On thermal softening and adiabatic shear failure of dynamically compressed metallic specimens in the Kolsky bar system

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The issues of thermal softening and adiabatic shear failure, in dynamically compressed metals, are revisited through experiments in the Kolsky bar system. Various materials were compressed by single- and multi-step loadings and the results were analysed through a new approach to the issue of instability strain, which is based on the temperatures existing in the specimens just prior to the onset of instability. These temperatures are compared with the threshold temperatures, which mark the steep decrease in the strength–temperature curves. This approach accounts for most of the materials we tested. However, the brittle behaviour of the titanium and magnesium alloys, which fail at a very low strain, should be treated by a different approach.

1. Introduction

High-strength metallic alloys are known to undergo a shear localization process at a relatively low strain, when subjected to high rate impulsive loading. This localization is manifested by the appearance of very narrow shear bands (of the order of 10–100 μm). Zener & Hollomon [1] were the first to analyse this phenomenon in terms of the adiabatic nature of the high rate loading process, by which the heat generated within the localized plastic flow cannot diffuse fast enough away from the flow site. These, so-called, ‘adiabatic shear bands’ are the precursors for fracture through cracks which form and grow along the bands. Adiabatic shear bands have been the subject of intense research over the past 70 years owing to their occurrence in metalworking and metalcutting, as well as in terminal ballistics.
exploding cylinders and other high rate processes. Much of the experimental observations and the theoretical approaches, which were developed in order to account for them, are summarized in the book by Dodd & Bai [2]. The narrow shear bands in impulsively loaded specimens are caused by either localized deformations or crystallographic phase transitions which occur as a result of the high temperatures. In steels, the deformed bands are typically 100–200 µm thick while the thickness of the transformed bands is of the order of 10–30 µm (see Giovanola [3]).

Owing to the diversity of the processes in which these shear bands are formed, a comprehensive model, which accounts for all the experimental finding, is still lacking. We should emphasize another important issue concerning the occurrence of adiabatic shear bands. Several experimental configurations employ specimens with sharp edges, as with the hat-shaped configuration where the stresses are geometrically forced in these corners. Thus, it is easy to obtain shear bands in specimens which, otherwise, do not show a tendency for such localizations, as noted by Chen et al. [4] for tantalum specimens. Obviously, the propensity of a given material to true adiabatic shear banding should not be determined by such configurations with ‘forced localized shear’.

In this work, we further explore the nature of adiabatic heating in disc-like specimens subjected to compressive loading in the Kolsky bar system, by comparing the response of specimens to a single load and to multi-step loading, as we have done in Ashuach et al. [5]. In addition, we compare the dynamic stress–strain curves with those obtained by static compression in an Instron machine, in order to delineate the adiabatic features of the failure under dynamic loading. The different responses of a specimen to these various types of loading can be explained by the different temperature increases in the specimens. In particular, the thermal softening of the specimen, under a single loading experiment, can be much higher than under a multi-step loading. This temperature rise can lead to bulk softening, which accounts for the loss of strength of some of the materials which we tested. Thus, their shear failure can be explained in a rather straightforward manner. On the other hand, there are materials (such as the Ti-6Al-4V alloy) which fail at very low strains under dynamic loading, with a negligible increase in bulk temperature. These materials fail in a brittle way, both statically and dynamically, and the mechanism behind their failure should be further investigated.

2. Experimental set-up

The Kolsky bar system in our laboratory includes maraging steel bars 25.4 mm in diameter, instrumented with interferometers which follow their velocities, as described in Avinadav et al. [6]. We have shown there that the inferred strain histories in the bars are identical with those measured by conventional strain gauges. We performed two types of compressive loadings on the specimens, by a single step, which was typically in the $5 \times 10^3$ s$^{-1}$ range of strain rates, and by multi-step loading, where we loaded the specimen several times at strain rates of the order of $5 \times 10^2$ s$^{-1}$. The specimens had diameters in the range of 7–10 mm and their thicknesses were half their corresponding diameters. Figure 1 shows our experimental set-up with the PDV interferometric system for the velocity histories of the two bars.

A comparison between the stress–strain curves obtained by a single- and multi-step loadings, with specimens of 6061-T6 aluminium, is shown in figure 2. The continuous line corresponds to the single-step loading and the broken curve to the multi-step loading. These dynamic curves are practically identical, showing very low strain hardening for this alloy and no failure up to strains of about 1.0. Moreover, published data for the stress–strain curve under static loading, of this aluminium alloy, are very similar to the curves in figure 2, except for a somewhat lower value for the flow stress, at high strains, owing to some strain rate sensitivity.

