

A DIFFERENT VIEWPOINT ON ADIABATIC SHEAR LOCALIZATION

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ABSTRACT

This short review addresses the adiabatic shear failure mechanism (ASB) in metals subjected to high strain-rate deformations. ASB is usually considered as an instability resulting from thermal softening effects. We present a different viewpoint, based on experimental observations in which we identify the stored energy of cold work as the driving force for microstructural re-arrangement by dynamic recrystallization (DRX). DRX is an athermal phenomenon that is observed at the early stages of the dynamic deformation process, while thermal effects are insignificant, long before an ASB has formed in the specimen. We tentatively identify DRX as the softening mechanisms whose growth in a localized area will lead to final failure.

1. INTRODUCTION

Material failure has long interested the Physics and the Engineering communities in an attempt not only to better understand, but also control the series of events leading to this point. The sudden creation of a discontinuity at the micro(nano) scale is indeed quite intriguing and not yet fully understood, even if recent sophisticated computer atomistic simulations have shed considerable light on this phenomenon. Among the typical failure mechanisms, one can consider two main phenomena: fracture (creation and separation of new surfaces), and the localization of the deformation into a specific spatial pattern as a precursor to fracture. These phenomena can be observed in the quasi-static regime for which the prescribed loads (displacements) are slowly applied to the solid so that inertial effects are negligible. They can also be observed when dynamic loading conditions prevail such as impact, for which wave propagation plays an important role. The first subject, dynamic fracture, has been, and still is, a subject of considerable interest with many open questions concerning the physics of the process at all scales, from the atomic to the continuum (for a review, see e.g. Cox et al. (2005)). Dynamic fracture will not be addressed further in the present paper.

The localization of the (plastic) deformation into a band (plane) or a surface has been analyzed in mathematical terms as a bifurcation phenomenon and the emergence of a localized solution for the displacement field for quasi-static situations (see e.g. Rice (1976)). From a physical point of view, the homogeneous plastic deformation localizes into a narrow band in which the shear strain can reach very high values leading to ultimate failure. This is referred to as the emergence of a shear band, the rest of the structure being essentially unloaded from that point thereon. When dynamic loading conditions prevail, the same phenomenon can be observed, namely formation of a shear band with an intense shear strain, with the difference that one must now consider thermomechanical coupling effects. Indeed, it has long been known that part of the strain energy gets dissipated as heat during a plastic deformation process (Farren and Taylor, 1925; Taylor and Quinney, 1934; Tresca, 1879). When the time scale is small, as in the case of impact, the heat cannot flow out of the solid, causing a temperature rise and the process can be qualified, to a first approximation of adiabatic (Boley and Weiner, 1960). The phenomenon is consequently referred to as adiabatic shear banding (ASB). ASB is very often associated with the dynamic cracking processes, when the latter follows

shortly the shear band formation. This paper will address some aspects of this phenomenon.

In order to understand the emergence of an adiabatic shear band, Zener and Hollomon (1944) proposed to consider the opposite influence of strain-rate hardening and thermal softening effects, the latter bringing to a point where the material can no longer harden, thus loses its stability and allows the formation of a localized failure mode. A large body of experimental evidence can be found in the book of Bai and Dodd (1992), while mathematical solutions are compiled in the book of Wright (Wright, 2002). Thermal softening has been the prevailing assumption since its inception, and the modern analyses rely on perturbation analysis of the thermal or mechanical fields (Molinari and Clifton, 1987).

In order to complete the picture, it must be mentioned that while a huge body of microstructural evidence can be found about adiabatic shear bands in many materials, the experimental mechanical characterization of the phenomenon has been much less investigated since the seminal contribution of Marchand and Duffy (1988), and related works. Since thermal evolutions are crucial in this problem, experimental emphasis has been put on real-time monitoring of the temperature of impacted specimens, using e.g. infrared radiometers (Hartley, et al., 1987; Hodowany, et al., 2000; Kapoor and Nemat-Nasser, 1998; Trojanowski, et al., 1997). The outcome of these various works was a better understanding of the thermomechanical conversion and its efficiency. In other words, while it had long been assumed that the thermal to mechanical conversion efficiency is a constant of the order 0.9, recent work showed that it is not only strain but also strain-rate dependent, so that this efficiency has a constitutive equation of its own (Mason, et al., 1994; Rittel, 1999). As a result, if adiabatic conditions are assumed, the energy balance can be written as:

$$\beta_{\text{int}} \int_{\varepsilon_y}^{\varepsilon_f} \sigma_{ij} d\varepsilon_{ij}^p = \rho C_p \Delta T \quad (1)$$

