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Experimental investigation of transient thermoplastic effects in dynamic fracture

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Abstract

In a recent paper, we investigated the extent to which transient *thermoelastic* effects could affect dynamic crack initiation [Rittel, D., 1998a. Experimental investigation of transient thermoelastic effects in dynamic fracture. Int. J. Solids and Structures 35(22), 2959–2973]. In the present paper, we pursue this line of investigation by investigating transient *thermoplastic* effects. Three characteristic experiments are described which address the assumptions of the isothermal nature of dynamic crack initiation in two representative materials: 'ductile' polycarbonate and 'brittle' polymethylmethacrylate.

First, by varying the global (baseline) temperature of dynamic fracture experiments, it is shown that these two materials react quite differently. The crack-tip temperature of the PC specimen rises significantly as evidenced from the failure mode at both the macroscopic and microscopic scales. By contrast, the failure mode of PMMA is apparently unaffected by temperatures below its Tg. This experiment demonstrates further that the *local* crack-tip temperature is likely to be different from the *baseline* temperature.

Next, we investigate the bulk thermomechanical behavior of these two materials to quantitatively assess the temperature rise which accompanies transient adiabatic deformations. This behavior is deemed to be characteristic of that of the process zone material ahead of the crack-tip.

Finally, we present thermal and fractographic results about transient temperature recordings ahead of an adiabatic shear band in a PC specimen loaded in mode II.

These results show that thermomechanical couplings must be taken into account in dynamic fracture investigations. These couplings are related to the *loading mode*, to the *strain levels* reached in a typical process zone and also to the *thermomechanical response of the specific material* at such strain levels.

Keywords: Thermoelastic; Thermoplastic; Impact; Dynamic fracture; Polymers

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1. Introduction

It has long been acknowledged that fracture is a dissipative process during which *either* plastic work is converted partly into heat (thermoplastic coupling, Taylor and Quinney, 1934) *and/or* heat is released during the fracture process itself (as mentioned e.g. in Griffith, 1920). When the process is adiabatic, the released heat causes a local temperature elevation which can markedly alter the mechanical characteristics of the material. As an example, consider adiabatic shear band formation during which the local temperature in the band can rise quite substantially (Marchand and Duffy, 1988). It is also established that the assumptions of adiabaticity during dynamic crack initiation are reasonable, as shown experimentally by Zhou et al. (1996).

While the previously mentioned *thermoplastic* coupling is the effect which is more generally investigated, one additional effect—the *thermoelastic* effect—is seldom mentioned and is generally overlooked in dynamic fracture studies. Both thermoelastic and thermoplastic couplings have been studied at relatively low strain rates and cyclic loading by Dillon and Tauchert (1966) and by Dillon (1963). Considering cracks, the nature of the crack-tip fields indicate that the linear thermoelastic crack-tip material should *cool down* upon (adiabatic) mode I loading. While very few mentions of this dilatancy related effect are found in the literature (Zehnder and Rosakis, 1991), Fuller et al. (1975) indeed reported crack-tip cooling ahead of the propagating crack tip in their study of polymers. Rittel (1998a) studied mode I dynamic crack initiation in a 'brittle' commercial polymethylmethacrylate and reported a noticeable crack-tip cooling phase during which the crack started propagating.

At this point it is clear that whereas the *mechanics* of dynamic crack initiation may be treated assuming isothermal concepts (no thermomechanical couplings), the criterion (a) for dynamic fracture which are based upon attainment of local material properties (e.g. toughness) *must* take into account the existence of thermomechanical couplings. Such couplings are dictated both by the nature of the crack-tip fields, (opening, shear), by the strain levels involved, and also by the thermomechanical behavior of the material itself. While the previous statement may seem obvious, this physical fact is seldom taken into account and experimental investigations of the subject are still scarce.

In our previous work, we described experiments whose aim was to show that thermomechanical couplings during dynamic fracture are not negligible. We now present additional new results, in the continuation of these works, in an attempt to investigate the high strain rate thermomechanical behavior of the bulk material which represents the material found in the crack-tip process zone. A better characterization of this behavior is thus used to understand dynamic crack initiation at the structural (specimen) size. Therefore, we report in this paper three distinct and complementary experiments which are aimed at shedding additional light on the nature of the thermomechanical couplings associated with dynamic crack initiation.

The first issue addressed is that of the *local crack-tip temperature elevation* vs. variations in the *global temperature*. The underlying question is that of the validity of the isothermal assumptions in dynamic crack-initiation.

The second issue is that of the *thermomechanical constitutive behavior* of two representative polymeric materials (one 'brittle' and the other 'ductile') and its relation to dynamic crack initiation. Given two materials, one more 'brittle' than the other, how do these materials react under adiabatic transient bulk deformations?

The last point addressed is that of *direct crack-tip transient temperature sensing* in correlation with fracture micromechanisms. Here, following a previous thermal study of mode I loading in PMMA (Rittel, 1998a), we present new results about real time temperature sensing at the tip of an adiabatic shear band in PC.

Throughout this work, two representative polymeric materials are used: commercial polycarbonate,

PC, (representative of a 'ductile' material) and commercial polymethylmethacrylate, PMMA, (representative of a more 'brittle' material). These polymers are glassy and they are characterized by their glass transition temperature (Tg, of 150°C and 100°C for PC and PMMA respectively).

