Metal flow behaviour and processing maps of high heat resistant steel during hot compression

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Abstract
This article reports the flow stress behaviour of ASTM A335 P92 steel. Uniaxial isothermal compression experiments were conducted to examine the hot deformation behaviour of P92 steel in a Gleeble® 3500 thermal–mechanical simulator. The test conditions were 0.01–10 s⁻¹ strain rate and 850–1000 °C deformation temperature. Constitutive equations and processing maps developed were used to describe the hot deformation process. The results showed that the flow stress–strain curves exhibited a dynamic recovery (DRV) behaviour as the dominant softening mechanism. The flow stress decreased with an increase in the deformation temperature or a decrease in strain rate. Using the Arrhenius equation, the stress exponent and the activation energy values were 8.0 kJ.mol⁻¹ and 487.56 kJ.mol⁻¹, respectively. The correlation between the constructed processing maps and microstructure showed that the optimal process parameters occurred at a lower strain rate in the region of 0.1 s⁻¹ and deformation temperatures of 900–950 °C and 1000 °C for the steel investigated.

Keywords
Flow stress · Arrhenius equation · Hot deformation · Processing map · DRV

1 Introduction
The characterisation of the metal forming process by using a constitutive equation has formed the basis to study the hot deformation behaviour of most metals and alloys. These equations model the relationship between the flow stress and the deformation conditions such as temperature, strain and strain rate. Moreover, the constitutive equation provides information on the flow stress characteristics. These equations act as input codes in the computer for computational simulation of the metal forming process [1–6]. The finite element method (FEM) simulation reduces the production cost and time [7]. The FEM tools assist in developing new metal forming techniques and process parameters optimisation of the existing ones, such as the Mannesmann industrial process used for the production of seamless tubes.

Broadly, the constitutive models describing the flow stress behaviour can be phenomenological or physically based. This classification depends on the computational parameters involved in quantifying the flow stress behaviour [8]. Because of the complexity of physical models, the Arrhenius constitutive model and the modified Arrhenius model have become popular in studying the relationship between flow stress and deformation parameters [8–13]. Several researchers have used the Arrhenius-type equation to investigate the flow stress behaviour of a variety of materials such as modified 9Cr-1Mo steel [3], P92 steel [7, 12, 13], 35CrMo steel [18], 20CrMo alloy steel [13], nickel-based superalloy [22], aluminium alloys [23] and magnesium alloys [24] and titanium alloys [26]. The flow stress behaviour is predicted by fitting experimental data into the constitutive numerical equations to determine material constants.

Most power plant structural components are made from P92 steel, especially the boiler pipes and tubes [27]. These power plant components operate at high temperatures and pressure to increase efficiency and reduce carbon dioxide
Therefore, the suitability of the plant components to operate under such severe working conditions require that the material has high strength and creep resistance. P92 steel exhibits superior properties such as corrosion resistance, high strength, high creep resistance and good fabricability for long-term power plant applications [16–20]. Compared with the other 9–12% Cr steels such as P91, E911 and P122 [14, 17, 21], P92 steels exhibit extended creep life and a high strength-to-weight ratio, which is especially useful for the production of boiler pipes and turbines sections. Literature reports that P92 steel has high strength, about 30% higher than P91 steel in prolonged power plant operating applications [23–25]. Thus, P92 steel has become a promising material for manufacturing components in modern power plant structural components such as boiler pipes.

The processing route of most plant components, such as boiler pipes, involves forming processes such as rolling, forging, drawing and extrusion [31]. The selection of the forming techniques largely depends on the required quality of the final product. Understanding metal flow behaviour during deformation is of great concern to engineers and designers [4]. The deformation behaviour of metals and alloys during forming is complex [32]. The complexity of deformation occurs due to deformation mechanisms caused by process parameters such as strain, strain rate and deformation temperature [33]. Therefore, there is a need to optimise the forming parameters to achieve high-quality products. Quantitative analysis of material flow response under different deformation conditions is therefore of interest. Constitutive equations that accurately predict the flow stress behaviour of a given material under given loading conditions must be developed.

More study efforts have been directed towards the deformation behaviour of P92 steel under creep conditions [30, 34, 35]. However, the hot workability behaviour of P92 steel is scarce in the literature. Therefore, there is a need to investigate the hot deformation characteristics of P92 steel over a wide range of loading conditions. The study will provide the required data for future analyses and in designing industrial schedules. In the current study, an attempt was made to investigate the effect of the metal forming parameters on the flow stress behaviour of P92 steel over a wide range of deformation conditions. A constitutive equation for predicting the flow stress behaviour of P92 steel was developed. The suitability of the equation in predicting the flow stress of P92 steel under the investigated conditions was verified using statistical tools: the correlation coefficient (R) and the average absolute relative error (AARE). Furthermore, forming parameters were optimised using processing maps and compared with the microstructure of the deformed samples.