Where σ_{ij} and ε_{ij}^p indicate stress and plastic strain respectively, the subscripts y and f stand for yield and failure, ρ is the density, C_p is the heat capacity, ΔT is the temperature rise and β_{int} is the efficiency of the thermomechanical conversion (Rittel, 1999; Taylor and Quinney, 1934). It must be noted that assuming a value of 0.9 or 1 for

β_{int} over the whole range of strains will result in an overestimation of the temperature rise.

To conclude this introductory survey, it will be noted that a well accepted criterion for the onset of ASB formation is expressed in terms of a critical strain, based on the Zener-Hollomon analysis for a specific constitutive model of the dynamic mechanical behavior of the material of interest.

So far, there is no clear connection between the profusion of microstructural observations and mechanical quantities, such as a critical strain for failure, so that the physical picture is still incomplete.

The purpose of this paper is to bring recent results on the phenomenon of adiabatic shear banding. The results are taken from the author's recent work and presented as a synthesis. Firstly, we will examine the extent to which thermoplastic coupling effects lead to an important temperature rise, showing that the latter is not significant in many materials until the late failure stages. Next, we present a new failure criterion based on mechanical energy rather than a critical strain. Once the physical nature of the energy is identified, the microstructure of the material appears as a natural extension of the criterion. Additional experimental results will be presented and discussed to clarify the role of microstructural evolutions by outlining a particular phase transition called dynamic recrystallization (DRX), which can be tentatively identified as a key factor in the generation of adiabatic shear bands.

2. RECENT EXPERIMENTAL RESULTS ON ASB FORMATION

2.1 On the extent of the temperature rise during dynamic deformation

Thermal fluctuations are viewed as instrumental in the generation of adiabatic shear bands. Here one must distinguish between two spatial scales. The first scale is very microscopic in essence, of the order of the square micron or less, and it is essentially inaccessible to modern instrumentation, e.g. infrared detectors whose pixel size is of the order of $30\mu^2$. The other scale is dictated by the pixel size, for which resolution limitations can be partly compensated by a large number of experimental datasets for which one may expect that some of these "hot spots" will eventually be detected. One is therefore necessarily limited to a more macroscopic characterization of the local temperature changes. Several experiments were recently carried out in order to characterize the elevation of temperature in the gauge section of an impacted specimen

for various metals. For a pure α -iron specimen deformed at $\dot{\varepsilon} \approx 8000\text{s}^{-1}$, the recorded temperature reached 431K at $\varepsilon \approx 0.7$ (Rittel, et al., 2006a). This temperature rise corresponds to a homologous temperature of 0.24. It is important to note that the recorded temperature rise was more moderate at lower strain rates, illustrating the point that β_{int} is rate-sensitive, as mentioned before. Later work on pure polycrystalline tantalum showed that this material could reach 378K at a plastic strain $\varepsilon_p \approx 0.3$, irrespective of the strain-rate this time (Rittel, et al., 2007). This temperature rise corresponds to a moderate homologous temperature of 0.15. Considering now an annealed commercial Ti6Al4V alloy, it was observed that the measured temperature at the macroscopic onset of localization ($\varepsilon \approx 0.27; \dot{\varepsilon} \approx 3000\text{s}^{-1}$) was of the order of 350K, corresponding to a homologous temperature of 0.19 (Figure 1). In the same paper (Rittel and Wang, 2008), the authors reported a localization temperature of 323K ($\varepsilon \approx 0.17; \dot{\varepsilon} \approx 3000\text{s}^{-1}$) for a cast magnesium (AM50) alloy, which corresponds to a homologous temperature of 0.46, which is indeed significant. For these two materials, the temperature measurement was carried out until full localization and fracture of the ASB, showing here very significant temperature rises of the order of several hundreds of degrees. At this stage, it appears that the extent of temperature rise is indeed material dependent, as dictated by the stress and strain levels at failure. The severity of the temperature rise, as expressed by the homologous temperature is also material dependent and will of course be more significant for materials that possess a low melting point such as the AM50 alloy. Otherwise, the measured temperature rise remains modest for many materials. However, apart from a comparison of the homologous temperatures involved in the process, additional information can be obtained by considering the mechanical energy at failure, as explained in the next section.

