

Behavior of high-density polyethylene at high strain rates

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Abstract

The Hopkinson split pressure bar (HSPB) was used for the testing of three polymers at strain rates between 10^2 to 10^3 s^{-1} . Higher strain rates were achieved using the direct Hopkinson test. Experimental data were evaluated in time as well as in the frequency domain. A more detailed analysis in the frequency domain showed that the description of tested polymers can be described in the framework of the linear viscoelasticity. The use of the direct Hopkinson test showed the occurrence of a permanent strain.

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Keywords: dynamic loading, strain rate, split Hopkinson pressure bar, direct Hopkinson test, viscoelasticity

1. Introduction

Polymeric materials with low mechanical impedance represent potential materials in light weight armour systems. Therefore, it is needed to study the behavior of polymers at strain rates corresponding to the impact loading of materials. These data on compressive [1, 2, 12, 14, 20], tensile [3, 13, 21, 22] and shear [8] stress-strain properties at high rates of strain of many polymers have been determined with the split Hopkinson pressure bar (SHPB) technique [9] and using a drop-weight apparatus [4, 15, 17, 18]. The main results are summarized in [16]. The experiments using the SHPB technique were performed mostly using loading stress pulses of long duration to achieve stress equilibrium in the specimen. In the present paper, the problem of specimen response to the stress pulse of short time of its duration is studied.

The paper is focused on the mechanical behavior of three polymers at strain rates which are typical in the course of a common impact loading. Three copolymers of high-density polyethylene are investigated. These copolymers differ in their structure. Experimental results are evaluated both in time and in the frequency domain. The performed analysis shows that there is no difference in the mechanical behavior of the tested polymers. The behavior of the polymers tested by the classical SHPB method can be described as viscoelastic. The use of the direct impact test [7] in connection with the high-speed camera shows a permanent strain of the tested materials. The constitutive model of the tested polymers is, thus, viscoelastic-viscoplastic.

2. Details of the experiments

Three copolymers of high density polyethylene have been investigated. The production of these copolymers (designated as HDPE – 1, HDPE – 2, and HDPE – 3) is described in our previous paper [19]. The main acoustic properties of these copolymers obtained in the paper are briefly

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reviewed in Table 1. These acoustic properties are characterized by velocities of the longitudinal wave c_L and transversal wave c_T . Assuming elastic behavior of the tested materials, both aforementioned velocities make it possible to evaluate the elastic properties such as the Young's modulus E and others.

Table 1. Acoustical properties of the tested polymers

polymer	density ρ [kg · m ⁻³]	c_L [m/s]	c_T [m/s]	E [MPa]	$c_0 = \sqrt{\frac{E}{\rho}}$ [m/s]
HDPE – 1	954	2 452	1 029	2 814.5	1 717.6
HDPE – 2	962	2 438	1 021	2 795.2	1 704.6
HDPE – 3	960	2 416	1 010	2 730.5	1 686.5

The velocity c_0 is the longitudinal wave velocity in the bar under the assumption of uniaxial stress state (1-D) known as the bar velocity. The wave propagation under the 1-D state is then characterized by the specific acoustic impedance

$$Z = \rho c_0.$$

For the experiment, specimen of cylindrical shape (15 mm in diameter and 16.5 mm in length) were prepared. These specimens were loaded using two methods of high strain loading.

2.1. Split Hopkinson pressure bar method

The general arrangement of the SHPB set-up is shown in Fig. 1. The set-up consists of two steels bars of 1000 mm in length and 15 mm in diameter. A striker bar made of the same material is 33 mm in length and 14 mm in diameter. The mechanical properties of the steel are as follows: Young's modulus $E = 2.105$ MPa, bar velocity $c_o = \sqrt{\frac{E}{\rho}} = 5 270$ m/s and acoustic impedance $Z = \rho c_o = 40.8425$ MPa · s/m. The specimen is sandwiched between the input and output bars. Specimen in the form of a cylinder is loaded by the stress pulse $\sigma_I(t)$ for a limited time.

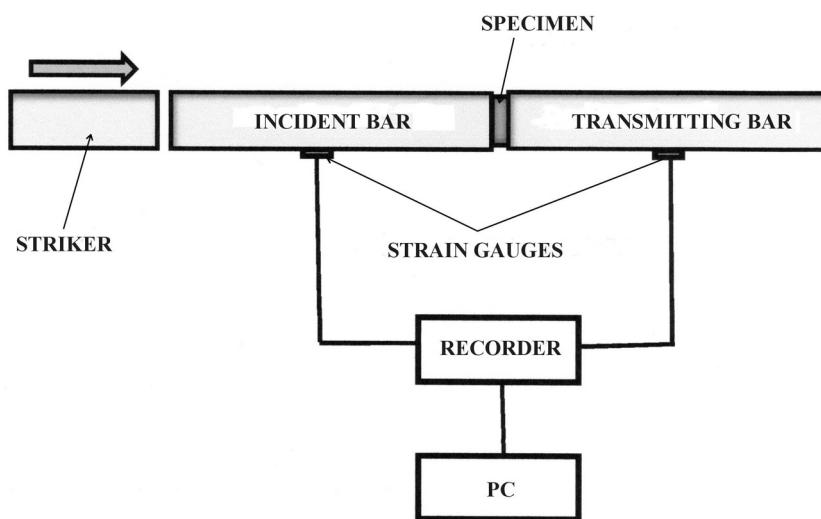


Fig. 1. Schematic drawing of the split Hopkinson pressure bar method

After the impact of the pulse on the specimen, its one part is reflected as a stress pulse $\sigma_R(t)$ and the other one is transmitted through the specimen into the transmission bar as stress pulse $\sigma_T(t)$. The time histories of all stress pulses are recorded using strain gauges. The maximum strain rate $\dot{\varepsilon}_{\max}$ of the SHPB technique depends on the yield strength of the incident pressure bar σ_y . If the striker and the incident pressure bar are made of the same material, the maximum impact velocity V_M , which can be applied without plastically deforming the incident pressure bar, is imposed by

$$\sigma_y = \frac{1}{2} \rho c_o V_M.$$

The maximum strain rate can be estimated as

$$\dot{\varepsilon}_{\max} = \frac{V_M}{2l_0},$$

where l_0 is the length of the specimen.

The higher strain rates can be achieved using the experimental arrangement described, e.g., in [10]. The technique, known as the direct impact Hopkinson pressure bar (DIHPB) technique, makes it possible to reach strain rates up to 10^5 s^{-1} . Its brief description is given in the following section.