The paper is divided into three sections, each reporting and discussing a typical experiment, followed by a discussion and concluding section.

2. Results

2.1. Experiment 1: mode II dynamic loading at various temperatures

Ravi-Chandar (1995) showed that mode II impact of a notched polycarbonate plate can be accompanied by a failure mode transition according to the impact velocity. The crack which propagates at a well defined kink angle at low impact velocities (typically < 50 m/s) no longer propagates when the impact velocity exceeds 50 m/s. This author observed the formation of adiabatic shear band at the tip of the non propagating crack. The failure mode transition (from opening to shear) was observed earlier in maraging steel (Kalthoff, 1988) and the concept was later extended to general mixed-mode loading (Rittel et al., 1997a, 1997b).

These experiments clearly showed that the crack-tip material heated adiabatically and softened to an extent which was sufficient to delay or suppress dynamic crack initiation. In order to assess the magnitude of the *temperature rise* involved we repeated these experiments by varying the global temperature of the specimen while keeping a relatively (high) constant impact velocity ($v_{\text{imp}} > 50$ m/s).

The experimental procedure has been described in several previous papers and will not be repeated in detail here (Rittel et al., 1997a, 1997b). The specimen (a rectangular, fatigue precracked plate) is brought in contact with an instrumented steel bar (Kolsky bar). A striker is fired against this bar, which induces a compressive stress wave in the bar. The wave reaches the specimen, interacts with the crack until fracture, causing essentially mode II loading. The specimen is unsupported and fractures by inertia only.

PC specimens were tested in the range of temperatures from -120°C to $+70$°C. The PMMA specimens were tested in the rage of temperatures from room temperature to about $+123$°C. The range of test temperatures was chosen as a function of the known material response at room temperature, including the failure mode transition in PC (Rittel, 1998d) which was not observed for PMMA.

Fig. 1 shows identical specimens of PC and PMMA which were impacted at various global (baseline) temperatures at the time of impact. For the PC specimen, a reduction of the kink angle from over 60° at low temperature to less than 40° at -25°C (0.59 Tg) is clearly evidenced. As the specimen temperature is increased from room temperature and above, the crack no longer propagates and a shear band is clearly distinguished in accord with previous reports. Fig. 2 is a scanning electron fractograph of the fracture surface of the specimen impacted at 0.59 Tg (-25°C). In this figure, signs of ductile shear fracture (elongated dimples) are clearly observed in a band roughly 200 μm wide ahead of the crack-tip. Such band is in fact an adiabatic shear band (Ravi-Chandar, 1995) which indicates significant heating up to near Tg. Consequently, a simple calculation indicates that the temperature elevation at the tip of the crack is of the order of 175°C.

By contrast, the behavior of PMMA is quite different. This material fractures in all experiments at a high angular value (typically 60° and above). This value is characteristic of local opening failure mode, governed by a maximum normal stress criterion (brittle material). Fracture is significantly suppressed or delayed only when the global temperature exceeds Tg (at 1.1 Tg).

This observation clearly indicates a lack of significant adiabatic crack-tip heating in this experiment (combination of material and loading mode), contrary to PC. It also shows that two glassy polymers reacted quite differently under identical experimental conditions. The 'ductile' PC crack-tip material

