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The effect of specimen dimensions on the propensity to adiabatic shear failure in Kolsky bar experiments

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ABSTRACT

The issue of adiabatic shearing is discussed in this work and a new interpretation is given to some failure phenomena which are usually termed as adiabatic shears. We propose that only a few materials undergo a truly inherent failure which is due to thermal softening at the shearing zone and that the interplay between microvoids, cracks and narrow shear bands should be taken into account through the temperature rise at the front of advancing cracks. Also, the size of the plastic zone ahead of a crack plays an important role in determining the brittleness of a given specimen and should be taken into account when specimens of different sizes are tested. Experimental results for several alloys in the Kolsky bar system support our approach.

Keywords: Kolsky bar, SHPB, adiabatic shear banding.

1 INTRODUCTION

Adiabatic shear banding is one of the major failure mechanisms of solids subjected to high rate loading. Extensive research on this subject has been conducted over the past 50 years, both experimentally and theoretically, much of which is summarized in BAI and DODD [1]. The most common explanation of this dynamic instability is based on the competing tendencies of the solid to strengthen at high strains (and strain rates) and, on the other hand, to soften with the temperature increase under adiabatic loading conditions. Although many materials seem to behave accordingly, the basic mechanism which causes this softening is still unresolved. Recent work on the subject is focused on the basic physics behind this process, as can be found in [2-5], for example. These works emphasize the link between adiabatic shearing and ductile fracture by the coalescence of microvoids, since the shear zones act as precursor sites for eventual failure by cracks (see [2]). GIOVANOLA [3, 4] conducted a careful experimental study on 4340 steel specimens, in order to follow the evolution of strain localization and failure which he finds to proceed in two stages. He found that the first stage of localization (deformed bands) results in the formation of local perturbations for the strain field, such as machining marks. These perturbations grow as a result of the imbalance between strain hardening and thermal softening. In the second stage, softening is due to the nucleation and growth of microvoids, leading to the shear fracture within the band. GIOVANOLA [3, 4] suggested that this process can take place under other conditions, including quasi-static loading. He also emphasized the role of the compressive stress which acts on the shear plane to reduce the microvoid nucleation process.

FLOCKHART *et al.* [5] used numerical simulations to follow the shear band in dynamically compressed specimen, by analyzing the loci of velocity discontinuities within the specimen. They also claim that "the distinction between shear fracture and adiabatic shear failure is not clear", and that "material susceptibility to adiabatic shear alone does not guarantee that failure occurs by this mechanism, as shear fracture may intervene". This interplay between shear instabilities, microvoids and microcracks which lead to an almost simultaneous shear and fracture process, complicates the phenomenon immensely, and makes it still unresolved as many workers realize.

O'DONELL and WOODWARD [6] give the results of dynamic compression experiments on cylindrical specimens of an aluminum alloy (2024-T351) with a diameter of 4.76mm. They found that shear failure appeared only in specimens with thickness larger than 4mm, defining a threshold aspect ratio (L/D) of about 1. A similar L/D effect was also reported by WALLEY *et al.* [7], using the Kolsky bar system, with specimens made of armor steel and tungsten alloy. For these materials, specimens with L/D>1 showed a clear indication of shear fracture while specimens with smaller ratio did not. From these works, as well as others, one may conclude that the dynamic failure of cylindrical specimens by such compressive loading is very

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sensitive to their shape and dimensions. Thus, geometry of the specimen plays a role in their failure in addition to their inherent physical properties.

The purpose of the work presented here was to investigate this L/D effect and, in particular, to check whether the dimension ratio is the important parameter or is that the thickness of the specimen (L). We shall present a series of compression tests conducted with the Kolsky bar apparatus on several materials with varying dimensions and L/D ratios. The results of these experiments will be analyzed by considering their propensity to adiabatic shearing. We also consider the relevance of the plastic zone ahead of the crack (ry), as defined by fracture mechanics for ductile materials, through their flow strength and fracture toughness.

2 MATERIALS AND METHODS

Our experiments were conducted with the well-known Kolsky bar apparatus (which is wrongly termed the split Hopkinson bar). The system in our laboratory consists of two maraging steel bars 25.4mm in diameter and 3m long, instrumented with strain gages at about 60cm from their ends. The length of the striker bar is 40cm and it is made of the same steel. All our specimens were straight cylinders ranging in diameter from 2-10mm and similar thicknesses.

