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The 50th anniversary of the death of Adolf Gustav Smekal (1895–1959), a pioneer in materials physics

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Abstract The Austrian physicist Adolf G. Smekal (1895–1959) was one of the most creative and active materials scientists of the twentieth century. This paper reports about the contributions of Smekal and co-workers to the physics of fracture, fractography, fracture dynamics, micro-plasticity, and the comminution of brittle materials.

Biographical

Adolf Smekal was born in Vienna on 12 September 1895, as only son of an artillery officer. He received his primary education at schools in Brünn and Ölmütz (now situated in the Czech Republic). At the latter school he passed Matura with excellent results in 1912. Between 1912 and 1913, he studied physics, mathematics, chemistry, and astronomy at the TH Wien (Vienna University of Technology), but he moved to the University of Graz, where he was promoted a Dr. Phil. in 1917. His doctoral advisor was Michael Radakovic (1866–1934), a theoretical physicist. Because of severe problems with his eye-sight, Smekal was exempted from military service. Between 1917 and 1919, he spent time in Berlin, Germany, in order to complete his studies in mathematics and physics. He went back to Austria and became the assistant to Heinrich Mache (1876–1954) at the Institute of Physics of the TH Wien. In 1920, he was the assistant to Gustav Jäger (1865–1938) at the Universität Wien (University Vienna), and he habilitated in the area of theoretical and experimental physics at this university. One year later he moved to the TH Wien, but went back to the Universität Wien in 1927 as a professor of theoretical physics. His scientific interest was focused on the application of the new young quantum theory to different areas of physics. An important result of that period was the theoretical forecast of the inelastic scattering of light from a gas, liquid, or solid with a shift in wavelength from that of the usually monochromatic incident radiation [1]. Smekal assumed that light has a quantum structure and showed that scattered monochromatic light would consist of its original wavelength as well as of higher and lower wavelengths. This effect is now named after the Indian physicist C.V. Raman (1888–1970), who could experimentally prove Smekal's theory. Today, Raman spectroscopy is a standard method for materials analysis. Unfortunately, Smekal's name is almost forgotten in that context. (In German speaking countries only, the effect is sometimes called "Smekal–Raman Effect".) A well-arranged review about Smekal's work in Vienna is provided by Mehra and Rechenberg [2].

In 1928, Smekal followed a call at the Martin-Luther-University in Halle/Saale, Germany, as a Chair Professor for theoretical physics, where he was working till the end of the Second World War. He rejected calls at the Universities of München (Munich) and Würzburg. During the 17 years he spent in Halle, Smekal and his co-workers made numerous lasting contributions to certain areas of fracture physics and materials science, which will be reported about in the present article. In June 1945, shortly before the Soviet Army occupied Saxony-Anhalt, the Americans evacuated Smekal (and many other scientists and engineers from the Universities of Halle and Leipzig as well as from the optics manufacturers Schott and Carl Zeiss of Jena) from Halle to Darmstadt in South Germany. At the University of Technology of that place, Smekal worked as a provisional professor for physics between 1946 and 1948.

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Fig. 1 Adolf Gustaf Smekal (1895–1959) in the 1930s (Photography: American Institute of Physics, College Park, Maryland, USA)

In 1949 he followed a call in his native country to the University of Graz. There he lived and worked till the end of his live, by March 7, 1959. Adolf Smekal was married twice (in 1924 and in 1942). He had one child with his second wife. Smekal is portrayed in Fig. 1.

