Measurements of the mechanical response of Indium and of its size dependence in bending and indentation

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Abstract

Tension, compression, three-point bending and indentation experiments are conducted on high purity Indium at room temperature and low strain rates. The material displays a ductile viscoplastic response, found to be size-independent in tension and compression. Simple analytical models are constructed to aid interpretation of the test results and detection of a size effect in bending and indentation, associated to a length-scale of order 50-100 µm.

Keywords: A. creep; plastic collapse; strengthening mechanisms; B. elastic-viscoplastic material; C. mechanical testing.

1. Introduction

Indium has wide applications in the electronics industry and potential uses as a propeller in Field-Emission Electric Propulsion micro-thrusters; it was a strong contender for use in the ESA Pathfinder mission [1]. Pure polycrystalline Indium at room temperature is a soft and ductile metal with a low melting point (157°C) and strength of order of few MPa. This metal is not suitable for structural applications at mm scale and above, and as a consequence there are only a few published studies focusing on the mechanical properties of Indium. At room temperature (22°C) Indium is at a homologous temperature of 0.69 and deforms by power-law creep. This paper will report measurements of the mechanical response of the material in its power-law creep regime and of the dependence of such response upon specimen size.

Ashby [2] has proposed that geometrically necessary dislocations (GND) associated with gradients of plastic strain enhance material strength and lead to size effects in the micron range. This has led to a large number of investigations into size effects in plasticity. For example, an increased yield strength with decreasing specimen size has been observed in torsion of thin metal wires by Fleck et al. [3] and in microbending experiments Stolken and Evans [4]. Likewise, an increase in the hardness of metals with decreasing indentation depth has been observed by several authors ([5]-[8]). Strain gradient plasticity theories have also been developed by several authors to account for the observed size effects (e.g. [9]-[13]). These theories contain one or more material length scales and typically predict a strength elevation when the specimen size is comparable with such length scale. The studies above focused on metals at either room temperature; some studies included the effects of higher...
temperatures, but not sufficiently high to activate creep mechanisms (e.g. [14]); other studies investigated the effect of ageing ([15]) on material length-scales, reporting small variations of these.

Existing research has focused on metals displaying an approximately rate-independent plastic response; investigations of the size dependence of the plastic response of metals in the creep regime are lacking. This paper explores the size dependence of the mechanical response of Indium at room temperature, and therefore in its power-law creep regime.

Indium is difficult to test, machine and process thus experimental studies on this material are scarce. Weertman [16] conducted constant load tensile tests on indium wires at room temperature, and observed a creep exponent of $N \approx 0.2$. Lucas and Oliver [17] performed instrumented indentation on indium with a Berkovic indenter, again determining a creep stress exponent of $N \approx 0.2$. Galli and Oyen [18] conducted room temperature indentation tests with a spherical indenter to a maximum depth of order 10 µm and determined values of the creep exponent between 0.1 and 0.2. Sapozhnikov et al. [19] focused on internal friction measurements of Indium, while Lee et al. [20] conducted compression tests on Indium nano-pillar, measuring a flow stress of order 100 MPa at this length-scale.

This study presents the results of a comprehensive investigation including uniaxial compression and tension, bending and indentation tests on high-purity Indium, exploring the dependence of the mechanical response upon strain rate and specimen size. The paper is organized as follows: details of the experimental methods are provided in section 2. To aid interpretation of test results, simple analytical models are constructed in section 3. Results are presented and discussed in section 4.