The first thing we wish to emphasize is that other specimen shapes can induce shear failure in materials which do not experience adiabatic shear banding with straight cylindrical specimens. Such are the geometries of the hat-shaped specimen (see TENG *et al.* [8] and CHEN *et al.* [9], for example), the notched specimens (see RITTEL *et al.* [10]), and the truncated cone which was used by LI *et al.* [11]. All of these special geometries introduce localized instabilities through stress concentrations which enhance the shear failure of these specimens. Thus, as is clearly stated in [8], the shear failure in these specimens is not caused by material propensity to this mode of failure but because of the geometrical details of the specimens. In fact, specimens of any material are bound to fail by shearing due to stress concentrations at these specimens, as was pointed out in [9]. This point may be a trivial one but we find it important as some researchers assign these geometrical instabilities to material properties which, as we just pointed out, is incorrect. Thus, compression experiments with the Kolsky bar should be done on straight cylinders which experience a uniform strain throughout their volume. The work of CHEN *et al.* [9] demonstrates this issue very effectively since they tested two shapes of tantalum specimens in their Kolsky bar. The straight cylinders did not fail even under high loading rates and large strains, while the hat-shaped cylinders failed easily by shearing along the planes where the stresses were concentrated by the geometry. Similar failures were obtained in the hat-shaped specimens of a tungsten alloy in the work of TENG *et al.* [8]. These alloys do not exhibit adiabatic shear failure when the specimens are straight cylinders and the shear failure found with hat-shaped specimens is the result of geometric discontinuities and sharp corners. Li *et al.* [11] also state that their tungsten alloy does not experience adiabatic shear localization when the specimens are straight cylinders. Thus, the failure of their truncated cones was due only to geometric effects and is not a material property.

The second point which is worth noting is the observation that many materials experience a shear failure at the same strain level in both static and dynamic compression. Thus, the term adiabatic shear banding for the failure of these materials should be questioned. For example, we performed static compression of the magnesium alloy AM50, as well as dynamic tests on similar specimens of this material. The resulting stress-strain curves from these experiments have shown that the specimens failed at about the same strain under the two loading rates. This material is often referred to as one which undergoes an adiabatic shear failure and, to our best understanding, the failure is by nucleation and coalescence of microcracks with no dependence on the temperature rise during loading. We suggest that only materials which exhibit a much lower strain to failure under dynamic conditions should be considered as materials which are failing by adiabatic shear. An excellent example for such materials is the alloy Ti-6Al-4V. Quasi-static compression of this material, which we performed with an Instron machine, resulted in a failure at a strain of about 50% while under dynamic conditions we found that it fails at a strain of less than 20%, as we show later on. Under these low strains the bulk of the specimen could not have been heated to temperatures which can cause a significant softening. Thus, it must be a local heating, acting in an avalanche mode within the narrow shear bands, which causes this truly adiabatic shear failure.

3 RESULTS

As we stated above, our main purpose was to investigate the effect of specimen dimensions on the shear failure of dynamically loaded cylinders in the Kolsky bar system. In particular, we were motivated by the results presented in [6] and [7], which showed that an aspect ratio of about L/D=1 turns out to be the threshold for the appearance of such a failure in several materials. Thus, O'donell and Woodward [6] found that their 2024-T351 specimens (diameter of 4.76mm) did not fail adiabatically when their thickness was 3.75mm or less, while specimens with thickness of at least 4.25mm did fail in their drop weight machine.

WALLEY *et al.* [7] found that specimens made of rolled homogenous armor steel (RHA) as well as from a tungsten alloy, failed in their Kolsky bar apparatus whenever their aspect ratio was larger than one and did not fail otherwise. On the other hand, all specimens made of the titanium alloy Ti-6Al-4V failed irrespective of their dimensions and aspect ratio. They relate this difference in behavior to the density of the specimens which, for the denser materials, add inertia effects to the dynamic compression process. The threshold aspect ratio, L/D=1, according to [7], is the limiting case for the intersection of the shear planes with the cylinder wall, as was suggested by ZHANG *et al.* [12] for the failure behavior of their Zr-based bulk metallic glass specimens. Moreover, the titanium specimens failed through a large number of planes inclined at about 45° to the specimen faces. On the other hand, the recovered steel and the tungsten alloy cylinders showed two shear planes emanating from the outer area of the cylinders. This was interpreted by Walley et al. as a clear indication of geometrical effects taking a role in the process of shear banding. The main conclusion of WALLEY *et al.* [7] is that "the adiabatic shear phenomenon is a mixed material/structure problem". In other words, the geometric effects have to be carefully analyzed in order to separate them from inherent material properties. This statement enhances the difficulty in isolating the true propensity of a given material to fail by adiabatic shearing. In the next sections we bring our results for the materials tested – Ti-6Al-4V and three aluminum alloys. All the tests were performed under similar strain rates, in the range of (1-5)·10-3 s-1, thus, we shall not deal with this parameter in the following sections.

