Adiabatic heating and energy absorption capability of an advanced high strength steel during drop weight impact testing

Peikang Xia \textsuperscript{a,b,*}, Mario Sanchez-Palomar \textsuperscript{b}, Ilchat Sabirov \textsuperscript{a}

\textsuperscript{a} IMDEA Materials Institute, Calle Eric Kandel 2, Getafe, 28906 Madrid, Spain
\textsuperscript{b} Universidad Politécnica de Madrid, E.T.S. de Ingenieros de Caminos, 28040 Madrid, Spain

Abstract

The manuscript focuses on the energy absorption capability of an advanced high strength steel during drop weight impact testing. The \textit{in-situ} measurements of temperature during impact resistance testing of samples revealed intensive adiabatic heating effect with the peak temperature of 225 °C at the top of the dome resulting in the local softening effect. The drop weight impact resistance of the material under present testing condition is 90 J. Microstructural characterization demonstrated that dislocation glide and formation of substructure is the main deformation mechanism. Analysis of fracture surface of cracked samples (tested with the impact energy of > 90 J) revealed ductile failure mode. The energy required for formation of fracture surface was calculated by quantitative analysis of the 3D digital models generated using microscopy images of the fracture surface taken from different angles. This energy is negligibly lower compared to the total energy spent for plastic deformation of the material.

**Keywords**: Advanced high strength steel; adiabatic heating; impact resistance; biaxial stress; fracture surface analysis

1. Introduction

1.1 Advanced high strength steels

Nowadays, decreasing the weight of vehicles (i.e. car body light-weighting) is one of the priorities for automobile manufacturers due to the imposed safety and carbon emission regulations and limited fossil energy \cite{1}. However, another car performance that should definitely not be weakened during light-weighting is passengers’ safety, which is also a...
dilemma between making a light car and making a safe car. One promising possible solution for this challenge is to embrace the advanced high strength steels (AHSSs), which have a good combination of strength and ductility [2]. Although carbon fiber composites [3] and aluminum alloys [4] do have preferable specific strength and have been used in a few car models, their extremely high cost handicapped their wide application in civilian vehicle market. It is expected that in the near future, AHSSs will be the main contributor to car body light-weighting due to its low cost and admirable mechanical properties. Over the last decades, a serials of AHSSs have been developed, and some of them have already been utilized in automobile components, such as transformation induced plasticity (TRIP) steel [5], twinning induced plasticity (TWIP) steel [6], complex phase (CP) steel [7], dual phase (DP) steel [8], quenching and partitioning (Q&P) steel [9], etc. Among these candidates, DP steel is an up-and-coming competitor because of its high strength as well as lean alloy addition; the former is favorable for reducing vehicle weight while the latter is advantageous for cutting cost. Typically, the microstructure of DP steel produced using a simple thermomechanical processing route is composed of the ferritic matrix with embedded martensitic grains providing a wide range of excellent mechanical properties, which can be tuned via modifying the morphology and volume fraction of microconstituents. It has been shown that the ultimate tensile strength (UTS) of DP steel can vary from ~500 MPa to more than 1000 MPa with varying martensite fraction via heat treatment or tailoring alloying elements [10–13]. Benefited from martensite and ferrite, high tensile strength and low initial yield stress (YS) can be easily achieved in DP steel. These mechanical advantages, where high UTS is desirable for weight-lighting and low YS is profitable for sheet-forming process, make DP steel a preferable choice in the automobile-related application market. Moreover, depending on the alloying elements and applied thermomechanical processing route, bainite [14], retained austenite [15] and carbides [16] can also form in the microstructure, that provides an extra tool to control its mechanical behavior.

1.2 Impact resistance

The significant body of research done on DP steels can be classified into two realms: 1) Optimization of chemical composition and thermo-mechanical processing routes (i.e. microstructure and mechanical properties); and 2) Study of their deformation, fatigue and
fracture behavior under different stress states/testing conditions. In the first category, various DP steel grades satisfying a broad scope of engineering requirements have been developed [8,11,12,17–19]. On the other hand, the features of their mechanical behavior (such as distribution of stresses/strains between microconstituents) during plastic deformation, work hardening, shear banding, local crack initiation and growth, etc. have also been thoroughly investigated on both macro- and micro-scale, via experiments and simulation [20–27]. Impact resistance is one of the most important properties of AHSS for automotive applications. For DP steels, their impact resistance have been studied mainly by means of Charpy impact testing [13,16,28–37], which is a standardized method to evaluate material impact toughness. An overview of Charpy impact energy values at room temperature reported for different DP steels is presented in Table 1. As can be seen, the measured values vary dramatically in the range of 1.7 to 235 J. It was shown that Charpy impact toughness of Vanadium-strengthened DP steel can be improved via replacement of pearlite by more ductile martensite [16]. Asim Bag et al. found that the Charpy V-notched impact energy of a Fe-0.16C-1.32Mn-0.44Si (wt. %) DP steel is maximum at 60% of martensite volume fraction [13]. Simulation conducted by Y. Prawoto et al. using Johnson-Cook model led to the similar optimum martensite volume fraction for enhanced Charpy impact resistance [28]. Computational modelling confirmed that both the constituent fraction and its morphology have a great influence on the Charpy impact resistance of martensite-ferrite DP steel [28]. Detailed investigation on DP 590 steel revealed that the ductile-brittle transition temperature (DBTT) is -95 °C, which is far below general automobile service temperature [31]. Another study on a hot-rolled DP 590 steel grade concluded that: 1) the Charpy impact energy in ductile-brittle transition range can be increased by splitting caused by silicate and carbide inclusions; 2) splitting reduces DBTT as well [36,37]. The survey on AISI 4340 steel with different microstructure composition via adjusting heat treatment demonstrated that the combination of bainite and martensite has better Charpy impact performance and tensile ductility than martensite-ferrite or full bainite microstructure [33]. The study on AISI 3115 steel pointed out that the impact strength of this steel decreases with the increasing intercritical annealing temperature, which affects the microstructure significantly [34]. It should be noted that, bars with dimensions of 10×10×55 mm are used for Charpy impact testing [38,39], so this method cannot be utilized for characterization of common automobile
AHSS steel sheets. Moreover, Charpy test is performed on notched samples, so deformation process is not relevant to the scenario, where automobile components are confronted during crash accidents.