2.2. Direct impact Hopkinson pressure bar technique

The experimental device for the loading of the specimen by the direct impact consists of a striker, the specimen and the Hopkinson bar as shown in Fig. 2. The impact velocity V of the striker is typically between 30 and 150 m/s. The loading history $\sigma_I(t)$ is recorded with the help of a strain gage placed on the Hopkinson bar. If the striker mass is large enough to provide kinetic energy much greater than the energy dissipated by the plastic deformation in the specimen, then the impact velocity may be considered to be constant during the test. The strain in the specimen can be expressed as

$$\varepsilon(t) = \frac{u_2(t) - u_1(t)}{l_0}, \quad (1)$$

where $u_1(t)$ is the specimen-impactor interface displacement and $u_2(t)$ is the specimen-bar interface displacement, and l_0 is the specimen length. As described, for example, in [7], these

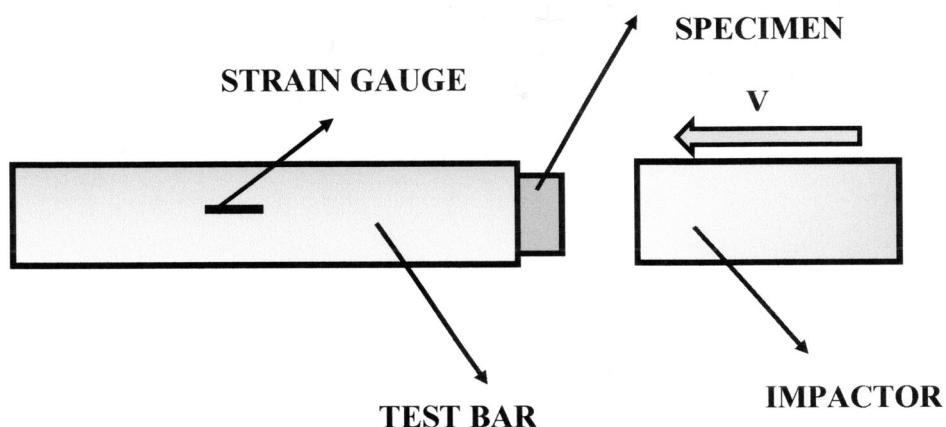


Fig. 2. Schematic drawing of the direct impact Hopkinson bar technique

displacements can be expressed as

$$u_1(t) = Vt - \frac{1}{Z_b} \int_0^t \sigma_T(t) dt, \quad u_2(t) = -\frac{1}{Z_b} \int_0^t \sigma_T(t) dt. \quad (2)$$

Then, the strain rate is given as

$$\varepsilon(t) = \frac{2 \int_0^t \sigma_T dt}{Z_b l_0}, \quad (3)$$

where l_0 is the specimen length, σ_T is the stress recorded in the bar and Z_b is the bar acoustic impedance. The stress in the specimen is expressed as

$$\sigma(t) = \sigma_T(t) \frac{A_T}{A_0}, \quad (4)$$

where A_0 and A_T are the cross-sectional areas of the Hopkinson bar and the specimen, respectively. Some corrections of (4) owing to the radial inertia can be found, e.g., in [5, 10].

3. Experimental results

3.1. Split Hopkinson pressure bar method

Fig. 3 shows experimentally recorded time histories of the incident σ_I , reflected σ_R , and transmitted σ_T stress pulses. For all the tested materials, the experimental records of the given stress pulses demonstrated the same qualitative features.

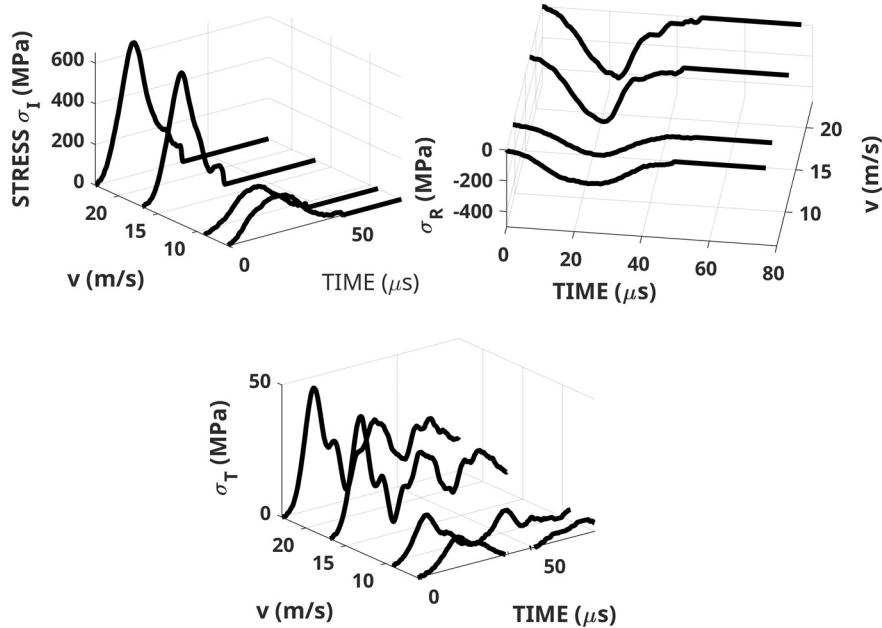


Fig. 3. Polyethylen HDPE – 3: Experimental records of the stress pulses, where V is the striking velocity

The knowledge of the stress pulses makes it possible to evaluate the strain $\varepsilon(t)$ and strain rate $\frac{d\varepsilon}{dt}$ in the specimen using the following equations [9]:

$$\frac{d\varepsilon}{dt} \equiv \dot{\varepsilon}(t) = \frac{\sigma_I(t) - \sigma_R(t) - \sigma_T(t)}{Z_b l}, \quad (5a)$$

$$\varepsilon(t) = \int \dot{\varepsilon} dt, \quad (5b)$$

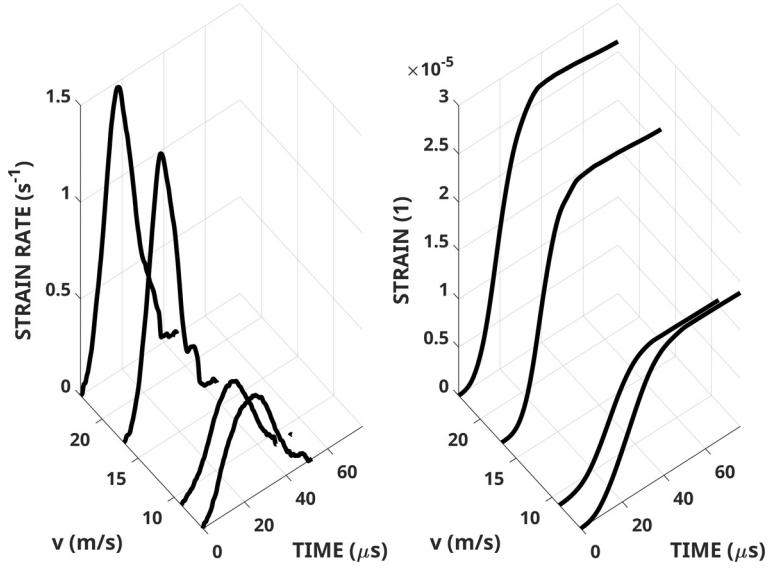


Fig. 4. Strain rates and strains evaluated for the specimens of the polymer HDPE – 3

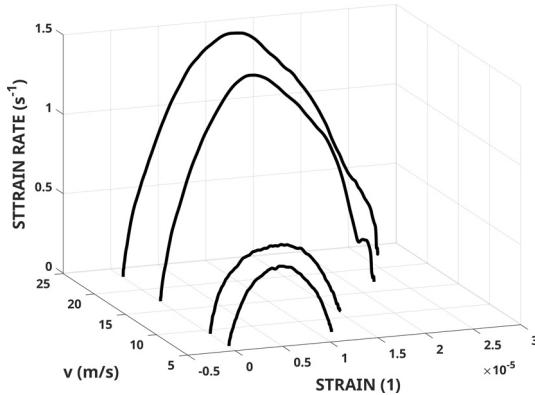


Fig. 5. Polymer HDPE – 3: Strain rate-strain dependence

where l is the specimen length. The time histories of the strain rate and strain are displayed in Fig. 4. The strain rate is not constant as shown in the Fig. 5.