3.1 Ti-6Al-4V Specimens

The first set of experiments was performed with this alloy of titanium which is the classical example of a material which fails by adiabatic shear. Figure 1 shows our experimental stress-strain curves for the different specimens with aspect ratios in the range of L/D=0.5-1.5. It is clearly seen that all specimens failed at a strain of about 0.15. Under static compression of a cylindrical specimen (in our Instron machine) this material failed at a strain of about 0.5. This is a clear indication that the dynamic failure is different than the static one, because of this material's propensity for adiabatic shear failure. As we summarized above, the physics behind this failure is very complex, involving several processes and we shall highlight some of them in the discussion section.

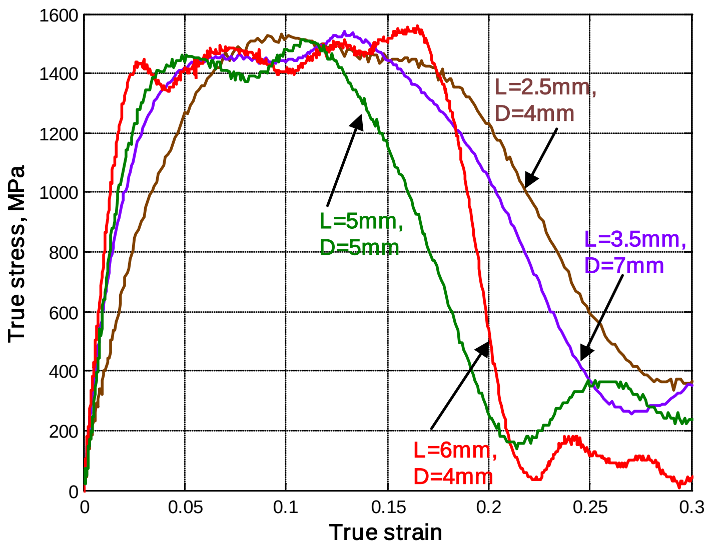

Figure 1: The results for the Ti-6Al-4V alloy.

3.2 Aluminum alloys

The most interesting results were obtained for the aluminum alloys which we tested here: 6061-T651, 2024-T351 and 7075-T651 which, for brevity reasons, will be termed 6061, 2024 and 7075, respectively. Figure 2 shows that the dynamic stress-strain curves for the 6061 alloy do not show any sign of failure, or even strain softening, up to a strain of 0.8. We also did not find any evidence for shear failure in the recovered specimens. This alloy exhibits an almost ideal elastic-perfectly-plastic behavior, as was found by many workers (see Forrestal *et al.* [13], for example). Moreover, the quasi-static compression curve for this alloy, as measured in [13], does not show any softening or failure even for strains as high as 1.0. The flow stress of this alloy, at these strain rates, is about 0.42GPa and the specimen dimensions, as well as their aspect ratios in the range of L/D=0.5-1.5, do not play any role in their compression behavior.

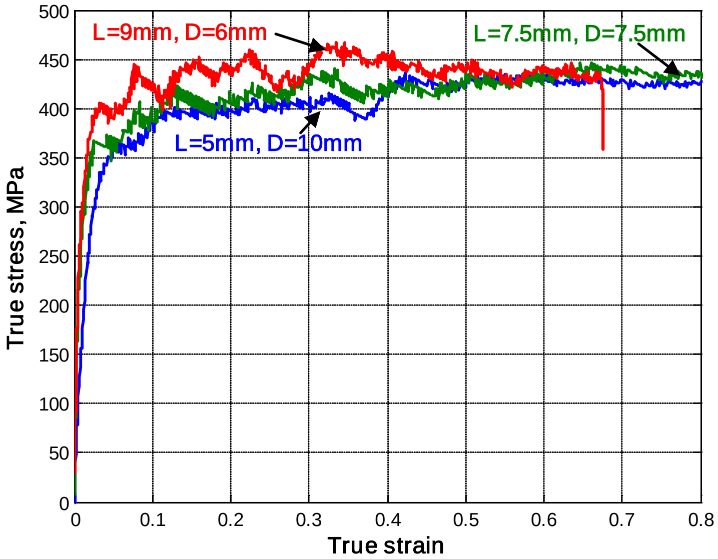

Figure 2: The results for the aluminum alloy 6061-T651.