Table 1. Values of Charpy impact energy measured on various DP steels.

| Chemical composition (wt. %) | T (°C) | Sample section size (mm) | Impact energy (J) | Ref. |
|------------------------------|--------|--------------------------|-------------------|------|
| Fe-0.16C-1.32Mn-0.44Si-0.09Mo | 25     | 10×10                    | 96                | [13] |
| Fe-0.14C-1.23Mn-0.34Si-0.10V  | 25     | 6.25×10                  | 189               | [16] |
| Fe-0.08C-1.24Mn-0.09Si-1.23Cr | 20     | 5.5×10                   | 70                | [31] |
| Fe-0.15C-0.90Mn-0.25Si-0.10Mo-0.80Cr-1.20Ni | 25 | 10×10                     | 109       | [34] |
| Fe-0.08C-1.24Mn-1.14Si-0.87Cr  | 31     | 5.5×10                   | 70                | [37] |
| Fe-0.16C-1.23Mn                | 25     | 10×10                    | 150               | [40] |
| Fe-0.17C-1.49Mn-0.22Si        | 25     | 3×4                      | 1.7               | [41] |
| Fe-0.07C-1.52Mn-0.34Si-0.05Nb-0.05V | 25 | 10×10                     | 235       | [42] |

Note: Information on the microstructure (phase volume fraction, grain size, morphology, etc.) is not provided in the table.

There have also been studies focused on impact performance of thin-walled [43–47] and flat sheet [48,49] specimens. Thin-walled columns composed of various cross-section shape and having various size were examined due to their similarity with automobile mid-rails. Experimental crush testing of hydroformed thin-wall steel tubes combined with simulation exhibited that DP steels had higher energy absorption capability compared with the high strength low alloy steel (HSLA) 350 and deep drawing quality steel [45]. Modeling and experiments on DP 800 steel manifested that the axial crushing performance of thin-walled specimen has great relationship with the shape of the sample (top-hat or square section) [44]. The study on DP steels with tensile strengths from 270 MPa to 1470 MPa by K. Sato et al. showed that yield strength is the most influencing factor on the bending moment in bending crash [50]. James R. Fekete et al. estimated that energy absorption capability of thin-walled samples increases by 10% when DP 340 is substituted by HSLA 340 steel [46].

Finally, just a few reports on drop weight impact resistance of flat-sheet specimen can be found in the literature. Odd Sture Hopperstad et al. studied the low velocity impact deformation process of monolithic and multi-layered DP 600 steel plates. Digital image correlation (DIC)
technique revealed that the plates’ resistance against perforation is closely connected with the shape of the impactor (e.g. blunt-ended or ogival-ended) [48]. Another successive study completed by that team on single DP 600 sheet demonstrated significant effect of strain rate and steel grade on the strain distribution during impact deformation [49].

From this literature overview, a few gaps exist in the area of energy absorption capability of advanced high strength steels. No systematic studies on impact resistance of flat sheets tested in the conditions relevant to crash accidents have been carried out up-to-date. The research done in this field has ‘engineering’ character, and no much attention has been paid to microstructure evolution and adiabatic heating during dynamic plastic deformation, despite their significant role. Therefore, the main objective of the present work is to study the drop weight impact resistance of an advanced high strength steel with respect to the microstructure evolution and adiabatic heating during dynamic plastic deformation.

2. Material and experimental procedures

2.1. Material

The material studied in this work is a commercial advanced high strength zinc galvanized DP 1180 steel. The chemical composition of the steel is shown in Table 2. As a basic alloying element in ferrous alloys, 0.12 wt.% carbon was added mainly for martensite strengthening and avoiding significant reduction in weldability [2]. Manganese is highly beneficial for hardenability and austenite stabilization. To prevent formation of banded microstructure, its content was limited to 2.48 wt. %. Chromium also stabilizes austenite and increases hardenability, but its content was confined to 0.59 wt. % because of cost control. Microalloying by 0.023 wt. % of niobium was done for grain refinement effect. The material was supplied in form of 1 mm thick sheets.
Table 2. Chemical composition (in wt. %) of the studied DP 1180 steel.

| C  | Si | Mn | Cr | P  | S   | Al  | Nb | Ti | Cu |
|----|----|----|----|----|-----|-----|----|----|----|
| 0.12 | 0.28 | 2.48 | 0.59 | 0.0085 | 0.002 | 0.035 | 0.023 | 0.026 | 0.01 |

2.2. Mechanical property characterization

Dog-bone tensile samples with a gauge length of 25 mm according to ASTM standard [51] were machined via electric spark cutting along the rolling and transverse direction. Their geometry is shown in Fig. 1. Tensile tests were carried out using a universal electromechanical testing machine (Instron 3384) with a constant crosshead speed of 1.5 mm/min, which corresponds to the initial strain rate of $\sim 10^{-3}$ s$^{-1}$.

![Fig. 1. Geometry of tensile samples.](image)

Flat samples with dimensions of 50×50×1 mm were used for drop weight impact tests, which were conducted on a drop weight impact testing system (Instron CEAST 9350). The punch had a hemispherical end with a diameter of 16 mm. The samples were firmly clamped by a pneumatic system under pressure of 6 bars. A schematic diagram of holder and fasten parts is shown in Fig. 2. To minimize the possible frictional effect, the punch tip was greased by a multipurpose lubricant (Super Lube, Syncolon®). The impact energy varied in the range of 20–100 J by adjusting the height of the drop weight of 5.3 kg. The K-type thermocouples were welded in the center on the backside of the sample to measure the adiabatic heating effect during testing.
Fig. 2. Schematic drawing of sample fixture and punch sectioned by a quarter.

The dome-shaped tested samples were cross-sectioned through the center, as shown schematically in Fig. 3. The true plastic strain accumulated at the top of the dome was determined as [52]:

$$ \varepsilon = \ln \frac{h_o}{h_f} $$

where $\varepsilon$ is the true strain, $h_0$ is the initial thickness of the sheet and $h_f$ is the thickness at the top of the dome after testing, as schematically shown in Fig. 3.

Fig. 3. Schematic presentation of a cross-sectioned sample [53]. The area marked by white square refers to the region selected for EBSD and TEM characterization.