### 2 Experimental procedure

The chemical composition of the P92 steel studied is given in Table 1. In the as-received condition, the steel exhibited a tempered martensitic microstructure and well-defined grain and lath boundaries, as shown in Fig. 1. The SEM micrograph shows precipitates along prior austenite grain boundaries. The visible carbides in the SEM-BSE micrograph are M23C6 (M = Fe, Mo, W, Cr) carbides. These carbides enhance solid solution strengthening, hence pinning dislocation movement during the deformation process [29]. A cylindrical specimen of 8-mm diameter and 15-mm

| Steel | C   | Mn  | Si  | Cr  | Mo  | Ni  | Cu  | Al  | V    | Nb  | W   | Co  |
|-------|-----|-----|-----|-----|-----|-----|-----|-----|------|-----|-----|-----|
| P92   | 0.11| 0.51| 0.22| 9.37| 0.50| 0.17| 0.27| 0.006| 0.19 | 0.130| 1.76| 0.028|

Fig. 1 SEM-BSE micrograph of P92 steel
length was subjected to a uniaxial compression test using a Gleeble® 3500. This equipment can physically simulate a hot compression test. The experimental data obtained can be used to design industrial processes.

Andrew’s model [36], as reported by Kim et al.[37] is given in Eq. (1) and was used in this study to determine the critical upper transformation temperature (Ae3) of the investigated steel. The Ae3 value of the investigated steel was 917 °C. This value (Ae3) is below the deformation temperature (850–1000 °C) used in this study. Hence, the hot deformation condition was at a single-phase austenite region during forming.

\[ \text{Ae}_3 \text{ (°C)} = 910 - 203 \sqrt{\text{C}} + 44.7 \text{Si} - 15.2 \text{Ni} + 31.5 \text{Mo} + 104.4 \text{V} + 13.1 \text{W} \]  

(1)

The uniaxial hot compression parameters were as follows: deformation temperature range of 850–1000 °C and strain rate range of 0.1–10 s⁻¹ under vacuum. The K-type thermocouples were welded at the midspan of the sample to measure test temperature and monitor the preset temperature during testing. Friction between the specimen and the ISO-T anvil was minimised by using graphite and nickel paste. All the tested samples were first heated to the austenitisation temperature of 1100 °C at a heating rate of 5 °C/s and held for 180 s for microstructure homogenisation to occur. Then, the specimen was cooled to the deformation temperature at a rate of 10 °C/s and soaked for 60 s to reduce the thermal gradient before compression. The samples were deformed to 60% (a true strain of 0.5), then rapidly air-cooled to room temperature.

3 Results and discussion

3.1 Flow behaviour

The flow stress–strain curves of hot uniaxial compression tests under different deformation conditions are shown in Fig. 1. The following observations were made.

- The study showed that the flow stress–strain curves were sensitive to the deformation conditions. At a given strain rate, the flow stress increased as the deformation temperature decreased and vice versa. At constant temperature, the flow stress increased as the strain rate increased.
- The flow stress–strain curve has two major stages: work hardening and dynamic softening. The flow stress–strain curves have characteristic points that indicate the change in the deformation mechanism.
- At the initial stage of deformation (strain ~ 0.2), the flow stress increased rapidly due to work hardening until the steady-state condition was reached caused by softening mechanisms such as dynamic recovery (DRV). The flow stress–strain curves for all the deformation conditions exhibited a rapid increase in flow stress up to 0.2 strain. The rapid increase in the flow stress shows high energy consumption for dislocation formation. Work hardening caused by accumulative dislocation density hinders deformation, resulting in higher flow stress. A rapid increase in the flow stress at earlier stages of deformation was due to the generation and multiplication of dislocations [38, 39]. Work hardening was the dominant deformation mechanism during the initial deformation.
- With a further increase in strain above 0.2, the slope of the flow stress–strain curves decreased until the flow stress attained a steady-state condition. This flow behaviour indicates that a dynamic softening had occurred. During this stage, flow curves show that work hardening and the dynamic recovery are in equilibrium, resulting in a relative constant dislocation density. This characteristic behaviour of the flow stress–strain curve suggests that the dominant softening mechanism is the dynamic recovery (DRV).
- The flow curves reflect the microstructural changes occurring during deformation, which are closely related to deformation mechanisms controlling the flow stress [40]. The deformation mechanisms such as work hardening (WH) and softening mechanisms (DRV and DRX) depend on the forming parameters. Steels such as P92 have high stacking fault energy, thus exhibiting DRV flow softening behaviour. The reason is they readily re-arrange into a polygonal sub-grain structure by cross-slip and climb dislocation during forming [41]. DRV becomes more pronounced as the temperature increases since high dislocation and grain boundary movements occur.
- From the flow stress–strain curves (Fig. 2), the study determined that dynamic recovery was the dominant softening mechanism under the conditions investigated for P92 steel.