Fracture physics

Historic aspects of fracture physics, respectively, fracture mechanics, are covered in the papers of Rossmanith [3] and Cotterell [4], whereby Smekal was just briefly mentioned in Rossmanith's contribution. Smekal was (like his colleague in Vienna, Karl Wolf [5]) one of the first scientists who critically argued the fracture theory of Griffith [6], which was introduced in 1921. In an early paper in *Die Naturwissenschaften*, he reviewed the results of Griffith (as well as those of Michael Polanyi who, at the Kaiser Wilhelm-Institute for Fibre Chemistry, Berlin, did research on the mechanical behavior of crystals) [7]. In the result of his studies, Smekal introduced the term *molecular tear resistance*, which he quantified in a later comprehensive paper as the so-called *internal fracture criterion* [8]. This parameter reads as follows (symbols in modern style):

$$\sigma_{\mathrm{M}} = \left(\frac{E_{\mathrm{M}} \cdot \gamma_{\mathrm{M}}}{r_0} \right)^{1/2}. \tag{1}$$

In that equation, E_{M} is the Young's modulus, γ_{M} is the specific surface energy, r_0 is the molecular effective range. The counterpart of the *molecular tear resistance* is, according to Smekal, the *technical tear resistance* (respectively, *external fracture criterion*), which reads as follows:

$$\sigma_{\mathrm{T}} = \left(\frac{4 \cdot E_{\mathrm{M}} \cdot \gamma_{\mathrm{M}}}{\pi \cdot r_0} \right)^{1/2} \cdot \left(\frac{R}{L_{\mathrm{R}}} \right)^{1/2}. \tag{2}$$

In that equation, R is the curvature at the crack tip, and L_{R} is the crack length. It can be shown that the widely cited Griffith equation is valid for the special case $r_0 = R$ of this general solution only. In fact, Griffith mentioned in his paper [6], that the curvature at the crack tip shall be of the same order as the molecular dimensions of the materials. (He mentioned $R = 5 \times 10^{-8}$ cm, which agrees well with the range provided by Smekal [8]: 5×10^{-8} to 10^{-7} cm.) Smekal formulated a fundamental, physically sound criterion. The molecular *tear resistance* always exceeds the *technical tear resistance*, and it is independent upon specimen geometry. The ratio between the square of the failure strength and Young's modulus—today referred to as *elastic strain energy density*—was also considered by Smekal on the molecular level. He called this parameter *cohesion energy*. Further, he defined a second necessary fracture criterion for isotropic brittle materials, which can be read as follows [8]:

$$\frac{\gamma_{\mathrm{M}}}{E_{\mathrm{M}} \cdot r_0} \approx 10^{-2}. \tag{3}$$

Smekal also tried to systematically arrange brittle solids according to their structures [9]. He distinguished solids according to their flaw density into: (i) solids with a *high flaw density* (e.g., salt crystals), (ii) solids with a *moderate flaw density* (e.g., quartz), and (iii) solids with a *low flaw density* (mica). The first two groups can be summarized as *non-homogeneous* solids, while the latter case characterizes *homogeneous* solids. Solids belonging to groups (i, ii) were further subdivided according to the geometry of the flaws into solids with *single non-homogeneity* and solids with *composed non-homogeneity*, and according to the orientation of the flaws into *isotropic* non-homogeneous solids and *anisotropic* non-homogeneous solids.

Between 1931 and 1936, numerous investigations about the fracture of glass rods were conducted in the laboratory in Halle, and the results were published in a series of papers in the famous *Zeitschrift für Physik* [8–14]. The experimental results were summarized and analyzed by Smekal in comprehensive papers [8, 9]. Smekal introduced the term *stress-thermal fracture characteristic* [8], which is closely related to two effects: the effects of internal flaws in the materials, and the combination of mechanical loading and diffusion. The Griffith model, which does not consider the contribution of internal flaws to the fracture process, cannot describe the associated features. It, therefore, describes a *non-thermal fracture characteristic*. Stress-thermal features are based on a combination of mechanical loading and diffusion. This combined effect causes the drift of near-surface elements off the flaw tip. The number of drifting elements depends on temperature (θ) and on the time available (respectively, loading rate, v):

$$\sigma_T = f(v, \theta). \quad (4)$$

If surface elements drift away, crack tips become blunt. As can be seen from Eq. 2, this process counteracts crack propagation. However, drifting requires time and, therefore, has a notable effect in the range of low crack propagation velocities only. Smekal considered the case of a *primary flaw*, and he found that the size of the *primary fracture area* (see Fig. 2 for an example: mirror plane is considered primary fracture area) varied if loading rate and temperature changed (see Sect. 3). In case of a circular glass rod, the *primary fracture area* is the so-called mirror plane (s); this is illustrated in Fig. 2. For this case, and for tensile loading, Eq. 4 can be rewritten as follows [8, 9]:

$$\sigma_T = f(s). \quad (5)$$

Here, s is the mirror plane area. Equation 5 combines fracture process features and fracture plane features. The field of fracture physics, which deals with such relationships, is fractography. Fractography was a favoured working area of Smekal for many years.