2.3. Microstructure characterization

To reveal the microstructure of the as-received material, samples with mirror-like surface
were prepared following standard metallography grinding and polishing procedures using 0.25 μm paste at the final stage and etched using 12 % Na₂S₂O₅ water solution. A systematic manual point counting method [54] was used to determine the phase fraction on SEM images of etched sample. For EBSD characterization, specimens were polished using colloidal silica suspension (OP-S, Struers®) at the final stage. The region selected for EBSD characterization is marked by white quadrat in Fig.3. Both SEM and EBSD characterization were carried out on a FEI Helios NanoLab 600i field emission gun scanning electron microscope (FEG-SEM). Kikuchi patterns were recorded by a NordlysNano detector controlled by the Aztec Oxford Instruments Nanoanalysis software (version 4.2®). A voltage of 18 KV, a current intensity of 2.7 nA and a step size of 50 nm were used in EBSD analysis. Orientation data were post-processed and analyzed by HKL Channel 5 software (version 5.1®). Kernel Average Misorientation (KAM) maps were calculated with respect to the third nearest neighbors using the HKL software. The KAM maps allow to analyze distribution of plastic microstrain over the microstructure by means of calculating the orientation difference between point clusters [55]. Histograms of statistical distribution of plastic microstrain in the analyzed microstructure were generated based on the KAM map for each analyzed specimen.

A FEG S/TEM microscope (Talos F200X, FEI) was employed to conduct TEM examination of the as-received and tested samples operating at 200 KV. For the as-received material, thin foils were prepared on a TenuPol 5 (Struers®) using twin-jet electron polishing method. For the tested samples, lamellar specimens were milled from EBSD scanned area using a FEI Helios NanoLab 600i DualBeam focused ion beam (FIB) system. Observations were made in both bright and dark field imaging modes. Convergent beam (CBED) and selected area electron diffraction (SAED) patterns were recorded from interested areas. Additionally, X-ray energy-dispersive spectrum (EDS) mapping analysis was carried out using a nanoprobe in Scanning TEM (STEM) mode.

2.4. Fracture surface reconstruction

A detailed quantitative analysis of fracture surfaces formed after cracking of samples tested with impact energy of > 90 J was performed by employing a SEM-based microscopic surface reconstruction technique [56–58]. This microscopic topography technique can be
briefly described as following: a pair of SEM images, which were eucentricly tilted to different angles (0–10°), were used to create a 3D digital elevation model (DEM) by an algorithm which quantifies the surface height variation via searching and comparing the homologous points in the image pair [58]. Thus, one more dimension (depth, z axis) is granted to the two dimensional image, providing the possibility to measure dimple depth. In present study, SEM image pairs consisting of 1536×1103 pixels were taken at 0° and 5° tilting. A commercial MeX software (Alicona, Graz, Austria) was used to reconstruct the DEMs of fracture surface and its further quantitative analysis. The fracture surface profiles were extracted from the obtained DEMs and the depth of at least 40 dimples was measured. The outcomes of these measurements were used for estimation of the energy consumed for formation of fracture surface during impact.

3. Results

3.1. Microstructure and tensile mechanical properties of the as-received material

The microstructure of the as-received material is shown in Fig. 4a. Martensite and ferrite are main microstructural constituents. Martensite has lathy and blocky shape (e.g., blue and yellow arrows in Fig. 4b), while ferrite has quasi-equiaxial grains (selectively marked by red arrows in Fig. 4a). The volume fraction of ferrite is 25.5±4.5%. The spatial distribution of both martensite and ferrite is random, and the grain size of ferrite is in the range of ~1 µm to ~5 µm (Fig. 4a and 4b). There is also very low amount of ultra-fine grained austenite (0.23% determined by EBSD), as marked in green color in Fig. 4b. It should also be noted that the as-received sheets are coated by a very thin hot dip galvanized zinc layer (Fig. 4c), having a thickness of 15.5±1.6 µm. Both retained austenite and zinc layer have a negligible effect on the mechanical properties of the material, so their role are not considered further in the paper.
Fig. 4. Microstructure of the as-received DP 1180 steel: (a) typical SEM image of microstructure (ferrite is marked by red arrows) of the as-received material; (b) band contrast map (austenite is marked by green color); (c) SEM image of the coated zinc layer on DP steel obtained using backscatter electron (BSE) detector.

The typical tensile engineering stress-strain curves of the studied DP steel are presented in Fig. 5. The difference of tensile curves between rolling (RD) and transverse direction (TD) is quite small, indicating the minor anisotropy and excellent quality of this commercial DP steel. The basic mechanical properties of this steel are summarized in Table 3. In RD, the steel has YS of 1062 MPa, UTS of 1244 MPa, elongation to failure of 8.2% and strain hardening exponent of 0.1, featured with ultra-high strength and limited ductility. The obtained mechanical properties in TD are very close to that of RD, as shown in Table 3.

Fig. 5. Typical engineering stress-strain curves for the studied DP 1180 steel in rolling and transverse directions.
Table 3. Mechanical properties of the studied DP steel. (YS: yield strength, 0.2% proof stress, UTS: ultimate tensile strength, EL\textsubscript{uniform}: uniform elongation, EL\textsubscript{total}: total elongation, n: work hardening exponent, K: constant).

|    | YS / MPa    | UTS/ MPa  | EL\textsubscript{uniform}/ % | EL\textsubscript{total}/ % | n       | K       |
|----|-------------|-----------|-------------------------------|----------------------------|---------|---------|
| RD | 1062±16     | 1244±19   | 4.2±0.3                       | 8.2±0.2                    | 0.097±0.002 | 1775±26 |
| TD | 1101±10     | 1254±13   | 4.5±0.9                       | 8.4±1.3                    | 0.089±0.005 | 1736±34 |

3.2. Drop weight impact response of the DP 1180 steel

The force-displacement curves recorded during drop weight impact testing with different impact energy are presented in Fig. 6a. It is seen that the maximum force increases with increasing impact energy and all curves from tests with impact energy of $\leq$ 90 J have similar shape. The force linearly decreases to zero after reaching the peak value. No cracks were observed in samples tested with impact energy of $\leq$ 90 J. A clear oscillation phenomenon is observed on the curve for the specimen impacted with 100 J energy, which can be related to the formation and growth of crack. Additional tests were carried out with the impact energy of 95 J. Therefore, it can be concluded that the drop weight impact resistance of this 1 mm thick DP steel is 90 J.

The true plastic strain values at the top of dome are plotted vs. impact energy in Fig. 6b. A clear linear relationship is seen with R-square of 0.99 obtained after fitting. The maximum true strain reaches up to 81.1% (at 90 J impact energy), revealing high formability of the material in high strain rate deformation under quasi-biaxial stress mode.