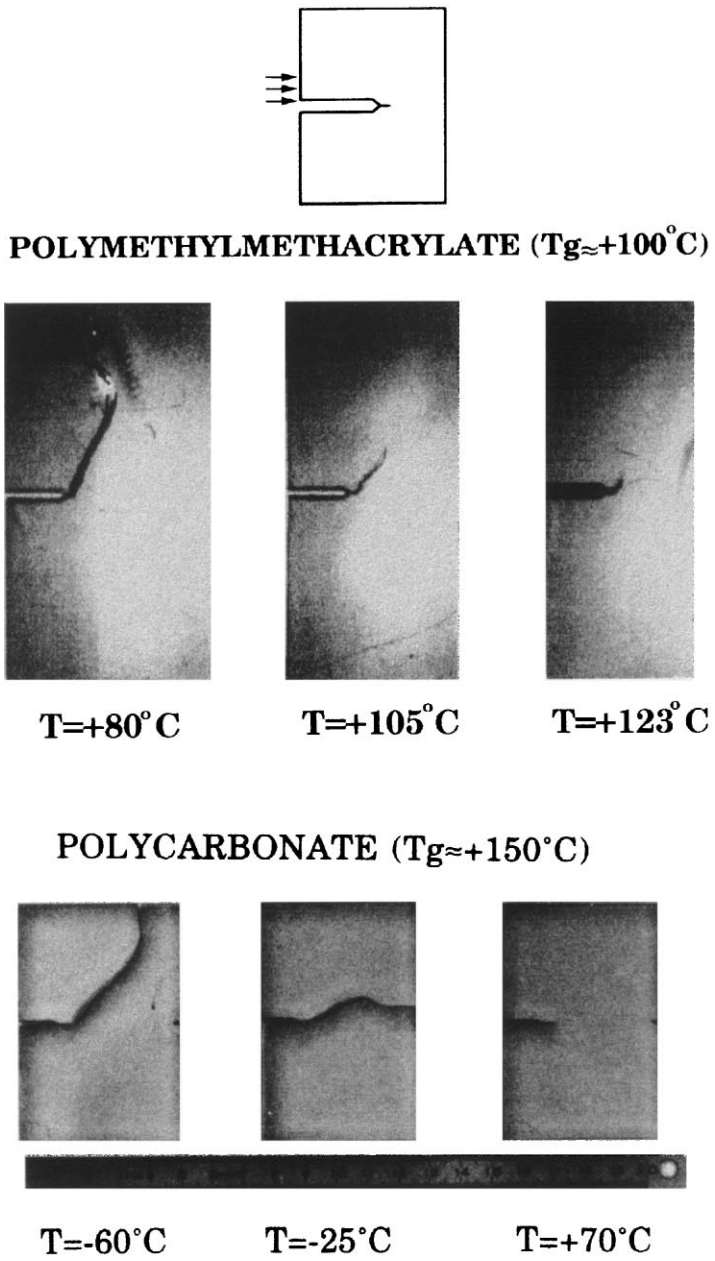

Fig. 1. Side impact experiments of fatigue precracked plates made of PMMA and PC, all impacted at velocities exceeding 50 m/s. The global test temperature is indicated under each specimen. Note that cracking is significantly reduced (suppressed) when the PMMA specimen is heated above its Tg temperature. By contrast, the PC specimen tested at −25°C exhibits a noticeable reduction of the crack kink angle. This experiment indicates that a different failure mode at initiation in this specimen, as a result of crack-tip heating.

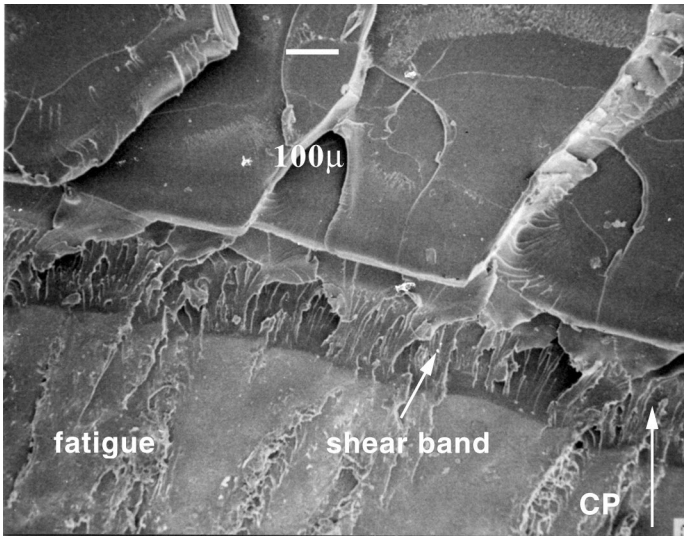

Fig. 2. Scanning electron fractograph of the PC specimen tested at −25°C. A 200-μm wide adiabatic shear band, characterized by elongated dimples is clearly visible ahead of the fatigue precrack. This band indicates noticeable adiabatic crack-tip heating. Beyond the band, fracture proceeds by an opening mechanism. CP indicates crack propagation direction.

(self) heated significantly while the PMMA remained unaffected until the global softening temperature was reached.

These experiments illustrate the *thermomechanical autonomy* of the crack-tip material, in the sense that it may or may not reach temperatures which are quite different from the baseline temperature. Such observation bears direct implications to the ductile to brittle transition testing during which the global specimen's temperature is varied while the local failure mechanisms are assessed, assuming isothermal conditions. It is therefore felt that additional *quantitative* information about the thermomechanical response of the crack-tip material (process zone) is necessary to provide further insight into the failure mechanism. Such information is provided my measuring the magnitude of the temperature rise which accompanies high strain rate deformation of polymeric disks. This result is presented in the following section.

2.2. Experiment 2: transient temperature sensing during dynamic deformation

The extent of the temperature rise in the two polymers was investigated by subjecting small disks to high velocity stress wave loading in a standard Kolsky apparatus (split-Hopkinson). The transient temperature was monitored using a small (wire diameter 127 μm) T-type thermocouple embedded in the polymeric disk. The thermocouple was sealed by means of home made liquid PC and PMMA according to the investigated material. The experimental procedure and the physics underlying the fast transient response of the embedded thermocouple have been described in detail and will not be repeated here (Rittel, 1998b). For the sake of brevity, it will be mentioned that the heat transfer problem of the embedded sensing tip is related to heat conduction rather than to heat convection (Carslaw and Jaeger, 1959). Consequently, the response of the sensor is extremely fast, of the order of a few microseconds. The embedded thermocouple technique was thus used to investigate the conversion of mechanical into thermal energy at high rates of strain in commercial PC (Rittel, 1999). We now report additional new results about the response of PMMA.