3.2 Constitutive modelling

The relationship between the flow stress and deformation conditions (flow stress, temperature and strain rate) during high-temperature forming is widely described by using the Arrhenius-type equations [42] as follows:

\[ \dot{\varepsilon} = A f(\sigma) \exp \left( \frac{-Q}{RT} \right) \]  

(2)
where

\[ f(\sigma) = \sigma'' \] for low flow stress \((a \sigma < 0.8)\)

\[ f(\sigma) = \exp(\beta' \sigma) \] for high flow stress \((a \sigma < 1.2)\)

\[ f(\sigma) = [\sinh (a \sigma)]^n \] for all flow stress

In which \(\dot{\varepsilon}\) is the strain rate in \(s^{-1}\), \(T\) is the temperature (K), and \(Q\) is the activation energy of hot deformation in \(kJ.mol^{-1}\). The stress function \(f(\sigma)\), which is the maximum flow stress value (peak stress \(\sigma_p\), saturation flow stress \(\sigma_{sat}\) and steady-state stress \(\sigma_{ss}\)), \(n\), \(n'\), \(\beta\) and \(\alpha\) are material constants and \(\alpha \approx \beta'/n'\) [17].

However, the general case of the constitutive equation for analysing hot deformation behaviour developed by Sellars and Tegart [42] in Eq. (2) is commonly used to determine the material constants:

\[
\dot{\varepsilon} = A[\sinh (a \sigma)]^n \left[ \frac{-Q}{RT} \right]^{\frac{1}{n}}
\]  

By solving the universal hyperbolic sine function (Eq. (3)) determines the material constants (activation energy \(Q\) and stress exponent \(n\)) from linear graphs.

At a constant temperature,

\[
\frac{1}{n} = \frac{\partial \ln[\sinh (a \sigma)]}{\partial \ln \dot{\varepsilon}}
\]  

Similarly, at a constant strain rate \(\dot{\varepsilon}\),

\[
Q = Rn \frac{\partial \ln[\sinh (a \sigma)]}{\partial \frac{1}{T}}
\]  

The constitutive equations are used to calculate the material constants and activation energy for deformation using the peak stress or the steady-state stress [43]. However, the flow stress–strain curves in this study (Fig. 2) did not exhibit peak stress or a steady-state region after the peak. In this study, therefore, saturation flow stress \(\sigma_{sat}\) has been used to calculate the material constants and the activation energy [44]. The saturation flow stress occurs when an increase in the flow stress is limited by dynamic recovery, hence reaching a steady-state value. The saturation flow stress value for each deformation condition is taken directly from the stress–strain curve, as reported by Laasraoui and Jonas [44]. The saturation flow stress can be applied to model the flow stress behaviour over a wide range of deformation conditions [45].

From these equations, \(\alpha\) was determined as follows.
For power law (low flow stress equation)

\[ n' = \frac{\partial \ln \dot{\varepsilon}}{\partial \ln \sigma_{\text{sat}}} \]  

(6)

For the exponential law (high flow stress equation)

\[ \beta' = \frac{\partial \ln \dot{\varepsilon}}{\partial \sigma_{\text{sat}}} \]  

(7)

Then, \( \alpha \approx \beta' / n' \) is obtained by solving the averages of the material constants \( n' \) and \( \beta' \) in Eqs. (5) and (6) by using linear regression.