An adiabatic thermodynamic environment can be presumed during drop weight impact tests because of the very short deformation time ($<2$ ms). Fig. 6c illustrates the outcomes of \textit{in-situ} temperature measurements drop weight impact testing with the impact energy of 90 J. From the curve, it is seen that the temperature at the top of the dome surges to the peak value of $\sim$180 °C within $\sim$50 ms followed by its steady decrease down to room temperature within $\sim$5 s. However, it should be noted that the deformation time in the experiment does not exceed 1.5 ms, as shown on the load-impact time (Fig. 6e) curve from the given experiments. This time lag can be ascribed to (1) heat transfer from the DP steel to the welded thermocouples, and (2) a thermal inertia of thermocouple itself [59]. The peak temperature tends to increase with
increasing impact energy (Fig. 6d) due to higher amount of plastic deformation induced into sample (Fig. 6b).

![Typical force-displacement curves recorded during drop weight impact testing of the DP 1180 steel with varying impact energy.](image1)

**Fig. 6.** (a) Typical force-displacement curves recorded during drop weight impact testing of the DP 1180 steel with varying impact energy. (b) The relationship between true strain at the top of dome and impact energy. (c) The temperature variation with time recorded by thermal couples during impact testing (90 J). (d) The peak temperature measured during drop weight impact tests vs. impact energy. (e) Force-time cures for drop weight impact testing.

### 3.3. Microstructure evolution during drop weight impact testing

A thoroughly microstructural analysis of the material before and after testing was performed. Fig. 7 illustrates typical KAM maps of the material before and after impact test with 90 J energy. In the as-received material (Fig. 7a), blue regions (indicating near-zero
misorientation angle) correspond to the non-deformed ferrite, while the areas in green color (indicating misorientation of 2–3°) correspond to martensite. The amount of non-indexed pixels is very low (9.5%). Impact testing with 90 J significantly increases the local misorientations all over the microstructure (Fig. 7b) and the fraction of non-indexed pixels (to 48.7%), indicating significant accumulation of lattice defects. The statistic KAM distribution is presented in Fig. 7c, evidently showing the shift of the curve from lower to higher KAM angle.

Fig. 7. KAM maps of (a) as-received material and (b) sample impacted with 90 J energy. The white spots in (a) and (b) are non-indexed points. (c) Statistic distribution of KAM maps.

Fig. 8a and 8b show the inverse pole figure (IPF) maps of the specimens before and after drop weight impact, respectively. It is seen that the microstructure of the as-received DP steel consists of equiaxial grains with diameter of up to ~6 μm (Fig. 8a). Biaxial stretching combined with plastic bending during impact testing results in flaser-like grains with aspect ratio of 0.46. Orientation gradients in the interior of individual elongated grains can be noticed, as shown in the magnified IPF map in Fig. 8c. The detailed misorientation along the black lines in Fig. 8c are plotted in Fig. 8d, where low angle grain boundaries can be seen directly, demonstrating the formation of substructures in the grain interior.
Fig. 8. Inverse polar figure (IPF) map of (a) the as-received material, (b) sample tested with 90 J energy and (c) magnified IPF map of the tested sample. Non-indexed areas are marked with white. (d) Misorientation profiles along the black lines drawn in (c).

The microstructure evolution during impact loading under biaxial stress was investigated in more detail by means of TEM. Typical TEM images of the as-received material are shown in Fig. 9. The martensite laths having a width of 100–300 nm, and ferrite, which has blocky shape, are clearly seen (Fig. 9a). Microalloying by Nb also results in formation of nanoscale spherical NbC precipitates having a size of 10–20 nm, though their volume fraction is very low (Fig. 9b). Their presence was confirmed by analysis of CBED patterns, as shown in inset in Fig. 9b, where a spot corresponding to the NbC precipitate is marked by red circle. Local EDS analysis also detected Nb segregations on those nanoprecipitates (Fig. 9c). The sample after impact testing shows markedly different microstructure. It is characterized by presence of dislocation tangles and irregular dislocation cell structure in the interior of individual grains (Fig. 9d). Size of cells is in the range of 0.3–1.0 μm, which is in a good accordance with the outcomes of the EBSD analysis (Fig. 8c and 8d). As reported previously, cell substructure can form at only 2% of true plastic strain in DP steel under uniaxial stress mode [60]. And the strain
distribution inside the martensite-ferrite microstructure is normally inhomogeneous because of the higher hardness of martensite compared to ferrite. As in the present study, the accumulated plastic strain reaches up to 81.1% in the 90 J impacted specimen (Fig. 6b), well pronounced cell substructure is formed at such high plastic strain. Its formation can be related entirely to high strain rate deformation upon impact, while the effect of adiabatic heating on the microstructure can be ruled out due to low maximum homologue temperature (which reaches just 182 °C as seen in Fig. 6d) and very short time at the peak temperature (which does not exceed 50 ms, Fig. 6c and 6e). The increased KAM angle in the impacted specimen suggested the much more accumulated local deformation, mainly derived from generation of dislocation multiplication as dislocation glide is the major deformation mechanism.
**Fig. 9.** (a) TEM image of the as-received material (general view). (b) TEM image of the NbC (marked with red arrows) in the as-received material. The inserted image is a convergent beam electron diffraction pattern, in which the red circle highlighted spot corresponds to the NbC, the blue rectangle and green parallelogram match the diffraction patterns of bcc iron from zone axis of [013] and [012]. (c) The corresponding EDS map of area (b) showing the segregations of niobium. (d) TEM image of the sample after 90 J impact test.

3.4. Fracture surface analysis

To understand failure of the DP steel under impact loading, fracture surface of cracks formed after testing with >90 J energy was carefully examined. Fig. 10a shows typical SEM images of 95 J impacted specimen. It is seen that the cracks show ductile fracture surface with well pronounced dimples (Fig. 10a). Dimples with size varying in the range of 1–6 μm are homogeneously distributed over the fracture surface, suggesting the failure process of void nucleation, growth, and coalescence. It is not possible to distinguish the origins (martensite or ferrite) of the dimples because they are homogenous from an overall view despite the local difference. This means there is excellent deformation compatibility between hard martensite and soft ferrite, and both of them served as nucleation cites under quasi-biaxial stress mode. The hardness difference between martensite and ferrite should be compensated by the abundant dislocations multiplied in the deformation process (Fig. 9d), as the increased dislocation density strengthened the ferrite phase. Some of the very coarse dimples were formed at manganese sulfide (MnS) inclusions were also found to be as the huge dimple initiation sites, leading flat cleavage surfaces, as shown in Fig. 10b. As this is a commercial steel and MnS inclusions are quite difficult and costly to be completely removed when the size is below 20 μm [61], the appearance of slight MnS inclusions with a size of a few micrometers is not surprising. Their influence on deformation process and energy absorption capability of the studied DP steel will not be further discussed because of their negligible fraction.