The next set of experiments, on the 2024 alloy, revealed some interesting results. Let us consider first the results for L/D=0.5 specimens shown in Figure 3. The flow stress of this alloy is near 0.65GPa. For the two smaller specimens we see indication for thermal softening at a strain of about 0.7. This softening was not observed in the 6061 alloy with the lower strength and the question is whether it is related to this strength difference. The most interesting feature in these curves is the failure observed for the large specimen with L=5mm. This specimen showed a clear shear failure when recovered after the test. Thus, a threshold for dynamic shear failure seems to exist for specimens with a thickness between 3mm and 5mm. This is in excellent agreement with the results of Ref. [6] where the threshold was at L=4mm. This threshold is clearly not due to the aspect ratio of the specimens since L/D=0.5 for all the specimens in this group. Thus, instead of an "L/D effect" we have to consider an "L effect". In order to check whether this threshold depends on the diameter of the specimen, rather than its thickness, we performed several tests with equal diameter specimens (D=10mm), with thicknesses in the range of 3-7mm. The results are shown in Figure 4 and one can see that with increasing thickness the failure is more pronounced. All of these specimens have aspect ratios well below 1, which means that the threshold of L/D=1 found in [6] and [7] should be reinterpreted as a thickness threshold.

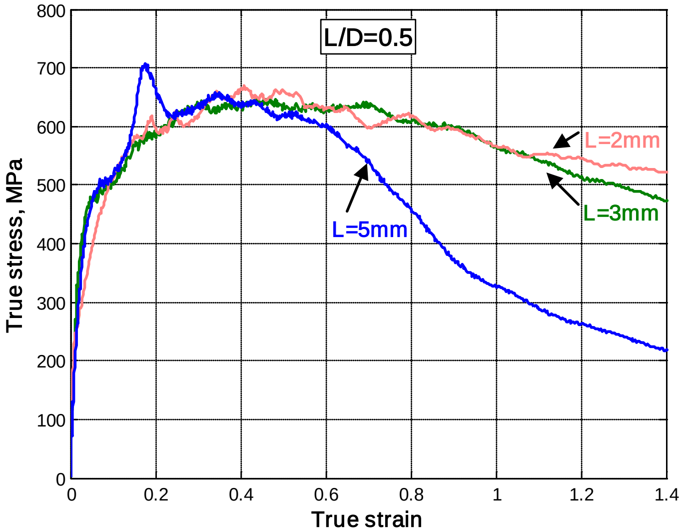

Figure 3: Stress-strain curves for L/D=0.5 specimens of aluminum alloy 2024-T351.

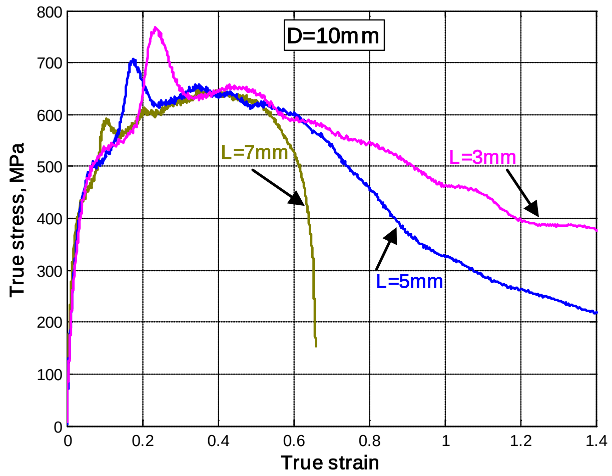

Figure 4: Results for aluminum alloy 2024-T351 specimens of different thicknesses and the same diameter (D=10mm).

Once we established that for this alloy the threshold for dynamic shear failure is ≈4mm we performed several tests on specimens with L>5. The results of these tests are shown in Figure 5. The interesting thing to note here is that as the thickness of the specimen increases its strain to failure decreases. For L=9mm we find a failure strain of 0.4 while the L=5mm specimen failed at a strain of about 0.6. This difference in failure strains can be interpreted through differences in ductility due to specimen size, as we discuss later on.