Substituting the saturation flow stress into the hyperbolic sine equation,

\[ \dot{\varepsilon} = A \sinh (\alpha \sigma_{\text{sat}})^n \frac{-Q}{RT} \]  

(8)

Then, using Eqs. (3) and (4) for solving the stress exponent \( n \) (Eq. (8)) and activation energy \( Q \) (Eq. (9)) becomes

\[ \frac{1}{n} = \frac{\partial \ln \sinh (\alpha \sigma_{\text{sat}})}{\partial \ln \dot{\varepsilon}} \]  

(9)

\[ Q = Rn \frac{\partial \ln \sinh (\alpha \sigma_{\text{sat}})}{\partial \frac{1}{T}} \]  

(10)

The Zener-Hollomon parameter \( Z \) provides information on the combined effect of deformation temperature and strain on the flow stress behaviour, as given in Eq. (11) [23, 46]. In this study, the Zener-Hollomon parameter was determined using the saturation flow stress \( \sigma_{\text{sat}} \) as follows:

\[ Z = \dot{\varepsilon} \exp \frac{Q}{RT} = A \sinh (\alpha \sigma_{\text{sat}})^n \]  

(11)

The natural logarithm of Eq. (11) and a plot of \( \sinh (\alpha \sigma_{\text{sat}}) \) vs. \( \ln Z \) can be used to determine the material constants \( A \) and \( n \) in the hyperbolic sine law by solving Eq. (12):

\[ \ln Z = \ln A + n \sinh (a \sigma_{\text{sat}}) \]  

(12)

The saturation flow stress \( \sigma_{\text{sat}} \) is expressed as

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

(13)

The materials constants \( n' \), \( \beta' \), \( \alpha \) and \( n \) and the activation energy \( Q \) in the constitutive equation were determined by using linear regression of Eqs. (6)–(10), as shown in Figs. 3 and 4 and summarised in Table 2.

From the flow stress data, the stress exponent \( n \) and activation energy \( Q \) were calculated using the hyperbolic sine equation proposed by Sellars and Tegart [42], referred to as the Arrhenius equations. The stress exponent \( n \) value obtained in this study was 8.0, which is higher than those reported in the literature ranging from 4.76 to 6.70 for P92 steel [15–17, 47]. The differences in stress exponent can be due to the forming conditions and the steel chemistry. A higher \( n \)-value may be due to the interaction between the precipitates, such as the \( \text{M}_2\text{C}_6 \) carbides and MX carbonitrides, with the mobile dislocations, thus hindering the plastic deformation [48]. From an industrial perspective, hot forming operations used to produce structural components, such as forged pipes, are typically done above 950 °C, and strain rates above 10 s\(^{-1} \) [49]. At lower temperatures, for example, below the austenite phase, different phases coexist in equilibrium. These phases hinder dislocation movement, resulting in higher stress exponents. The stress exponent values depend strongly on the deformation temperature. Process parameters influence microstructure evolution, hence affecting the flow stress. The effect of forming conditions was observed in the log–log plots of strain rate versus hyperbolic sine (\( \sinh (\alpha \sigma) \)) flow stress, where the curves were not parallel to each other (Figs. 3 and 4).

In this study, the activation energy was measured to be 488 kJ.mol\(^{-1} \). This \( Q \) value is consistent with the values...
reported in literature of 390 kJ.mol\(^{-1}\) [47], 437 kJ.mol\(^{-1}\) [16], 498.9 kJ.mol\(^{-1}\) [17] and 565 kJ.mol\(^{-1}\) [50]. The \(Q\) value obtained in this study was higher than those for self-diffusion of iron atoms in austenite of 270 kJ.mol\(^{-1}\) [51] and 250 kJ.mol\(^{-1}\) for \(\alpha\)-Fe in ferrite [52]. However, the calculated activation energy indicates the resistance to deformation under a specific set of deformation conditions [53].

Compared with the \(Q\) values reported in the literature, the variation in this type of steel can be due to alloying elements, especially chromium. The chromium content increases the stacking fault energy of the steel. A higher stacking fault energy affects the dissociation of dislocations, thus affecting the ability of dislocation cross-slip.

The higher activation energy can also be due to the high values of the stress exponent. The stress exponent is affected significantly by an increase in the flow stress. As the flow stress increases, the activation energy increases. From the literature, the chemical composition and the activation energies of P92 steels reported show some variability. According to Sun et al. [50], the critical phase transformation temperatures for P92 steel lie between 900 and 1000 \(^{\circ}\)C, which is in the austenite phase region. The accurate activation energy value is obtained when the deformation temperature is within the austenite region. During deformation, high deformation temperatures (~1300 \(^{\circ}\)C) cause dynamic strain softening, resulting in incorrect \(Q\) values. According to Carsi et al. [47], deformation in the austenite region experiences less resistance to dislocation motion due to the absence of carbides. Carbides impede dislocation motion, hence increasing the stress exponent. An increase in stress exponent increases the activation energy.