2.2 On the critical failure strain and thermal softening

The concept of a critical failure strain, or the strain at which the adiabatic shear band forms, is indeed simple and convenient as a failure criterion. One would therefore expect that irrespective of the previous thermomechanical history of a given material, the critical failure strain will characterize the onset of adiabatic shear failure in all situations. To verify this point, a series of experiments was carried out on the two above-mentioned alloys, namely Ti6Al4V and AM50. The first set of experiments consisted of applying a

static pre-strain to the material, the extent of which could be normalized with respect to the quasi-static failure strain. The specimens were subsequently impacted to failure and the failure strain was thus recorded. The first remarkable result of these experiments is that the total (static+dynamic) strain to failure increases with the amount of pre-strain (Rittel, et al., 2006b). This observation shows that there is no strain to failure properly speaking, as the latter must take into account the overall thermomechanical history of the material. However, as shown in Figure (2), when the dynamic mechanical energy is calculated, which is the integral of the dynamic stress-strain curve until the onset of failure, it is noted that this mechanical quantity remains remarkably constant over large levels of normalized pre-strain. In other words, the prior quasi-static phase has little influence on the subsequent dynamic loading step in terms of failure energy. This observation emphasizes the importance of the dynamic phase for the adiabatic shear failure process, at the expense of the previous quasi-static history.

Coming back to thermal considerations, additional information can be gained by performing monotonic dynamic tests with these materials at various temperatures. Here, instead of looking for a specific softening effect, one records again the dynamic mechanical energy until the onset of adiabatic shear. In the same paper, the authors report that this mechanical energy remains remarkably constant for test temperatures that correspond to a maximal anticipated value obtained by assuming $\beta=1$, which is an upper bound in most cases (Figure (1)). The significance of this result is that even if the AM50 alloy reaches a relatively high homologous temperature during the test, the amount of energy needed to break it (and the Ti alloy as well) is not significantly affected, again on an spatially averaged level. These observations cast a doubt on the efficiency of the thermal softening process in relation to adiabatic shear band formation. Without obliterating this assumption, it appears that there are exceptions of a nature that justifies additional investigation into this phenomenon.

At this stage, one can write a simple balance of energy, noting that the deformation energy splits into (equals) the thermally dissipated energy and the remaining stored energy. The latter, also referred to as stored energy of cold work (Bever, et al., 1973) is stored in the material, e.g. through microstructural re-arrangements at the dislocation level. Noting that the thermal part plays what seems to be a minor role in the process, it comes out that the *dynamically* stored energy of cold work is a key factor in the generation of ASB's. Such a concept is certainly different from that of a strain (or stress)

based criterion, but it also has a different physical meaning. Namely, since the stored energy of cold work is a direct consequence of the microstructure, a seamless link is thus established between a mechanical phenomenon and the material microstructure, which was not noted previously.

The next step is therefore to investigate this microstructural evolution that will eventually lead to shear localization.

2.3 On dynamic recrystallization

Dynamic recrystallization (DRX) is a structural rearrangement whereby the initial grain structure of a material gets refined to a point where new nanograins form and eventually grow. Dynamic recrystallization bears an analogy with the usual recrystallization phenomenon, which is driven by accumulated strain and temperature considerations. However, in the present case, DRX is unique in the sense that this is an athermal phenomenon, as observed e.g. by Hines and Vecchio (1997) in copper that was deformed in liquid nitrogen. The operating mechanisms consist of a successive grain refinement, as described in the recent literature (Meyers, 1994; Meyers, et al., 2000; Perez-Prado, et al., 2001; Rittel, et al., 2002; Xu, et al., 2008). It is interesting to note that dynamically recrystallized grains are observed in almost each shear band forming metal, and it is understood that DRX is a result of the large strains and temperatures that develop locally in the shear band (Murr, et al., 2002; Xu, et al., 2008). Toth et al. (2000) modeled the DRX in adiabatic shear bands and they successfully reproduced the corresponding texture of a typical adiabatic shear band in steel. More recently, an attempt was made to systematically couple DRX to the adiabatic shear band in a numerical simulation mimicking the evolution of the microstructure in the band (Medyanik, et al., 2007). However, these two works considered DRX as a result of shear localization, in the sense that it is thermally-driven, whereas experimental evidence has clearly shown that thermal activation is not a pre-requisite.

