COMPOSITE ARTIFICIAL TENDONS WITH HYDROGEL MATRIX

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Dedicated to Dr Blahoslav Sedlacek on the occasion of his 70th birthday.

The brief review is concerned with synthetic model tendons consisting of water-swollen hydrogel matrix and poly(ethylene terephthalate) fibres. Texturized fibres were employed in tendon construction in order to imitate stress–strain and other mechanical properties of natural tendons. Stiffness, creep behaviour, strength, and elongation-at-break of model tendons are predetermined by those of incorporated fibre bundles. Thus, by varying the fibre volume fraction, it is possible to achieve mechanical properties required for various types of tendon prostheses.

Natural tendons are complex composite structures consisting of collagen fibres and of a matrix containing gel-like acid mucopolysaccharides and fibroblast cells[1–3]. The matrix surrounds fibre bundles and protects their integrity while the collagen fibres, oriented along the tendon axis, allow for high tensile stiffness and strength[3–6]. As the primary role of tendons is the load transfer between muscle and bone, the stress–strain dependences have received most attention[7,8]. In the toe (initial) region, the tensile stress S increases slowly (Fig. 1) with the strain up to $e = 0.02$ or so. Low initial modulus, as found, e.g., for rat tail tendons, is a result of the straightening of the zig-zag waveform of collagen fibres into a parallel pattern[9–11]. The stress–strain behaviour of straightened collagen fibres is characterized by the stress–strain linearity limit 3–6% (after the subtraction of the toe region) and tensile modulus E = 1–1.5 GPa. At higher strains, the collagen fibres are irreversibly damaged, which prevents restauration of the original waveform after stress release. Natural tendons show strain-at-break e_b = 6–14% and tensile strength S_b = 10–60 MPa.

Complex problems associated with the healing[1,2,12–17] or replacement of damaged tendons soon initiated attempts at the preparation of various artificial tendons[1,2,14–24] or substitutes based on, e.g., carbon fibres[25–27], Kevlar fibres[28], or poly(tetrafluoroethylene) fibres[29] which have very good biocompatibility and make the reconstruction of tendons or ligaments possible[28,30], though they are mechanically and compositionally quite different from host tissues and natural tendons. If an artificial tendon is used only

during healing of an injured sheath and then replaced with a natural graft[12–14,17,31,32], its mechanical properties and durability play a secondary role. In the case of a long-term use, the synthetic tendons should duplicate the mechanical properties of the substituted material and remain reliable throughout the required period of time. In the first attempts, various synthetic tendons were prepared as composites consisting of continuous poly(ethylene terephthalate) fibres (PET, e.g. Dacron) and of the matrix of silicone rubber (Silastic)[1,2,15,16,18,19]. Both components are known for their good biocompatibility, which can be attributed, among other things, to the fact that they can be prepared in a very pure form[33].

Later on, silicone rubbers were replaced by hydrophilic gels which are crosslinked structures absorbing up to 40–80% of water. Also hydrogels, such as poly(2-hydroxyethyl methacrylate) (PHEMA) and its derivatives, rank among materials with very good biocompatibility[33–37]. Their applications are frequently limited due to their inferior mechanical properties[38]. However, some hydrogels prepared by copolymerization of HEMA with some other methacrylates possess remarkable mechanical properties which are presumably related to the block structure of the copolymers[38,39]. Nonetheless, suitable composite structures have to be prepared[37], even in less demanding situations than those of synthetic tendons. Recently, biodegradable composite tendon prostheses have been prepared[40] which consist of water-swollen hydrogel matrix and poly(lactic acid) fibres. Their construction is quite analogous to that of previous artificial tendons reinforced with crimped PET fibres[20].

The objective of this brief review is to summarize our results related to (i) preparation and development of poly(ethylene terephthalate) fibres and their bundles convenient for artificial tendons; (ii) preparation (on laboratory scale) of model tendons consisting of PET fibre bundles and PHEMA matrix; (iii) mechanical testing of the tendons; (iv) control of the tendon structure and composition in order to mimic mechanical properties of natural tendons.