2. Experimental Investigation

2.1 Material and specimen preparation

All specimens were manufactured from a high-purity (99.999%), isotropic Indium ingot with initial grain size of order 0.5 mm. The material was dense but soft and the ingot could be cut, with a heavy duty knife, in small pieces. Such pieces were slowly hot-pressed, at a temperature of 80 °C and in presence of abundant lubrication of the loading platens, between two microscope glass slides, to obtain flat indium plates of smooth surface finish and different thickness in the range 3-10 mm. These plates served as indentation specimens and as a base material for production of the other specimen types. In preliminary experiments, with the help of microhardness tests, it was found that the large plastic deformation imposed during hot-pressing hardened the material substantially. An annealing procedure was developed by trial and error in order to restore the material hardness to its original value; the optimal procedure comprised gradual heating from room temperature to 80 °C in 30 min, holding this temperature for 4 hours, then cooling to room temperature (22 °C) in 1 hour.

Specimens for compression tests were produced starting from Indium plates of thickness 10 mm and 5 mm. Two circular punches with diameters 6 mm and 3 mm were used to extract from the plates small cylindrical specimens with axis along the through-thickness direction. Subsequently these cylinders, of slightly irregular shape, were slowly pressed into cylindrical nylon moulds (of diameters 3.5 and 8 mm, respectively) by hand; the excess material left outside the mould was removed with a razor blade to
obtain a flat and smooth test surface. After this manufacturing step, the annealing procedure described above was repeated prior to testing.

Specimens for tension and bending required thinner Indium plates; these were manufactured by slowly hot-rolling (80 °C) the 5 mm Indium plates between two polished stainless steel rolls, allowing production of indium sheets of thickness in the range 0.25 – 2 mm; such sheets were used to manufacture bending and tension specimens. Rectangular specimens for bending tests were cut with a razor blade from Indium sheets of different thicknesses. Dogbone shaped specimens were used in the tension tests, with width 2 mm and other dimensions as in Fig. 1. These specimens were extracted from the Indium sheets with a Zwick/Roell ZCP punch. Bending and tension specimens were also annealed prior to testing. All tests took place at temperature in the range 22 - 23 °C.

2.2 Uniaxial tension tests
Tensile tests were conducted with a Deben M200 micro-tester of 200 N capacity. The specimens were gripped by flat steel plates and forced to elongate, while the tensile force was measured by a resistive load cell. Two reflective strips were glued in the gauge portion of the dogbone specimen to measure strain via a laser extensometer. All tests were conducted at a nominal strain rate of $1.7 \times 10^{-4} \text{s}^{-1}$.

2.3 Uniaxial compression tests
Circular cylinders of diameter 3.5 and 8 mm were tested in compression using both an Instron 5969 machine, the micro-tester described above and a DMTA analyzer (TA Instruments, RSA-G2). Strain was measured via either image analysis of a video footage or a laser extensometer. Tests were conducted at nominal strain rates of $10^{-1}$, $10^{-2}$ and $10^{-3} \text{s}^{-1}$ on the two different specimen sizes.

2.4 Three-point bending tests
A bespoke three-point bending fixture was designed and 3D-printed in ABS plastic; this consisted of two rollers of radius 1 mm at a span of $L = 20$ mm, and of a loading roller of the same radius. These were connected to an Instron tensometer by mechanical fastening and were used to load in bending beams of Indium with thicknesses ranging from 0.25 to 2 mm. The bending force was measured by a resistive load cell of capacity 10 N, the deflection was measured via a laser extensometer of resolution 1 μm. Beams of different thickness were tested and force versus deflection histories were recorded. The rate of deflection of the loading roller was chosen as $\dot{\delta} = \dot{\varepsilon}_{\text{max}} \frac{L^2}{6t}$ such to guarantee, in the initial phase of loading, a constant maximum elastic strain rate $\dot{\varepsilon}_{\text{max}} = 10^{-3} \text{s}^{-1}$ for all beam thicknesses $t$.