High amounts of chromium and other ferrite-forming elements such as W, V and Nb influence the microstructure of creep-resistant steels [54, 55]. The presence of solute atoms contributes to solution strengthening. Hence, this causes the formation of numerous precipitates that prevent dislocation movement during deformation resulting in higher activation energies [52, 56]. A study by Medina and Hernandez [57] and Suikkanen et al. [58] found that interstitial elements such as carbon and boron reduce activation energy since they enhance diffusion while substitutional alloying elements increase \(Q\). An increase in the number of substitutional alloying elements causes an increase in the activation energy [58] due to an accumulation of internally stored energy [59]. The variation in the activation energy can be due to the effect of alloying elements, which affect the formation of precipitates, as reported by Yang et al. [14]. Minor alloying and impurity elements in steels, such as Mn, Ti, Si, Al and Nb, contribute to high \(Q\) values [43, 60, 61] as they affect the diffusivity of Fe in gamma iron and delay the dynamic recrystallisation process [47].

Chromium can change the phase equilibrium of precipitates, hence affecting the number and composition of carbides or the amount of alloying elements in solid solution. The precipitate formation will affect the deformation process. The phases in P92 steels can be predicted using Cr-Ni equivalents and the Schaeffler diagram [62]. The Cr-Ni equivalents (in wt\%) are calculated from Eqs. (14) and (15) developed by Klueh and Maziasz [62] as follows:

\[
\text{Ni}_{eq} = \text{Ni} + 0.5\text{Mn} + 0.3\text{Cu} + 25\text{N} + 30\text{C} \quad (14)
\]

\[
\text{Cr}_{eq} = \text{Cr} + 2\text{Si} + 1.5\text{Mo} + 5\text{V} + 5.5\text{Al} + 1.75\text{Nb} + 1.5\text{Ti} + 0.75\text{W} \quad (15)
\]

| Steel | \(n\) | \(Q\) | \(\ln A\) | \(n'\) | \(\beta'\) | \(\alpha\) |
|-------|---|---|---|---|---|---|
| P92   | 8 | 488 | 47.63 | 10.62 | 0.058 | 0.0055 |
In steel with no nickel, when the $\text{Cr}_{eq} < 10$, $\delta$-ferrite will not form and the steel contains only martensite; for $\text{Cr}_{eq} = 10–12$, the steel contains both martensite and $\delta$-ferrite; and for $\text{Cr}_{eq} > 12$, only $\delta$-ferrite will form [63]. Using Eq. (14), the calculated $\text{Ni}_{eq}$ is 4.58, while Eq. (15) gives the $\text{Cr}_{eq}$ value to be 13.10. From the Schaeffler diagram (Fig. 5), the investigated steel lies on the $M/M + \delta$ boundary line. The Schaeffler diagram confirms the presence of $\delta$-ferrite. Similar to the numerical value of calculated $\text{Cr}_{eq}$ which was higher than 12. A higher $\text{Cr}_{eq}$ value causes lower ductility of the steel due to the presence of $\delta$-ferrite, hence lower hot workability. The formation of $\delta$-ferrite may have been the reason for the high-stress exponent and the activation energy obtained in this study.

Substituting the calculated material constants and activation energy values given in Table 2 into the hyperbolic sine law in Eq. (3), a constitutive equation of the saturation flow stress for the tested steel can be described by a hyperbolic sine equation as follows:

$$\dot{\varepsilon} = 4.85 \times 10^{20} \sinh (0.0055\sigma_{\text{sat}})^8 \exp \left[ \frac{-487560}{RT} \right]$$  \hspace{1cm} (16)

From Eq. (13), the saturation flow stress under different conditions for the steel investigated can be determined as

$$\sigma_{\text{sat}} = \frac{1}{0.0055} \ln \left[ \left( \frac{Z}{4.85 \times 10^{20}} \right)^{\frac{1}{8}} + \left( \frac{Z}{4.85 \times 10^{20}} + 1 \right)^{\frac{1}{8}} \right]$$  \hspace{1cm} (17)

By substituting the activation energy in Eq. (11), the $Z$ value was obtained. The $Z$-parameter shows the combined effect of deformation temperature and the strain rate. Figure 6 shows the linear relationship between $\ln Z$ and $\ln (\sinh (a\sigma_{\text{sat}}))$. The correlation index value was 0.993. This value ($R^2$) showed that $\ln Z$ had a high linear relationship with $\ln (\sinh (a\sigma_{\text{sat}}))$. The intercept of the $\ln Z$-$\ln (\sinh (a\sigma_{\text{sat}}))$
plot gave the ln A-value of 47.63. Therefore, A-value was $4.88 \times 10^{20}$, as summarised in Table 2.