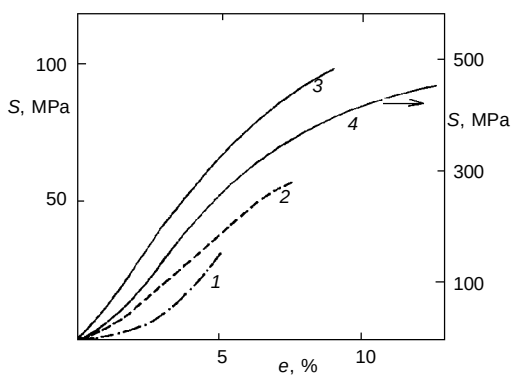

FIG. 1
Stress (*S*)–strain (*e*) curves for horse tendon (*1*), rat tail tendon (*2*), artificial tendon (*3*), and poly(ethylene terephthalate) fibre bundle (*4*) (ref.[20])

Preparation of Fibre Bundles and Model Tendons

We have used texturized fibres for several reasons: (i) they are very "voluminous", which ensures their uniform distribution throughout the cross-section of the tendons; (ii) the fibres located near the surface impart enhanced abrasion resistance to the tendons (smooth untexturized fibres are not distributed uniformly in the tendon cross-section and the surface layer formed by neat hydrogel has poor mechanical properties); (iii) bundles of texturized fibres allow us to imitate stress–strain behaviour of natural tendons, i.e., in the toe region the fibres undergo uncrimping (straightening) which is marked with a low apparent modulus; at larger strains, the modulus increases rapidly due to actual tensile deformation of the present fibres.

For model tendons, PET fibres (about 10 μm in diameter) with false twist were used, which is one of the commonest ways of fibre texturization. A laboratory procedure consisting of thermal treatment and mechanical conditioning was developed[20] in order to raise the tensile modulus and strength and to suppress the plastic component of the fibre creep. Later on, a procedure of fibre upgrading was implemented on a semi-pilot plant scale[22,41]: after forming a large number of twists (hundreds per 1 m of fibre length) on a bundle of tens of fibres (36 in our case), the twists were fixed at a temperature close to 220 °C; finally, the fibres were cooled, untwisted, and subject to postdrawing (additional drawing at a temperature below that of the fixation so as to preserve sufficient texturization of the fibres) in order to suppress the plastic component of tensile strain.

Prepared bundles of fibres were washed three times in ethanol in order to remove lubricant, dried at 60 °C, and pulled through a silicone rubber hose 2, 3, or 4 mm in diameter. The fibres were slackened by 0.5% (samples A0, B, C; Table I), 1.3% (sample A1) and 2.5% (sample A2) of their fully straightened length so that owing to their texturization they could uniformly fill the cross-section of the hose (Fig. 2). Using an oil pump, a mixture containing 60 vol.% HEMA, 0.09 vol.% ethylene dimethacrylate as the crosslinking agent, 40 vol.% glycerol diacetate as the diluent and 0.1 wt.% azobisisobutyronitrile as the polymerization initiator was sucked into the hose. (The concentration of the diluent approximately corresponds to the equilibrium swelling of PHEMA in water so that extraction and substitution of glycerol diacetate with water does not lead to a change in the volume of the PHEMA matrix.) Then both ends of the hose were sealed and polymerization was carried out in a water bath (65 °C, 2 h). The water bath contained some 0.05% of sodium disulfite as the deoxidizing agent because oxygen diffusing through the wall of the silicone rubber hose may negatively affect the polymerization and mechanical properties of PHEMA in the surface layer of tendons. Model tendons taken out from silicone hoses were extracted with water for one month in order to remove initiator residues and to replace the used diluent by water. The tendons were stored at room temperature in water containing traces of sodium azide which prevents formation of molds.

Mechanical Testing of Artificial Tendons

Creep of fibres and tendons was determined[20,22] by using a conventional apparatus with the specimen between clamps in vertical position. The lower clamp was fixed, the imposed load acted via a pulley on the upper mobile clamp through the core of an indicator which detected the displacement with an accuracy of 0.01 mm. The sample length was between 7 and 9 cm, the maximal displacement was about 4 mm, which means deformation of about 5%. Before measurements, the samples were subjected to mechanical conditioning[42], i.e., they were exposed for 2 h to a stress corresponding to about 130% of the maximal stress used in the experiments and then fully recovered within 24 h.