2.5 Vickers indentation tests
Three different Vickers indentation studies were performed, as detailed below.

2.5.1 Vickers micro-hardness tests
A Tukon 1202 Micro-Hardness Tester with a Vickers indenter was employed to measure hardness at loads in the range 98.1 mN to 19.61 N. This machine operates by progressively releasing a dead weight over the sample via a simple mechanism, and then holding the load constant for a prescribed amount of time before removing it. Preliminary measurements with an Indium sample mounted on a button load cell showed that the load ramping phase lasts approximately $t_0 = 2s$ for all values of maximum load; the
load ramping occurs at an approximately constant load rate, i.e. \( P(t) = \left( \frac{P_0}{t_0^2} \right) t \). The loading history associated with this experiment is sketched in Fig. 2. The duration of the dwelling phase was chosen as 5 s, giving a total test time \( t_1 = 7 \) s.

Upon removal of the load, the indentation footprint was observed via an optical microscope; the indentations were fairly square and regular at all loads. Image analysis was used to measure the projected area of indentation to determine the hardness \( H \). In a second set of measurements, the contact area was determined from measurements of the indentation diagonals. The difference between the hardness values calculated via the two different methods was found to be less than 7\% and in the following sections we report hardness measurements obtained from the average indentation diagonal.

2.5.2 Indentation creep tests

The experimental setup described in section 2.5.1 was also used to conduct additional sets of microhardness experiments at loads in the range 0.245 N to 1.962 N. For each value of the load, indentation experiments were repeated several times (at different locations) on the sample surface; in each experiment the load was held for a different time, ranging from \( t_1 = 10 \) s to 1000 s.

2.5.3 Instrumented indentation tests

Instrumented indentation was performed using both an Instron tensometer (Instron 5969, with a 10 N Instron load cell) and a DMTA analyzer (TA Instruments, RSA-G2) at room temperature. A Vickers indentation tip was displaced, at a constant velocity, against a stationary, flat, 5 mm thick Indium sample, whilst recording the indentation force and the indenter displacement at a frequency of 100 Hz. The indentation speed was chosen as 7.5 \( 10^{-3} \) mm/s in all tests.

3. Analytical models

This section presents the construction of simple analytical models to aid the interpretation of experimental results, particularly with respect to size-dependence of strength. The models assume that the material is viscoplastic and its response is size-independent; a comparison to the experimental data can therefore highlight the size dependence of the measured response.

3.1 Indentation of a power-law creeping material

The following model helps interpreting the microhardness measurements presented in Section 4. We formulate first-order predictions of the indentation response of a power-law creeping solid, absent any material size effects, when subjected to the load history sketched in Fig. 2.

The average plastic strain rate under the indenter can be considered proportional to \( \dot{h}/h \), i.e. \( \dot{\varepsilon} = C_{\varepsilon} \dot{h}/h \) ([21],[22]) where \( C_{\varepsilon} \) is a constant to be determined, \( h \) is the indentation depth and the dot denotes differentiation with respect to time. In the creep regime the flow stress \( \sigma \) in the isotropic solid depends upon the axial strain rate via a power-law expression
\[
\sigma = C_\sigma \dot{\varepsilon}^N \tag{1}
\]

where \( C_\sigma \) and \( N \) are material constants for a given temperature, to be determined experimentally. The material hardness \( H \) is defined as the average pressure under the indenter, proportional to the flow stress via

\[
H = C_H \sigma \tag{2}
\]

where \( C_H \) is suggested by Tabor for rate-independent metals (Tabor [23]). The true contact area is assumed to scale quadratically with indentation depth \( h \) according to the simple geometric relation

\[
A = C_A h^2 \tag{3}
\]

This assumption neglects any pile-up or sink-in effects, which were negligible in the indentation tests performed, as well as roundness of the tip; the effect of this is also negligible at indentation depths higher than 10 \( \mu \)m. The constant \( C_A \) is known and only a function of the indenter shape.