The stress in the specimen is highly non-homogeneous as illustrated by the example in Fig. 6. In this figure, the input stress $\sigma_{\text{input}} = \sigma_I(t) + \sigma_R(t)$, the average stress $\sigma_{\text{average}} = \frac{1}{2}(\sigma_I + \sigma_R + \sigma_T)$ and the transmitted stress $\sigma_T(t)$ are plotted.

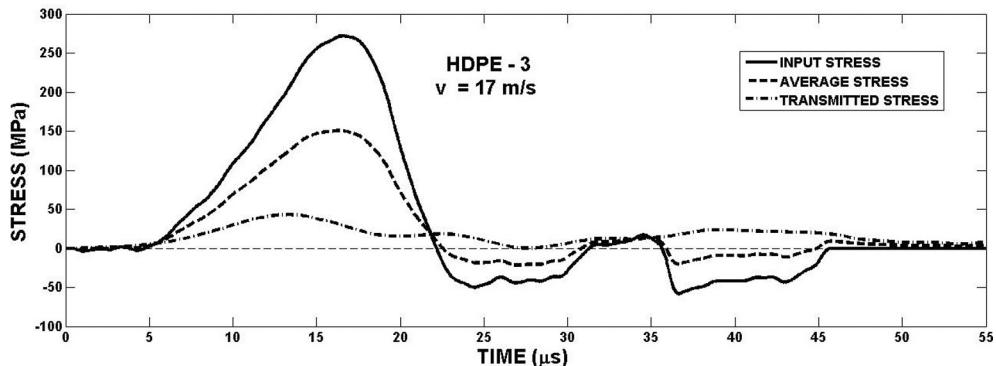


Fig. 6. Time histories of the input average and transmitted stresses in the specimen

Similar results were obtained for all tested samples, meaning significant gradients of the stress are present. Owing to this fact, the specimen behavior cannot be described using the stress strain dependence $\sigma(\varepsilon)$. The stress σ can be evaluated only if the stress equilibrium in the specimen is achieved, i.e., $\sigma = \sigma_I + \sigma_R = \sigma_T$. The stress equilibrium is reached for loading stress pulses σ_I with sufficiently high values of the time of their duration, generally about $200\ \mu\text{s}$ and more. For shorter stress pulses, another procedure for the description of the specimen response is used.

The incident stress pulses are characterized by the following quantities:

- the maximum value of the stress (amplitude) σ_{Im} ,
- the impulse $I_I = \int_0^{\lambda_I} \sigma_I(t) dt$
- the energy $w_I = \frac{1}{Z} \int_0^{\lambda_I} \sigma_I^2(t) dt$, where Z denotes the acoustic impedance of the test bar σ_I .

The response of the tested polymers is described using the following parameters:

- values of the maximum stress (amplitude) of the transmitted pulse σ_{Tm} , value of the reflected pulse σ_{Rm} ,
- impulses of the reflected I_R and transmitted I_T pulses,
- energies of the reflected w_R and transmitted w_T pulses.

Fig. 7 presents the dependence of the parameters describing the specimen response on the input (loading) stress pulse parameters.

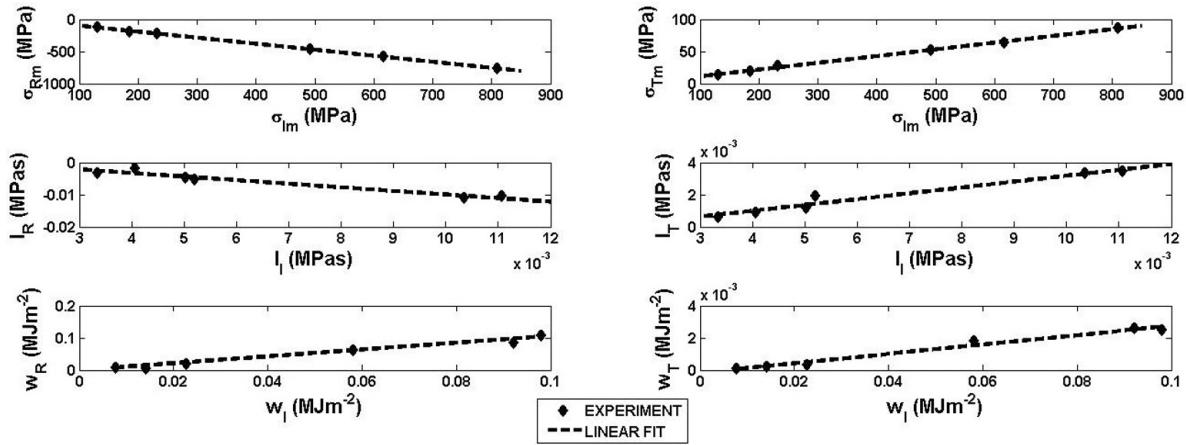


Fig. 7. Polymer HDPE – 1: Parameters describing the specimen response to the stress pulse loading

The experimental points can be fitted by a linear function. The remaining two polymers exhibit the same behavior. An overview of the parameters of the used linear functions is presented in Table 2.

The tested specimens did not exhibit any permanent strain. It means the behavior of the tested polymers is either elastic or viscoelastic. If the behavior of the specimen is elastic, then the transmitted and reflected pulses can be expressed [5] as

$$\sigma_T(t) = \frac{4Z_s Z_b}{(Z_s + Z_b)^2} \sum_{n=0}^{n=\infty} \left(\frac{Z_s - Z_b}{Z_s + Z_b} \right)^{2n} \sigma_I(t - (2n + 1)T_0), \quad (6a)$$

$$\sigma_R(t) = \frac{Z_s - Z_b}{Z_s + Z_b} \sigma_I(t) + \frac{(Z_s - Z_b) Z_b^2}{(Z_s + Z_b)^3} \sum_{n=0}^{n=\infty} \left(\frac{Z_s - Z_b}{Z_s + Z_b} \right)^{2n} \sigma_I(t - 2nT_0), \quad (6b)$$

