a value of $\theta = 0$. As stated in section 1, the $\alpha$ can be calculated according to $\alpha \approx \beta/n'$, where the $n'$ and $\beta$ cannot be determined by employing the relatively high stress ($\alpha \sigma > 1.2$) and low stress ($\alpha \sigma < 0.8$), respectively. In order to divide the peak stress into relatively low and high stresses and thus determine the $n'$ and $\beta$, the approximate value of $\alpha$ is estimated as 0.0075 MPa$^{-1}$ according to the value of $\alpha$ of other austenitic stainless steels reported by other researchers [13, 15, 16, 38]. Finally, by employing the peak stresses at 1100 °C–1150 °C and 0.05–1 s$^{-1}$ and at 900 °C–1000 °C and 0.05–10 s$^{-1}$ the material constants $n'$ and $\beta$ can be determined, respectively, in this study.

According to equations (20) and (21), it can be obtained:

\[
\ln \dot{\varepsilon} = \ln A' + n' \ln \sigma_p - \frac{Q}{RT} \tag{23}
\]

\[
\ln \dot{\varepsilon} = \ln A'' + \beta \sigma_p - \frac{Q}{RT} \tag{24}
\]

It is clear that the material constants $n'$ and $\beta$ can be acquired to be 7.565 27 and 0.054 493 MPa$^{-1}$ from the slopes of the lines in the $\ln \dot{\varepsilon} \text{--} \ln \sigma_p$ plot at the deformation temperatures of 1100 °C–1150 °C and strain rates of 0.05–1 s$^{-1}$ (figure 9(a)) and in $\ln \dot{\varepsilon} \text{--} \sigma_p$ plot at 900 °C–1000 °C and 0.05–10 s$^{-1}$ (figure 9(b)), respectively, and thus the stress multiplier $\alpha$ is 0.0072 MPa$^{-1}$.

According to equation (19), it can be obtained:

\[
\ln \dot{\varepsilon} = \ln A + n \ln[ \sinh(\alpha \sigma_p)] - \frac{Q}{RT} \tag{25}
\]

After differentiating equation (25), it can be obtained:

\[
Q = R \cdot \frac{\partial \ln \dot{\varepsilon}}{\partial \ln[ \sinh(\alpha \sigma_p)]} \bigg|_T \cdot \frac{\partial \ln[ \sinh(\alpha \sigma_p)]}{\partial (1/T)} \tag{26}
\]

It is clear that the activation energy $Q$ can be acquired to be 512.497 236 65 kJ mol$^{-1}$ from the slopes of the lines in the $\ln \dot{\varepsilon} \text{--} \ln[ \sinh(\alpha \sigma_p)]$ plot (figure 9(c)) and $\ln[ \sinh(\alpha \sigma_p)] = 1000/T$ plot (figure 9(d)), and then the $Z$ can be acquired at different strain rates and temperatures. The activation energy $Q$ for experimental steel are high, although it is close to those reported for ASSs which are typically in the range of 350–520 kJ mol$^{-1}$ [13, 15]. This can be ascribed to the higher content of solid solution elements such as C, Cr, Ni, and Mo [15].

3.4.2. Determination of strain-dependent material constants $A$ and $n$

For the compensation of strain, the material constants $A$ and $n$ in the Arrhenius type model are considered as the strain-dependent constants which are determined by establishing the polynomial relationship between strain and the $A$ and $n$. The following are the determination of the $A$ and $n$ at the strain of 0.3.

According to equations (19) and (22), it can be obtained:

\[
Z = \dot{\varepsilon} \cdot \exp\left( \frac{-Q}{RT} \right) A[ \sinh(\alpha \sigma)]^n \tag{27}
\]

According to equation (27), it can be obtained:

\[
\ln Z = \ln A + n \ln[ \sinh(\alpha \sigma)] \tag{28}
\]

It is clear that the $\ln A$ and $n$ can be acquired to be 42.795 25 and 6.584 52 from the intercept and slope of the line in the $\ln Z \text{--} \ln[ \sinh(\alpha \sigma)]$ plot (figure 10(a)), respectively.
Subsequently, the $n$ and $\ln A$ at different strains ($0.1 \sim 0.7$ at an interval of 0.05) are acquired by using the above-mentioned solution procedure, and the strain-dependent constants $n$ and $\ln A$ are very well described by polynomial fitting (equation (29)), as shown in figures 10(b), (c). The coefficients of the polynomial are given in table 4.

\[
\begin{align*}
    n &= E0 + E1\varepsilon + E2\varepsilon^2 + E3\varepsilon^3 + E4\varepsilon^4 + E5\varepsilon^5 + E6\varepsilon^6 \\
    \ln A &= F0 + F1\varepsilon + F2\varepsilon^2 + F3\varepsilon^3 + F4\varepsilon^4 + F5\varepsilon^5 + F6\varepsilon^6
\end{align*}
\]  