Consider the load history presented in Fig. 2. In the first loading phase the load is ramped according to \( P = t(P_0/t_0) \). Combining this with the definition of hardness and eqns. (1), (2) and (3) one obtains

\[
H = \frac{P}{C_A h^2} = C_H C_\sigma \left( \frac{\dot{h}}{h} \right)^N \tag{4}
\]

which integrated provides the time history of the indentation depth in the ramping phase

\[
h_R(t) = K_R t^{(N+1)/2} \quad \text{where} \quad K_R = \sqrt{\frac{P_0}{C_H C_\sigma C_A t_0}} \left( \frac{2}{N+1} \right)^{\frac{N}{2}} \tag{5}
\]

When \( t > t_0 \) the load is held constant, \( P = P_0 \). Substituting \( P = P_0 \) in (4) and integrating with the initial condition \( h(t_0) = h_R(t_0) \), the time history of the indentation depth in the dwelling phase is calculated as

\[
h(t) = \sqrt{\frac{P_0}{C_H C_\sigma C_A \left( \frac{2t[(N+1)] - t_0}{N(N+1)} \right)^{\frac{N}{2}}}} \tag{6}
\]

and the variation of hardness as

\[
H(t) = C_\dot{\varepsilon} C_H C_\sigma N^N \left[ N t_0^{\frac{N^2+N-2}{2N}} + 2(t - t_0) \right]^{-N} \tag{7}
\]

The model therefore predicts that the measured Vickers hardness does not depend on load.

In the instrumented indentation tests the indenter is displaced into the specimen at a fixed velocity; under the assumptions above, imposing this boundary condition provides
\[ H = C_H C_\sigma C_\varepsilon N \bar{h}^N h^{-N} \]  

(8)

For the case of instrumented indentation, the hardness is expected to decrease with increasing indentation depth, as a consequence of the rate sensitivity of the material.

### 3.2 Bending of a power-law creeping metal

An analytical model is developed to aid interpretation of the bending test results. The material response is assumed to be rigid-viscoplastic with flow stress \( \sigma_y \). It is assumed, in line with observations, that the beams collapse by forming a plastic hinge at mid-span, and that this hinge extends along the beam for a length equal to the beam thickness \( t \). The yield stress is taken as

\[ \sigma_y(\varepsilon, \dot{\varepsilon}) = \sigma_0(\varepsilon) + A|\dot{\varepsilon}(z)|^N \]  

(9)

and it is assumed that plane sections of the beam remain plane during bending, i.e. \(|\varepsilon(z)|^N = k|z|^N\) and \(|\dot{\varepsilon}(z)|^N = \dot{k}|z|^N\), where \( z \) denotes a spatial coordinate along the thickness of the beam, with origin at the neutral axis, and \( k \) is the curvature. The moment carried at mid-span can be found by integration of the stress distribution:

\[ M = \frac{bt^2}{4} \sigma_0(\varepsilon) + \frac{bt^{N+2}}{2N+1(N+2)} A \dot{k}^N \]  

(10)

where \( b \) is the width of the beam; this allows calculations of the force at midspan \( F = 4M/L \). The force versus displacement histories recorded in the experiments will be presented, in Section 4, in terms of the maximum (elastic) bending stress \( \sigma_{max} = 3LF/2bt^2 \) versus the maximum (elastic) bending strain \( \varepsilon_{max} = 6t\delta/L^2 \).

In order to ensure an identical maximum elastic strain rate when testing beams of different thicknesses, in the tests the speed of the loading roller was chosen as \( \dot{\varepsilon} = \varepsilon_{max}L^2/6t \), with \( \varepsilon_{max} = 10^{-3} \text{ s}^{-1} \) (the prescribed maximum elastic strain rate in all tests). The history of maximum bending stress versus maximum bending strain, \( \sigma_{max}(\varepsilon_{max}) \), can be computed by

\[ \sigma_{max} = \frac{3}{2} \sigma_0(\varepsilon) + \frac{3}{N+2} A \left( \frac{\varepsilon_0 L}{t} \left[ 1 + 4 \left( \frac{L^2 \delta}{3(L^2 + 4\delta^2)\varepsilon} \right)^2 \right] \right)^N \]  