To estimate the energy consumed during fracture process, the dimple depth was quantitatively analyzed. Fig. 10c presents the reconstructed 3D DEM of the fracture surface in Fig. 10a. In Fig. 10d, typical elevation (depth) profiles on fracture surface are plotted. Their corresponding positions are marked with red lines in Fig. 10a. From these profiles, it is clear
that the largest and deepest dimples have the depth of about 2 µm. The fracture process begins with their formation and growth. The shallowest dimples having a depth of just dozens of nanometers are formed during coalescence of the coarse dimples.

The depth of dimples was statistically counted from the reconstructed DEMs. For each dimple considered, its deepest elevation value was measured even though the depth inside every dimple varies from one site to another. Dimples with diameter > 4 µm should absorb significantly more energy than finer ones during cracking process. Over 40 dimples with size > 4 µm were analyzed. The averaged depth obtained for this DP steel is 1.33 µm. According to the Stüwe model [62,63], the specific energy necessary to form a unit micro-fracture surface, $R_{surf}$, can be described using the following equation after approximation:

$$ R_{surf} = 2S\bar{\sigma}h_o $$

(2)

Where $S=0.25$ was adopted as the fracture surface is mainly consisted of dimples [62]. $h_o$ is the average dimple depth and $\bar{\sigma}$ is the mean flow stress of material, described as [64]:

$$ \bar{\sigma} = \sigma_{UTS} \frac{e^n}{(1+n)n^n} $$

(3)

where $\sigma_{UTS}$ and $n$ are the ultimate tensile strength and the strain hardening exponent of the material.

The obtained $R_{surf}$ values for the DP steel under high-speed biaxial stress mode is about 1.14 kJ/m$^2$. The estimated energy consumed for formation of the fracture surface in the 95 J impacted sample is about $2.33\times10^{-3}$ J, and for 100 J impacted specimen, the value is about $2.23\times10^{-2}$ J. This indicates that the energy spent for dimples formation is negligible compared to the energy spent for plastic deformation.
Fig. 10. (a) and (b) Typical SEM images of fracture surface for 95 J impacted specimen. MnS inclusions are marked with red arrows in (b). (c) Corresponding 3D view of (a) after DEM reconstruction. (d) Depth profiles along the red lines drawn in (a).

4. Discussion

The local stress state on the specimen is changing continuously during drop weight impact tests, resulting in multi-axial stress state (tension, shearing, bending, etc.) in most area on the hemisphere sample. For the sake of simplification, the top of the dome-shape sample was assumed to have quasi-biaxial stress state. In addition, the energy spent for possible slight friction was also neglected.

4.1. Deformation mechanisms during drop weight impact tests

Analysis of experimental results shows a dramatic difference in ductility of the material tested in uniaxial tensile mode (8% of true strain, Fig. 5) and quasi-biaxial mode (81% of true strain, Fig. 6b). This observation can be rationalized based on two effects. First, the DP steel was softened by the abundant heat converted from the mechanical work. As the drop weight impact test was completed in only a few milliseconds and the heat generated by deformation had no time to dissipate into environment, the whole testing process can be considered as adiabatic. The adiabatic heating at the top dome-shaped sample reaches 182 °C, which is sufficient to weaken the tensile strength of the steel. According to the previous study, the tensile strength of the DP600 steel at 200 °C decreases by 10–15% compared to the room temperature [65]. Similar results can be found for other steel grades in [66,67]. Therefore, the softened DP steel showed better quasi-biaxial stretching formability. Second, dislocation multiplication reinforced the ductility of the DP steel. After deformation under equi-biaxial stress state, the
equiaxial grains were ‘compressed’ into ‘flattened’ ones (Fig. 8a, 8b and 8c), demonstrating the improved plasticity. As dislocation glide is the only deformation mechanism in the DP steel derived from its martensite and ferrite microconstituents, the ‘flattened’ grains should be the result of dislocation movement and multiplication of different slip systems. On the other hand, the softened material and increased temperature, both are caused by adiabatic heating, are two promoters for dislocation nucleation and motion [68].

4.2. Energy absorption capability

The work done by the drop weight mass is transformed into two parts: (1) heat, which caused temperature changing of the material and then dissipated into the environment; (2) strain energy, which was mostly stored in the material in the form of lattice defects. For the first part, it was detected directly by the thermocouples, as shown in Fig. 6c and 6d, obviously manifesting the fact that there is a lot of energy has converted into heat. The exact ratio of mechanical work transformed into heat (designated as $\beta$) has been reported previously in detail in [69–71], where Kolsky pressure tensile bar and infrared camera were used to satisfy an adiabatic condition and to catch the temperature rising, respectively. The ratio value measured experimentally is usually in the range of 0.8–0.9, which means most of the mechanical work during plastic deformation (i.e. 72–81 J) is spent for material heating. Consequently, the strain energy stored during plastic deformation, i.e. the second part, consumes a small portion of the total work (i.e. 9–18 J). However, the strain energy, determined by the plastic strain which limited by the ductility of the material, is mutually influenced with the total plastic work. Because improved plasticity increases total strain as well as total mechanical work. And the total work finished before failure during drop weight impact tests equals to the energy absorbed by the tested material. Of course, the work is also influenced by the strength of the material. For the studied DP steel, the tensile strength is above 1200 MPa, which is excellent. Nevertheless, the ductility is not so attractive. However, this deficiency was enhanced to a large extent under dynamic biaxial loading, resulting in improved formability, as well as total plastic work, namely overall absorbed energy.
5. Conclusions

The microstructure evolution and energy absorption capability of a commercial high strength martensite-ferrite dual phase (DP) steel during drop weight impact testing were thoroughly investigated. The main conclusions are:

1. Intensive adiabatic heating effect was detected via in situ temperature measurements resulting in the peak temperature of 225 ℃. The drop weight impact resistance of the DP steel under present testing conditions is 90 J. The ductility of the material was improved by the softening effect due to the large amount of heat converted from plastic work.