3. Thermal softening analysis

The traditional treatment of adiabatic shearing assumes that the shear stress $\tau$ is a function of the shear strain $\gamma$, the strain rate $\dot{\gamma}$ and its instantaneous temperature $T$. Thus, one can write:
\[ \tau = \tau(\gamma, \dot{\gamma}, T). \] The load instability is assumed to be initiated when \( d\tau = 0 \), thus

\[ d\tau = \left( \frac{\partial \tau}{\partial \gamma} \right)_{\gamma, T} d\gamma + \left( \frac{\partial \tau}{\partial \dot{\gamma}} \right)_{\gamma, T} d\dot{\gamma} + \left( \frac{\partial \tau}{\partial T} \right)_{\gamma, \dot{\gamma}} dT = 0. \] (3.1)

An instability strain can be obtained from this equation by considering a certain form for the constitutive relation. Most workers use a power-law relation between the stress and the strain (work hardening), thus:

\[ \tau = K \gamma^n. \]

The strain rate in a given test is practically constant and the temperature rise can be obtained from the deformation work through the well-known Taylor–Quinney relation

\[ \rho C_V dT = \beta \tau \cdot d\gamma, \] (3.2)

where \( \rho \) and \( C_V \) are the density and heat capacity of the material, respectively, and \( \beta \) is a constant with a value between 0.9 and 1.0.

With these assumptions, equation (3.1) results in the following expression for the instability strain:

\[ \gamma_i = -\frac{n \rho C_V}{(d\tau/dT)_{\gamma, \dot{\gamma}}}, \] (3.3)

where the value of \( \beta \) is taken as 1.0 for simplicity.

One should note that this relation has been obtained through several assumptions, as noted by Culver [7], who pointed out that the whole purpose of the analysis is to identify the metals, such as titanium alloys, which have a low value for \( \gamma_i \), where the thermal instability may be a major factor in their failure under dynamic loading. Equation (3.3) predicts that low-density materials which behave in an elasto-perfectly plastic manner (low \( n \)) should have low...
instability strains, in contrast with many experimental findings. Moreover, a close examination of experimental values for $\gamma_i$ shows that high-strength materials have low values of $\gamma_i$, a fact which is not borne out by equation (3.3), which does not include the strength of the specimen. Most importantly, the actual temperature of the specimen, just prior to its shear instability, does not enter in this analysis. Thus, to our best understanding, the idea of equating the strain-hardening slope with the strength decrease owing to thermal softening is not the main issue here. We suggest that one should consider the specific dependence of strength on temperature of the specimen, $\sigma = \sigma(T)$, particularly at those critical temperatures ($T_c$), where a sharp decrease is evidenced in this curve. Our basic assumption is that once the temperature rise in the specimen reached $T_c$, further deformation is unstable because of the enhanced softening for $T > T_c$.

In order to demonstrate this approach, we show in figure 3 the variation of the ultimate tensile strength of the 6061-T6 alloy, as taken from [8]. One can clearly see that up to a temperature of about 370 K (an increase of 80°C), the strength does not decrease appreciably while at higher temperatures it drops quite significantly. One should note that, owing to lack of enough data points, the location of the ‘knee’ is not very well defined. However, we chose this value by observing the relatively large drop in strength at temperatures higher than 370 K.

We should note that under dynamic loading of 6061-T6 aluminium, the corresponding ‘knee’ is shifted to higher temperatures (about 200°C), as shown by Rosenberg et al. [9]. This is a very important point for our analysis because it relies on the critical temperature ($T_c$), at the onset of strength loss under dynamic loading. Integrating the $\sigma(\epsilon)$ curve in figure 2, and inserting the relevant coefficients for the 6061 alloy in equation (3.2), shows that up to a strain of 1.0 the temperature increase is about 160°C. This value is higher than the apparent ‘knee’ in figure 3, but it is lower than the corresponding ‘knee’ in the strength–temperature curve under dynamic loading [9]. Thus, we may conclude that under dynamic loading to a strain of about 1, this material does not reach the critical temperature $T_c$.

Our approach to the issue of thermal softening is based on calculating the temperature of the specimen, through the deformation work, and equating it with the threshold temperature ($T_c$) where the strength of the specimen decreases sharply. This equality should result in the instability strain which has been reached by the thermal softening mechanism. This idea can be expressed by the following equation, which applies for the temperature rise within the bulk of the material, rather than in the localized areas:

$$\rho C_V (T_c - T_0) = \int_0^{\gamma_i} \tau \, d\gamma_i.$$  (3.4)
Conversely, one can compare the experimental values of $\gamma_i$ as obtained by a dynamic test, with the expected values from this equation, when the values of $T_c$ are given for the critical temperatures. According to equation (3.4), a high-strength material will have a lower value for $\gamma_i$ when compared with a similar material with lower strength. For the experiments discussed here, dynamic compression in the Kolsky bar system, we replace the shear stress $\tau$ and the shear strain $\gamma$ by the flow stress $\sigma$ and the compression strain $\varepsilon$, respectively.