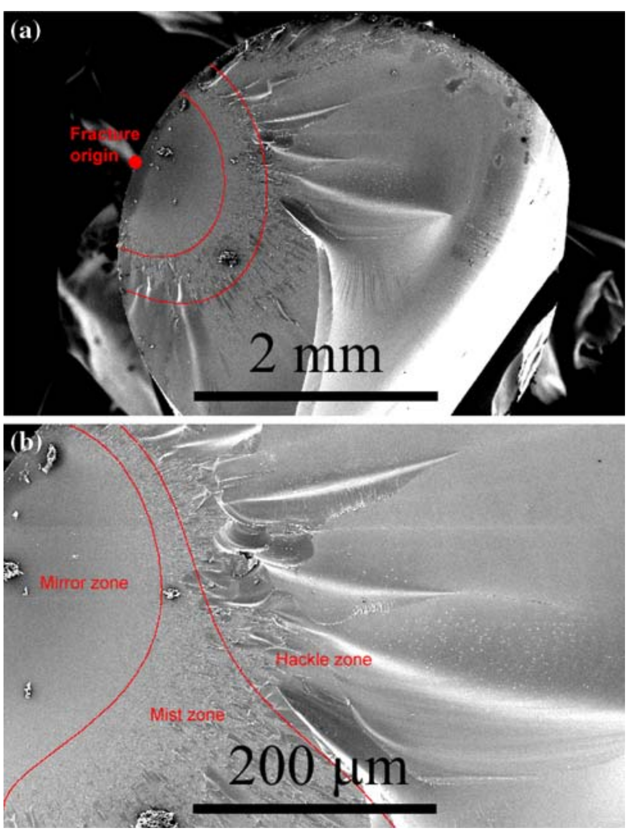

Fig. 2 Fracture surface of a glass rod broken during a tensile test. **a** General view with fracture origin; **b** topographical features (Photographies: University of Cambridge, Cambridge, UK)

Fractography

Fracture plane analysis

Fractographic analyses of fracture planes can deliver phenomenological as well as quantitative information about the fracture process. Smekal early recognized and employed the resultant possibilities.

In three papers in 1936 and 1937, Smekal discussed the importance of fracture plane features, and he suggested ways how to employ their appearance for fracture analysis in great detail [8, 9, 15]. A typical appearance of a tensile fracture of a glass rod is shown in Fig. 2. It basically consists of a mirror plane, which is surrounded by hackle and mist regions. Smekal and his co-workers in Halle have experimentally described all essential properties of such fracture planes in detail, and they have interpreted them in terms of Smekal's fracture theory [8–12]. The institute in Halle had an excellent reputation in terms of accurate experimentation. A basic result of the work in Halle was the introduction and construction of so-called *material curves*. A material curve can be constructed by plotting the rupture strength versus the relative mirror area of the fracture plane. An example is shown in Fig. 3. The straight line drawn in the graph can be expressed as follows:

$$\sigma_T = \sigma_0 \cdot (1 - s_0). \quad (6)$$

Here, σ_0 is the so-called *reduced rupture strength*, and s_0 is the relative mirror area (ratio between mirror area to total