TABLE I
Mechanical properties of synthetic tendons[20]

Code	d^a, mm	V_f^b	E^c, GPa	$S_b{}^d$, MPa	$e_b{}^e$, %	$L_b{}^f$, N	E/ν_f, GPa	S_b/ν_f, MPa
A0	2	0.189	1.60	95.2	9.6	299	8.46	504
A1	2	0.192	1.43	95.2	8.6	299	7.44	496
A2	2	0.194	1.53	92.1	7.3	289	7.88	475
B	3	0.168	1.42	70.8	7.4	500	8.45	421
C	4	0.190	1.39	76.5	6.9	961	7.31	403
PET fibre	–	1.00	5.92	475	12.8	–	5.92	475

[a] Diameter; [b] volume fraction of fibres; [c] tensile modulus in the fibre direction; [d] tensile strength; [e] strain at break; [f] load at break.

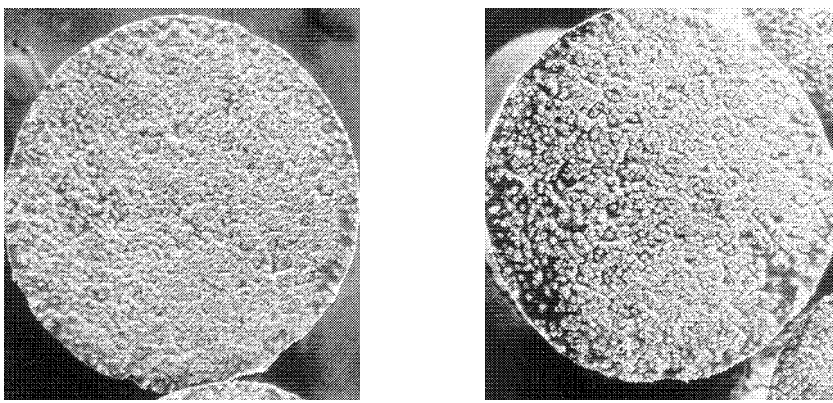

FIG. 2
SEM micrographs of the cross-section of two artificial tendons consisting of texturized poly(ethylene terephthalate) fibres and hydrogel matrix (diameter 2 mm, dried)

Stress–strain dependences were determined with the aid of an Instron tester[20] (strain rate 10%/min). Due to high pressures produced in clamping, the tendons usually broke in the jaws. To determine the strength of fibres or fibre bundles, their loops were put on spools. The transverse modulus of tendons was determined[21] from the force needed for the indentation of a bead[43] (5 mm in diameter). All the experiments described above were carried out in distilled water at 37 °C.

Mechanical Properties of Fibre Bundles

The stress–strain curves (Fig. 1) show that modified PET fibres have a tensile modulus E_f = 7–9 GPa, average tensile strength S_{fb} = 0.55–0.65 GPa, and strain-at-break e_{fb} = 16–19%. In creep testing (Fig. 3), fibres or model tendons were loaded for 1 min and the resulting strain was used to calculate the creep tensile modulus. The stress–strain dependences were linear up to strains about 4% (Fig. 4), so that a reliable value of the creep tensile modulus can be calculated. To achieve a complete recovery of the viscoelastic part of deformation, a time longer by an order of magnitude than that of creep is needed[42]; irreversible (plastic) deformation was therefore assessed after 10 min of recovery (Fig. 3). The creep (1 min)–recovery (10 min) cycle was repeated five times, the cycles following immediately one after another. The imposed stress rose from the first to the third cycle (up to 40% of the sample strength); the fourth and fifth cycles were a repetition of the first and second cycles in order to detect changes in the mechanical properties (if any) due to the highest stress imposed. Table II which contains data for post-drawn fibres (drawing ratio 1.184 at 140 °C) shows that the total plastic deformation in three cycles amounts to about 0.5%. No further plastic deformation was identified in cycles 4 and 5. This result suggests that modified PET fibres have mechanical properties which make them suitable for the use in artificial tendons.

The tensile modulus of composites with unidirectionally oriented continuous fibres is given for the fibre direction by the rule of mixing[44–46]:

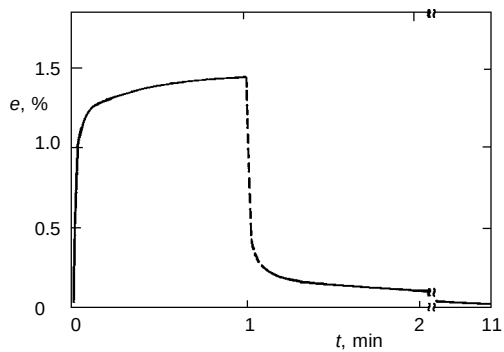

FIG. 3
Strain (*e*) of an artificial tendon[20] as a function of time (*t*): creep 1 min, recovery 10 min

$$E = E_f v_f + E_m v_m , \quad (1)$$

where E_m is the modulus of the matrix; v_f and v_m are the volume fractions of the fibre and matrix, respectively. If the tensile behaviour of a fibre composite is linear up to fracture, its tensile strength in the fibre direction can be calculated by using a formally similar equation[44–46]:

$$S_b = S_{fb} v_f + S_{mb} v_m , \quad (2)$$

where S_{mb} is the stress in the matrix at the strain-at-break of fibres. As the modulus and strength of hydrogels are usually by three orders of magnitude lower than those of PET fibres, the second term in Eqs (*1*) and (*2*) can be neglected.

Equation (*2*) is based on an unrealistic assumption that the strength of all fibres in a bundle is the same. In fact, however, the strength of individual fibres is determined by the most severe defect (flaw) in their structure which statistically varies from fibre to fibre. In fibre bundles, weaker fibres break first with increasing load, so that the stress acting upon the remaining (stronger) fibres rises faster than the loading. In composites

TABLE II
Mechanical properties of drawn fibres in repeated creep[20]

Cycle number	1	2	3	4	5
Stress, MPa	75	176	252	75	176
Modulus E_f, GPa	5.56	4.94	5.34	5.66	5.00
Plastic deformation, %	0.27	0.17	0.09	<0.01	<0.01

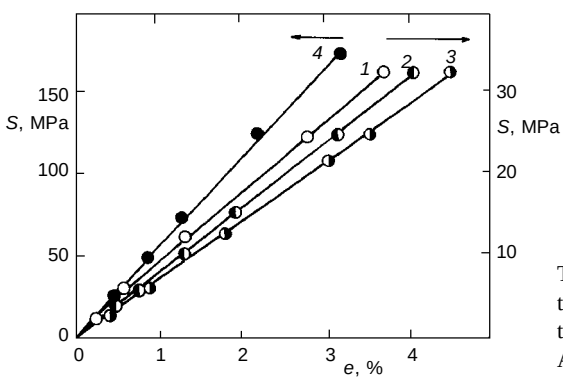

FIG. 4
The stress–strain dependences of artificial tendons determined by using the data for the 1 min creeps[20] (*4*) fibre bundle; (*1*) A0; (*2*) A2; (*3*) A3

with a good interfacial fibre/matrix adhesion, the broken fibres can also carry a part of the load transmitted upon them from the matrix through shear forces. In synthetic tendons, however, the adhesion between fibres and water-swollen hydrogel matrix is very low so that the parts of broken fibres are easily pulled out from the matrix. For this reason, the strength of the model tendons may be approximated by the strength of present fibre bundles. The calculations of the fibre bundle strength by using the Weibull statistical approach[47] were in a great detail described elsewhere[22,45,48]. They show that the strength of fibre bundles (related to a cross-section unit) is always lower than the average fibre strength, but both values are of the same order of magnitude. The decrease in the bundle strength is the larger, the broader is the fibre strength distribution. If the real cumulative distribution function[22] of the tensile strength of PET fibres (Fig. 5) is taken into account for tendons containing from 1 800 to 7 400 fibres, the real tendon strength can be expected to be lower by approximately 20% than that predicted from Eq. (*2*) when an average value of fibre tensile strength is considered.

Mechanical Properties of Artificial Tendons

The stress–strain dependences of synthesized tendons are less S-shaped in their toe region than those observed for parent fibre bundles (Fig. 1). The observed dependences were not noticeably affected by different percentage of the slackening of built-in fibre bundles. This result suggests that incompressible hydrogel inside the fibre bundles impedes straightening (uncrimping) of the fibres before the onset of their viscoelastic extension and thus reduces the toe region.