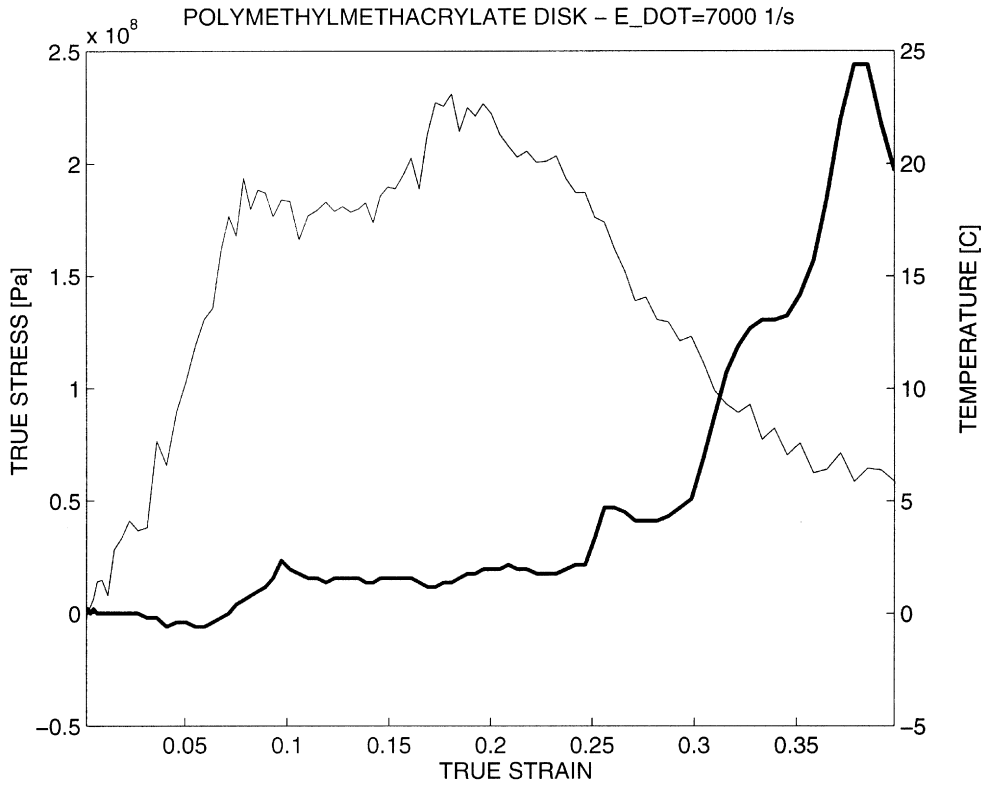

Fig. 3. True stress, true strain and temperature recording during transient deformation of a PMMA disk tested at an average strain rate of 7000 s^{-1}. Note that the temperature rises moderately in the nonlinear domain, up to $\varepsilon \approx 0.2$, and jumps abruptly in the strain softening regime until final failure.

PMMA disks (2–5 mm thick, 10 mm diameter) were machined out of a stock plate. Two groups were subsequently formed. The first group was left untouched for subsequent testing. The second group of specimens were annealed at 100°C for 20 min followed by slow furnace cooling to minimize the effects of residual stresses due to machining. Such specimens were prepared to allow for comparison with previously reported results at lower strain rates of annealed disks (Arruda et al., 1995). The following results were compared with those obtained from specimens without prior annealing treatment. We did not observe a substantially different response as a result of the annealing treatment.

In Figs. 3 and 4 we have plotted the dynamic true-stress, true-strain and the associated temperature rise of two characteristic specimens which were deformed at a very high average strain rate (from 6500 to 7000 s^{-1}). While the dynamic mechanical properties of PMMA and PC are different, as expected, the thermal signals bear a certain degree of similarity. It is first noted that in the 'macroscopically' elastic domain (up to about 0.05 for PMMA and 0.1 for PC), little temperature rise is evidenced for both polymers. The temperature increases in the nonlinear domain, as expected. For PC an abrupt temperature jump is observed in the strain softening regime at strains in the range 0.15–0.35. By contrast, the temperature rise of PMMA is relatively small (less than 5 K) until the stress has dropped quite significantly (from 0.25 and above). A common point to these polymers is that the temperature does not rise significantly for strains inferior to 0.15 in PC and 0.25 in PMMA. A similar observation was reported by Trojanowski et al. (1997) who investigated the response of an epoxy resin compressed