Figure 9. Plots used to determine the material constants $\alpha$ and $Q$ of modified Arrhenius type model for the experimental steel. (a) $\ln \dot{\varepsilon} = \ln \sigma$ plot at 1100 °C–1150 °C and 0.05–1 s$^{-1}$, (b) $\ln \dot{\varepsilon} = \sigma$ plot at 900 °C–1000 °C and 0.05–10 s$^{-1}$, (c) $\ln \dot{\varepsilon} = \ln [\sinh(\alpha\sigma)]$ plot, (d) $\ln [\sinh(\alpha\sigma)] = -1000/T$ plot.

According to equation (27), it can also be obtained [3]:

\[
\sigma = \frac{1}{\alpha} \ln \left( \left( \frac{Z}{A} \right)^{1/n} + \left[ \left( \frac{Z}{A} \right)^{2/n} + 1 \right]^{1/2} \right)
\]  

Finally, the modified Arrhenius type model with partial strain-dependent constants and key constant determined by division of stress region can be expressed as follows:

\[
\begin{align*}
    \sigma &= \frac{1}{0.0072} \ln \left( \left( \frac{Z}{A} \right)^{1/n} + \left[ \left( \frac{Z}{A} \right)^{2/n} + 1 \right]^{1/2} \right) \\
    Z &= \dot{\varepsilon} \cdot \exp \left( \frac{512497.23665}{RT} \right) \\
    n &= 13.63659 - 86.26788\varepsilon + 473.04879\varepsilon^2 - 1437.72335\varepsilon^3 + 2474.81615\varepsilon^4 - 2254.89062\varepsilon^5 + 841.76677\varepsilon^6 \\
    \ln A &= 46.77732 - 48.88114\varepsilon + 267.28813\varepsilon^2 - 830.98392\varepsilon^3 + 1501.58207\varepsilon^4 - 1440.68418\varepsilon^5 + 561.30581\varepsilon^6
\end{align*}
\]
3.5. Verification of developed constitutive models

For verifying the developed constitutive models in this study, the comparison between the flow curves measured and flow stresses predicted is shown in figure 11. Overall, the flow stresses predicted on the basis of all the developed constitutive models, especially the modified Arrhenius type model, can very well track flow curves measured, although some deviations occur at some deformation conditions, especially at lower temperatures and higher strain rates such as at 900 °C in 10 s⁻¹ and 950 °C in 10 s⁻¹. At 900 °C in 10 s⁻¹, the flow stresses predicted on the basis of all the developed constitutive models are almost always higher than flow stresses measured, while at 950 °C in 10 s⁻¹ those are almost always lower than flow stresses measured. The following are factors that generate the above-mentioned deviations.

(1) The errors formed during the acquisition of experimental data;

(2) The cumulative errors formed during the determination of the material constants and/or strain-dependent material constants such as the establishing of polynomial relationship between strain and material constants;

Table 4. Coefficients of the polynomials for relationships between \( n \), \( \ln A \) and strain.

| \( n \) | \( E_0 \) | \( E_1 \) | \( E_2 \) | \( E_3 \) | \( E_4 \) | \( E_5 \) | \( E_6 \) |
|-------|--------|--------|--------|--------|--------|--------|--------|
| 13.636 59 | -86.267 88 | 473.048 79 | -1437.723 35 | 2474.816 15 | -2254.890 62 | 841.766 77 |

| \( \ln A \) | \( F_0 \) | \( F_1 \) | \( F_2 \) | \( F_3 \) | \( F_4 \) | \( F_5 \) | \( F_6 \) |
|-------|--------|--------|--------|--------|--------|--------|--------|
| 46.777 32 | -48.881 14 | 267.288 13 | -30.983 92 | 1501.582 07 | -1440.684 18 | 561.305 81 |

Figure 10. Plots used to determine the strain-dependent material constants \( n \) and \( A \) of modified Arrhenius type model for the experimental steel. (a) \( \ln Z - \ln [\sinh(\alpha \sigma_{0.3})] \) plot at the strain of 0.3, (b) relationship between \( n \) and strain, (c) relationship between \( \ln A \) and strain.

3.5. Verification of developed constitutive models

For verifying the developed constitutive models in this study, the comparison between the flow curves measured and flow stresses predicted is shown in figure 11. Overall, the flow stresses predicted on the basis of all the developed constitutive models, especially the modified Arrhenius type model, can very well track flow curves measured, although some deviations occur at some deformation conditions, especially at lower temperatures and higher strain rates such as at 900 °C in 10 s⁻¹ and 950 °C in 10 s⁻¹. At 900 °C in 10 s⁻¹, the flow stresses predicted on the basis of all the developed constitutive models are almost always higher than flow stresses measured, while at 950 °C in 10 s⁻¹ those are almost always lower than flow stresses measured. The following are factors that generate the above-mentioned deviations.

