The use of the direct impact Hopkinson pressure bar technique to describe thermally activated and viscous regimes of metallic materials

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The influence of strain rate over domains involving the thermal activation and the viscous drag behaviour of the dislocations is discussed. While it is recognized that the Koslky–Hopkinson technique or split Hopkinson pressure bar technique can generate data up to the upper strain-rate limit of the thermal-activated regime, it is necessary to use a direct impact Hopkinson pressure bar technique to access the viscous regime. Data generated with this technique are presented for a series of metals, including steel, nickel, copper and tungsten alloys. The motivation to generate such data is provided through three industrial applications.

1. Introduction

The study of structures subjected to impact requires taking into account the influence of loading rate on the strength of the materials. When dealing with low-impact velocity up 10 m s⁻¹ for metals, the associated bulk strain rate does not exceed 10³ s⁻¹. From quasi-static strain rates, 10⁻⁴ s⁻¹, to these dynamic strain rates the deformation mechanisms of the dislocations are thermally activated. It results a moderate increase of the strength that is linearly dependent of the strain rate. From 10⁻³ to 10³ s⁻¹, the resulting increase for metals is about 10–40% for face-centred cubic (FCC) systems, 10–20% for hexagonal closed packed systems and 50–300% for body-centred cubic (BCC) systems. These results come from data generated with conventional testing machines up to 1 s⁻¹, and on the other hand, from data

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generated with the Koslky–Hopkinson technique or split Hopkinson pressure bar technique (SHPB) up to $5\times 10^3 \, s^{-1}$ [1,2]. Numerical simulations of impacted engineering structures at velocities less than $10 \, m \, s^{-1}$ provide satisfactory results when associated material models take into account this thermal dependence such as the classic Johnson–Cook model [3]. This is the case for the majority of the crashworthiness studies.

For impact velocities exceeding $10 \, m \, s^{-1}$, plastic deformation occurs at strain rates ranging from $10^3$ to $10^5 \, s^{-1}$, where the motion of dislocation is slow down through a viscous drag phenomenon [4]. It results in a tremendous increase of the flow stress. When material models do not take into account this strengthening, plastic deformation is overestimated.

The SHPB technique is not capable of establishing this increase of the flow stress in the $10^3$–$10^5 \, s^{-1}$ strain-rate regime. This is because the maximum strain rate, $\dot{\varepsilon}_M$, of the SHPB technique is directly related to the yield stress of the incident pressure bar, $\sigma_y$. When considering an impactor and incident pressure bar of the same impedance, the maximum impact velocity, $V_M$, which can be applied without plastically deforming the incident pressure bar is imposed by

$$\sigma_y = \frac{1}{2} \rho C_o V_M,$$

where $\rho$ and $C_o$ are the density and sound velocity of the impactor and incident bars, respectively.

This maximum velocity implies a maximum strain rate, $\dot{\varepsilon}_M$, given by

$$\dot{\varepsilon}_M = \frac{V_M}{2L_o},$$

where $L_o$ is the specimen length. With an impactor and an incident bar having a yield stress of 1300 MPa and a specimen length of 5 mm, a maximum impact velocity of 67 $m \, s^{-1}$ can be applied resulting in a maximum strain rate of $7\times 10^3 \, s^{-1}$.

One can increase this maximum strain rate by miniaturizing the SHPB technique with 1 mm thick specimens as proposed by Jia & Ramesh [5]. However, the use of such size limits to materials with a characteristic microstructural length less than to $50 \, \mu m$ in order to respect the ratio of 20 between this characteristic length and the specimen size as representative.

In order to access the $10^3$–$10^5 \, s^{-1}$ strain-rate regime, it is necessary to apply higher velocity to the specimen. For this purpose, a modified version of the Hopkinson bar technique has been developed which consists of the direct impact of the specimen [6]. This technique, referred as the direct impact Hopkinson pressure bar (DIHPB) technique, permits strain rates to be generated from $3\times 10^3$ to $10^5 \, s^{-1}$. With this technique, a precise description of the strengthening originating from the viscous behaviour of the dislocations can be obtained.

This work recalls the procedure of the DIHPB and provides data that have been generated for a series of metals. The motivation to generate such data is provided through several industrial applications for which the strengthening occurring in the viscous regime plays a major role.