### 3.3 Constitutive equation verification

The constitutive equation (Eq. (17)) developed in this work was verified by comparing experimental versus predicted flow stress data under hot deformation conditions, as shown in Fig. 7. The graph shows a linear relationship between experimental saturation flow stress and predicted saturation flow stress. The value of the correlation index ($R^2$) was 0.994. This value indicates a close relationship between the two parameters. The statistical parameters, Pearson’s correlation coefficient $R$ (Eq. (18)) and the average absolute relative error $AARE$ (Eq. (19)), were used to check the validity of the constitutive equation derived in Eq. (18). These parameters ($R$ and $AARE$) show the linear relationship and the ability of the developed constitutive model to predict flow stress [13, 14, 64].

$$R = \frac{\sum_{i=1}^{N} (M_i - \bar{M})(P_i - \bar{P})}{\sqrt{\sum_{i=1}^{N} (M_i - \bar{M})^2 \sum_{i=1}^{N} (P_i - \bar{P})^2}}$$ (18)

$$AARE(\%) = \frac{1}{N} \sum_{i=1}^{N} \left| \frac{M_i - P_i}{M_i} \right|$$ (19)

where $M$ is the experimental (measured) saturation flow stress, $P$ is the predicted saturation flow stress, $\bar{M}$ is the average value of $M$, and $\bar{P}$ is the average value of $P$.

The Pearson’s correlation coefficient ($R$) for the tested steel was 0.996, where 1.000 indicates perfect correlation. However, a good correlation between two variables does not necessarily depend on Pearson’s correlation coefficient. The predicting constitutive model may be biased towards the extremes (higher or lower) values, hence giving higher values of $R$ [2]. Therefore, statistical AARE analysis was used, which gives the accuracy of the developed constitutive model by determining the relative error of each flow stress term. The smaller the AARE values, the higher the accuracy of the constitutive model in predicting flow stress [65]. The percentage error was 1.91%. The low AARE values show the high accuracy of the constitutive model in predicting the flow stress [66]. Thus, the developed constitutive equation had sufficient statistical integrity to predict the flow stress for steel tested under the investigated hot deformation conditions.

### 3.4 Processing map

A processing map is an analysis tool used to determine the optimal processing parameters during hot forming. The technique applies a dynamic material model (DMM) [67]. DMM provides a way of identifying the optimum processing window to produce a defect-free component. This technique links the deformation mechanics with the microstructure evolution, hence describing the dynamic response of the microstructure during forming. The processing map delineates the ‘safe’ or ‘stable’ and ‘unsafe’ or ‘unstable’ deformation processing conditions. The construction of the processing map is done by superimposing the power dissipation map and the instability map [68]. The DMM has two major components: the power dissipation $G$ due to plastic deformation and the energy dissipation $J$ due to microstructural changes such as DRX, DRV and phase transformation [69, 70]. According to Narayana Murty and Nageswara Rao [71], Eq. (20) gives the total power dissipation:

$$P = G + J = \sigma \dot{\varepsilon} \int_{0}^{\varepsilon} \sigma d\varepsilon + \int_{0}^{\sigma} \dot{\varepsilon} d\sigma$$ (20)

In Eq. (20), $\sigma$ is the flow stress and $\dot{\varepsilon}$ is the strain rate. The power dissipation of the material is largely described by its constitutive response to the deformation conditions. The flow stress distribution at any given deformation temperature and strain can be expressed as a power law, as in Eq. (21) [72].

$$\sigma = K \dot{\varepsilon}^m$$ (21)

where $K$ is the stress coefficient, $\dot{\varepsilon}$ is the strain rate, $\sigma$ is the flow stress, and $m$ is the strain rate sensitivity of flow stress for the material. Taking natural logarithms on both sides of Eq. (21)
The strain rate sensitivity \( m \) is obtained from the slope of \( \ln \sigma \) against \( \ln \dot{\varepsilon} \):

\[
\ln \sigma = \ln K + m \ln \dot{\varepsilon} \tag{22}
\]

The strain rate sensitivity \( m \) is obtained from the slope of \( \ln \sigma \) against \( \ln \dot{\varepsilon} \):

\[
m = \frac{\partial \ln \sigma}{\partial \ln \dot{\varepsilon}} = \left( \frac{\partial J}{\partial G} \right)_{\varepsilon,T} \tag{23}
\]