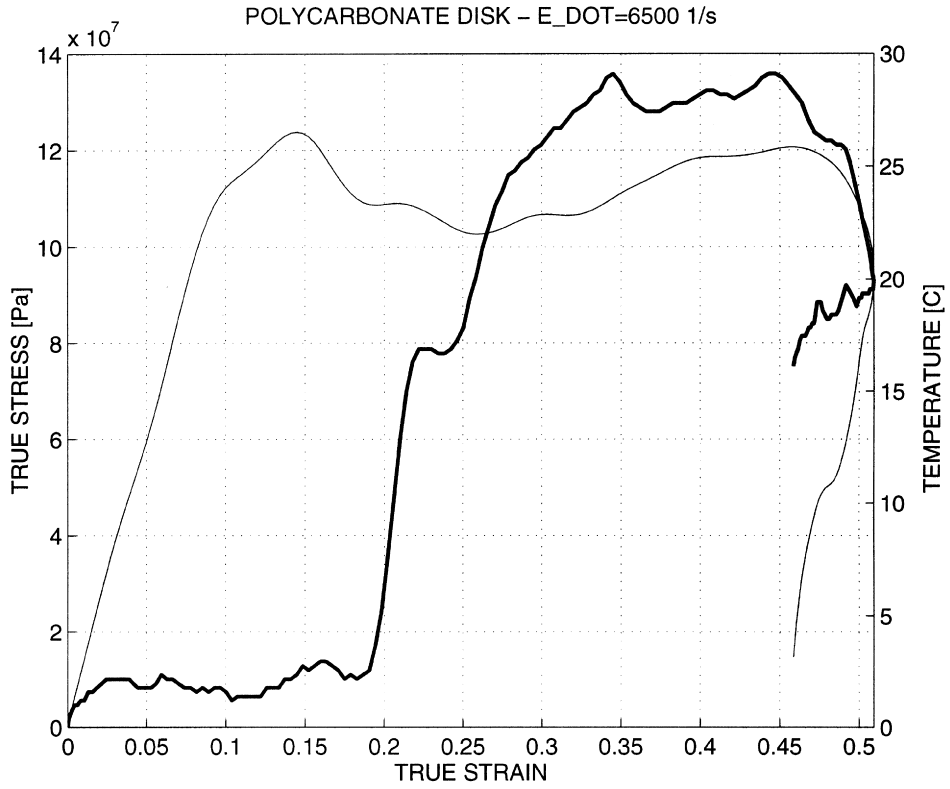

Fig. 4. True stress, true strain and temperature recording during transient deformation of a PC disk tested at an average strain rate of 6500 s^{-1}. Note the very small temperature rise up to $\varepsilon \approx 0.2$ followed by the steep increase in the strain softening regime at $\varepsilon \approx 0.2$–0.35.

at relatively high strain rate (2500 s^{-1}). These authors used infrared monitoring of the specimen and they reported negligible heating for strains inferior to 0.15. In their work, the dominant temperature jump was observed in the post-peak regime ($\varepsilon \geqslant 0.6$). Such real time temperature monitoring in these materials is seldom carried out. However indirect techniques (heat sensitive films) have been used to assess the maximum temperatures and their spatial distribution. Swallowe et al. (1986) and Walley et al. (1989) used high speed photography together with heat sensitive film to investigate the dynamic failure of polymers. Their results show that PC is capable of sustaining fairly large strains ($\varepsilon \geqslant 1$) in compression prior to fracture by shear banding. By contrast, the "brittle" PMMA shatters at relatively small strains ($\varepsilon \approx 0.05$). These authors also report that plastic deformation causes temperature rises inferior to 200 K whereas cracking and fracture cause a temperature rise in the range of 500–700 K. It should be noted that the heat sensitive film technique employed in these works is not capable of detecting temperature rises inferior of 200 K so that this figure is a kind of 'upper bound'. By contrast our results indicate much smaller temperature rises both in PC and in PMMA. This result concurs with the results of Trojanowski et al. (1997).

Additional comparison can be established with those of Arruda et al. (1995) who investigated the thermomechanical coupling in annealed PMMA disks loaded in compression at various strain rates, using infrared temperature monitoring. These authors report that for a strain rate of 0.1 s^{-1}, the mechanical process is adiabatic. In Fig. 5 we have superimposed the mechanical results of Walley et al.

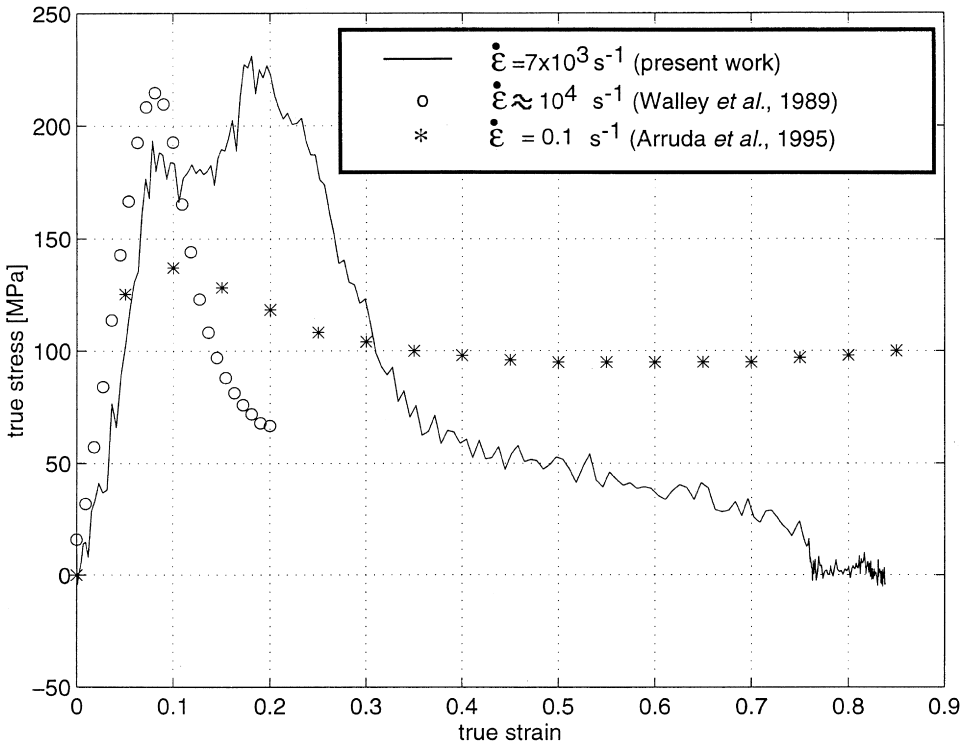

Fig. 5. True stress vs. true strain results for PMMA specimens tested in compression at a strain rate of 0.1 s^{-1} (Arruda et al., 1995), ca. 10^4 s^{-1} (Walley et al., 1989), and of 7000 s^{-1} in the present work. As expected, the flow properties are affected by the strain rate.