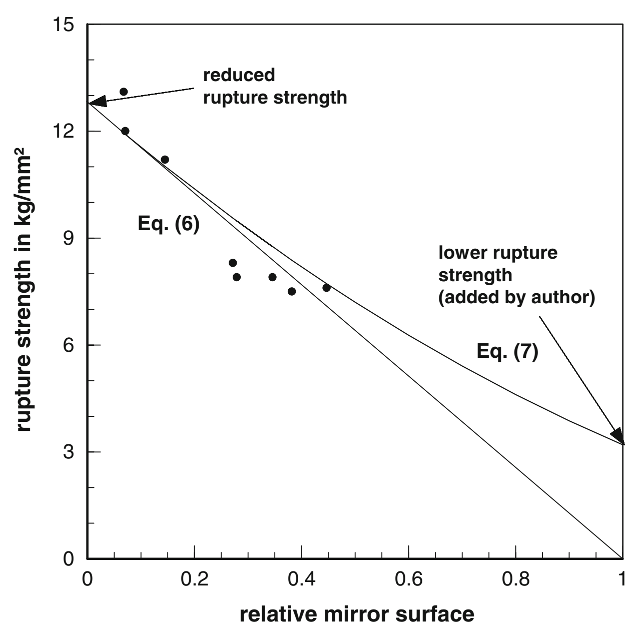

Fig. 3 Example of a *material curve* according to Smekal's definition. Measurements (at room temperature) taken from Apelt [10]. The *dotted line* is illustrative only. Its calculation is based on the strength values listed in the last line of Table 1

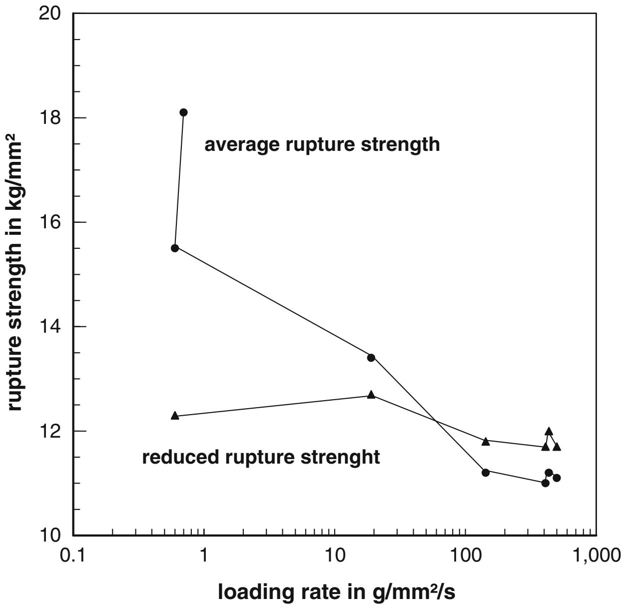

Fig. 4 Stress-thermal fracture, evidenced due to the dependence of the average rupture strength on loading rate. The *reduced rupture strength*, however, does not depend on loading rate, but is a material parameter. Temperature 400–445 °C. Measurements taken from Eichler [13]

rod cross section). Equation 6 is one solution to the more global Eq. 5. It is valid for small relative mirror areas up to about 40%. For very small values for s_0 (<0.1) only, the *material curve* depended on temperature [12]. The parameter σ_0 can be found at the intersection between material curve and ordinate ($s_0 = 0$). In Fig. 3, its value is about 12.8 kg/mm^2. The parameter σ_0 is a material parameter, and its scattering can characterize the anisotropy of a material [10]. The scattering would be an expression of the amount and influence of *internal flaw* on the fracture process [12]. Results plotted in Fig. 4 show that the reduced rupture strength is independent on loading rate, while the average rupture strength drops with an increase in loading rate. This latter trend, however, depended on the environmental temperature. There existed a transition temperature of about 140 °C. For lower temperatures, an opposite trend between loading rate and average rupture strength was found [10]. Experimental results showed that Eq. 6 failed for larger relative mirror areas ($s_0 > 0.4$), and that it should be replaced by a quadratic relationship between relative mirror surface and rupture strength [11]:

$$\sigma_T = \sigma_0 - \sigma_0 \cdot s_0 + \sigma_U \cdot s_0^2 \qquad (7)$$

This procedure would deliver an intersection between the function and $s_0 = 1$, which produced a second rupture strength σ_U, called *lower rupture strength* by Smekal and co-workers [9]. This strength parameter had notably lower