2. Dislocation glide and formation of dislocation cell substructure is the main deformation mechanism active in the DP steel during high strain rate quasi-biaxial stretching.

3. Ductile failure mode features the fracture surface of the DP steel under high strain rate biaxial stress loading and the energy consumed during cracking presents very low fraction of the total plastic work.

Acknowledgements

Peikang Xia acknowledges gratefully the financial supporting from China Scholarship Council (No. 201606890031, Beijing, China). Miguel Castillo-Rodriguez is gratefully acknowledged for help with TEM characterization. The authors also would like to express their gratitude to José Luis Jiménez for the help in drop weight impact testing.

References

[1] Y. Li, Z. Lin, A. Jiang, G. Chen, Use of high strength steel sheet for lightweight and crashworthy car body, Mater. Des. 24 (2003) 177–182. https://doi.org/10.1016/S0261-3069(03)00021-9.

[2] R. Kuzziak, R. Krawall, S. Waengler, Advanced high strength steels for automotive industry, Arch. Civ. Mech. Eng. 8 (2008) 103–117. https://doi.org/10.1016/S1644-9665(12)60197-6.

[3] Q. Liu, Y. Lin, Z. Zong, G. Sun, Q. Li, Lightweight design of carbon twill weave fabric composite body structure for electric vehicle, Compos. Struct. 97 (2013) 231–238. https://doi.org/10.1016/j.compstruct.2012.09.052.

[4] J. Hirsch, Aluminium in Innovative Light-Weight Car Design, Mater. Trans. 52 (2011) 818–824. https://doi.org/10.2320/matertrans.L-MZ201132.

[5] P.J. Jacques, Transformation-induced plasticity for high strength formable steels, Curr. Opin. Solid State Mater. Sci. 8 (2004) 259–265. https://doi.org/10.1016/j.cossm.2004.09.006.

[6] O. Bouaziz, S. Allain, C.P. Scott, P. Cugy, D. Barbier, High manganese austenitic twinning induced plasticity steels: A review of the microstructure properties relationships, Curr. Opin. Solid State
A. Karelova, C. Krempaszky, E. Werner, P. Tsipouridis, T. Hebesberger, A. Pichler, Hole Expansion of Dual-phase and Complex-phase AHS Steels - Effect of Edge Conditions, Steel Res. Int. 80 (2009) 71–77. https://doi.org/10.1007/BF00356631.

M. Sarwar, R. Priestner, Influence of ferrite-martensite microstructural morphology on tensile properties of dual-phase steel, J. Mater. Sci. 31 (1996) 2091–2095. https://doi.org/10.1007/BF00356631.

J. Speer, D.K. Matlock, B.C. De Cooman, J.G. Schroth, Carbon partitioning into austenite after martensite transformation, Acta Mater. 51 (2003) 2611–2622. https://doi.org/10.1016/S1359-6454(03)00059-4.

R.G. Davies, Influence of martensite composition and content on the properties of dual phase steels, Metall. Trans. A. 9 (1978) 671–679. https://doi.org/10.1007/BF02659924.

C. Peng-Heng, A.G. Preban, The effect of ferrite grain size and martensite volume fraction on the tensile properties of dual phase steel, Acta Metall. 33 (1985) 897–903. https://doi.org/10.1016/0001-6160(85)90114-2.

P. Movahed, S. Kolahgar, S.P.H. Marashi, M. Pouranvari, N. Parvin, The effect of intercritical heat treatment temperature on the tensile properties and work hardening behavior of ferrite-martensite dual phase steel sheets, Mater. Sci. Eng. A. 518 (2009) 1–6. https://doi.org/10.1016/j.msea.2009.05.046.

A. Bag, K.K. Ray, E.S. Dwarakadasa, Influence of martensite content and morphology on tensile and impact properties of high-martensite dual-phase steels, Metall. Mater. Trans. A. 30 (1999) 1193–1202. https://doi.org/10.1007/s11661-999-0269-4.

A. Kumar, S.B. Singh, K.K. Ray, Influence of bainite/martensite content on the tensile properties of low carbon dual-phase steels, Mater. Sci. Eng. A. 474 (2008) 270–282. https://doi.org/10.1016/j.msea.2007.05.007.

A.K. Sachdev, Effect of retained austenite on the yielding and deformation behavior of a dual phase steel, Acta Metall. 31 (1983) 2037–2042. https://doi.org/10.1016/0001-6160(83)90021-4.

R. G.Davies, The deformation behavior of a vanadium-strengthened dual phase steel, Metall. Trans. A. 9 (1978) 41–52. https://doi.org/10.1007/BF02647169.

N.J. Kim, G. Thomas, Effects of morphology on the mechanical behavior of a dual phase Fe/2Si/0.1C steel, Metall. Trans. A. 12 (1981) 483–489. https://doi.org/10.1007/BF02648546.

H.-C. Chen, G.-H. Cheng, Effect of martensite strength on the tensile strength of dual phase steels, J. Mater. Sci. 24 (1989) 1991–1994. https://doi.org/10.1007/BF02385411.

Z. Jiang, Z. Guan, J. Lian, Effects of microstructural variables on the deformation behaviour of dual-phase steel, Mater. Sci. Eng. A. 190 (1995) 55–64. https://doi.org/10.1016/0921-5093(94)09594-M.

M. Luo, T. Wierzbicki, Numerical failure analysis of a stretch-bending test on dual-phase steel sheets using a phenomenological fracture model, Int. J. Solids Struct. 47 (2010) 3084–3102. https://doi.org/10.1016/j.ijsolstr.2010.07.010.

H. Ghassemi-Armaki, R. Maass, S.P. Bhat, S. Sriram, J.R. Greer, K.S. Kumar, Deformation response of ferrite and martensite in a dual-phase steel, Acta Mater. 62 (2014) 197–211. https://doi.org/10.1016/j.actamat.2013.10.001.

H. Toda, A. Takijiri, M. Azuma, S. Yabu, K. Hayashi, D. Seo, M. Kobayashi, K. Hirayama, A. Takeuchi, K. Uesugi, Damage micromechanisms in dual-phase steel investigated with combined phase-
and absorption-contrast tomography, Acta Mater. 126 (2017) 401–412. https://doi.org/10.1016/j.actamat.2017.01.010.