(1989), those of Arruda et al. (1995) and ours. A broad agreement is noted between the high strain rate data in the sense that a higher strain rate increases the flow stress while reducing the ductility.

In Fig. 6 we have superimposed the corresponding thermal results of Arruda et al. (1995) and ours. This comparison shows that at the lower strain rate, the temperatures rises markedly for strains between 0.2 and 0.6. At the higher strain rate, the temperature rises steeply when the strain exceeds 0.3. This significant temperature rise can thus be attributed to specimen fracture rather than plastic deformation on the basis of the high speed photographic observations of Walley et al. (1989).

Finally the comparison of results from three different sources and different strain rates clearly shows that, irrespective of the strain rate, the temperature rise relatively insignificant for strains inferior to 0.2. This observation was consistently made for three different polymers using two different temperature monitoring techniques (infrared and thermocouples) and one indirect technique.

The reported temperature rises are obtained for dynamic compression. It is well known that larger strains can develop in compression than in tension as the necking instability is prevented. One can reasonably assume that the tensile thermomechanical behavior is the identical counterpart of the compressive one with a smaller strain range.

On these premises it can be concluded that the 'ductile' polycarbonate will markedly heat up at moderate nonlinear strains (i.e. $\varepsilon \geqslant 0.2$) by contrast with the more 'brittle' polymethylmethacrylate, which will fail in tension at strains of this order (or smaller) for which heating is still negligible. This result complements and corroborates the observations reported in the previous set of experiments, namely the local (crack-tip material) temperature rise in the PC sample as opposed to the apparent lack

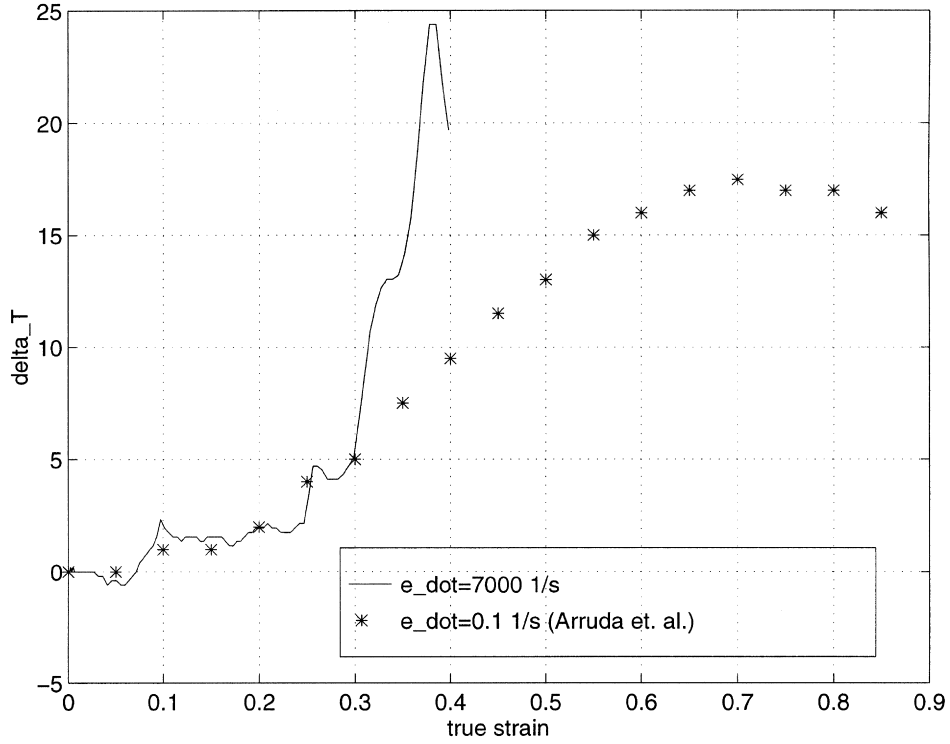

Fig. 6. Temperature rise vs. true strain for the experiments of Fig. 5. The temperature rise is very similar regardless of the strain rate up to strains of about 0.3. Beyond this strain, the rates of heating differ markedly.

of local heating in the PMMA sample. These observations can be directly transposed to the material in the process zone. Therefore, when applied to dynamic crack initiation testing, these results emphasize the importance of the thermomechanical constitutive behavior of the (crack-tip) tested material.

2.3. Experiment 3: transient temperature sensing during mode II dynamic fracture

In the last set of experiments, small thermocouples were embedded at a distance inferior to 10^{-3} m from the front of a fatigue precracked mode II specimen. The experiment was identical to the experiment of Section 2.1). The idea was to monitor in real time the temperature variations associated with dynamic crack initiation (as in Rittel, 1998a). In a typical experiment, the thermocouple recorded a temperature rise shortly after the stress wave had reached the specimen. Fracture was indicated by a single wire fracture gage. As previously stated, fracture is a dissipative phenomenon, whether it is accompanied by plastic deformations or not (see for instance the experiments on fracture of glass plates by Weichert and Schönert, 1978). Consequently, we will present here one (of the few) particularly successful experiments in which the temperature rise was actually recorded *at the tip of the shear band* during its formation. The success of this experiment depended on the fracture path. Success was achieved when the adiabatic shear band grew and reached the thermocouple prior to cracking at a kink angle which most often avoided the thermocouple.