Consequently, one intriguing question relates to the exact nature and role of the dynamic recrystallization phenomenon that almost inevitably comes along with the adiabatic shear band. Rittel et al. (Rittel, et al., 2008) addressed this issue by performing interrupted dynamic tests on the Ti6AL4V alloy that is well known for its propensity for adiabatic shear failure. The interrupted test consisted of impacting a shear specimen while carefully controlling the amount of imparted strain, to keep it to roughly one half

of the dynamic failure strain value. Since the specimens contain a natural stress raiser (fillet), where shear bands always develop, the fillet region was selected for transmission electron microscopic investigation of the microstructure, in both interrupted specimens and also in specimens that were loaded to failure. The outcome of these investigations was the striking observation that dynamic recrystallization was clearly observed in those interrupted specimens that had not failed or developed a shear band, revealing a very similar microstructure to that of the shear band of the failed specimens. The meaning of this result is that dynamic recrystallization, in this material at least, is occurring and clearly preceding any sign of adiabatic shear localization (Figure (3)). It must also be noted that this happens at a stage where the temperature rise is clearly insignificant, just in accord with the athermal nature of the phenomenon. At this stage, one should take a closer look at the thermal evolution in the gauge section of impacted specimens (Figure (2)), and realize that while the final temperature rise may reach a homologous temperature of 0.2 or perhaps more, this only occurs very close to the macroscopic localization point (the maximum stress/strain from which the material starts to soften noticeably). For the investigated alloy, the temperature rise until $\varepsilon \approx 0.15$ is actually insignificant but DRX is readily observed. One can now propose that DRX develops at an early stage of the deformation process, independently of the thermal evolution in the gauge section, so that its multiplication which is analogous to a phase transformation results upon its completion into the adiabatic shear band. It should also be noted that the initial stages of the DRX are not accompanied by a visible strain softening of the material. This observation confers a different meaning to the concept of a mechanical instability. Yet, it is important to assess the exact mechanical influence of the development of clusters of nanograins that are virtually dislocation free at that stage. One would naturally suspect that this nanophase possesses superior mechanical properties based on a Hall-Petch type of hardening rule. However, it has been shown on several occasions that for nanograins, the Hall-Petch relationship no longer applies and in fact the contrary just occurs, namely that the ultrafine grain size does indeed weaken the material (Trelewicz and Schuch, 2008). Mercier et al. (2007) addressed the grain-size dependence of the flow strength of copper using an elastic-viscoplastic model, and in addition to the grain-size dependence, they introduced a strain-rate dependence for the critical grain size for which a reverse Hall-Petch effect should be observed. From their results, it appears that this critical size decreases as the strain rate increases. However,

Mercier et al. (2007) came to the conclusion that at high strain-rates, as those typical of adiabatic shear localization, the reverse Hall-Petch effect does not exist as diffusional processes are no longer operative. This observation requires additional experimental work. Alternatively, it was also proposed that grain boundary sliding might be the operative mechanism, by analogy with superplasticity, as opposed to dislocation motion related deformation (Chokshi and Meyers, 1990). Therefore, it appears that the formation of dynamically recrystallized enclaves just weaken the material locally, and as such act as local destabilizing (softening) nuclei whose dynamics remains to be studied in detail.

As a last remark, one should establish a connection between the previous mention of a critical energy storage and the current DRX phenomenon. Here it can be proposed that dynamic recrystallization is driven by the stored energy of cold work, which upon reaching a critical level, triggers the recrystallization phenomenon. Once initiated, the latter is likely to increase the local temperature by additional plastic strain which in turns will surely contribute to additional recrystallization. However, at that stage, the phenomenon has just been identified as a potential key contributor to shear localization, and its kinetics remain to be studied in detail through methodical testing in various materials.

3. DISCUSSION

This paper discussed adiabatic shear failure, as commonly observed mechanisms of dynamic failure, with emphasis of the physics of the phenomenon. Until recently, two main classes of observations and models were available, namely those of a metallurgical character and the (few) mechanical experiments. These two classes bear little in common, and a well accepted criterion is that of a critical strain to failure, the exact physical causes of the phenomenon remaining largely unexplored.