Table 1 Limiting rupture strength values for two glasses [11] (see also Fig. 3)

Glass type	Temperature (°C)	σ_0 (kg/mm^2)	σ_U (kg/mm^2)
Gundelach glass (pre-stressed)	20	9.0	3.2
Gundelach glass	20	11.0	3.1
	102	10.7	3.9
Schott glass	106	12.8	3.2

values than the reduced rupture strength, and it should be used in order to define the fracture resistance of the material for practical applications. Typical values for the two strength parameters are listed in Table 1. Large relative mirror areas could be detected in particular, if the glass rods were provided with an artificial external flaw, as considered in Griffith's experiments [6]. At $s_0 = 0.4$, Eq. 7 would approach Eq. 6. However, developments in fracture mirror size analysis have superseded this particular procedure, and namely Eq. 6 shall be replaced by other relationships. The reader may refer to the work of Quinn [16].

Wallner lines

A further fractographical discovery, which is considered to be classical today, is closely related to Smekal's name. In 1939, a visiting assistant to Smekal in Halle, Helmuth Wallner, published a paper in the *Zeitschrift für Physik*, where he discussed certain line structures on the mirror planes of fractured glass samples [17]. An example for such lines is provided in Fig. 5a. Today, these lines are known as *Wallner* lines. However, Smekal firstly reported about this phenomenon to the scientific community already in December 1938 during the "Hallenser Physikalisches Kolloquium". He immediately recognized the possibility to use these lines for the graphical estimation of the fracture propagation velocity, and he also provided a physical explanation. Small, internally stressed areas in the vicinity of the fracture origin (which can be generated artificially due to grinding or etching) emit ultrasound waves, which interact with the moving crack front and leave marks at the fracture surface. The frequency of these short waves was estimated by Smekal [18] to be up to 10^{10} Hz. These marks can be utilized to follow the fracture process. The complete procedure is illustrated in Fig. 5b. A very nice photograph of an original fracture plane, where such assessment lines are drawn at, can be found in one of Smekal's review papers [19]. Later, Kerkhof, who was very much aware of Smekal's work, refined this method by using artificially generated ultrasound waves [20]. The most eminent advantage of the Wallner–Smekal method over any other method was the possibility to document the fracture propagation velocity along the entire fracture propagation path (see Sect. 4).