Table 2. Overview of the dependences describing the specimen response

polymer HDPE – 1	a	b	R^2
$\sigma_{Rm} = a + b\sigma_{Im}$ [MPa]	−5.792	−0.9379	0.9996
$\sigma_{Tm} = a + b\sigma_{Im}$ [MPa]	0.7439	0.1048	0.9964
$I_R = a + bI_I$ [MPa · s]	0.001197	−1.117	0.9358
$I_T = a + bI_I$ [MPa · s]	−0.000471	0.3643	0.9459
$w_R = a + bw_I$ [MJm ^{−2}]	−0.004806	1.066	0.9713
$w_T = a + bw_I$ [MJm ^{−2}]	−0.0001657	0.02915	0.9732

polymer HDPE – 2	a	b	R^2
$\sigma_{Rm} = a + b\sigma_{Im}$ [MPa]	2.286	−0.9624	1
$\sigma_{Tm} = a + b\sigma_{Im}$ [MPa]	2.36	0.1032	0.9984
$I_R = a + bI_I$ [MPa · s]	−0.001476	−0.6349	0.9803
$I_T = a + bI_I$ [MPa · s]	−0.000581	0.2581	0.911
$w_R = a + bw_I$ [MJm ^{−2}]	2.839e−018	1	1
$w_T = a + bw_I$ [MJm ^{−2}]	−5.893e−005	0.01016	0.9834

polymer HDPE – 3	a	b	R^2
$\sigma_{Rm} = a + b\sigma_{Im}$ [MPa]	−1.482	−0.9362	1
$\sigma_{Tm} = a + b\sigma_{Im}$ [MPa]	−0.9604	0.1128	0.9993
$I_R = a + bI_I$ [MPa · s]	0.001654	−1.232	0.8853
$I_T = a + bI_I$ [MPa · s]	−0.0003481	0.2815	0.9506
$w_R = a + bw_I$ [MJm ^{−2}]	−0.01136	1.214	0.9799
$w_T = a + bw_I$ [MJm ^{−2}]	−0.0003448	0.02652	0.9764

where Z_s is the acoustic impedance of the specimen, Z_b is the acoustic impedance of the bar, $T_0 = \frac{l_0}{c_s}$, l_0 is the specimen length and c_s is the bar velocity in the specimen. If we neglect the stress wave reflections in the specimen, the relations reduce to

$$\sigma_T(t) = \frac{4Z_s Z_b}{(Z_s + Z_b)^2} \sigma_I(t - T_0) = T \sigma_I(t - T_0), \quad (7a)$$

$$\sigma_R(t) = \frac{Z_s - Z_b}{Z_s + Z_b} \sigma_I(t) = R \sigma_I(t). \quad (7b)$$

The quantities mentioned above are listed in Table 3.

Table 3. Elastic properties of the tested materials

polymer	Z_s [MPa · s/m]	Z_b [MPa · s/m]	c_s [m/s]	T [1]	R [1]
HDPE – 2	1.1448	40.8425	1 192.5	0.1061	−0.9455
HDPE – 3	1.1595	40.8425	1 205.3	0.1074	−0.9448
HDPE – 1	1.1587	40.8425	1 214.5	0.1073	−0.9448

Taking (7a) and (7b), the pure elastic response of the specimen is

$$\sigma_{Tm} = T\sigma_{Im}, \quad \sigma_{Rm} = R\sigma_{Im}, \quad I_T = TI_I, \quad I_R = RI_I, \quad w_T = T^2w_I, \quad w_R = R^2w_I, \quad (8)$$

where the parameters R and T should be equal to the parameters in Table 2. It is obvious that this condition is approximatively valid only for the dependences $\sigma_{Tm} = f(\sigma_{Im})$ and $\sigma_{Rm} = g(\sigma_{Im})$, meaning that the behavior of the tested polymers is viscoelastic. The viscoelastic properties of the specimens can be obtained if we use the representation of the stress pulses in the frequency domain, see, e.g., [18]. This approach is based on the integral transformation, when, for example, the time function assigns its spectral function by means of

$$S_{I,R,T} = \int_{-\infty}^{\infty} \sigma_{I,R,T}(t)e^{i\omega t} dt,$$

where ω is the angular frequency. The spectral function is generally complex and can be expressed as

$$S(\omega) = \text{Re}S(\omega) + \text{Im}S(\omega) = \sqrt{\text{Re}S(\omega)^2 + \text{Im}S(\omega)^2}e^{i\phi} = S e^{i\phi},$$

where S is the magnitude and ϕ the phase of the spectral function. An example of the spectral function is displayed in Figs. 8 and 9.

The same qualitative features were observed in all experimental results. The knowledge of the spectral function makes it possible to define the transmission T and reflection R functions, respectively,

$$T(\omega) = \frac{S_T(\omega)}{S_I(\omega)}, \quad R(\omega) = \frac{S_R(\omega)}{S_I(\omega)}.$$

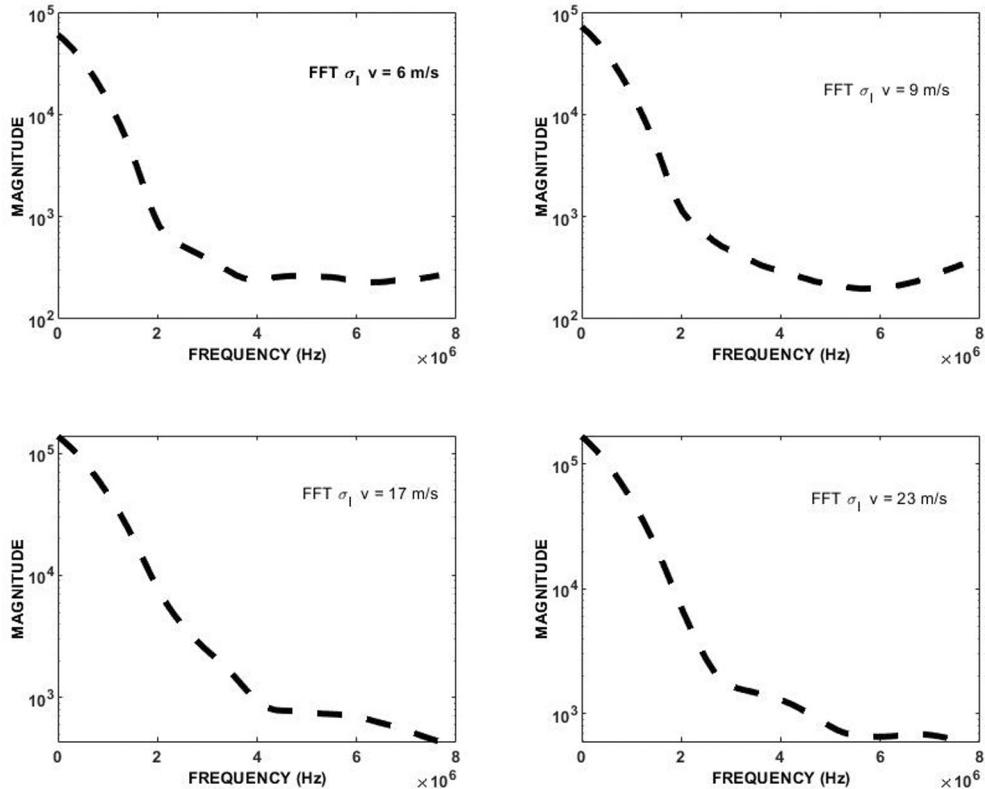


Fig. 8. Polymer HDPE – 2: Magnitude of the spectral function of the incident stress pulse