4. Results and discussion

We shall now present our experimental results for the stress–strain curves of several materials, both under single- and multi-step loadings, and analyse their instability strains by considering their $\sigma = \sigma(T)$ curves. It is clear that the onset of instability is by the maximal shear stresses, along diagonal lines in the disc-shaped specimens of these experiments.

(a) 2024-T351 and 7075-T6 aluminium alloys

Figure 4 shows our results for a single loading of these alloys, from which it is clear that the instability strain of the stronger alloy (7075) is lower than that of the weaker one (2024). This is an important observation which can be safely related to the difference in their stronger flow stresses which should lead to higher temperatures in the corresponding specimens. Such a difference is not borne out by equation (3.3) while, by contrast, our approach anticipates such differences. Figure 4b shows the loss of the ultimate tensile strength of both alloys owing to temperature, as obtained from [8]. Both curves show a ‘knee’ at a temperature of about 370 K, which means that $T_c = 100^\circ$C for both alloys, as far as static loading is considered.

Considering the fact that the lower strength 6061 alloy did not soften at these experiments (figure 2), we find a logical ordering of the softening strain according to the strength of the alloy. Note that equation (3.3) does not include the strength of the specimen, so this observation cannot be anticipated by this equation. Moreover, the strain exponent $n$ for the three materials seems to increase with the strength of the alloy, as seen by the multi-step loading results. Thus, using equation (3.3) one may expect a higher value of the instability strain for stronger alloys, which is clearly not the case here.

Inserting the values of $\varepsilon_i = 0.55$ for the 2024 alloy (from figure 4) into equation (3.4), we find that it corresponds to a temperature rise of about 130$^\circ$C, which is somewhat higher than the corresponding $T_c$ for this alloy, probably because of the strain rate effect on this threshold temperature. With the same procedure, we find that for the 7075 alloy, with an instability strain of $\varepsilon_i = 0.4$, from figure 4, the expected value for temperature rise is, up to this strain, 100$^\circ$C, in good agreement with the value of $T_c$ for this alloy. For the two aluminium alloys, we used tabulated $C_V$ and $\rho$ values from [8]. To summarize, the expected values for the thermal instability, strains of the aluminium alloys can be accounted for by the corresponding values of the threshold temperatures, as determined by the ‘knees’ in their $\sigma = \sigma(T)$ curves.

Figure 5 compares the stress–strain curves of the 2024 alloy under both single- and multi-step loading. It is clear that the quasi-isothermal step loading of this alloy does not exhibit a thermal softening. Thus, by eliminating the temperature increase, with the multi-step loading, the specimen can reach high values of compressive strains without failure. It is interesting to note that the data from Forrestal et al. [10] for the static compression of 7075-T6 aluminium show no failure up to a strain of 0.75. The statically loaded 7075 specimen reached a flow stress of nearly 0.7 GPa which is close to the dynamic value, as shown in figure 4.

(b) Stainless steel 304L

The stainless steel 304L exhibits a very high strain-hardening behaviour, as is clearly shown in figure 6, which shows our experimental results under single- and multi-step loadings. Both curves show a tendency towards flattening at their corresponding high strains. The fact that the
Figure 4. (a) Dynamic stress–strain curves for the two aluminium alloys and (b) the temperature dependence of the ultimate tensile strength of these alloys (adapted from [8]). (Online version in colour.)

Figure 5. The stress–strain curves of the 2024 alloy. (Online version in colour.)
maximal flow stress reached by the multiply loaded specimen is higher than that of the specimen which experienced a single load can be considered as a manifestation of the effect of temperature increase in the single loading test. In both tests, the specimens did not fail even at high strains of the order of 1.

Using the corresponding values of $C_V$ and $\rho$, from [8], we calculate an increase of about 220°C at a strain of 1.0. Obviously, this value is much lower than the $T_c$ value of 800 K which is shown in figure 7, for the temperature dependence of the ultimate tensile strength of this stainless steel. Thus, our simple approach accounts for the fact that we do not see an appreciable thermal softening in this test. We should note that the specimens in these tests had an aspect ratio $L/D = 0.5$. In other tests with $L/D$ larger than 1, we found that the specimens failed at high strains of about 0.7. This $L/D$ effect, in dynamically compressed specimens, has been discussed in [11] and in the references therein.

We should note that the materials tested, up to this point, show a rather gradual decrease in their stress–strain curves right after the instability strain. This trend is different than the much steeper decrease which is evident in the corresponding curves of some of the other materials we tested, as will be shown below.
Figure 8. The dynamic stress–strain curves for the magnesium alloy. (Online version in colour.)