In Fig. 7, the incident pulse, temperature rise and the indication of fracture are shown for one high velocity impact experiment ($v_{\text{imp}} \approx 50$ m/s). The stress wave reaches the specimen at $t \approx 140$ μs, typically 70 μs after the beginning of the incident pulse. This delay corresponds to elastic wave propagation in the

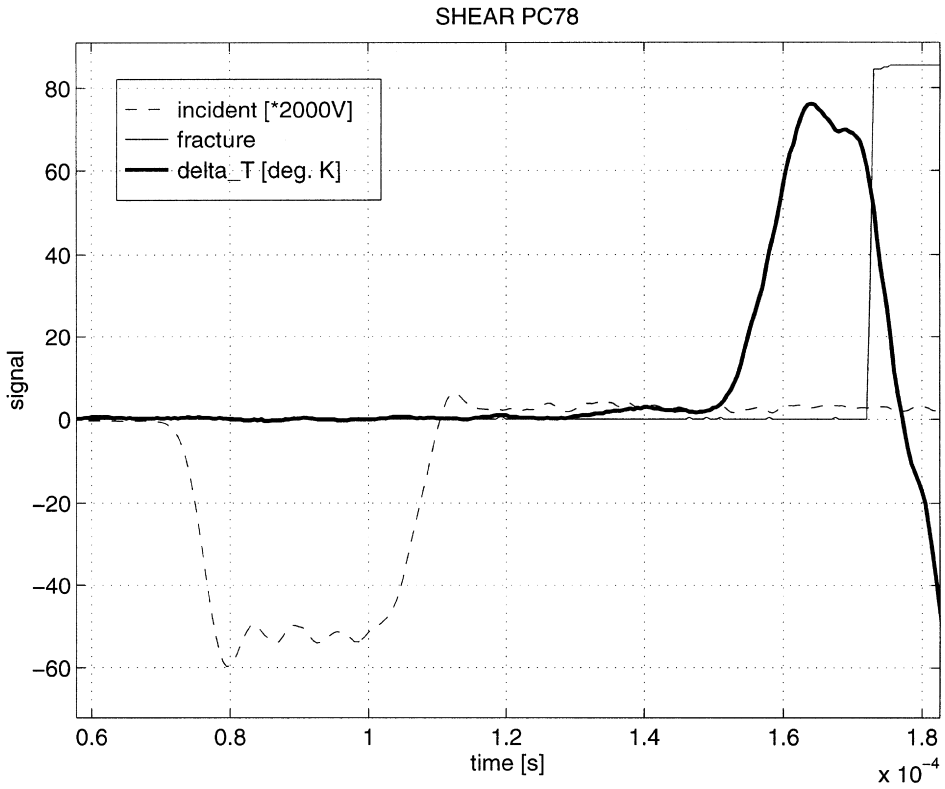

Fig. 7. Dynamic fracture experiment with PC fatigue precracked plate (as in Fig. 1) impacted at 52 m/s. The dashed line indicates the incident pulse which loads the specimen about 70 μs after its onset, at $t \approx 140$ μs. The stress wave interacts with the crack tip at $t \approx 150$ μs and the thermocouple indicates significant temperature rise. Fracture is detected by the single wire fracture gage at $t \approx 170$ μs, 30 μs after specimen loading. The temperature rise is of about 75°C which is of the order of the estimated 175°C from the first experiment.

steel bar. The temperature rise occurs shortly after the wave loads the crack at about $t \approx 150$ μs. Therefore, the temperature rises noticeably some 10 μs after the stress wave loads the specimen until final fracture which is indicated at $t \approx 170$ μs by the fracture gage. This indication corresponds to fracture occurring 30 μs after the stress wave loads the specimen. Beyond fracture time, the thermal signal oscillates strongly and the physical mechanism responsible for these oscillations is not clear at this point.

Previous simulations of the crack-tip loading state have been carried out for similar experiments by Rittel et al. (1997a, 1997b). These calculations showed that during about the first 130 μs, the crack-tip experiences almost pure mode II loading later followed by mixed mode loading with a pronounced positive mode I component. While these calculations did not account for the inclusion of the embedded thermocouple in the vicinity of the crack-tip, they nevertheless indicate that the initial 30 μs correspond to mode II loading.

Referring back to the first experiment reported in this paper, the fracture surface in the vicinity of the thermocouple is now shown in Figure 8. This figure clearly shows the familiar elongated dimples which characterize the adiabatic shear band, in the very vicinity of the embedded thermocouple. For this successful experiment, a temperature rise of the order of 70°C is detected in real time at about 1 mm from the initial crack-tip. It must be remarked that the adiabatic shear band grows toward the

