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Inelastic behavior and fracture of high modulus polymeric fiber bundles at high strain-rates

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ABSTRACT

The tensile behavior and fracture mechanisms of poly(phenylene terephthalamide) (PPTA) and high modulus polyethylene (HMPE) fiber bundles were studied at high strain-rates using a tension Kolsky bar. For all fiber bundles investigated, a significant amount of strain energy was found to be dissipated by inelastic processes in addition to that due to fracture. The differences in microstructure and properties between the fibers were shown to have a noticeable influence on the inelasticity and fracture behavior in PPTA fiber bundles. No significant strain-rate effect on inelastic behavior and maximum strength was found in HMPE fiber bundles. Scanning electron micrographs of the fracture surfaces of PPTA fiber showed that the failure occurs mainly by fibrillation resulting in pointed breaks, and showed no fundamental difference in fracture mechanism at quasi-static and high strain-rates. However, the fracture mechanism in the HMPE fiber was different at quasi-static and high strain-rates, crazing was dominant at high strain-rates and plate formation under quasi-static conditions. This difference was more substantial in HMPE fibers with lower degree of crystalline order, which suggested that the inelastic behavior is governed by a precise mechanism of load transfer between the crystalline and amorphous phases present in HMPE fibers as a function of loading rate. At high strain-rates, HMPE fibers appear to be able to dissipate more strain energy than PPTA fibers due to this intrinsic change of deformation mechanism. Our results also support the idea that the mechanical behavior of a PPTA fiber bundle is inherently statistical including variations in strength distribution and alignment of the individual filaments.

1. Introduction

The specific tensile strengths of high-performance fibers produced from poly(phenylene terephthalamide) (PPTA) and high modulus polyethylene (HMPE) exceed that of steel, and appear to dissipate large amounts of energy during ballistic impact. These two fibers are marketed commercially, for example PPTA as Kevlar® (DuPont) and HMPE as Spectra® (Honeywell), respectively. Although the structure of these fibers is now well-characterized [1,2], the relationship between microstructure and macroscopic mechanical behavior is not fully understood. In particular, there is little literature on the correlation between morphology and fracture behavior in these polymers at high strain-rates. In this study, we attempt to characterize the strain-rate dependence of the tensile strength and fracture of PPTA and HMPE fiber bundles, and explore the underlying deformation mechanisms at high strain-rates. The strain-rates ∼200–1500 s^{-1}, which are estimated to be present when a projectile at ∼500 Km/h hits, for instance, a planar plain-woven fabric [3,4], are readily attainable in a Kolsky bar apparatus [5].

Both PPTA and HMPE are highly crystalline. The chemical formula for the PPTA monomer is shown in Fig. 1a. Molecules including aromatic structures or amide groups are usually strong; PPTA contains both. The aramid chains form rigid planar sheets with the chain-extended molecules hydrogen bonded together. The sheets are stacked to form a crystalline array but there is only weak van der Waals forces between the sheets, which are arranged in a radial system of axially pleated lamellae as illustrated in Fig. 1b [6,7]. Microscopic, scattering and spectroscopic techniques have been extensively used to study the radial orientation of PPTA fibers [1]. The structure of pedigreed PPTA fibers (including Kevlar® 149, Kevlar® 49 and Kevlar® 29) is known to be radial lateral with crystallinity of the order of 90–95% for Kevlar® 149 and Kevlar® 49 and 80–85% for Kevlar® 29 [6]. Using wide angle X-ray diffraction, it was found that PPTA forms orthorhombic crystals with two chains per unit cell. The apparent crystallite size is 50 nm, 4.4 nm

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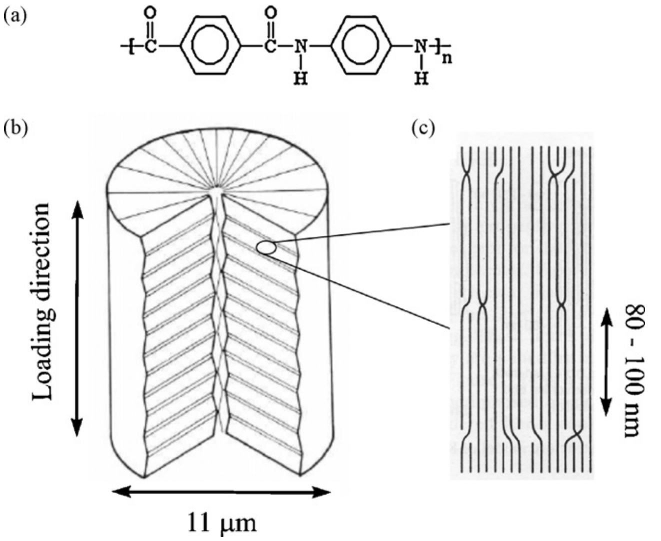

Fig. 1. (a) Chemical formula of PPTA; (b) radial pleated-sheet structure of PPTA filaments and (c) crystalline structure model of PPTA [1].