For ideal power dissipation, \( m = 1 \) and the value of \( J = J_{\text{max}} = \frac{\sigma \dot{\varepsilon}}{2} \). The power dissipation efficiency \( \eta \) which defines the ability of the material to dissipate energy has been described by Narayana Murty et al. [73] as

\[
\eta = \frac{J}{J_{\text{max}}} = \frac{2m}{m + 1} \tag{24}
\]

Equation (24) indicates that the parameter \( \eta \) reflects the microstructure evolution mechanisms. Equation (24) implies that the percentage of power dissipated due to the microstructural changes to the total power dissipation during deformation. The power dissipation efficiency is obtained by considering several deformation conditions [66]. Then, the construction of the power dissipation map is made by

\[
\ln \sigma = \ln K + m \ln \dot{\varepsilon}
\]

plotting the strain rate against the deformation temperature at various power dissipation efficiency values [39].

During forming, defects may arise due to localised deformation [73]. Hence, to obtain high-quality products, deformation parameter optimisation is required, thus avoiding the instability regions [74]. Naryana Murty et al. [73] discussed several instability criteria. For stable materials, the Murty instability criterion is recommended [75].

\[
2m < \eta \tag{25}
\]

Solving the inequality of the stability condition in Eq. (25) gives

\[
\xi(\dot{\varepsilon}) = \frac{2m}{\eta} - 1 < 0 \tag{26}
\]

The instability values \( \xi(\dot{\varepsilon}) \) (Eq. (26)) are obtained using a cubic spline function under the deformation conditions. The temperature and strain rate plot at different values of \( \xi(\dot{\varepsilon}) \) develop the instability map [76]. A negative value of \( \xi(\dot{\varepsilon}) \) shows the instability regimes, which describe the flow...
instabilities such as adiabatic shear band formation, wedge cracking microvoids and flow localisation.

The DMM technique has received criticism and rejection by Montheillet et al. [77]. These authors argued that the model is not based on material laws but on a heuristic approach. They recommended strain rate sensitivity as a technique for optimisation. This parameter has a relationship with workability and not power dissipation efficiency. However, the DMM technique for studying the workability of metals and alloys is used widely for process parameter optimisation [70, 78, 79]. Many researchers have used this technique to approximate the optimal conditions over a wide range of deformation conditions. In the current study, the DMM technique was used.

The construction of processing maps is done by superimposing the instability maps onto the power dissipation maps. However, no marked differences between the dissipation and instability maps were observed in the present study. Hence, the contour maps are called deformation maps. The contour line numbers indicate the power dissipation efficiency value ($\eta$) in the dissipation maps and the flow instability value ($\xi$) in instability maps. The deformation maps obtained for the temperature range of 850–1000 °C and strain rates of $10^{-1}$–$10$ s$^{-1}$ at a true strain of 0.5 are shown in Fig. 8.

The highest power efficiency of 32% occurred at 1000 °C and 0.1 s$^{-1}$ deformation conditions, as shown in Fig. 8a,b. For 900–950 °C and 0.1 s$^{-1}$, the power efficiency was 17–26%. At a strain rate of 10 s$^{-1}$, there was a lower power dissipation efficiency of 8–14% at 850–900 °C and 1000 °C (Fig. 8a,b). The low power efficiency (850–900 °C, and strain rate of 10 s$^{-1}$) gives undesirable deformation conditions for this steel. The instability map, Fig. 8c,d, shows that

![Fig. 9 Sketch of a strain distribution and b stress distribution in a deformed sample using FEM (regions: I – intense shear zone, II – dead metal zone and III – moderate deformation zone)](image)

![Fig. 10 SEM-BSE micrographs of deformed samples under different deformation conditions (region I). The yellow arrows show the grain boundaries, while the light blue arrows show the laths](image)

![a) 850 °C/0.1 s$^{-1}$](image)

![b) 850 °C/10 s$^{-1}$](image)

![c) 1000 °C/0.1 s$^{-1}$](image)

![d) 1000 °C/10 s$^{-1}$](image)
this region had the lowest instability value of ~0.04–0.06. A lower value of power efficiency and instability is an indication that the material has low ductility. Metalworking in this region (850–900 °C and strain rate of 10 s⁻¹) may cause a ductile fracture. The deformed sample microstructure in the instability region (850–900 °C and strain rate of 10 s⁻¹) is shown in Fig. 10b. The instability region exhibited flow localisation. As the deformation temperature increased to 950 °C and the strain rate of 10 s⁻¹, the power efficiency was slightly higher at 14–20%. During deformation at a strain rate of 1 s⁻¹, the power efficiency was 14–23% for all the deformation temperatures.