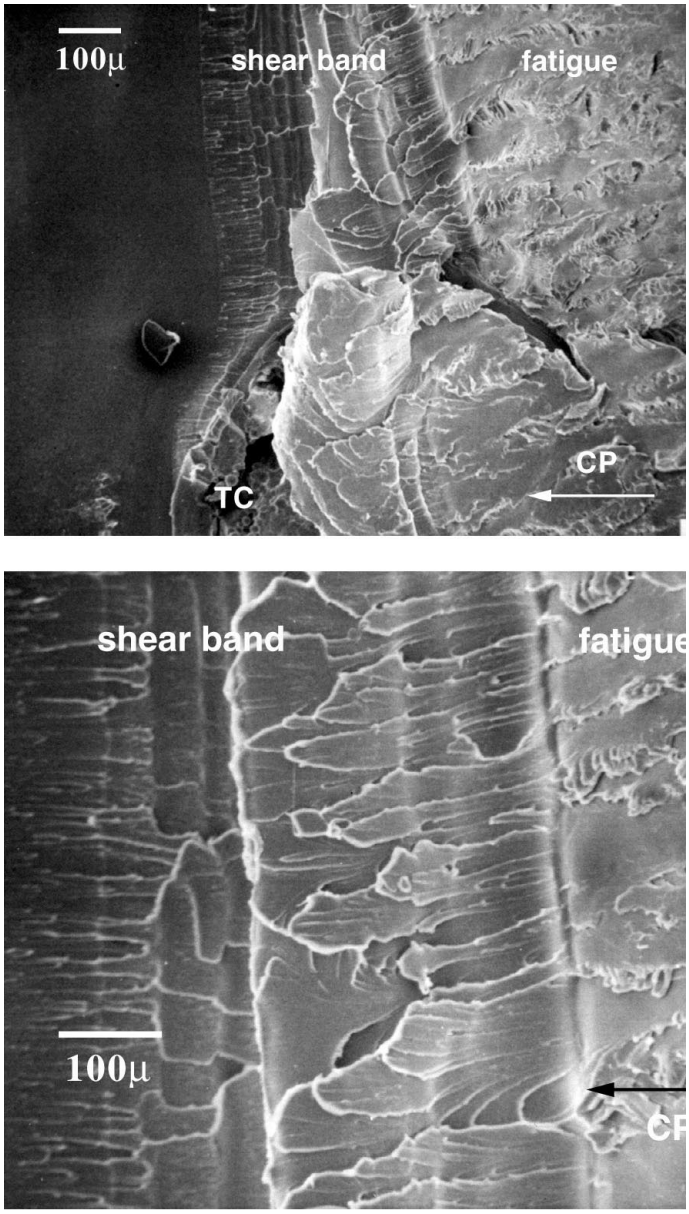

Fig. 8. Scanning electron fractograph of the PC specimen whose recording is shown in Fig. 7. A 200-μm wide adiabatic shear band, characterized by elongated dimples (as in Fig. 2) is clearly visible ahead of the fatigue precrack in the immediate vicinity of the thermocouple's location. This band corresponds to adiabatic crack-tip temperature rise sensed by the thermocouple. CP indicates crack propagation direction.

thermocouple so that the recorded temperature is that of the moving band-tip. Keeping this point in mind, it may nevertheless be noted that the extent of the observed temperature rise is of the same order of magnitude than that assessed in the first series of experiments.

Monitoring of the temperature in adiabatic shear bands has been mostly carried out in metals using

infrared detectors (Marchand and Duffy, 1988; Zhou et al., 1996). This experiment complements these works by providing original results for a polymer using thermocouple sensing.

3. Discussion

In this work, we have tried to present a consistent framework for the investigation of the nature of thermomechanical couplings at the tip of dynamically loaded cracks. In this context, the dynamic fracture behavior of the specimen (structural scale) is studied *along with* the thermomechanical characteristics of the same bulk material which is deemed to represent the material adjacent to the crack-tip (local scale). Deformations were carried out in the adiabatic regime such as to cause (whenever possible) crack-tip heating. The investigation of two distinct materials (PMMA and PC), each of which representative of a broad class of materials has shown the following facts:

- The global temperature of the specimen and the local crack-tip material's temperature are likely to be different. The crack-tip is therefore autonomous from a thermomechanical point of view.
- Only for a material which remains essentially (thermo)elastic or does not heat up noticeably during inelastic deformation (e.g. PMMA in the present work), can it be assumed that the crack-tip temperature is equal to the global temperature. In such case the usual isothermal assumptions of dynamic crack initiation are justified.

  The nature of the crack-tip temperature changes are dictated by the loading mode (shearing vs. opening), by the magnitude of the overall strains involved and the tendency of the material to strain soften, as experienced with the present polymers. The joint knowledge of these factors will determine the response of the crack-tip material in a given test. Beyond the specific interest of polymeric behavior, it is felt that the present observations are not restricted to these materials and apply equally well to other materials.

  Consequently, taking into account the present results might shed additional light on two issues which are seemingly related to thermomechanical couplings.
- The first concerns the reported variations of dynamic fracture (initiation) toughness which do not seem so far to obey any logical criterion.
- The second related point is that of the ductile to brittle transition in materials. This phenomenon is generally assessed by subjecting metallic specimens (Charpy) to impact testing at various *global* temperatures. Without questioning the invaluable information about the transition itself, the exact crack-tip temperature at initiation and the corresponding failure mechanisms should provide additional information on the *local conditions* at initiation.

4. Conclusions

The following conclusions can be drawn from the present work:

- Dynamic (or more generally adiabatic) crack initiation is not necessarily an isothermal event.
- As a result, the crack-tip temperature is likely to be different from the global specimen temperature at initiation.
- The dynamic failure criterion should take thermomechanical couplings into account.
- The nature of the thermomechanical couplings associated with dynamic crack initiation are dictated by the loading mode(s), the strain levels (solid mechanics) and by the thermomechanical response of the crack-tip material (materials science).

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