(11)

and the definition of maximum elastic strain \( \varepsilon_{max} = 6t\delta/L^2 \). The model predicts that viscous effects induce a dependence of \( \sigma_{max} \) upon beam thickness, due to the second term in eq. (11), once plasticity is initiated. This is true at any imposed deflection (or equivalently, surface strain). The second term in eq. (11) can be used to predict the elevation of the curve \( \sigma_{max}(\varepsilon_{max}) \) with varying beam thickness, which will be discussed below (Section 4.3).
4. Results and discussion

4.1 Uniaxial tension

The measured true stress vs strain curves are presented in Fig. 3 for specimens of different thickness. After an initial elastic response, the specimens display plastic deformation and associated strain hardening, whose magnitude reduces with increasing applied tensile strain. Partial or complete fracture of the specimen intervened at strains in the range 5-15%; the stress versus strain curves are plotted until the first major or total loss in load. The results indicate no obvious size dependence of the tensile response as it can also be observed in Fig. 4, showing the tensile flow stress at strains of 0.1% and 1% as a function of beam thickness. A best-fit through the initial portions of the stress versus strain curves provided an estimate of the elastic modulus of 12.03 GPa, in line with the value (12.74 GPa) reported by ASM International [24] for Indium in tension at 20 °C. Analysis of the tensile specimen via optical microscopy revealed shear bands of width of a few µm and spacing of order 10 µm, as shown in the insert of Fig. 3. These were oriented approximately at 45% with respect to the specimen’s axis. The material displays a substantial intrinsic scatter, possibly due to the relatively large crystals compared to the specimen size. Due to the rolling process, tensile and bending specimens had grain size of order 1 mm in-plane and variable, smaller dimension along the thickness.

4.2 Uniaxial compression

The measured compressive response is shown in Fig. 5 for three values of the imposed strain rate, in terms of histories of true stress versus true strain. Test repetitions are shown to highlight the scatter in the response. The compressive response is slightly dissimilar from the tensile and on this note we remind the reader that tensile and compressive specimens are cut in different (perpendicular) directions and have different textures corresponding to the different manufacturing routes. In addition, compressive specimens were subject to smaller strains during manufacturing and their texture is more isotropic than bending and tension specimens. A pronounced strain rate sensitivity is evident from the compression measurements, with the response becoming stronger at higher rates of strain. Again the material displays scatter in the response of order 15% at all strain rates. Following a brief elastic phase of the response (not possible to resolve in the tests, due to the very small elastic strains) the material displays plastic deformation at very low stress, below 0.5 MPa. The material then shows a strain-hardening phase followed by a saturation phase, with the stress displaying a plateau response. The length of the strain-hardening phase was a function of the imposed strain rate. The load oscillations during the saturation phase were not due to noise in the measurements but were, quite possibly, associated to micro-cracking event, whose acoustic signature could be detected during the tests.

The value of the compressive flow stress at strains of 5%, 20% and 30% was extracted from the data in Fig. 5 and fitted by the power-law assumed in eq. (1), giving the values of the material parameters $C_\sigma$, and $N$ indicated in Table 1. When the material is in the strain hardening phase (5% strain), the creep exponent is 0.14; if the flow stress is measured in the saturation phase (20% and 30% strain), the creep exponent is of order 0.2, in line with previously reported measurements. Fig. 6 summarizes the measured compressive flow stress (at a strain of 5%) for cylindrical specimens of two different size. The compressive strength is clearly size-independent; a higher scatter is associated to measurements on the smaller samples, as expected.
4.3 Three-point bending

