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Inelastic deformation micromechanism and modified fragmentation model for silicon carbide under dynamic compression

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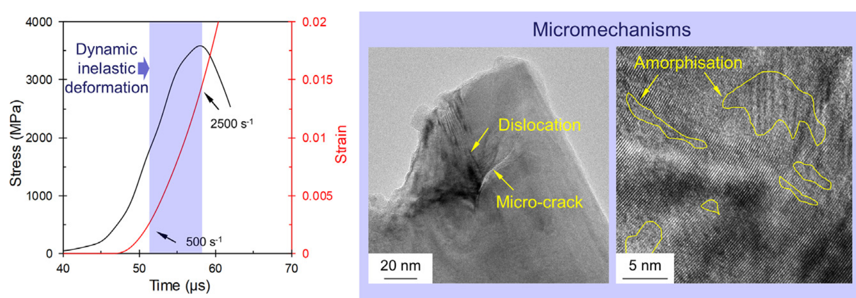
HIGHLIGHTS

- Inelastic deformation followed by transgranular fracture occurs at dynamic strain rates.
- Dislocation motion and localised amorphisation determine the bulk inelastic behaviour.
- Dynamic fragmentation model modified with damage accurately predicts the fragment size.

GRAPHICAL ABSTRACT



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ABSTRACT

The underlying micromechanism remains to be clarified for the bulk inelastic behaviour of specific ceramics under impact loads. In this study, the silicon carbide materials were subjected to the split-Hopkinson pressure bar compression in which the strain rate was not constant but increased to the dynamic level at high stresses. The inelastic deformation occurs in the high strain rate stage in compression, followed by the final transgranular fracture. The post-test fragments were examined in both the SEM and high resolution TEM. It was found that macroscopic inelastic behaviour is dominated by the dislocation motion and the localised amorphisation that arise at high strain rates. The damage and thus the degraded modulus in the dynamic inelastic deformation were incorporated to modify a dynamic fragmentation model to evaluate the fragment size as a function of strain rates. The modified model more accurately predicts the size of fragments produced at high strain rates.

1. Introduction

Silicon carbide ceramics possess the excellent properties for high speed applications, such as high strength-to-weight ratio, high hardness, low compressibility and superior ballistic performance [1–5]. They have been utilised in the layered armour systems to dissipate impact energy [1, 2]. Evaluation of the energy dissipation capacity requires a better understanding of the underlying micromechanism of deformation and failure in the ceramic subjected to dynamic impact loads.

Plentiful experimental and numerical studies have been reported on the dynamic deformation and failure mechanisms in various length scales in different ceramics [6–15]. The split-Hopkinson pressure bar (SHPB) experiments and subsequent analysis of measured strain waves have been investigated to accurately determine bulk dynamic stress–strain responses of ceramic materials [6, 16]. A large portion of research focused on the phenomena in the microscopic scale such as micro-crack initiation and propagation that account for macro-cracking and fragmentation in the final fracture [3, 7, 10, 17–20]. The

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micro-crack propagation is significantly affected by the heterogeneities such as inclusions. The localised tensile stress concentrated near grain boundaries and inclusions drives micro-cracks to nucleate and grow to a subcritical state; the macroscopic failure finally occurs as a result of the coalescence of micro-cracks [10, 21]. The loading rate plays a crucial role in the failure process for a specific ceramic material [1, 9, 10, 22–24]. The fracture mode in ceramics can be intergranular, transgranular or a combination of both, depending on intrinsic heterogeneities and extrinsic loading rates [8, 10, 20, 22, 25–27]. The size of fragments collected in dynamic impact tests often follows a statistic distribution, and is strongly affected by the strain rate; the fragment size reduces with the increasing strain rate [7, 10]. Various analytical models have been proposed for the dynamic fragmentation in ceramics [7, 9, 23, 28]. However, some of the constitutive parameters (e.g., Young's modulus) have usually been assumed constant in these analytical models, thus leading to the inaccuracy of predictions. In fact, the material properties can vary as a function of loading conditions during the dynamic impact process, e.g., the degradation of the bulk Young's modulus at high rates in ceramics. A more reliable dynamic fragmentation model needs to incorporate the evolution of these properties.

Bulk inelastic behaviour and microscopic localised softening have been observed in various ceramic materials under shock wave loading [8–10, 16, 17, 25, 29]. The dislocation motion at room temperature gives rise to significant plastic deformation prior to the final fracture [3, 27, 30]. Dislocation activities near grain boundaries dominate the dynamic deformation mechanism in polycrystalline ceramics, often resulting in the intergranular fracture mode. At the very high loading rate, deformation twinning can become a prevalent mechanism, and the twinning interface serves as the preferred cleavage plane for the transgranular fracture [17, 25]. Numerical models at the atomic level such as molecular dynamics have been developed to explore the physical mechanism for inelastic behaviour [31]. However, only a very small domain can be simulated in the model, thus limiting the exploration of the inelastic phenomena. Therefore, the debate still continues on the underlying micromechanisms of the dynamic inelastic behaviour, even for ceramic materials of the same chemical compositions but fabricated by the different processes.