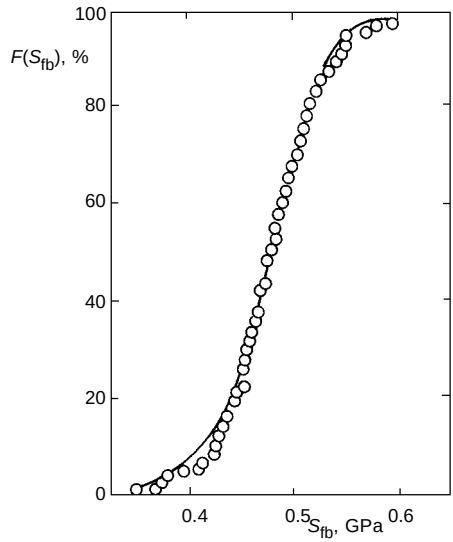

FIG. 5
Cumulative distribution function of the tensile strength S_{fb} of poly(ethylene terephthalate) fibres[22]. Solid line represents approximation of experimental data by the Weibull distribution function

As far as the creep behaviour is concerned, the observed dependences for model tendons are predetermined by those of the parent fibres (Fig. 6), which is in conformity with Eq. (*1*). (Obviously, the contribution of the hydrogel matrix itself to the tendon compliance $D(t)$ is negligible.) In the range between 1 and 100 min, the creep of tendons proceeds with an almost constant creep rate $\mathrm{d} \log D(t)/\mathrm{d} \log t = 0.025$. Though the compliance does not approach an equilibrium value, the observed extent of creeping is very small. Complete recovery was also observed for samples subject to 100 min creep (Fig. 6), but its time dependence was not recorded because of practical reasons.

In real situations, a long-term creep of tendons is rather unlikely. As much more frequent, a short-term creep can be expected followed by an interval during which the tendons recover. Such a regime was imitated in a very simplified manner: after 1 min of creeping (at tensile stress 12.5 MPa), the load was removed for 1 min, and the cycle was repeated 30 times without interruption (Fig. 7). As indicated in Fig. 3, 1 min was sufficient only for recovery of 90–95% of the induced elongation. Thus, the unrecovered strain e_u (during 1 min recovery) in cyclic creep rose with the number n of the cycles (Fig. 7). The dependence e_u–log n was found to be approximately linear for all investigated samples. It is significant that the strain produced during one minute of creeping attained an equilibrium value after about 10 cycles. These results reveal that the model tendons lose about 10% of their stiffness in the first decade of cycles, while further cyclic loading does not reduce it noticeably. In conformity with the static experiments, the recovery of the samples loaded 30 times was complete after a sufficiently long time (some 10 h).

Tensile modulus and strength of the composites with unidirectionally oriented continuous fibres can be approximated for the fibre direction by Eqs (*1*) and (*2*), respectively. Table I shows that the E/v_f values of tendons are by 20–40% higher than those of incorporated fibre bundles. This discrepancy can be ascribed to the effect of the fibre embedding in incompressible hydrogel matrix, which causes that straightening of the texturized fibres is partly replaced by their authentic tensile deformation.

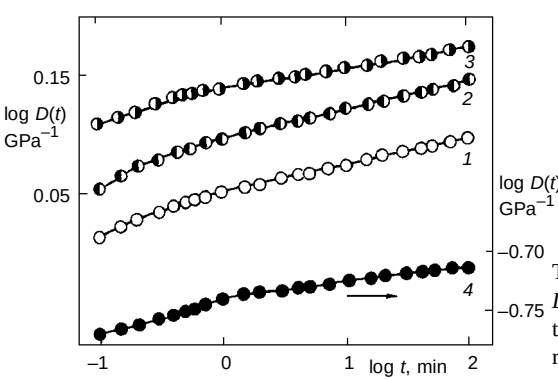

FIG. 6
Time dependence of the tensile compliance $D(t)$ for fibres (right) and tendons (left) at the stress 12.5 MPa (ref.[20]). For numbering see Fig. 4

The strength of the tendons approximately corresponds to the strength of built-in fibre bundles (Table I). Nevertheless, one can see a small decrease in S_b/v_f with increasing diameter of tendons, which is in accord with the generally recognized concept that with rising diameter of a specimen the strength diminishes due to increasing probability of the presence of larger (critical) defects in the structure of specimens. The strain-at-break is lower for tendons than for fibres, due obviously to the suppression of the straightening of texturized fibres inside the hydrogel matrix. The extensibility of the artificial tendons ranges in the interval 6.8–9.6%, which is close to that of collagen fibres. Equations (*1*) and (*2*) evidence that the stiffness and strength of the composite tendons can be controlled by the content of texturized fibres. For the PET fibres texturized with false twist, the volume fraction convenient for synthetic tendons varies within the range 0.15–0.25. Such volume fractions allow for an even distribution of fibres in the tendon cross-section and enable the mechanical properties of natural tendons to be closely imitated.