2. The direct Hopkinson pressure bar technique

The DIHPB technique, introduced by Dharan & Hauser [6] and revisited by Klepaczko [7], consists of impacting at a constant velocity a specimen placed against a Hopkinson pressure bar, as shown in figure 1. The impactor speed, $V_i$, typically between 30 and 150 $m \, s^{-1}$, is recorded just prior to impact. A strain gage is placed on the Hopkinson bar in order to provide the loading history of the specimen.

The impact velocity is assumed to be constant during the tests. This hypothesis holds when the impactor mass is large enough to provide a kinetic energy much greater that dissipated by plastic deformation in the specimen. Several configurations of the impactor have been used. In the original version, the diameter of the impactor bar was much larger than the transmitted bar diameter in order to be able to make the assumption of an undeformable impactor. Since then, another configuration has been developed by Gorham [8] with an impactor and a transmitted bar of the same diameter and impedance. This configuration provides a more precise measurement of the strain and strain rate, as it will be shown next [7].
The engineering deformation, \( e(t) \), is obtained from the specimen end displacements. The specimen/bar interface displacement, \( u_2(t) \), is deduced from the strain history, \( \varepsilon_T(t) \), of the Hopkinson bar

\[
\begin{align*}
    u_2(t) &= -C_o \int_{0}^{t} \varepsilon_T(t) d\tau 
    \quad \text{(2.1)}
\end{align*}
\]

with \( C_o \) the sound speed of the Hopkinson bar material. The specimen/impactor interface displacement, \( u_1(t) \), is deduced from the impactor speed \( V \), and the assumption of equilibrium at both specimen interfaces

\[
\begin{align*}
    u_1(t) &= Vt - C_1 \int_{0}^{t} \varepsilon_T(t) d\tau 
    \quad \text{(2.2)}
\end{align*}
\]

with \( C_1 \) the sound speed of the impactor material. For the case of the Hopkinson bar and impactor made of the same material, the engineering deformation is given by

\[
\begin{align*}
    e(t) &= u_2(t) - u_1(t) = \frac{2C_o \int_{0}^{t} \varepsilon_T(t) d\tau - Vt}{L_o}, 
    \quad \text{(2.3)}
\end{align*}
\]

where \( L_o \) is the specimen length. The true deformation is then deduced

\[
\begin{align*}
    \varepsilon(t) &= \ln(1 + e(t)) = \ln \left( \frac{L_o + 2C_o \int_{0}^{t} \varepsilon_T(t) d\tau - Vt}{L_o} \right). 
    \quad \text{(2.4)}
\end{align*}
\]

The engineering axial stress is obtained from the strain history, \( \varepsilon_T(t) \), of the Hopkinson bar

\[
\begin{align*}
    s_x(t) &= \frac{\rho C_o^2 \varepsilon_T(t) S_T}{S_o}. 
    \quad \text{(2.5)}
\end{align*}
\]

With the assumption that plastic deformation takes place at constant volume, the specimen true axial stress is deduced

\[
\begin{align*}
    \sigma_x(t) &= s_x(t)(1 + e(t)) = \frac{\rho C_o^2 \varepsilon_T(t) S_T(1 + e(t))}{S_o}, 
    \quad \text{(2.6)}
\end{align*}
\]

where \( \rho \) is the density of the Hopkinson bar, \( S_T \) and \( S_o \) are the cross-sectional area of the Hopkinson bar and specimen, respectively. Owing to inertia, the radial and tangential stresses, \( \sigma_r(t) \) and \( \sigma_\theta(t) \) are no-zero [7]:

\[
\begin{align*}
    \sigma_r(t) &= \sigma_\theta(t) = \left( \frac{3}{8} \right) \rho \left( \frac{R_o V}{L_o} \right)^2 (1 - e(t))^{-3}. 
    \quad \text{(2.7)}
\end{align*}
\]

The equivalent stress is then derived

\[
\begin{align*}
    \sigma(t) &= \sigma_x(t) - \sigma_r(t) = \frac{\rho C_o^2 \varepsilon_T(t) S_T(1 + e(t))}{S_o} - \left( \frac{3}{8} \right) \rho \left( \frac{R_o V}{L_o} \right)^2 (1 - e(t))^{-3}. 
    \quad \text{(2.8)}
\end{align*}
\]

The true strain rate is given by

\[
\begin{align*}
    \dot{\varepsilon}(t) &= \frac{2C_o \varepsilon_T(t) - V}{L_o(1 + e(t))}. 
    \quad \text{(2.9)}
\end{align*}
\]
Figure 2. (a,b) Stress and strain-rate responses versus strain for a copper specimen impacted at 18.7 and 50.7 m s$^{-1}$.