(1) The errors formed during the acquisition of experimental data;

(2) The cumulative errors formed during the determination of the material constants and/or strain-dependent material constants such as the establishing of polynomial relationship between strain and material constants;

Table 4. Coefficients of the polynomials for relationships between \( n \), \( \ln A \) and strain.

| \( n \) | \( E_0 \) | \( E_1 \) | \( E_2 \) | \( E_3 \) | \( E_4 \) | \( E_5 \) | \( E_6 \) |
|-------|--------|--------|--------|--------|--------|--------|--------|
| 13.636 59 | -86.267 88 | 473.048 79 | -1437.723 35 | 2474.816 15 | -2254.890 62 | 841.766 77 |

| \( \ln A \) | \( F_0 \) | \( F_1 \) | \( F_2 \) | \( F_3 \) | \( F_4 \) | \( F_5 \) | \( F_6 \) |
|-------|--------|--------|--------|--------|--------|--------|--------|
| 46.777 32 | -48.881 14 | 267.288 13 | -30.983 92 | 1501.582 07 | -1440.684 18 | 561.305 81 |
The limitation of the Johnson-Cook, Zerilli-Armstrong and Arrhenius type models. Some of these constitutive models can describe the flow behaviour on the basis of empirical information and are comprised of some simple mathematical expressions [7]. Furthermore, during the modeling of flow behaviour based on these models, the relationships between flow stress and deformation conditions are established by linear fitting according to the existing measured data. In fact, however, the change of the flow stress under different deformation conditions such as strain rates is nonlinear (figure 4), and the change of some factors which can influence the flow stress are also nonlinear due to the complex interaction between some metallurgical phenomena such as work hardening, DRX and DRV and so on during hot deformation [3]. Hence, the errors will occur in flow stress prediction on the basis of the Johnson-Cook, Zerilli-Armstrong and Arrhenius type models. The limitation of these models has also been found by other researchers [39, 40];

The occurrence of textural hardening under some deformation conditions. During deformation, the average Taylor factor possibly increases with the increase of strain in the occurrence of no or very limited DRX, indicating that the intensities of the ‘hard’ texture components increase, and this phenomenon will generate the hardening (textural hardening) and hinder the deformation [3, 41–43]. As discussed in the section 3.1, no or very limited DRX occur during hot deformation at lower temperature and higher strain rate. The microstructure before deformation and microstructure after deformation at 950 °C in 10 s⁻¹ are analyzed by employing EBSD, as shown in figure 12. It can be seen that the microstructure before
deformation consists of equiaxed grains, with many twins having been observed owing to low SFE of experimental steel [44]. After deformation at 950 °C in 10 s⁻¹, the microstructure is made up of elongated grains and a small number of fine new grains formed in the original high-angle grain boundaries, indicating the occurrence of a very limited DRX during this deformation conditions. Furthermore, the Taylor factor can be acquired from the EBSD analysis, and the fraction of grains in the Taylor factor range of 2–3 and 3–4 are 64.5% and 35.5% before deformation, respectively, while those are 48.8% and 51.2% after deformation at 950 °C in 10 s⁻¹. That is to say, the increase of Taylor factor occurs after deformation at some deformation conditions if no or very limited DRX occur. Therefore, the flow stresses predicted on the basis of all the developed constitutive models are underestimated under some deformation conditions such as at 950 °C in 10 s⁻¹, as shown in figure 11. Nevertheless, the developed constitutive models in this study still has high accuracy in predicting the flow stress at some deformation conditions even if the average Taylor factor possibly increases as well during deformation. Maybe at these deformation conditions the increase in
average Taylor factor has no so crucial influence on hardening because of the fairly high hardening at some deformation conditions (for example, relatively low temperatures) or the fairly high softening at other deformation conditions (for example, relatively low strain rates).

(5) The occurrence of thermal softening due to deformation heating at high strain rate. At relatively high strain rates, the deformation time is too short to allow for heat transfer, which can cause the occurrence of thermal softening in the specimen [4]. Therefore, the flow stresses predicted on the basis of all the developed constitutive models are overestimated under some deformation conditions with high strain rate such as at 900 °C in 10 s⁻¹, as shown in figure 11. Nevertheless, the developed constitutive models in this study still has high accuracy in predicting the flow stress at some deformation conditions with high strain rate even if the deformation heating possibly occurs as well during deformation. Maybe at these deformation conditions with high strain rate the deformation heating has no so crucial influence on softening due to fairly high available thermal activation energy that need to cause thermal softening at relatively high temperatures [4, 12].