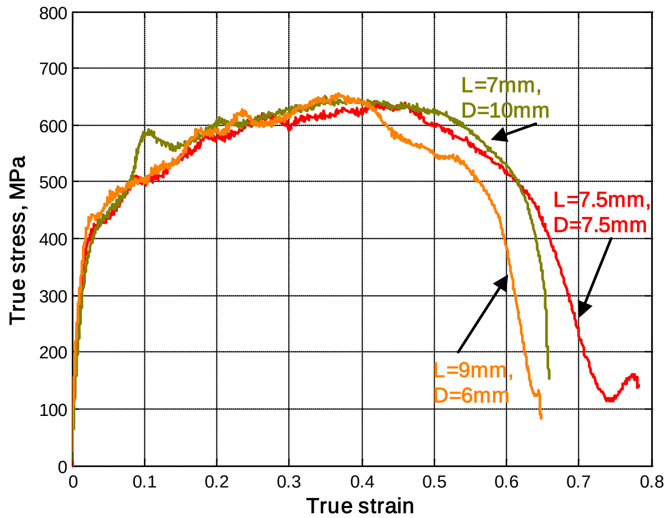

Figure 5: Results for aluminum alloy 2024-T351 specimens with L>5mm.

The strongest aluminum which we tested was the 7075 alloy for which the stress- strain curves are shown in Figure 6. We see that the flow stress of this alloy is about 0.7GPa and that all the specimens which we tested failed at a strain of 0.45-0.5. Considering the different shapes and sizes of this specimen we conclude that there is no dimensional effect for this material and that the failure is a regular shear failure which occurs also under static conditions for this strong alloy, which is known to be rather brittle. A quasi-static test in our instrumented Instron machine showed that cylindrical samples 8mm in diameter and 8mm thick, failed at a strain of about 0.4. This value is very close to the dynamic failure strain we found above, strongly enhancing our claim that this failure is not an adiabatic shear failure as we explained here.

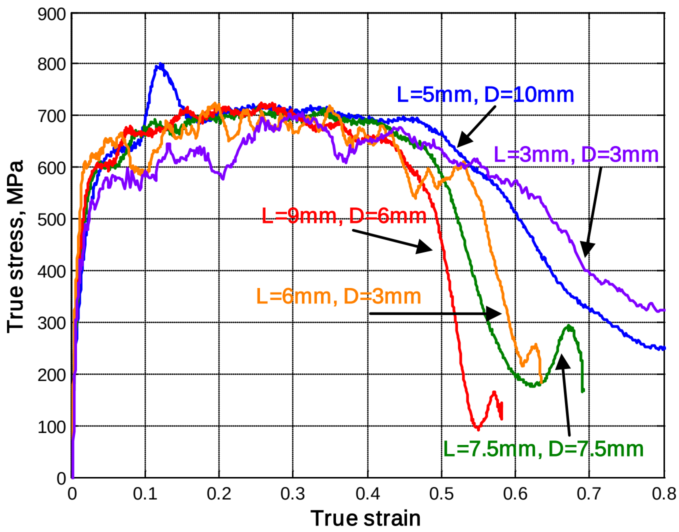

Figure 6: Stress-strain curves for the aluminum alloy 7075-T651.

4 DISCUSSION

We have emphasized the difference between the inherent propensity for adiabatic shearing and the geometrical constraints, due to the shape of the specimen and the loading conditions, which enhance the process of shear failure under dynamic loading. These constraints were termed by CHEN *et al.* [9] as forced localization and they are present in many real situations such as in terminal ballistics of blunt projectiles penetrating hard targets. MOLINARI and CLIFTON [14] analyze the different softening mechanisms which can operate under shear loading, either static or dynamic. These softening mechanisms are responsible to shear instabilities, which appear as shear bands. They list several sources such as geometric inhomogeneities, local damage in the specimens, phase transitions, thermal softening, etc.