The aim of the study reported here is to investigate the physical phenomena in the microscopic scale for the inelastic behaviour of silicon carbide, and then develop a dynamic fragmentation model. The SHPB experiments were performed on the silicon carbide specimens to quantitatively characterise the bulk deformation process and identify the inelastic deformation stage. The fragments collected in the SHPB tests were examined in the scanning electron microscope (SEM) to analyse the size distribution statistically. The thin fracture edge in the fragments was further examined in the high resolution transmission electron microscope (TEM) to explore the change in crystalline structures during dynamic deformation.

2. Experimental procedure

The dynamic uniaxial compression experiments of hot-pressed cylindrical silicon carbide specimens of the diameter 5 mm and the length 5 mm (Chair Man Hi-Tech Co. Ltd., Taiwan) were performed in an in-house split-Hopkinson pressure bar system using YAG300 maraging steel striker, input and output bars (diameter 20 mm). The silicon carbide material was >97% in purity with a density of 3100 kg m^{-3}. The SHPB system as well as the subsequent analysis of measured strain waves was developed for the accurate measurement of dynamic behaviour of ceramics as detailed in the previous work [6]. A pair of wave impedance-matched cylindrical tungsten carbide inserts (diameter 17 mm) was laterally confined by the steel sleeves and sandwiched between the bars and the specimen to prevent the indentation into the steel bars by the harder silicon carbide specimen. To achieve the highest possible strain rate at a level of $>10^3$ s^{-1}, no pulse shaping technique was applied in the SHPB tests. For comparison, the quasi-static uniaxial compression tests of specimens (diameter 5 mm and length 12 mm) were conducted in an INSTRON 5569 (INSTRON, MA, USA) electromechanical universal testing machine. The details on both the SHPB and INSTRON tests of high strength ceramics can be found in the previous work [6, 10].

Prior to the compression experiments, the specimen surfaces were examined in the optical microscope; and only the specimens without surface defects were tested. The specimen ends were lubricated with Castrol LMX grease to minimise the interfacial friction. In the tests, an in-house acrylic transparent enclosure box was used to encompass and protect the specimen setup and to collect the fragments after failure. At least ten specimens were tested for the compression at the quasi-static and dynamic rates.

After the uniaxial compression experiments, the collected specimen fragments were examined in a JEOL JSM-5600LV SEM (JEOL Ltd., Japan) to reveal the fracture features. The small fragments obtained from the SHPB tests were found to have the very thin edges that are electron transparent and thus ideal for inspection in TEM [17, 25]. Some of these small fragments were dispersed in isopropanol alcohol via sonication for 5 min. The dispersion was deposited onto the carbon coated copper grid via drop casting, and followed by the evaporation of the

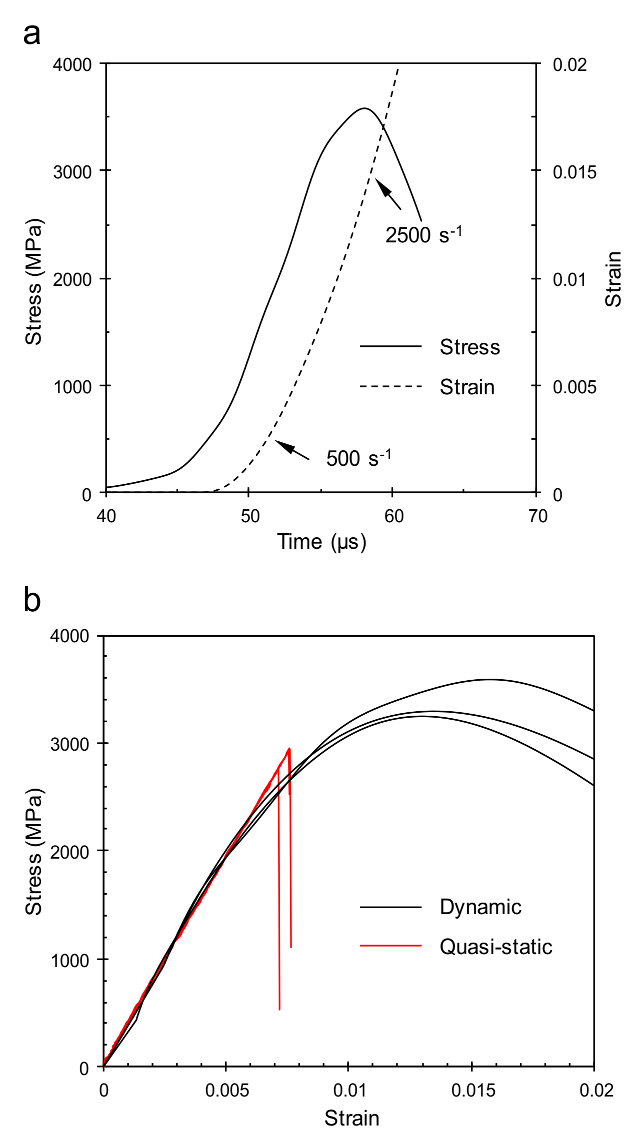

Fig. 1. (a) The stress and strain histories in a silicon carbide specimen as calculated from the strain waves measured in the split-Hopkinson pressure bar test, and (b) the representative stress–strain curves at quasi-static and dynamic loading rates.