Besides the visual examination, the verification of developed constitutive models is quantitatively performed using statistical analysis by introducing the correlation coefficient $R$ and average absolute relative error $AARE$. The $R$ can be expressed as follows:

$$R = \frac{\sum_{i=1}^{N}(M_i - \bar{M})(P_i - \bar{P})}{\sqrt{\sum_{i=1}^{N}(M_i - \bar{M})^2\sum_{i=1}^{N}(P_i - \bar{P})^2}}$$

where $M$ is the data measured; $P$ is the data predicted on the basis of the developed constitutive model; $\bar{M}$ and $\bar{P}$ are the mean of $M$ and $P$, respectively; $N$ is the number of used data. The correlation between the flow stresses measured and those predicted are evaluated, as shown in figure 13. It can be seen that a higher $R$ and corresponding good linear relationship exists between the flow stresses measured and those predicted on the
basis of all the developed constitutive models, especially the modified Arrhenius type models in this study. Sometimes, nevertheless, this good correlation possibly not necessarily reflect high accuracy of a model because all the data predicted on the basis of model may be higher or lower than those measured [8, 45, 46]. Therefore, the AARE also needs to be calculated for the verification of developed constitutive models. The AAREs acquired on the basis of the modified Johnson-Cook, modified Zerilli-Armstrong and modified Arrhenius type models developed in this study are found to be 3.058%, 3.482% and 2.539%, respectively, which are all relatively low especially that on the basis of the modified Arrhenius type models. Note that during the modeling of flow behaviour for experimental steel, besides the uncomplicated approach for compensating strain and the approach by division of stress region for determine key material constant, the modified Arrhenius type model is also developed by the conventional approach for determining key material constant and compensating strain, and the corresponding $R$ and AARE are calculated to be 0.993 and 3.176%, respectively.

The above-mentioned visual examination and statistical analyses indicate all the developed constitutive models can exhibit an accurate prediction of flow behaviour, and the modified Arrhenius type model has a better predictability than the other two developed constitutive models under investigated strain rate-temperature domain for 316H austenitic stainless steel in this study. In addition, the modified Arrhenius type model of 316H austenitic stainless steel developed by the method in this study has more satisfactory reliability and stability than that of 316H austenitic stainless steel established by the conventional approach and those of other austenitic steels and fcc alloys (such as ASS, high-Mn steel and nickel-base alloy) developed by the method proposed by other researchers [9, 12, 37, 38, 47] in terms of $R$ and AARE, as shown in figure 14. As a result, an approach by dividing low and high stress regions for determining the key material constant $\alpha$ and a uncomplicated approach for compensating strain in which the activation energy $Q$ can be determined from the peak stress and only material constants $n$ and $A$ can be considered as strain-dependent constants, are recommended to develop the Arrhenius type model with high accuracy and thereby describe the flow behaviour for austenitic steels and fcc alloys.

4. Conclusion

The flow behavior at the temperatures of 900 to 1150 °C and strain rates of 0.05 to 10 s$^{-1}$ of a 316H austenitic stainless steel is investigated using hot compression tests. The constitutive modeling for flow behaviour is performed based on the Johnson-Cook, Zerilli-Armstrong and Arrhenius type models. Main results are as follows:

1. The intense sensitivity of flow stress to deformation temperature and strain rate is exhibited. The flow stress decreases with increasing deformation temperature and decreasing strain rate. At the relatively low temperatures of 900 °C and 950 °C, the strain rate sensitivity is closely connected with deformation temperature and strain rate. The strain rate sensitivity decreases with decreasing deformation temperature
and increasing strain rate, which can be attributed to the low stacking fault energy for 316H austenitic stainless steel.

(2) An approach by dividing low and high stress regions and a uncomplicated approach for compensating strain in which the activation energy can be determined from the peak stress and only other material constants can be considered as strain-dependent constants, are utilized to determine the key material constant and incorporate the effect of strain in the Arrhenius type model, respectively and eventually the modified Arrhenius type model is developed. The modified Johnson-Cook and modified Zerilli-Armstrong models are also developed.

(3) The average absolute relative errors acquired on the basis of the modified Johnson-Cook, modified Zerilli-Armstrong and modified Arrhenius type models developed in this study are found to be 3.058%, 3.482% and 2.539%, respectively, and the correlation coefficients are 0.994, 0.992, and 0.996, respectively. The modified Arrhenius type model possesses a noticeably higher accuracy than the other two developed constitutive models by visual examination and statistical analyses. The modified Arrhenius type model with the partial strain-dependent constants and the key constant determined by division of stress region is recommended to describe the flow behaviour for austenitic steels and fcc alloys.

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