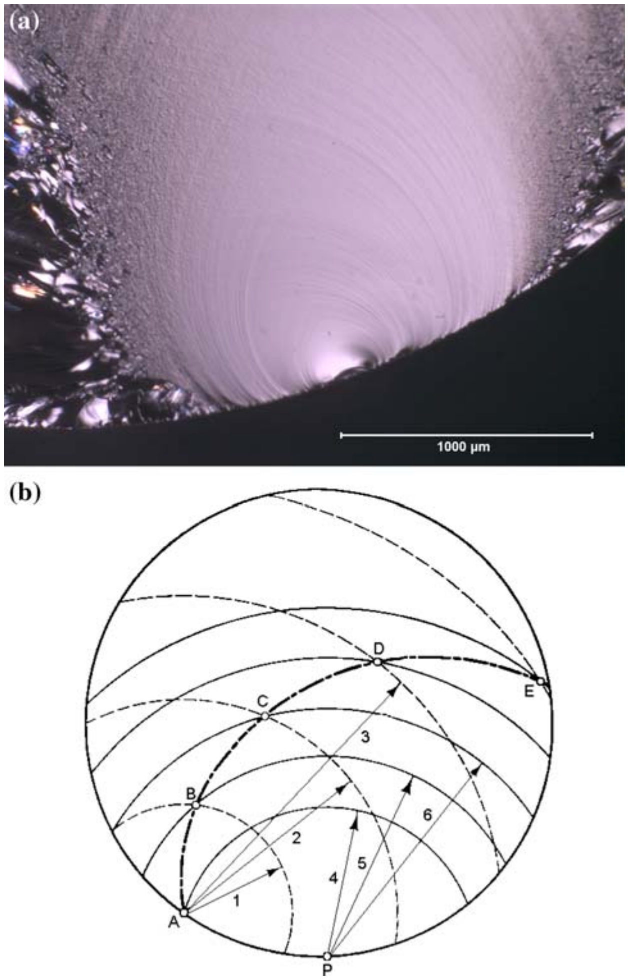

Fig. 5 Appearance and construction of a Wallner line. **a** Fused silica glass rod, broken in tension, showing Wallner lines and mirror markings (Courtesy of G.D. Quinn, NIST, Gaithersburg, USA). **b** Construction scheme (Ref. University of Freiburg, Freiburg i.B., Germany). *A*—starting point of the ultrasound wave; *1–3*—circular arcs of the ultrasound wave at three time periods; *P*—starting point of fracture; *4–5*—fracture front radii at three time periods; intersecting points *B–E*—Wallner line

Lower limit is a ratio between fracture propagation velocity to speed of sound of about 0.2, because no Wallner lines form below this value. Smekal has investigated these relationships very thoroughly, and he summarized his results in a comprehensive paper [19]. More recent theoretical considerations on the formation of Wallner lines are, among others, due to Hull [21] and Rabinovitch et al. [22].

Fracture lances

Under multiaxial stress condition, some very typical fracture surface features can be observed. A classical feature is the appearance of *fracture lances*. An early description of this type of fracture was due to Murgatroyd in 1942, who called them *hackle marks* [23]. A much more detailed description, along with the utilization of a special test arrangement, was reported by Sommer in his often cited publication about the superposition of tensile and anti-plane shear loading on glass rods [24]. (The cover image of the journal *Engineering Fracture Mechanics* is taken from this paper.) It seems, however, that a researcher at Smekal's institute in Halle, Gerhard Apelt, performed already in 1945 exactly such tests with an oil coat and reported about the formation of fracture lances under mixed loading conditions. Unfortunately, numerous results of investigations carried out in Smekal's institute during the last months of the Second World War could not be published. Not until many years later, during Smekal's time in Graz, these results could be analyzed and were partly published [25]. One of the last academic dissertations supervised by Smekal at the University of Graz was Kienle's dissertation on fracture investigations on glass rods, where the formation of fracture lances was described and analyzed in detail [26]. The appearance of curved lances, as portrayed in Sommer's paper, which are caused by a superposition of tensile and torsion stresses, was already noted in that dissertation, along with the formation of straight lances which must have been caused by different mechanisms. A qualitative explanation for the formation of fracture lances was provided by Smekal in 1953 [25]. According to this explanation, fracture lance formation is due to the local adjustment of the crack plane to changes in the direction of the maximum principle stress. In glass, cracks tend to propagate in a plane perpendicular to the axis of the maximum principal stress at the crack tip. The situation is illustrated in Fig. 6. The cleavage fracture, generated due to a tensile stress σ_I, propagates up to the line *P–Q*. At that point, principal tensile stress changes direction (σ_I'). This condition can be simulated experimentally by the superposition of a small amount of torsion, or due to ultrasound modulation. This stress is still perpendicular to the direction of the primary fracture, but no longer perpendicular to the already formed fracture areas (B_1 and B_2). A continuous adjustment of the crack plane along the entire crack front is not possible. Thus, the crack breaks into partial fronts (a–d) which can adjust to the new stress direction. The lines (hatched sections in Fig. 5) separating these partial fracture planes are the observed fracture lances. Sommer has shown that a critical rotation (of about 3°) of the principal stress field is required for the nucleation of lances [24]. More recent theoretical considerations on the formation of fracture lances are, among others, provided by Hull [28] and Lazarus et al. [29].