The bending response of the material is presented in Fig. 7 as histories $\sigma_{\text{max}}(\varepsilon_{\text{max}})$ for beams of different thickness. The initial slope of all such histories is equal to the material’s Young’s modulus; we remind the reader that the velocity of the loading roller was chosen to give identical strain rate histories in the elastic phase of the response. Clearly however, beams of smaller thickness display a stronger response in plasticity. Fig. 8 presents values of the maximum bending stress at bending strains of 0.1% and 0.2%, for all beams tested; smaller beams again appear stronger than thicker specimens. The bending stress increases by 50% when the beam thickness decreases from 2 to 0.25 mm. We note (i) that the simple model developed (eq. (11)) would predict an elevation in stress of order 1%*, suggesting that the stronger response is not a consequence of the strain rate dependence of the material; and (ii), that the same indium sheets, tested in tension at comparable strain rates, display a size-independent response. The above observations suggest a size dependence of the plastic response similar to that observed by various authors on different metals, as detailed in Section 1. However, in the experiments presented here the size effect is already evident at beam thicknesses just below 1 mm, while other authors have typically detected it on foil specimens of thickness of order 10 $\mu$m.

4.4 Vickers microhardness

The projected indentation area was calculated as $d^2/2$, where $d$ is the measured average indentation diagonal, while the penetration depth of the Vickers indenter is given by $h = d/(2\sqrt{2} \tan 68^\circ)$. The measured hardness versus indentation depth $h$ is plotted in Fig. 9; individual measurements as well as the average are shown to highlight the scatter in repeated tests. The data points obviously group along the hyperbola of eq. (4), with each group representing a different imposed indentation load.

At high loads, and correspondingly high indentation depths, hardness is independent of load. If the load is decreased sufficiently to give indentation depths below 80 $\mu$m, the hardness becomes size dependent and continues to increase as load and indentation depth decrease. The figure includes a best fit of eq. (7), with $t = 7$ s, through the measurements at the two highest loads. The data suggests a material size-dependence in presence of the plastic strain gradients induced in the indentation test. The size dependence is evident at indentation depths of around 100 $\mu$m, much larger than what previously observed in the literature for rate-independent metals.

4.5 Creep indentation tests

In this set of experiments repeated indentation tests were conducted at different locations, holding the load for different times. Results are presented in Fig. 10; at any load, the hardness decreases as a function of time, as a consequence of the increased indentation depth. It is also evident that at any time, the hardness increases with decreasing load $P_0$. Equation (7) predicts that for these experiments
hardness $H$ should be independent of load but only depend on time. Fig. 10 includes a best fit of eq. (7) through the data, for an indentation load of 200 g. The model fits the experiments reasonably well; the agreement between model and experiment becomes poorer as the load decreases. These observations again suggest a size dependence of the indentation response, evident at indentation depths below 100 $\mu$m.

4.6 Instrumented indentation experiment
Figure 11 presents load versus displacement histories recorded during instrumented indentation on Indium samples, displaying the typical quasi-parabolic shape. A moving average of 14 tests was computed from the data in Fig. 11, and converted into a hardness versus indentation depth history, presented in Fig. 12. The figure includes a fit of eq. (8) through the data, for $h > 100$ $\mu$m. The model agrees well with experiments in this range but the agreement becomes poorer as the indentation depth decreases. The model predicts an elevation in hardness of 63% as the indentation depth varies from 150 to 10 $\mu$m, however in the experiments the corresponding measured elevation is of 209%. This suggests, again, that the material displays size dependence of the response at indentation depths smaller than 100 $\mu$m.

4.7 Comparison of the observed size dependence with previous studies
The Fletch-Hutchinson theory (FH, [9],[10]) predicts that the yield stress $\sigma$ in a bending experiment is a function of the beam thickness $h$, a characteristic length scale $l_{FH}$ and the yield stress at large specimen size, $\sigma_{y,FH}$, as

$$\sigma = \sigma_{y,FH} \left( l_{FH} \frac{2}{h} + 1 \right)$$

(12)

A best fit of eq. (12) through the data in Fig. 8 (for a strain of 0.1%) provides $l_{FH} = 93.34$ $\mu$m and $\sigma_{y,FH} = 1.55$ MPa (and a corresponding yield strain $\varepsilon_{y,FH} = \frac{\sigma_{y,FH}}{E} = 1.4 \times 10^{-4}$).