Medical Testing of Synthesized Tendons

In vivo testing of the model synthetic tendons was carried out at the University of Naples[23,24]. The fibre bundles for the tendons were slightly modified in order to facilitate the suture with natural tissues. Tendons of appropriate dimensions were implanted in rabbits as a substitution of Achille's tendon. The implants underwent mechanical and histological tests after 6–12 weeks of implantation. The stress–strain measurements of the whole muscle–tendon system revealed that the mechanical properties of the implanted and as-prepared tendons were very similar. General examination confirmed the absence of any inflammatory reaction in the vicinity of implants (cf. refs[40,49,50]). This

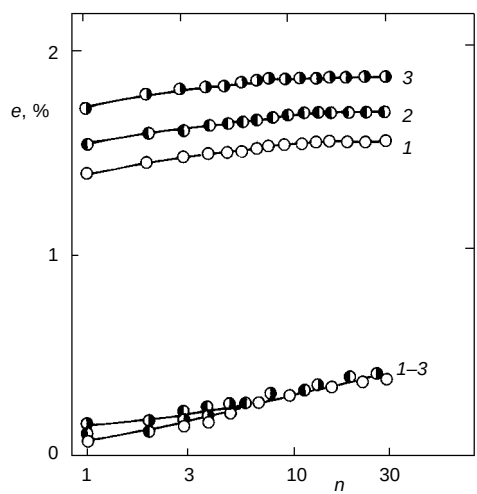

FIG. 7
Dependences of the strain (*e*) of tendons after 1 min creep (upper part) and of the strain after the following 1 min recovery (lower part) on the number (*n*) of creep cycles (at stress 12.5 MPa). For numbering see Fig. 4

finding proved a very good biocompatibility of the developed synthetic tendons. In periods longer than 60 days, proliferation of cells into the tendon hydrogel matrix was observed. Thus, it seems that the PET/hydrogel artificial tendons could be qualified as candidate materials for the substitution of natural tendons.

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REFERENCES

1. Dunn H. K., King R., Andrade J. D., DeVries K. L.: J. Biomed. Mater. Res. Symp. *4*, 109 (1973).
2. King R. N., McKenna G. B., Statton W. O.: J. Appl. Polym. Sci., Appl. Polym. Symp. *31*, 335 (1977).
3. Torp S., Maer E., Friedman B.: Colston Pap. *26*, 223 (1975).
4. Baer E., Gathercole L. J., Keller A.: Colston Pap. *26*, 189 (1975).
5. Scott J. E.: Chem. Br. *15*, 13 (1979).
6. Kastelic J., Baer E.: Polym. Prepr. *20*(1), 282 (1979).
7. Herrick N. C., Kingsbury H. B., Lou D. Y. S.: Biomed. Mater. Res. *12*, 877 (1978).
8. Kenedi R. M., Gibson T., Evans J. H., Bartenel J. C.: Phys. Med. Biol. *20*, 699 (1975).
9. Torp S., Arridge R. G. C., Armeniades C. D., Baer E.: Colston Pap. *26*, 197 (1975).
10. Baer E., Hiltner A., Friedman B.: Mekh. Polim. *6*, 1051 (1975).
11. Dale W., Baer E.: J. Mater. Sci. *9*, 369 (1974).
12. Hunter J. M., Steindel C., Salisbury R., Hughes D.: J. Biomed. Mater. Res. Symp. *5*, 155 (1974).
13. Hunter J. M., Subin D., Minkow F., Konikoff J.: J. Biomed. Mater. Res. Symp. *5*, 163 (1974).
14. Salisbury R. E., McKeel D., Praitt D., Mason A. D., Palermo N., Wade C. W. R.: J. Biomed. Mater. Res. Symp. *5*, 175 (1974).
15. Wade C. W. R., Oulletee J. E., Hodhe J. W., Urban J. J., Salisbury R. E.: J. Biomed. Mater. Res. Symp. *6*, 146 (1975).
16. King R. N., Dunn H. K., Bolstad K. E.: J. Biomed. Mater. Res. Symp. *6*, 157 (1975).
17. Amstutz H. C., Coulson W. F., David E.: J. Biomed. Mater. Res. *10*, 47 (1976).
18. Hunter J. M.: Am. J. Surg. *109*, 325 (1965).
19. Hoffman A. S.: J. Appl. Polym. Symp. *31*, 313 (1977).
20. Kolarik J., Migliaresi C., Stol M., Nicolais L.: J. Biomed. Mater. Res. *15*, 147 (1981).
21. Migliaresi C., Kolarik J., Stol M., Nicolais L. in: *Rheology* (G. Astarita, G. Marrucci and L. Nicolais, Eds), Vol. 3, p. 635. Plenum Press, New York 1980.
22. Kolarik J., Vanicek J., Migliaresi C.: J. Biomed. Mater. Res. *18*, 115 (1984).
23. Migliaresi C., Nicodemo L., Nicolais L. in: *Hydrogels in Medicine and Pharmacy* (N. A. Peppas, Ed.), Vol. III, p. 84. CRC Press, Boca Raton 1987.
24. Migliaresi C., Ambrosio L., Guida G., Kolarik J., Nicolais L., Ronca D. in: *Proceedings of the Second World Congress on Biomaterials, Washington D.C., April 27–May 1, 1984*, p. 137. The Society for Biomaterials, Washington D.C. 1984.
25. Mendas D. G., Iusim M., Angel D., Rotem A., Mordehovich D., Rottman M., Lieberson S., Boss J.: J. Biomed. Mater. Res. *20*, 699 (1986).
26. Makisalo S. E.: Biomaterials *10*, 499 (1989).
27. Makisalo S. E., Paavolainen P., Gronblad M., Holmstrom T.: Biomaterials *10*, 105 (1989).