During the uniaxial compression test, the cylindrical specimen undergoes barrelling due to the effect of friction at the specimen-die interface. Inhomogeneous deformation occurs due to friction between the die and the workpiece, causing strain variation in the deformed specimen. The metal flow pattern results in three distinct regions; Deform™ 3D finite element software can illustrate these regions, as shown in Fig. 9.

- Region I, located at the centre of the specimen, experiences intense shear. This region extends to the corners of the specimen (Fig. 9).
- Region II – dead metal zone (DMZ) occurs at the die-workpiece interface, which experiences low strain.
- Region III at the outer lateral surface of the specimen having drum-shaped geometry experiences moderate strains. However, the region experiences the maximum tensile stress during deformation, hence acting as the point of crack initiation and propagation. Each of these

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**Fig. 11** SEM-BSE micrographs at a constant temperature of 850 °C and different strain rates. The yellow arrows show grain boundaries, light blue arrows show the laths, and white arrows show a localised flow.
three regions was used to explain the microstructure evolution during deformation.

To this end, note that the microstructure evolution during forming varies, as shown by the different deformation regions in the FEM images (Fig. 9).

The SEM micrographs for the deformed steel samples for the region I after deformation temperatures of 850 °C and 1000 °C for strain rates of 0.1 s⁻¹ and 10 s⁻¹ are given in Fig. 9. The SEM-BSE micrographs show that the microstructures had a similar lath martensitic structure for all the deformation conditions except at 850 °C/10 s⁻¹ which showed a localised deformation (Fig. 10b). The microstructure instability occurs at a higher strain rate and lower deformation temperature during deformation, hence affecting flow behaviour. Flow instabilities may occur at low temperatures and higher strain rates [80]. Instability occurs due to the generation of localised heat during plastic deformation. At a higher strain rate, the time to conduct the heat generated to other specimen regions is insufficient and local temperature spikes occur [67]. The internal heat generated causes a decrease in the flow stress due to this temperature rise. This temperature change results in the possibility of localised plastic flow. The processing map indicates that the instability region had the lowest power energy dissipation efficiency of 8% (Fig. 8), indicating that a lower amount of energy was available for the microstructure evolution. The microstructures of regions II and III are as shown in Figs. 11 and 12, respectively. The SEM-BSE micrographs showed elongated lath martensite at deformation conditions of 850 °C and 1000

![Region II](image1.jpg)

![Region III](image2.jpg)

**Fig. 12** SEM-BSE micrographs at constant temperature 1000 °C under different strain rates. The yellow arrows show grain boundaries, and light blue arrows show laths
4 Conclusion

Uniaxial hot compression of creep-resistant steel ASTM A335 P92 steel was conducted over a wide range of deformation temperatures (850–1000 °C) and strain rate of 0.1–10 s⁻¹ for a strain of 0.6 in the Gleeble®3500 thermal–mechanical equipment. The flow stress data obtained from the experimental test was employed to develop the constitutive equations and processing maps. The following conclusions are drawn from this study.

1. The flow stress–strain curves showed that the deformation process was influenced by work hardening and dynamic recovery (softening). The flow stress decreased with an increase in deformation temperature or a decrease in strain rate.

2. The hot deformation parameters were stress exponent (8.0) and activation energy (487.56 kJ.mol⁻¹). The constitutive equation developed for P92 steel is as follows:

\[ \dot{\varepsilon} = 4.85 \times 10^{30} \sinh(0.0055\sigma_{\text{sat}}) \exp\left[\frac{-487560}{RT}\right] \]

\[ \sigma_{\text{sat}} = \frac{1}{0.0055} \ln \left[ \left( \frac{Z}{4.85 \times 10^{20}} \right)^{\frac{1}{8}} + \left( \frac{Z}{4.85 \times 10^{20}} + 1 \right)^{\frac{1}{8}} \right] \]

\[ Z = \frac{488000}{RT} = 4.85 \times 10^{20} \left(\sinh(0.0055\sigma_{\text{sat}})\right)^{8} \]

3. The developed constitutive equation was verified using statistical parameters: Pearson’s correlation coefficient \( R \) (0.996) and absolute average relative error AARE (1.91%). These parameter values show a good correlation between the predicted and the experimental data.

4. Processing maps help to describe the complexity of metal forming, thus providing a means of obtaining the optimal hot forming parameters. In this study, the optimal processing conditions occurred at a lower strain rate of 0.1 s⁻¹ and deformation temperatures of 900–950 °C and 1000 °C for the steel investigated.

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Availability of data and material The authors confirm that the data supporting the findings of this study are available within the article.

Code availability Not applicable.

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