(c) Magnesium alloy: AM50

Magnesium alloys are brittle materials which fail at strains of the order of 0.2 under static compressive loading, as shown by Hanina [12], for example. We tested the AM50 alloy under both single- and multi-step loading and the resulting stress–strain curves are shown in figure 8. It is clear that the specimens failed at very low strains, around 0.2, under the two loading configurations. Obviously, the specimens achieved very low temperatures in these tests, and it will be hard to relate the observed failure with a thermal softening mechanism. Moreover, the fact that the dynamic instability strains are close to the static value means that the failure of this material has nothing to do with the adiabatic loading conditions in the Kolsky bar test. Thus, we conclude that brittle materials, which fail at low strains under static conditions, do not experience thermal instabilities and their failure modes should be considered through other mechanisms.

(d) Titanium and its Ti-6Al-4V alloy

The stress–strain curves for unalloyed titanium (grade 2) specimens are shown in figure 9a. It is interesting to note that under multi-step loading the specimen did not fail up to strains of the order of 0.6, which is the expected result since we do not anticipate any heating in this experiment. We also performed static loading of an unalloyed titanium specimen and it showed no failure up to a strain of 0.4, which was the maximal strain reached in our Instron machine.

The temperature dependence of the flow stress of unalloyed titanium, at \( \varepsilon = 0.2 \), is shown in figure 9b. The data for this figure were taken from [13]. It is clearly evident that at a temperature of about 500 K there is a significant drop in the flow stress. Using our model, with the corresponding values of \( C_V \) and \( \rho \), together with \( \sigma = 1.3 \text{ GPa} \) and \( \varepsilon = 0.35 \), from figure 9a, we obtain an expected temperature rise of 190 K. This value is in excellent agreement with the ‘knee’ shown in figure 9b, strongly enhancing our model.

The most interesting material which we tested for this study is the alloy Ti-6Al-4V, which is considered by many workers as a material with a pronounced tendency to undergo adiabatic shear failure. The resulting stress–strain curves for single- and multi-step loadings of this alloy are shown in figure 10. For both loading conditions, we find a very low value for the instability strain of about 0.2. We should also note that under static compression of this alloy, we observed a shear failure at a compressive true strain of about 0.3 in our Instron machine. The multi-step loading of the specimen resulted in its failure at the last (fourth) step, so that we assign a value of about 0.3 for the failure strain of this alloy under multi-step loading. Clearly, the low values for the observed failure strains, under the various loading conditions of this alloy, should be viewed...
Figure 9. (a) The stress–strain curves for unalloyed titanium specimens and (b) the temperature dependence of dynamic flow stress at $\varepsilon = 0.2$ for unalloyed titanium (adapted from [13]). (Online version in colour.)

Figure 10. The dynamic stress–strain curves for Ti-6Al-4V specimens. (Online version in colour.)
as an indication of its low ductility, much like that of the magnesium alloy. Thus, our conclusion is that one should categorize the titanium and the magnesium alloys as a different class of materials, as compared with the other materials which we tested in this study.

The most important results of this work concern the close values for the failure strains, under the different loading conditions, of the magnesium and titanium alloys. Moreover, these strains are especially low (of the order of 0.2), in comparison with the other materials which we tested. We conclude that these alloys are quasi-brittle by nature and we relate their brittleness with the plastic zone ahead of a crack, $r_y$, as we have done in [11]. The plastic zone ahead of a crack is defined by

$$r_y = \frac{1}{2\pi} \left( \frac{K_{IC}}{Y} \right)^2,$$

where $K_{IC}$ and $Y$ are the fracture toughness and the strength of the specimen, respectively. As we discussed in [11], a specimen can behave either in a ductile or a brittle manner, depending on its size as compared with $r_y$. If the specimen is much larger than $r_y$, the material behaves in a quasi-brittle manner, and vice versa. Taking values from the literature for $K_{IC}$ and $Y$, we find that $r_y \approx 2$ mm for the magnesium alloy and $r_y \approx 0.5$ mm for the titanium alloy. These values of $r_y$ are lower than the specimen’s size, and this can explain the quasi-brittle behaviour for both materials. On the other hand, the values of $r_y$ for the other materials which we tested are of the order of 5 mm, accounting for their ductile behaviour (relatively large failure strains).

5. Summary

The purpose of this study was to revisit the issues of thermal softening and adiabatic shear failure in dynamically loaded materials. We presented experimental results for the dynamic compression curves of various materials compressed by the Kolsky bar system, both by single- and multi-step loadings, in order to highlight the effect of the temperature increase in these tests. We proposed a new approach to the instability strains in terms of the actual temperature increase in the dynamically compressed material, together with the loss of strength which it experiences with temperature increase. This analysis accounts for all the materials we tested except for the titanium and magnesium alloys. These alloys fail at relatively low strains, both under single- and multi-step loading. Moreover, these materials also fail at low strains under quasi-static compression tests. The brittle nature of their failure is attributed to the low values of the plastic zone ahead of a crack, $r_y$, in both materials.

References

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