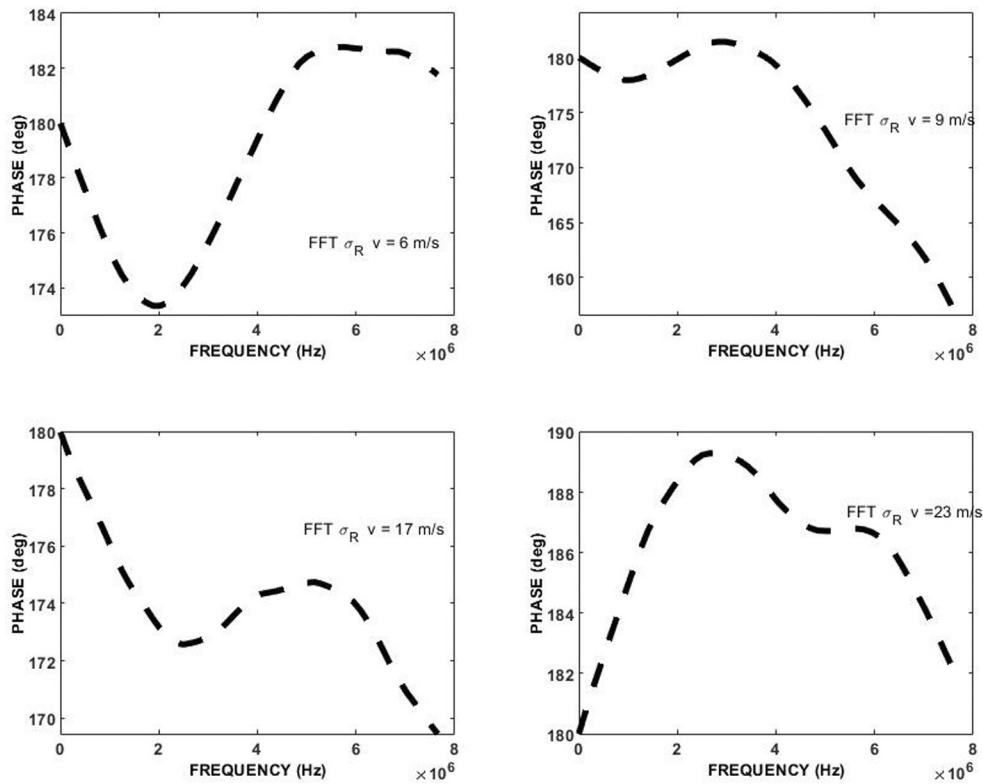


Fig. 9. Polymer HDPE – 2: Phase of the spectral function of the incident stress pulse

Examples of the frequency dependence of the transfer function amplitude and phase are shown in Figs. 10 and 11.

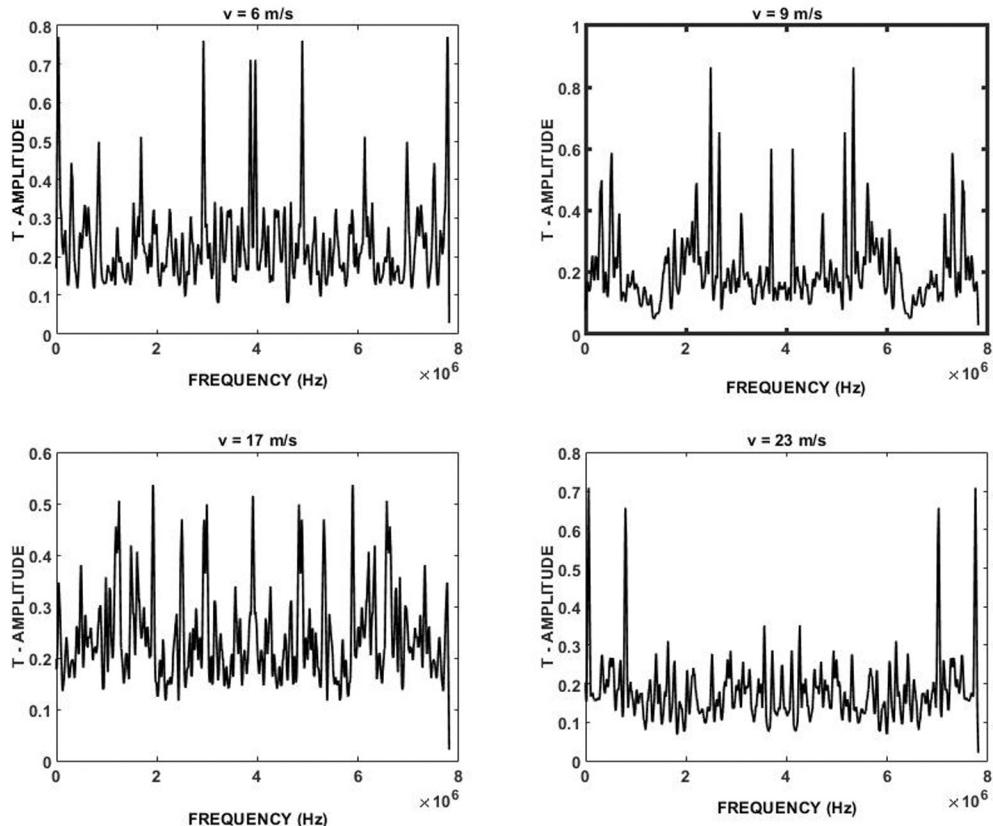


Fig. 10. Polymer HDPE – 2: The effect of the frequency on the transfer function amplitude