We have seen that of all the materials which we tested, as well as those studied in the references cited above, only the Ti-6Al-4V alloy can be considered as a material which experiences a true adiabatic shear failure. Thus, the quest for a physical understanding of this phenomenon, as in the work of GIOVANOLA [3] on this alloy, has to be carried out only on similar materials. Our basic assumption is that the physics of this process involves the complicated interplay between local heating due to strain inhomogeneities and the developments of microvoids or microcracks within the shear bands. Temperatures can rise to high levels in front of these microdefects, as many workers have calculated and even measured to some success (see [15-17], for example). A physically-based account for these local temperatures, ahead of a moving crack tip, has been given by RICE and LEVY [17]:

$$\Delta T = \frac{1.41(1-\nu)^2 \cdot K \cdot Y \cdot V^{0.5}}{E \cdot (\rho \cdot c \cdot k)^{0.5}} \tag{1}$$

where V is the crack velocity, Y is the strength of the material, c its sound speed, and k is its thermal conductivity. K, E, ν and ρ are the stress intensity factor, Young's modulus, Poisson's ratio and the density of the material, respectively. We see that this expression for the temperature increase contains the relevant mechanical and thermal parameters of the metal. One can use Eq. (1) to assess the propensity of a given material to develop adiabatic shear bands. Using published values for the thermal and elastic constants, as well as the yield strengths and stress intensity factors, RICE and LEVY [17] calculate the temperature rise at the crack tips for mild steel and the 2024 aluminum and Ti-6Al-4V titanium alloys. It turns out that for a given crack tip velocity this rise for the titanium alloy is higher by more than an order of magnitude than the corresponding rise for the two other materials. This large difference is the result of higher values for (KY) together with the lower values of (ρ·c·k) for the titanium alloy. Thus, it is not surprising that this alloy has such a large tendency for adiabatic shear failure.

In order to explain our results, for the different aluminum alloys, we have to realize that all of the physical and mechanical properties, which appear in Eq. 1, are very close (if not identical) for the three alloys. Thus, we cannot expect any meaningful differences in the temperatures ahead of cracks, microcracks or microvoids which develop in these materials during the dynamic loading. Moreover, we saw that the 2024 alloy exhibit a clear geometric effect, which we termed as an "L-effect". Thus, the shear failure which the specimens with L>5mm experienced here, and also in the work of O'DONELL and WOODWARD [6], has to have another source. We also note that the 6061 alloy, which is considered as a ductile material, did not fail in our tests at all specimen dimensions. In contrast, all the specimens of the more brittle 7075 alloy failed at about the same strain. These observations lead us to propose that the main cause for these differences is the different ductility of these alloys. A common measure of this property is the relation between the plastic zone ahead of the crack (ry) and the size of the specimen. This physical quantity is used to explain the ductile-to-brittle transition, which is often found in many materials. In particular, this transition takes place when specimens with different dimensions are tested under similar tests. For relatively ductile materials the plastic zone is defined by:

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

where KIC is the fracture toughness and Y the yield strength of the material.

Thus, specimens much larger than ry behave in a brittle manner while specimens with dimensions smaller than ry are more ductile. Our interpretation of the data for the 2024 aluminum alloy is that its ry value should be somewhat less than 5mm. Thus, in the Kolsky bar tests we see evidence for the ductile-to-brittle transition by using specimens with thicknesses smaller and larger than ry. Indeed, using published data (see [18], for example) of KIC and Y for the 2024 and 7075 alloys, KIC=44 and 24MPa·m0.5, respectively, we get values of ry=2.4mm for the 2024 alloy (with Y=0.345GPa) and ry=0.4mm for the 7075 alloy (with Y=0.5GPa). This value of ry=2.4mm for the 2024 alloy is very close to the value of the threshold thickness; L=4mm, which was found here and by O'DONELL and WOODWARD [6]. The sub-millimeter value of ry for the 7075 alloy explains the fact that specimens from this material failed under dynamic loading at all the sizes which we tested. This material is expected to behave in a more ductile manner only for specimens smaller than about 0.5mm. It is interesting to note that according to [18], the plastic zone ahead of the crack in the Ti-6Al-4V alloy is also very small, ry=0.6mm. Thus, one may claim that its dynamic failure at all sizes is also due to its brittleness, as for the 7075 aluminum alloy. However, as we emphasized above, the large difference between the static and dynamic failure strains for this alloy, are pointing towards a much more complicated process, like the adiabatic shear failure.

5 CONCLUSIONS

We discuss the issue of adiabatic shear failure by eliminating all the cases where geometrical inhomogeneities lead to stress concentrations which enhance the failure processes. We bring experimental results, with the Kolsky bar system, on several materials with which we find very different behavior as far as dynamic failure is concerned. These differences are analyzed in terms of the inherent physical properties of solids which influence their ductile-to-brittle transition, and also their propensity to develop high local temperatures in front of running cracks. These new observations should be considered more thoroughly in future theoretical accounts for the occurrence of adiabatic shearing.

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