Figure 2 provides the data generated with the DIHPB technique for a copper specimen 3.25 mm in diameter and 5 mm in height impacted at 18 and 50 m s$^{-1}$. The copper has quasi-static yield strength of 155 MPa and a grain size of 26 μm. The DIHPB technique used consists of an impactor and a Hopkinson bar, 20 mm in diameter, made of the same tungsten alloy, 17.5 g cm$^{-3}$ in density and 1500 MPa in yield stress. The impactor speed, $V$, was recorded using two lasers beams separated by 18 mm and positioned 30 mm prior impact. The strain rate in the specimen increases during the deformation. Such increase comes from the fact that the displacement rate is mainly imposed by the impactor speed. Because the impactor speed is constant, the increase of the strain rate is due to the specimen length reduction. When considering an SHPB test, generally no such increase in strain rate is observed. This comes from a displacement rate directly related to the reflected stress wave intensity which is itself dependent on the specimen strength. When the specimen work hardens, the intensity of the reflected stress wave decreases. This decrease compensates for the specimen length reduction resulting in an almost constant strain-rate test with the SHPB configuration.

With regard to radial and tangential stresses induced by inertia, there were found to be negligible when considering the copper test conducted at 50.7 m s$^{-1}$ (figure 2). Precisely, the induced radial and tangential stresses at a strain of 0.30 and a strain rate of $2.0 \times 10^4$ s$^{-1}$ are less than 1.2% of the axial stress. A strain-rate limitation $\dot{\varepsilon}_{\text{lim}}$ with regard to the ratio $r$ between the inertia stresses and the axial stress has been proposed by Jia & Ramesh [5] based on the work of Malinowski & Klepaczko [9] and Gorham [10]

$$\dot{\varepsilon}_{\text{lim}} = L^{-1} \left( \frac{4.36 r \sigma_y}{\rho} \right)^{1/2},$$

(2.10)

where $\rho$ and $\sigma_y$ are the density and yield strength of the specimen, respectively. Applied to the copper test conducted at 50.7 m s$^{-1}$ with a yield strength of 550 MPa, see figure 2, the strain-rate result for a ratio $r$ of 5% is $2.3 \times 10^4$ s$^{-1}$. Such limitation confirms that the tremendous increase of the flow stress for copper with the strain rate increasing from $10^{-3}$ s$^{-1}$ (155 MPa) to $2.0 \times 10^4$ s$^{-1}$ (550 MPa) is real.

3. Thermally activated and viscous regimes of body-centred cubic and face-centred cubic metals

Stress–strain-rate data generated in the $10^3$–$10^5$ s$^{-1}$ range for FCC, BCC and FCC/BCC metals with the DIHPB technique are provided in figure 3. The technique provides a precise description of the yield stress occurring in the $10^3$–$10^5$ s$^{-1}$ strain-rate regimes. The strain-rate threshold characterizing the strengthening of the FCC metals associated with the viscous behaviour of the dislocations was found to be within $1.8 \times 10^3$–$4 \times 10^3$ s$^{-1}$ (figure 3a,b). When considering the
copper, the threshold was found to be independent of the grain size. Again for nickel, with a grain size down to 130 nm, the threshold strain rate was found to be invariant.

There are several material models describing the upturn of the strength in the $10^3$–$10^4$ s$^{-1}$ regimes [11–16]. For example, a modification of the strain-rate term in the Johnson–Cook model has been used which includes a power strain-rate component added to the logarithm strain-rate term of the classic Johnson–Cook formulation, $D(\dot{\varepsilon}/\dot{\varepsilon}_1)^k$, with $D$ and $k$ two constants [11]. The model has been found to reproduce room temperature data up to $2 \times 10^4$ s$^{-1}$ and high temperature data ($1000^\circ$C) up to 1 s$^{-1}$ for a nickel. The strain-rate term is normalized by a reference strain rate, $\dot{\varepsilon}_1$, characterizing the transition between the thermally activated regime and the viscous regime, taken to be $10^3$ s$^{-1}$. The equivalent stress function of plastic strain, $\varepsilon_p$, strain rate, $\dot{\varepsilon}$ and temperature, $T$, is expressed as

$$
\sigma = (A + B\dot{\varepsilon}_p^n) \left(1 + C \ln \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_o} \right) + D \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_1} \right)^k \left(1 - \left[ \frac{T - T_r}{T_m - T_r} \right]^m \right) \right)
$$