This viewpoint aims at reconciliating the two classes of observations, based on specific experimental observations. The first one is that it is difficult to define a critical strain to failure for a given material that has been subjected to various thermomechanical treatments. Specifically, if one varies systematically the amount of quasi-static pre-strain prior to dynamic testing, the total strain to failure is observed to increase instead of keeping a constant value. This observation emphasizes the central role of the dynamic deformation phase irrespective of the prior history. One way to analyze the data consists

of looking at the mechanical energy invested in the process, for the dynamic phase, and observing that this quantity remains relatively constant, irrespective of the prior quasi-static amount of pre-strain. A parallel observation relates to the temperature rise during dynamic deformation, the latter being always considered as a softening factor leading to localization. Experimental recordings of the thermal evolution in several materials all point to the fact that the temperature rise is generally modest until the macroscopic onset of localization, and its relative extent is material dependent. In other words, if one considers a strong, ductile material that localizes at very large strains, one cannot overlook thermal activation. Yet for the investigated materials, the failure strain is relatively small and the temperature increases only at the later stages of the macroscopic pre-localization phase. Considering again the dynamic deformation energy, the latter is not really affected by the anticipated temperature rise. One might also wonder if the additional temperature rise does not weaken when the baseline temperature is higher, due to the higher dislocation mobility and annihilation phenomena, but this remains an open question so far. In any case, provided that the temperature rise may not be that large, another mechanism has to be identified which immediately suggests the importance of the stored energy in the material. The latter is directly connected to the material microstructure and as such provides an elegant transition between the mechanical and the microstructural characteristics of the material. At that stage, it was shown that while there is no significant temperature change in the specimen, a phase transformation (dynamic recrystallization) can be detected, which was always assumed to occur only past the shear band initiation and development. This result suggests that DRX actually precedes the ASB and to some extent might likely be its cause rather than its consequence. The DRX phenomenon is athermal and believed to be triggered by the energy that has been stored in the material, thus linking the two mentioned observations. Moreover, dynamic recrystallization is believed to locally weaken the material through dislocation free nanograins, so that this local perturbation is expected to grow and destabilize the deformation processes, resulting in an ASB for which most observations concur to show dynamically recrystallized grains.

This sequence of events is another physical scenario to interpret adiabatic shear failure. One cannot claim that this is the general and unique phenomenon, as discussed before for the case of highly ductile and shear-resistant materials. It nevertheless reveals a new perspective on a phenomenon that was apparently well understood. It is therefore felt that additional effort is now needed to identify initiation and growth criteria of dynamic

recrystallization, followed by proper numerical modeling and verification of the anticipated softening effects.

4. CONCLUSIONS

This viewpoint has presented and discussed another possible scenario leading to adiabatic shear localization. Instead on considering uniquely thermal softening as the driving force, it is believed that stored energy issues and their influence on the microstructure are key factors that lead to dynamic recrystallization. The latter is a precursor to adiabatic shear banding and not its outcome. Dynamic recrystallization is tentatively identified as a local softening factor, whose multiplication and growth leads to adiabatic shear band formation.