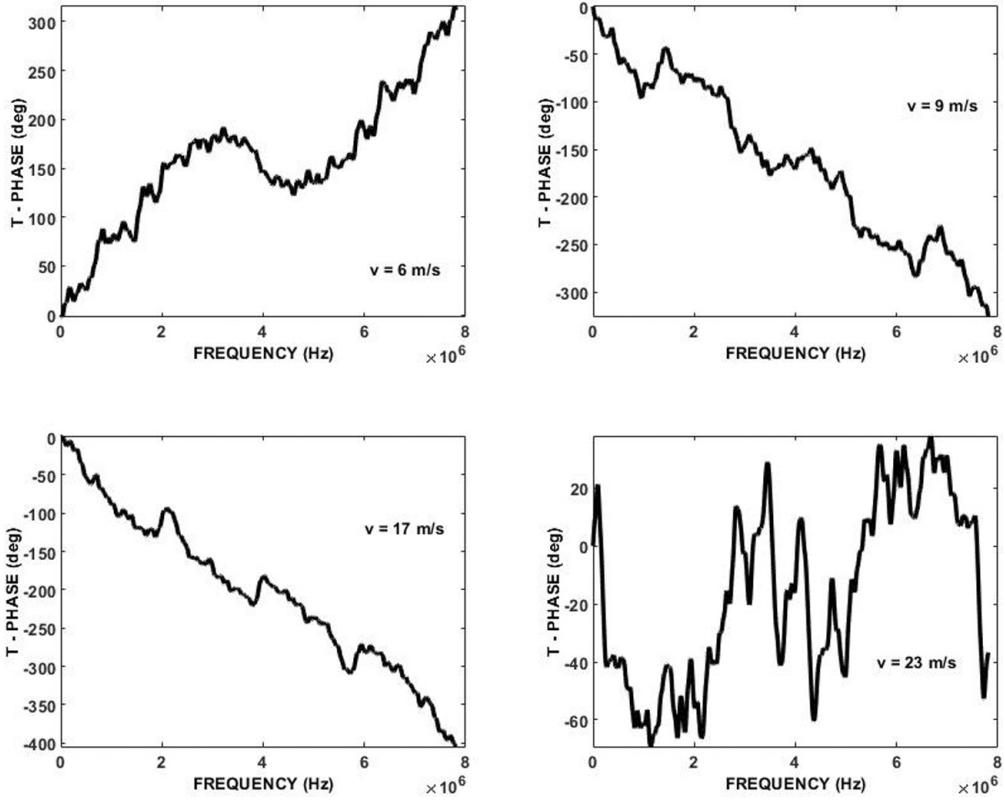


Fig. 11. Polymer HDPE – 2: The effect of the frequency on the transfer function phase
For a linear elastic material, the transfer function T can be expressed as [16]

$$T(\omega) = \frac{4Z_s Z_b}{(Z_s + Z_b)^2} \frac{\exp(-i\omega T_0)}{1 - \left(\frac{Z_s - Z_b}{Z_s + Z_b}\right)^2 \exp(-i2\omega T_0)}.$$

An example of the amplitude of the transfer function is shown in Fig. 12.

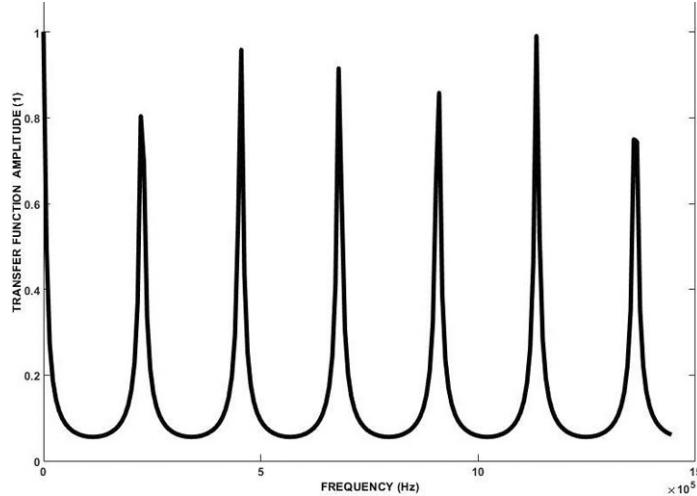


Fig. 12. Polymer HDPE – 2: Amplitude of the transfer function for the striker velocity $V = 23 \text{ m/s}$

One can see that the specimen behavior differs from the linear elastic one. Using the method of viscoelastic properties evaluation mentioned, e.g., in [11], the behavior of viscoelastic materials is described using of the complex modulus

$$E(\omega) = E'(\omega) + iE''(\omega),$$

where E' is the storage modulus and E'' is the loss modulus. The complex modulus can be expressed as

$$E(\omega) = E \exp(i\varphi),$$

where $E = \sqrt{E'^2 + E''^2}$ and $\tan \varphi = \frac{E''}{E'}$. Figs. 13–15 show the frequency dependence of the amplitude and phase of the complex modulus.