(3.1)

with $T_r$, $T_m$ the room and melting temperatures, $\dot{\varepsilon}_o$, the reference strain rate equal to 1 s$^{-1}$, and $A$, $B$, $C$, $n$, $m$ are the five constants of the classic Johnson–Cook model. This formulation enables one to return to the classical formulation when strain rates are lower than $10^3$ s$^{-1}$.

Figure 3. Yield stress versus strain rate in compression of (a) coppers, (b) a stainless steel, (c) nickels, (d) an aluminium alloy, (e) a tantalum and (f) tungsten alloys. (Online version in colour.)
The yield stress versus strain-rate data for a nickel can be reproduced precisely using the modified Johnson–Cook model, as shown in figure 4. Constants of this model for a nickel, aluminium alloy and a stainless steel are provided in table 1.

4. Evidences of strengthening at high-impact velocities

(a) Low-strength metallic projectiles

The necessity to take into account the strengthening occurring in the viscous regime for impacted structures involving low-strength metals, 200–300 MPa, at is the case in gas tank failure events. When a gas tank detonates, valve components made of low-strength steel projectiles are generated which can perforate neighbouring gas tanks. In the majority of perforation studies, the penetrator exhibits high strength, usually greater than 700 MPa allowing the hypothesis of a rigid penetrator to be made. With such hypothesis, the critical speeds of perforation can be reproduced numerically using material models limited to the thermally activated dependence of the yield stress with strain rate, such as the classic Johnson–Cook model.

When considering low-strength penetrators, the perforation speed and the final diameter of the projectile head are overestimated [17]. In this study, a stainless steel projectile 36 mm in length and 9 mm in diameter, 260 MPa in yield stress, was sent against an aluminium plate 6 mm in thickness, 280 MPa in yield stress, at velocities ranging from 100 to 200 m s\(^{-1}\). Using the classical formulation of the Johnson–Cook model [3], the perforation speed and the final diameter of the projectile head are overestimated (figure 5a). This is owing to the excessive plastic deformation occurring when strain rates exceed 3000 s\(^{-1}\).

With the modified version of the Johnson–Cook model, see table 1, the plastic deformation of the projectile head is reduced providing numerical diameters comparable to the experimental data (figure 5a). Along with these data, numerical prediction of the perforation speed and residual velocities of the projectile are found to fit the experimental data (figure 5b). Because

Table 1. Constants of the modified Johnson–Cook model for a series of metals.

<table>
<thead>
<tr>
<th>Material</th>
<th>Yield Stress (MPa)</th>
<th>A (MPa)</th>
<th>B (MPa)</th>
<th>C</th>
<th>D</th>
<th>k</th>
<th>m</th>
<th>n</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nickel</td>
<td>350</td>
<td>200</td>
<td>800</td>
<td>0.01</td>
<td>0.25</td>
<td>0.50</td>
<td>0.3</td>
<td>0.6</td>
</tr>
<tr>
<td>Aluminium alloy</td>
<td>280</td>
<td>350</td>
<td>10</td>
<td>0.02</td>
<td>0.20</td>
<td>0.45</td>
<td>3.0</td>
<td>0.5</td>
</tr>
<tr>
<td>Stainless steel</td>
<td>300</td>
<td>190</td>
<td>270</td>
<td>0.01</td>
<td>0.52</td>
<td>0.80</td>
<td>0.2</td>
<td>0.4</td>
</tr>
</tbody>
</table>

Figure 4. Yield stress versus strain rate for a nickel against two formulations of the Johnson–Cook model.
the loading conditions at the head of the projectile are mainly in the viscous regime, the plastic deformation is minimized owing to the strengthening effect that allows these perforation processes to be reproduced.