[23] C.P. Scott, B. Shalchi Amirkhiz, I. Pushkareva, F. Fazeli, S.Y.P. Allain, H. Azizi, New insights into martensite strength and the damage behaviour of dual phase steels, Acta Mater. 159 (2018) 112–122. https://doi.org/10.1016/j.actamat.2018.08.010.

[24] L. Schemmann, S. Zaefferer, D. Raabe, F. Friedel, D. Mattissen, Alloying effects on microstructure formation of dual phase steels, Acta Mater. 95 (2015) 386–398. https://doi.org/10.1016/j.actamat.2015.05.005.

[25] A.-P. Pierman, O. Bouaziz, T. Pardoen, P.J. Jacques, L. Brassart, The influence of microstructure and composition on the plastic behaviour of dual-phase steels, Acta Mater. 73 (2014) 298–311. https://doi.org/10.1016/j.actamat.2014.04.015.

[26] J. Kadkhodapour, A. Butz, S. Ziaei Rad, Mechanisms of void formation during tensile testing in a commercial, dual-phase steel, Acta Mater. 59 (2011) 2575–2588. https://doi.org/10.1016/j.actamat.2010.12.039.

[27] J. Liao, J.A. Sousa, A.B. Lopes, X. Xue, F. Barlat, A.B. Pereira, Mechanical, microstructural behaviour and modelling of dual phase steels under complex deformation paths, Int. J. Plast. 93 (2017) 269–290. https://doi.org/10.1016/j.ijplas.2016.03.010.

[28] Y. Prawoto, M. Fanone, S. Shahedi, M.S. Ismail, W.B. Wan Nik, Computational approach using Johnson–Cook model on dual phase steel, Comput. Mater. Sci. 54 (2012) 48–55. https://doi.org/10.1016/j.commatsci.2011.10.021.

[29] A.P. Modi, Effects of microstructure and experimental parameters on high stress abrasive wear behaviour of a 0.19wt% C dual phase steel, Tribol. Int. 40 (2007) 490–497. https://doi.org/10.1016/j.triboint.2006.04.013.

[30] A. Bag, K.K. Ray, E.S. Dwarakadasa, Influence of martensite content and morphology on the toughness and fatigue behavior of high-martensite dual-phase steels, Metall. Mater. Trans. A. 32 (2001) 2207–2217. https://doi.org/10.1007/s11661-001-0196-5.

[31] Y.J. Chao, J.D. Ward, R.G. Sands, Charpy impact energy, fracture toughness and ductile–brittle transition temperature of dual-phase 590 Steel, Mater. Des. 28 (2007) 551–557. https://doi.org/10.1016/j.matdes.2005.08.009.

[32] A. Hüseyin, K.Z. Havva, K. Ceylan, Effect of Intercritical Annealing Parameters on Dual Phase Behavior of Commercial Low-Alloyed Steels, J. Iron Steel Res. Int. 17 (2010) 73–78. https://doi.org/10.1016/S1006-706X(10)60089-1.

[33] N. Saeidi, A. Ekrami, Comparison of mechanical properties of martensite/ferrite and bainite/ferrite dual phase 4340 steels, Mater. Sci. Eng. A. 523 (2009) 125–129. https://doi.org/10.1016/j.msea.2009.06.057.

[34] M.A. Maleque, Y.M. Poon, H.H. Masjuki, The effect of intercritical heat treatment on the mechanical properties of AISI 3115 steel, J. Mater. Process. Technol. 153–154 (2004) 482–487. https://doi.org/10.1016/j.jmatprotec.2004.04.391.

[35] L. Shi, Z. Yan, Y. Liu, C. Zhang, Z. Qiao, B. Ning, H. Li, Improved toughness and ductility in ferrite/acicular ferrite dual-phase steel through intercritical heat treatment, Mater. Sci. Eng. A. 590 (2014) 7–15. https://doi.org/10.1016/j.msea.2013.10.006.

[36] M. Yang, Y.J. Chao, X. Li, D. Immel, J. Tan, Splitting in dual-phase 590 high strength steel plates: Part II. Quantitative analysis and its effect on Charpy impact energy, Mater. Sci. Eng. A. 497 (2008) 462–470. https://doi.org/10.1016/j.msea.2008.07.066.
[37] M. Yang, Y.J. Chao, X. Li, J. Tan, Splitting in Dual-Phase 590 high strength steel plates: Part I. Mechanisms, Mater. Sci. Eng. A. 497 (2008) 451–461. https://doi.org/10.1016/j.msea.2008.07.067.

[38] ISO 148-1:2016, Metallic materials—Charpy pendulum impact test—Part 1: Test method, International Organization for Standardization, 2016. https://www.iso.org/obp/ui/#iso:std:iso:148:1:ed:3:v1:en.

[39] ASTM A370-17a, Standard test methods and definitions for mechanical testing of steel products, ASTM International, 2017. https://doi.org/10.1520/A0370-17A.

[40] J. Lis, A.K. Lis, C. Kolan, Processing and properties of C–Mn steel with dual-phase microstructure, J. Mater. Process. Technol. 162–163 (2005) 350–354. https://doi.org/10.1016/j.jmatprotec.2005.02.105.

[41] M. Calcagnotto, D. Ponge, D. Raabe, Effect of grain refinement to 1μm on strength and toughness of dual-phase steels, Mater. Sci. Eng. A. 527 (2010) 7832–7840. https://doi.org/10.1016/j.msea.2010.08.062.

[42] Z. Sami, S. Tahar, H. Mohamed, Microstructure and Charpy impact properties of ferrite–martensite dual phase API X70 linepipe steel, Mater. Sci. Eng. A. 598 (2014) 338–342. https://doi.org/10.1016/j.msea.2014.01.052.

[43] K. Sato, T. Inazumi, A. Yoshitake, S.-D. Liu, Effect of material properties of advanced high strength steels on bending crash performance of hat-shaped structure, Int. J. Impact Eng. 54 (2013) 1–10. https://doi.org/10.1016/j.ijimpacteng.2012.10.012.

[44] V. Tarigopula, M. Langseth, O.S. Hopperstad, A.H. Clausen, Axial crushing of thin-walled high-strength steel sections, Int. J. Impact Eng. 32 (2006) 847–882. https://doi.org/10.1016/j.ijimpacteng.2005.07.010.

[45] N. Abedrabbo, R. Mayer, A. Thompson, C. Salisbury, M. Worswick, I. van Riemsdijk, Crash response of advanced high-strength steel tubes: Experiment and model, Int. J. Impact Eng. 36 (2009) 1044–1057. https://doi.org/10.1016/j.ijimpacteng.2009.02.006.