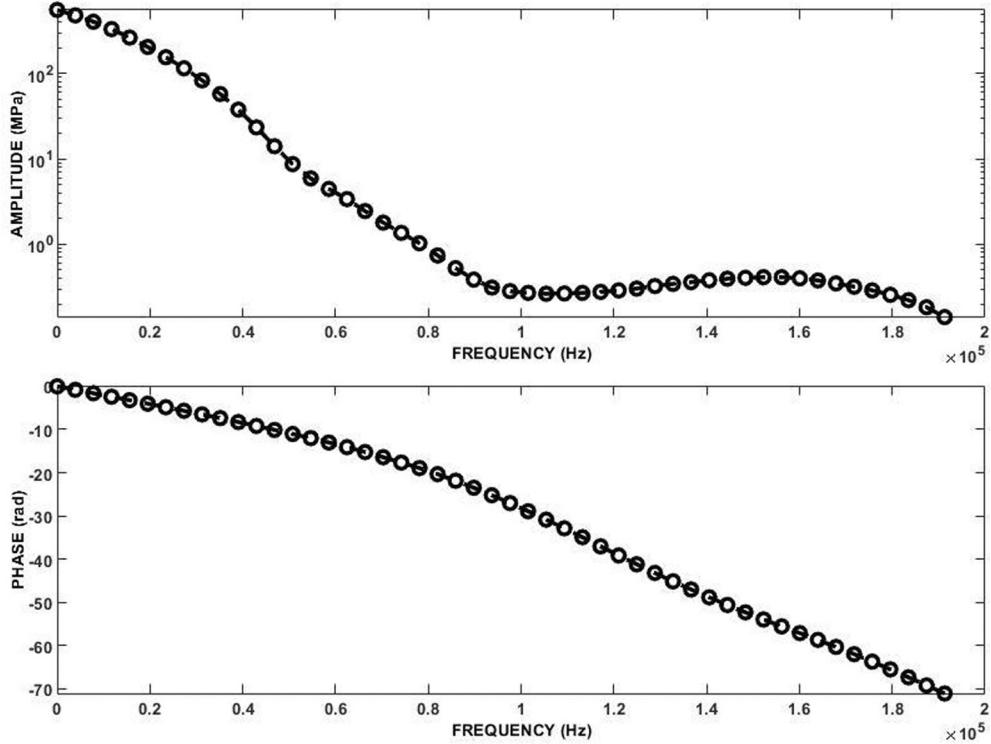


Fig. 13. Complex modulus of the HDPE – 1

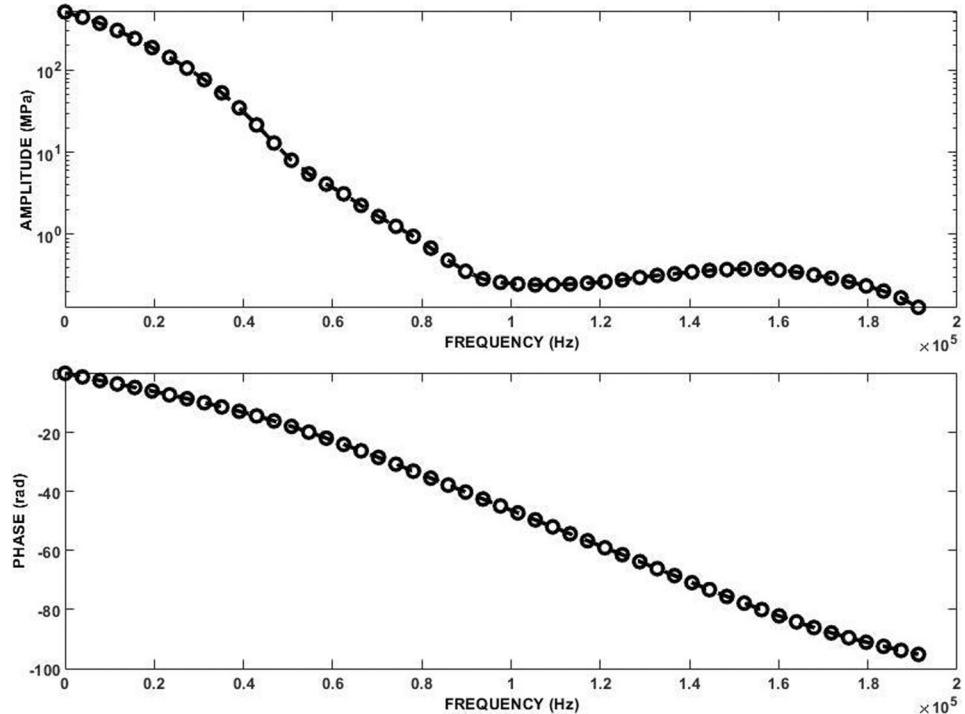


Fig. 14. Complex modulus of the HDPE – 2

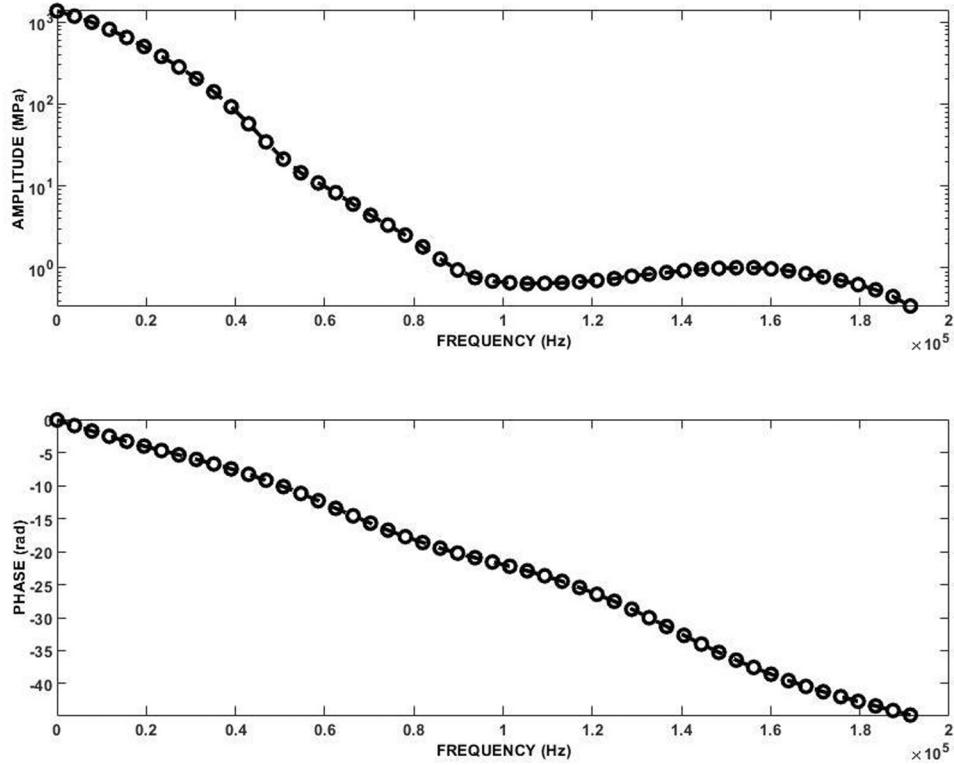


Fig. 15. Complex modulus of the HDPE – 3

Note that similar results were obtained, e.g., in [6]. The difference among the tested polymers is relatively small as documented by Figs. 16 and 17.

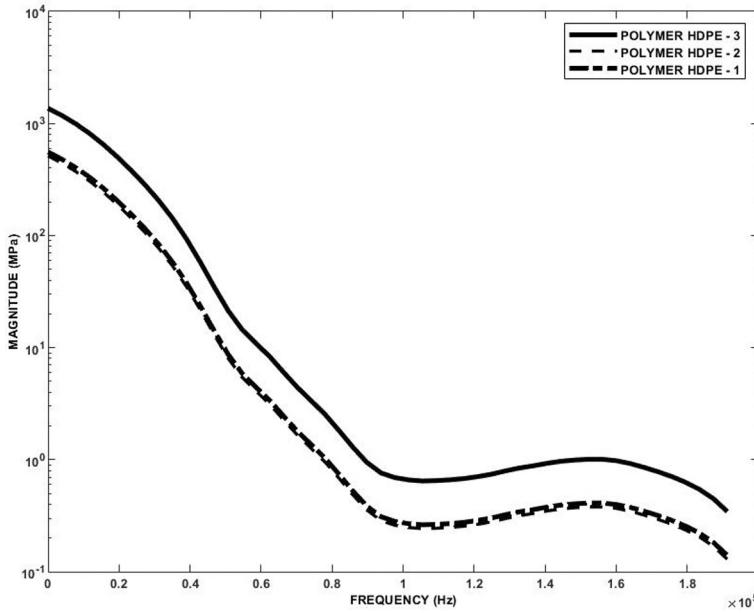


Fig. 16. Frequency dependence of the magnitude

3.2. Direct impact Hopkinson pressure technique

In order to obtain additional information, preliminary experiments using the direct Hopkinson pressure bar (DIHPB) technique were also performed. In Fig. 18, the time histories of the stress in the bar are displayed.