(b) Explosively formed projectile

Another example where the strengthening occurring at high strain rates plays a major role is with explosively formed projectiles (EFPs). EFPs consist of the loading of a concave metal disc usually made of tantalum, pure iron or nickel by detonating an explosive placed against the liner. The loading history starts with a shock phase followed by the shaping of a projectile occurring at strain rates ranging from $1 \times 10^4$ to $5 \times 10^4$ s$^{-1}$. Again, when dealing with a material model limited to the thermally activated regime, the plastic deformation is overestimated (figure 6a,b) [18]. These simulations result in an elongated projectile of length that is more than twice the length observed experimentally. With the modified version of the Johnson–Cook model described in figure 4 and table 1, plastic deformation is contained resulting in more realistic numerical results with a matching diameter and an underestimation of 55% of the length observed experimentally (figure 6c).

(c) Charpy energy at high-impact velocities

Very high Charpy energies have been recorded at impact velocities ranging from 30 to 160 m s$^{-1}$ for a series of tungsten alloys and a stainless steel using a round bar Charpy experiment [19]. The experiment consists of the dynamic loading of a three-point bend specimen 60 mm in length with an impactor bar 10 mm in diameter and 84 mm in length. The specimen is placed against two Hopkinson pressure bars 10 mm in diameter so as to record the contact loads. The impactor
guided by a Teflon sabot is fired using a gas gun. A specific set-up has been developed to arrest the Teflon sabot, and to guide the impactor up to impact through continuous sliding of the impactor in the Teflon sabot. The velocity at impact, $V_i$, and after impact, $V_f$, of the impactor is deduced from high-speed camera observations. The kinetic energy of the broken specimen after impact was found to be less than 1% of the total energy absorbed by the specimen [19]. Consequently, the difference between the kinetic energy of the impactor before and after impact provides a good estimate of the energy consumed by the Charpy-type specimen.

The Charpy energy as a function of the impactor impact velocity for the stainless steel whose yield stress is 620 MPa (figure 3b) is reported in figure 7. These data have been generated with notched round specimens 8.4 mm in diameter and conventional notched 5 × 5 mm$^2$ Charpy specimens. Up to 30 m s$^{-1}$, the failure energy is about constant. A completely different response is obtained when higher impact velocities are applied to both types of specimens, with a surprising increase of the failure energy exceeding by 10 times the conventional Charpy energy at an impact velocity of 140 m s$^{-1}$.

Numerical simulations of the Charpy experiment reveal tangential strain rates reached at the initiation failure site of the three-point bend exceeding $5 \times 10^3$ s$^{-1}$ at an impact velocity of 30 m s$^{-1}$, to reach $2.4 \times 10^4$ s$^{-1}$ at an impact velocity of 140 m s$^{-1}$ (figure 8). Based on the yield
stress–strain-rate data of figure 3b, the loading conditions of the Charpy specimens are well within the strengthening occurring in the viscous regime.

The failure process in the entire section of the Charpy specimen occurs via a spherical void and growth coalescence (figure 9a,b). The void size was found to be independent of the impact velocity. Such results indicate that the failure strain is constant, implying a Charpy energy increasing with increase of the material strength. Because the loading conditions of the Charpy specimens are well within the viscous regime, the associated strengthening contributes to the increase of the Charpy energy with increase of the impact velocity.

5. Conclusion

The DIHPB is a technique allowing the measurement of the strengthening occurring at high strain rates when the viscous behaviour of the dislocations prevails. Data generated with a series of FCC and BCC metals reveal a transition from the thermally activated behaviour of the dislocations to the viscous regime occurring at strain rates ranging from $1 \times 10^3$ to $5 \times 10^3 \text{ s}^{-1}$.

This material strengthening has to be taken into account when impact velocities exceed $10 \text{ m s}^{-1}$. Several material models are available to describe the upturn in the strength in the viscous regime. Using data generated with the DIHPB technique, these models permit numerical reproduction of events involving large plastic deformation in the $10^4 \text{ s}^{-1}$ regime. The use of such a model was found to reproduce the interaction of projectiles made of low-strength steels with a metallic target, and EFPs. The strengthening occurring in the viscous regime was found to be a means to interpret the high Charpy energies recorded at impact velocities exceeding $30 \text{ m s}^{-1}$ for a stainless steel.

References