With regards to the indentation response, following Evans and Hutchinson [25], we consider the model developed by Nix and Gao (NG, [11]), predicting the dependence of hardness upon indentation depth

$$H = H_0 \sqrt{1 + \frac{l_{NG}}{5.2h}}$$

(13)

A least-square fit of eq. (13) to the instrumented indentation measurements presented in Fig. 12 (for depths higher than 5 $\mu$m) provides $l_{NG} = 85.21$ $\mu$m and $H_0 = 5.12$ MPa. Making use of the relationship between hardness and yield stress given by Tabor [23], the yield strain can be calculated as $\varepsilon_{y,NG} = cH_0/E = 1.5 \times 10^{-4}$, with $c = 0.34$.

The length scales measured for bending and indentation of Indium in this study are compared to those measured by other authors on different rate-independent metals and summarised in [25] as a function of the material yield strain. This is shown in Fig. 13, while a summary of the experiments in the figure is provided in Table 2. The figure shows that the length scales measured for Indium are substantially higher than those previously reported; however, our measurements confirm the correlation, highlighted in [25], between measured length scales and material yield strain. This reinforces the
notion, proposed by Fleck-Hutchinson, that the length scale depends upon the distance travelled by dislocations, which in turn determines the yield strain.

As all tests in Fig. 13 are conducted on rate-independent metals, with the exception of those on Indium, it might be conjectured that the different mechanisms of dislocation motion active in the creeping material (such as dislocation climbing as opposed to sliding) may have an effect upon the measured length scales. To clear this doubt we conducted a preliminary set of instrumented indentation experiments at a temperature of 123 K (below the power-law creep regime for Indium), using a DMTA tester and a nitrogen vapour atmosphere. Again the hardness was found to increase with decreasing indentation depths, and the corresponding indentation length scale was measured as $l_{NG} = 72 \mu m$, only marginally lower than what measured at room temperature. This suggests only a mild influence of the mechanisms of dislocation motion upon the measured length scale. We leave a comprehensive investigation of the mechanical response of Indium at 123 K as a topic for a future study.

<table>
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<tr>
<th>Research paper</th>
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<th>Experiment type</th>
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<tr>
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<td>Indentation</td>
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<tr>
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Table 2: Summary of the research previously carried out on the length scale of different metals, summarised in the study carried out by Evans and Hutchinson (2009).

5. Conclusions

The mechanical response of high purity Indium was measured in tension, compression, bending and indentation at room temperature. The material displays a pronounced viscoplastic response with strength increasing with increasing imposed strain rate. The response was accurately described by a power-law creep relation with strain rate exponent ranging from 0.14 at small strains to 0.2 at large strains.

The response was approximately symmetric in compression and tension; in both tension and compression tests the material exhibited pronounced strain hardening; this was followed by a saturation phase in compression, while fracture intervened in the tension tests at strains of order 10%.

In tension and compression, the material response was insensitive to specimen size in the ranges explored. On the other hand, the bending and indentation response displayed an increase in strength with decreasing beam thickness or indentation depths, respectively. Such increase was much beyond what can be ascribed to viscoplastic effects and estimated by the simple analytical models presented.

In particular, the size effect was evident at indentation depths just below 100 $\mu$m and, in bending, for beam thicknesses just below 1 mm. The associated material length scales are therefore much higher
than previously measured for rate-independent metals. The measurements presented in this study confirm the correlation, previously highlighted in the literature, between measured length scales and material yield strain, reinforcing the notion, proposed by Fleck-Hutchinson that the length scale must depend upon the distance travelled by dislocations. Preliminary instrumented indentation experiments at 123 K suggest that such high length-scale is not due to creep mechanism but has to be ascribed to the microstructure of the material.

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**References**


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