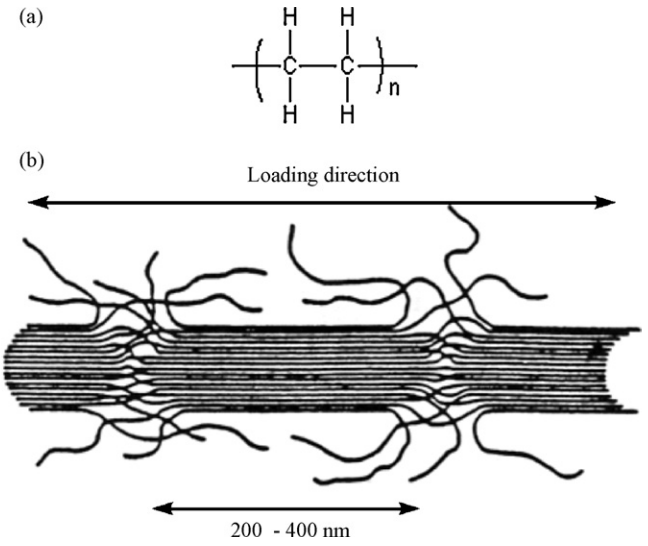

Fig. 2. (a) Chemical formula of HMPE and (b) longitudinal structure of HMPE microfilaments [2].

and 3.6 nm along [0 0 2], [1 1 0] and [2 0 0] directions, respectively. PPTA fibers have high degrees of axial and lateral crystalline order [1]. These fibers are formed by a dry-jet wet spinning process that produces liquid crystalline domains with chains oriented along the flow direction [7]. As a result, PPTA shows anisotropy in mechanical behavior with larger modulus and strength along the longitudinal orientation of the fiber.

Based on experimental data and simulation on single filaments, Cheng et al. [8] have observed that the tensile response of Kevlar® KM2 fibers is linear and elastic until failure. This behavior was related to fibrillation, i.e. decohesion of fibrils along the longitudinal axis of the fiber. The mechanical behavior was also found to be nonlinear along the transverse direction. Both longitudinal and transverse behavior showed no strain-rate dependency. In contrast, Wang and Xia [9] found a strain-rate dependence of mechanical behavior in Kevlar® 49 bundles investigated at strain-rates between $0.0001\,s^{-1}$ and $1350\,s^{-1}$. Using a Weibull failure analysis, these authors have observed that the Young's modulus and strength of the fiber bundles increase slightly with increasing strain-rate. However, Creasy [10] has argued that the effects of slack and variability in test procedure may account for the apparent strain-rate dependence of these parameters.

The chemical formula for the monomer of a HMPE fiber is shown in Fig. 2a. This polymer is processed by gel spinning followed by drawing to produce longitudinally oriented chains [2]. In the gel state, polyethylene is amorphous, and the undrawn fiber is characterized by the formation of crystalline spherulites in an amorphous matrix. During drawing, the spherulites deform into parallel plate-like structures [11]. HMPE fibers usually have a lower degree of X-ray crystallinity (∼75%) than PPTA [2]. The microstructure of HMPE filaments is schematically represented in Fig. 2b. The drawn fibers consist of microfibrils and intrafibrillar matter. The microfibrils are composed of crystalline and amorphous regions oriented along the fiber axis. The aligned crystalline regions are on the order of 60–400 nm in length [2]. Though intermolecular van der Waals forces between the PE chains are weak, the primary covalent bonds along the chain-axis contribute to the strength of the fiber.

Compared to PPTA, little has been published regarding the fracture of HMPE fiber. Within the fiber, the microfibrils are aligned longitudinally and linked together by the intrafibrillar chains [2]. Under tension the microfibrils can stretch through chain unfolding, giving these regions greater strength. Prevorsek et al. [3] used high-speed photography to observe the effects of ballistic impact on Spectra® fibers and composites. Since HMPE fibers melt at a lower temperature than PPTA, investigations of HMPE fibers have focused on heating effects in addition to deformation and failure. Another study by the same group using numerical simulation of projectile impact provided a model of heat effects during ballistic impact on polyethylene fiber composite armor [12]. It was concluded that heating effects were negligible in the armor performance of these composites.

Polyethylene is known to dissipate energy through chain unfolding and crazing in its amorphous domains, and through crystal separation and slippage in its crystalline regions [10,13]. Crazing is the primary yielding morphology seen in amorphous glassy polymers [14]. Crazes are localized zones of micro-cracking which grow perpendicular to the direction of maximum principal stress. Above the glass transition temperature, a polymer is more viscous, and thus allows energy to be absorbed through chain unfolding [15]. Because of the relatively high crystalline content, it is unclear if the same mechanisms operate in HMPE fibers.

This study describes an effort to understand the mechanical behavior and fracture of PPTA and HMPE fiber bundles under uniaxial tension at both quasi-static and high strain-rates. Atomic force microscopy (AFM) was used to characterize the cross-sectional structure of the filaments. Uniaxial tensile tests were conducted at high strain-rates using a tension Kolsky bar. The failure modes and mechanisms were investigated by scanning electron microscopy (SEM) and related to the structure of each of the fiber samples. Section 2 presents the details of the experimental procedure. The results obtained from mechanical testing and fracture surface analyses are presented in Section 3. The relationships between microstructure, mechanisms, and strain-rate dependence of deformation and fracture in PPTA and HMPE are discussed in Section 4.

2. Experimental

2.1. Materials

Three grades of PPTA fiber bundles (DuPont) labeled Fiber-A, -B and -C, were tested [16,17]. The first two types of fiber belong to the same family of PPTA, but were obtained from different processing conditions. Two HMPE fiber bundles (Honeywell) labeled Fiber-A and -B [18,19] were used in this study with the former having lower