[46] J.R. Fekete, A.M. Stibich, M.F. Shi, A Comparison of the Response of HSLA and Dual Phase Sheet Steel in Dynamic Crush, SAE International, Warrendale, PA, 2001. https://doi.org/10.4271/2001-01-3101.

[47] N. Peixinho, N. Jones, A. Pinho, Experimental and numerical study in axial crushing of thin walled sections made of high-strength steels, J. Phys. IV Proc. 110 (2003) 717–722. https://doi.org/10.1051/jp4:2002078.

[48] J.K. Holmen, O.S. Hopperstad, T. Børvik, Low-velocity impact on multi-layered dual-phase steel plates, Int. J. Impact Eng. 78 (2015) 161–177. https://doi.org/10.1016/j.ijimpacteng.2014.12.005.

[49] G. Gruben, M. Langseth, E. Fagerholt, O.S. Hopperstad, Low-velocity impact on high-strength steel sheets: An experimental and numerical study, Int. J. Impact Eng. 88 (2016) 153–171. https://doi.org/10.1016/j.ijimpacteng.2015.10.001.

[50] K. Sato, T. Inazumi, A. Yoshitake, S.-D. Liu, Effect of material properties of advanced high strength steels on bending crash performance of hat-shaped structure, Int. J. Impact Eng. 54 (2013) 1–10. https://doi.org/10.1016/j.ijimpacteng.2012.10.012.

[51] ASTM E8 / E8M-13a, Standard test methods for tension testing of metallic materials, ASTM International, 2013. https://doi.org/10.1520/E0008_E0008M-13A.

[52] T.H. Courtney, Mechanical Behavior of Materials: Second Edition, Waveland Press, 2005.

[53] P. Xia, I. Sabirov, J. Molina-Aldareguia, P. Verleyens, R. Petrov, Mechanical behavior and microstructure evolution of a quenched and partitioned steel during drop weight impact and
punch testing, Mater. Sci. Eng. A. 737 (2018) 18–26. https://doi.org/10.1016/j.msea.2018.09.015.

[54] ASTM E562 – 9, Test method for determining volume fraction by systematic manual point count, ASTM International, 2019. https://doi.org/10.1520/E0562-19.

[55] S.I. Wright, M.M. Nowell, D.P. Field, A Review of Strain Analysis Using Electron Backscatter Diffraction, Microsc. Microanal. 17 (2011) 316–329. https://doi.org/10.1017/S1431927611000055.

[56] D. Samak, A. Fischer, D. Rittel, 3D Reconstruction and Visualization of Microstructure Surfaces from 2D Images, CIRP Ann. 56 (2007) 149–152. https://doi.org/10.1016/j.cirp.2007.05.036.

[57] F. Marinello, P. Bariani, E. Savio, A. Horsewell, L.D. Chiffre, Critical factors in SEM 3D stereo microscopy, Meas. Sci. Technol. 19 (2008) 065705. https://doi.org/10.1088/0957-0233/19/6/065705.

[58] J. Stampfl, S. Scherer, M. Gruber, O. Kolednik, Reconstruction of surface topographies by scanning electron microscopy for application in fracture research, Appl. Phys. A. 63 (1996) 341–346. https://doi.org/10.1007/BF01567324.

[59] J.A. Dantzig, Improved transient response of thermocouple sensors, Rev. Sci. Instrum. 56 (1985) 723–725. https://doi.org/10.1063/1.1138214.

[60] D.A. Korzekwa, D.K. Matlock, G. Krauss, Dislocation substructure as a function of strain in a dual-phase steel, Metall. Mater. Trans. A. 15 (1984) 1221. https://doi.org/10.1007/BF02644716.

[61] X.F. Zhang, W.J. Lu, R.S. Qin, Removal of MnS inclusions in molten steel using electropulsing, Scr. Mater. 69 (2013) 453–456. https://doi.org/10.1016/j.scriptamat.2013.05.033.

[62] H.P. Stüwe, The work necessary to form a ductile fracture surface, Eng. Fract. Mech. 13 (1980) 231–236. https://doi.org/10.1016/0013-7944(80)90056-9.

[63] J. Stampfl, O. Kolednik, The separation of the fracture energy in metallic materials, Int. J. Fract. 101 (2000) 321–345. https://doi.org/10.1023/A:1007500325074.

[64] S. Nemat-Nasser, ed., Three-dimensional Constitutive Relations and Ductile Fracture, Elsevier Science Ltd, North-Holland, Amsterdam, 1981.

[65] F. Ozturk, S. Toros, S. Kilic, Tensile and Spring-Back Behavior of DP600 Advanced High Strength Steel at Warm Temperatures, J. Iron Steel Res. Int. 16 (2009) 41–46. https://doi.org/10.1016/S1006-706X(10)60025-8.

[66] Chen Ju, Young Ben, Uy Brian, Behavior of High Strength Structural Steel at Elevated Temperatures, J. Struct. Eng. 132 (2006) 1948–1954. https://doi.org/10.1061/(ASCE)0733-9445(2006)132:12(1948).

[67] J. Chen, B. Young, Stress–strain curves for stainless steel at elevated temperatures, Eng. Struct. 28 (2006) 229–239. https://doi.org/10.1016/j.engstruct.2005.07.005.

[68] D. Dong, Y. Liu, Y. Yang, J. Li, M. Ma, T. Jiang, Microstructure and dynamic tensile behavior of DP600 dual phase steel joint by laser welding, Mater. Sci. Eng. A. 594 (2014) 17–25. https://doi.org/10.1016/j.msea.2013.11.047.

[69] J. Hodowany, G. Ravichandran, A.J. Rosakis, P. Rosakis, Partition of plastic work into heat and stored energy in metals, Exp. Mech. 40 (2000) 113–123. https://doi.org/10.1007/BF02325036.

[70] A. Rusinek, J.R. Klepaczko, Experiments on heat generated during plastic deformation and stored energy for TRIP steels, Mater. Des. 30 (2009) 35–48. https://doi.org/10.1016/j.matdes.2008.04.048.

[71] G. Ravichandran, A.J. Rosakis, J. Hodowany, P. Rosakis, On the Conversion of Plastic Work into Heat During High-Strain-Rate Deformation, AIP Conf. Proc. 620 (2002) 557–562. https://doi.org/10.1063/1.1483600.