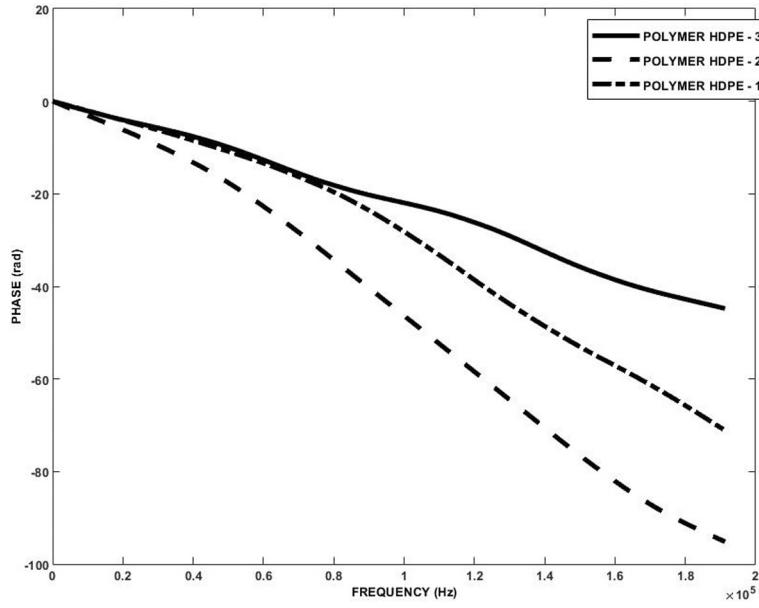


Fig. 17. Frequency dependence of the phase

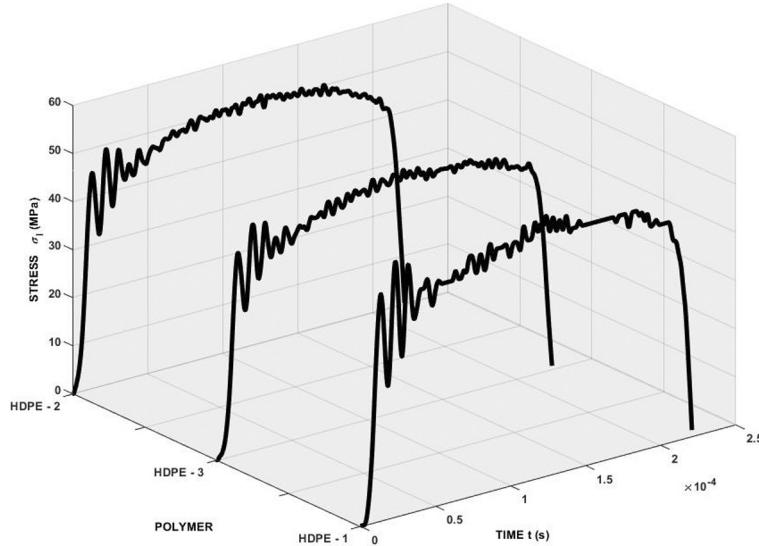


Fig. 18. Stress pulses recorded during the direct impact Hopkinson bar test

The impact velocity was the same as in the previous case, i.e., $V = 45$ m/s. Very small if not negligible differences in behavior between the tested polymers were observed. The behavior of a specimen during the loading is shown in Fig. 19. It is evident that there is a significant change in the specimen shape, which is permanent. Quantitatively, the response of the specimen can be described both in time and in the frequency domain. An example of the frequency analysis is shown in Fig. 20.

The main parameters describing the specimen response to the direct impact in the frequency domain show that there is no difference in behavior between the tested polymers during the direct impact.

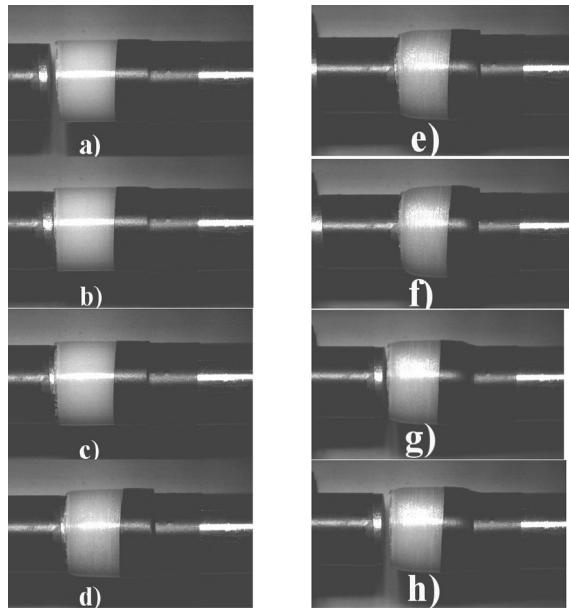


Fig. 19. High speed sequence of polymer BB54 – 002 direct impact test, the distance between the frames is $30 \mu\text{s}$

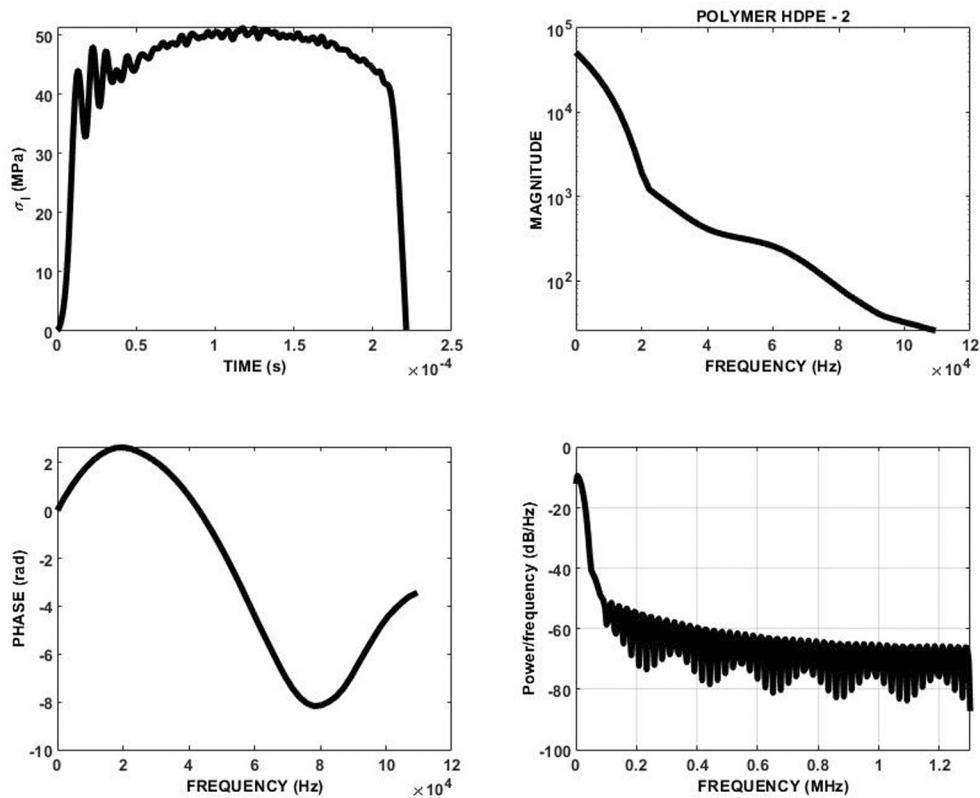


Fig. 20. Example of the response of the specimen in time and in the frequency domain

4. Conclusions

In the present paper, the response of three polymers to stress pulse loading was studied. First, the response was investigated using the split Hopkinson pressure bar technique. The obtained results showed that there is no stress equilibrium in the specimen and, at the same time, that the strain rate is non-homogeneous. The response of the specimen was evaluated using selected main parameters of the stress pulses. The analysis of the obtained data showed viscoelastic behavior of the tested polymers. The frequency analysis of the recorded stress pulses was used to obtain the complex modulus. The frequency dependence of the magnitude of this modulus exhibited very small difference among the tested polymers. A slightly more significant difference could be observed in the case of frequency dependences of the phases of the complex moduli.

The preliminary results obtained using the direct Hopkinson pressure bar technique showed that the increase in the strain rate led to the beginning of permanent strain. The behavior of all tested polymers could be described in terms of the viscoelastic-viscoplastic constitutive equation.

Acknowledgements

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