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FIGURE 1

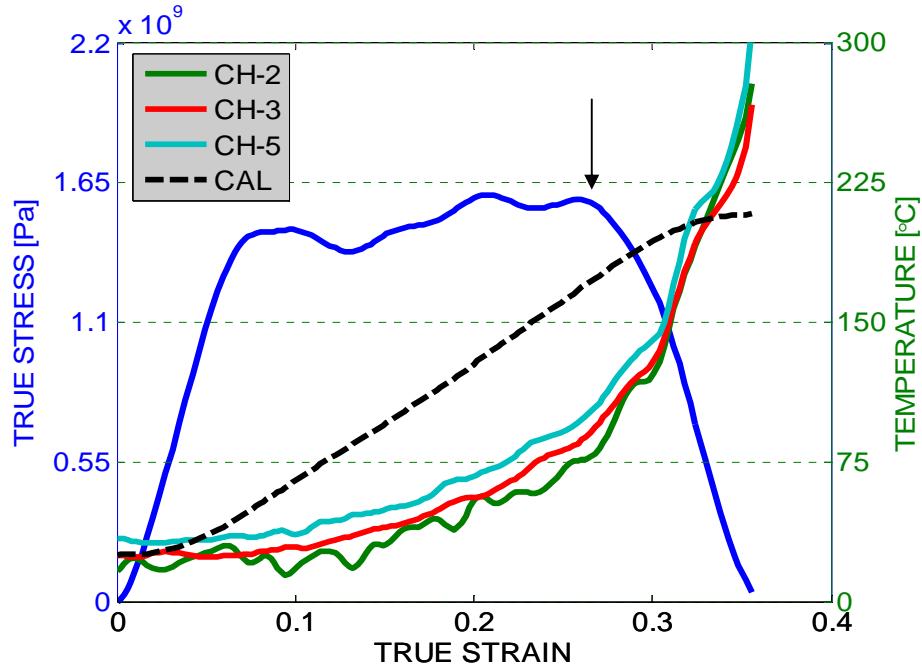


Figure 1: Typical stress and temperature vs. strain for dynamically loaded Ti6Al4V alloy. The strain rate is $\dot{\varepsilon} \approx 3000 \text{ s}^{-1}$. The temperature is measured on 3 infrared channels and calculated (dashed line) according to $\beta = 1$ (eqn (1)). The arrow indicates the strain at which the materials softens considerably (failure), indicating the presence of an adiabatic shear band. Note that the temperature rise is quite modest until $\varepsilon \approx 0.2$. It then increases slightly until the failure strain, reaching a value of $T \approx 75^\circ\text{C}$. Note that the measured temperature, until the failure strain, is markedly lower than the calculated one, showing that $\beta \neq 1$. (reprinted after Rittel et al. (2008)).

FIGURE 2

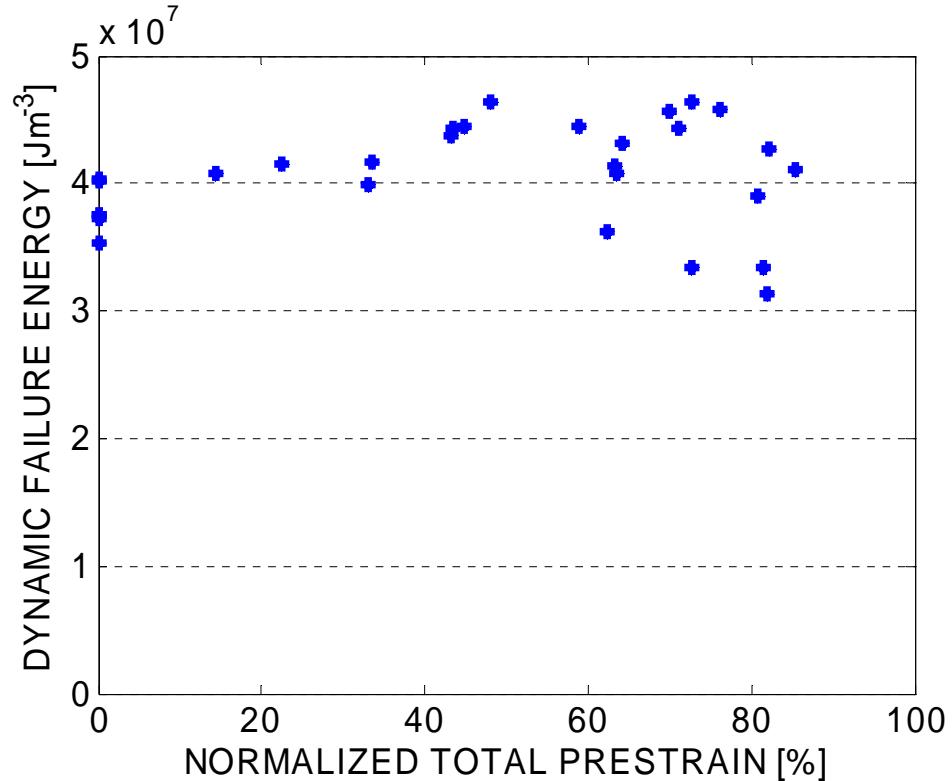


Figure 2: Plot of the dynamic mechanical energy of AM50 as a function of the normalized pre-strain $\frac{\varepsilon_{\text{pre-strain}}}{\varepsilon_{\text{FS}}} \times 100$. Note that the energy is remarkably constant up to normalized pre-strain levels of the order of 0.7, beyond which it decreases slightly. The overall influence of the quasi-static pre-strain on the dynamic mechanical energy is minor, up to deep pre-strain levels. (reprinted from Rittel et al. (2006b)).