28. Zimmerman M., Gordon K. E.: J. Biomed. Mater. Res. *22*, 339 (1988).
29. Williams R., August S.: Am. J. Surg. *107*, 913 (1964).
30. Ricci J. L., Gona A. G., Alexander H., Parsons J. R.: J. Biomed. Mater. Res. *18*, 1073 (1984).
31. McMaster W. C., Kouzelos J., Liddle S., Waugh T. R.: J. Biomed. Mater. Res. *10*, 259 (1976).
32. Hsu S. Y. C., Cheng S. C. Y., Chong Y. W., Leung P. C.: Biomaterials *10*, 258 (1989).
33. Lyman D. J. in: *Biomedical Polymers* (A. L. Bement, Ed.), p. 269. University of Washington Press, Seattle 1971.
34. Sprincl L., Vacik J.: J. Biomed. Mater. Res. *7*, 123 (1973).
35. Hoffman A. S. in: *Polymers in Medicine and Surgery* (R. L. Kronenthal, Z. Oser and E. Martin, Eds), p. 259. Plenum Press, New York 1975.
36. Bruck S. D.: J. Biomed. Mater. Res. *7*, 387 (1973).
37. Corkhill P. H., Hamilton C. J., Tighe B. J.: Biomaterials *10*, 3 (1989).
38. Kolarik J., Migliaresi C.: J. Biomed. Mater. Res. *17*, 757 (1983).
39. Kolarik J.: J. Polym. Sci., Polym. Phys. Ed. *21*, 2445 (1983).
40. Davis P. A., Huang S. S., Ambrosio L., Ronca D., Nicolais L: J. Mater. Sci., Mater. Med. *3*, 359 (1991).
41. Vanicek J., Kolarik J., Berka M.: Czech. 203 822 (1982).
42. Ward I. M.: *Mechanical Properties of Solid Polymers*, p. 217. Wiley-Interscience, London 1971.
43. Hrouz J., Vojta V., Ilavsky M.: Polym. Eng. Sci. *20*, 402 (1980).
44. Vinson J. R., Chou T. W.: *Composite Materials and Their Use in Structures*, p. 224. Applied Science, London 1975.
45. Broutman L. J., Krock R. H.: *Modern Composite Materials*, p. 124. Addison-Wesley, Reading, MA 1967.
46. Kolarik J.: *Vysokomodulova vlakna a vlaknove kompozity*, p. 47. Academia, Praha 1984.
47. Weibull W.: J. Appl. Mech. *18*, 293 (1951).
48. Coleman B. D.: J. Mech. Phys. Solids *7*, 60 (1958).
49. Guidoin R., Gosselin C., Domurado D., Marois M., Levaillant P. A., Awad J., Rouleau C., Levasseur L.: Biomater. Med. Devices Artif. Organs *5*, 177 (1977).
50. Eskin S. G., Trevino L., Chimosky J. E.: J. Biomed. Mater. Res. *12*, 517 (1978).