FIGURE 3

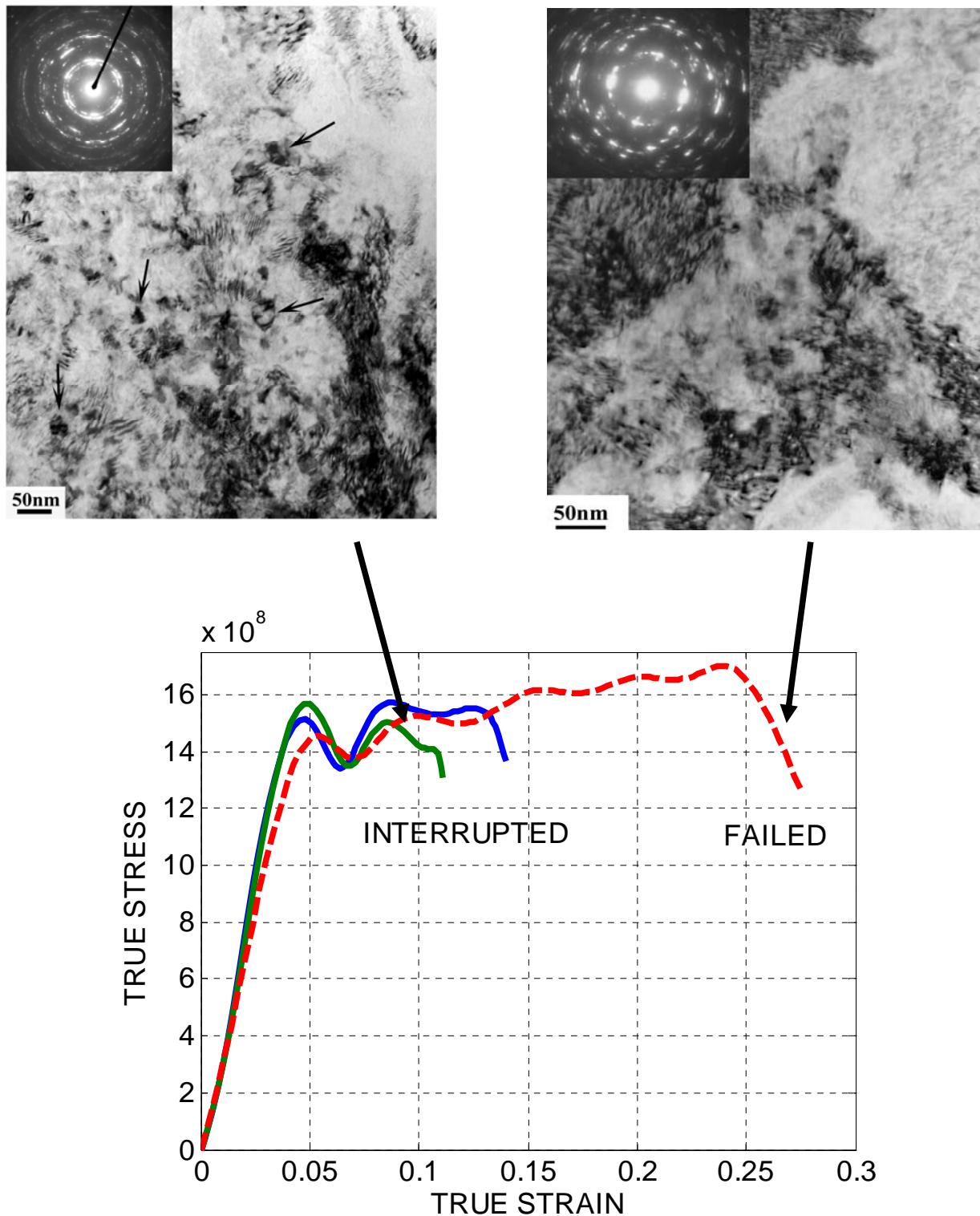


Figure 3: Dynamic shear tests of ti6Al4V specimens. The stress strain curves show 2 interrupted tests (no failure) and 1 test until failure. The corresponding transmission electron micrographs are indicated by arrows. The electron diffraction patterns are very similar and they both indicate a very fine microstructure comprised of recrystallized nanograins. Some of these grains are indicated by arrows in the left hand-side micrograph. The micrographs point to the fact that in these experiments, dynamic recrystallization was observed to develop significantly before any kind of failure by adiabatic shear developed in the specimen. (Reprinted after Rittel et al. (2008)).

FIGURE CAPTIONS

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