Fracture propagation velocity

In the July issue of the journal *Glastechnische Berichte* 1938, the famous work of Schardin and Struth

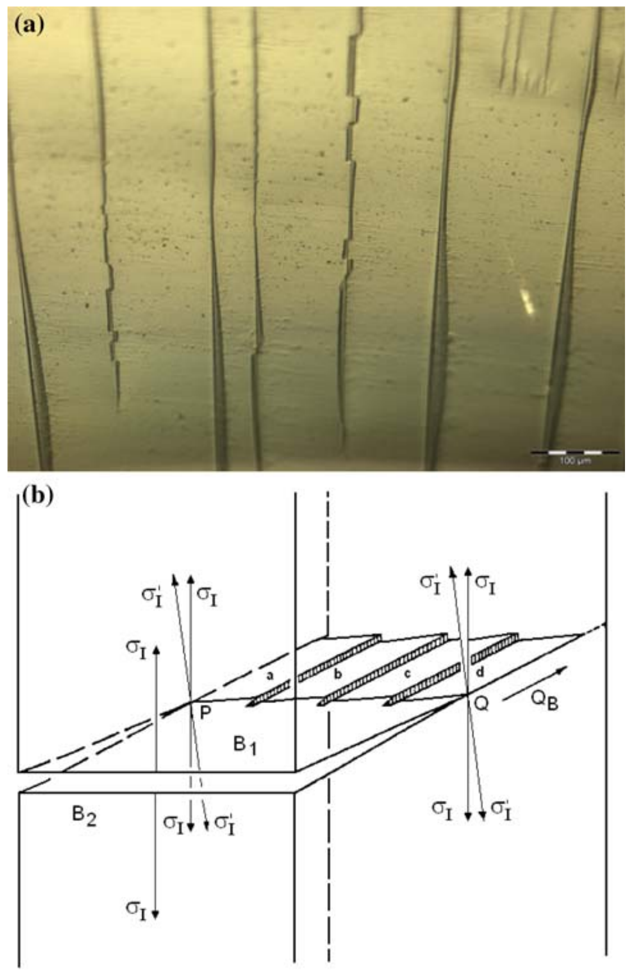

Fig. 6 **a** Fracture lances, formed during ball indentation of a soda-lime glass sample. *Scale bar*: 100 µm (Photograph: Andreas Momber). **b** Scheme of fracture lance formation in a brittle solid (adapted from Kerkhof [27])

"Hochfrequenzkinematographische Untersuchung der Bruchvorgänge in Glas" appeared, where the authors measured the fracture propagation velocities in various glass types [30]. They formulated the basic law that the fracture propagation speed approaches a constant terminal value, which is considered to be a characteristic material constant. Results of these measurements are provided in Table 2. Smekal performed an analysis of that work in the same issue of this journal [31]. He analyzed in detail the relationship between terminal fracture propagation speed and speed of sound in a material, and he showed that the speed of sound is the limiting terminal speed for crack propagation. Smekal made the assumption that the individual fracture at the macroscopic fracture front propagates from one flaw to the next if the molecular bond at the flaw tip is broken due to elastic stress. The elastic disturbances generated during this process must propagate with a velocity lower than the speed of sound in an undisturbed medium. The rather low fracture propagation velocity compared with the speed of sound is, therefore, a result of the existence, and the particular distribution, of microscopically small flaws in the material. This explanation was verified through the differences in the terminal fracture propagation velocity for mirror glass and quartz glass. These glass types had almost equal values for the sound of speed, whereas their fracture propagation velocities varied notably (see Table 2). However, this explanation was not supported by any thorough experimental result and remained ill-founded. Smekal also proposed that the terminal fracture propagation velocity in a given material shall be independent of the intensity of fracture excitement. This proposal was verified almost immediately through tests results obtained by Barstow and Edgerton in 1939 [32]. Also important to note is, that Smekal highlighted the role of secondary fractures which form if the terminal fracture speed is achieved. Notable contributions to the interpretation of the maximum fracture propagation velocity were made, among others, by Buehler et al. [33] and Fineberg et al. [34], whereby the first authors introduced a scenario that would allow for the exceedance of the speed of sound limit.

Table 2 Terminal fracture propagation velocities for different glass types [30]

Glass type (thickness)	Terminal fracture propagation velocity (m/s)
Window glass (2 mm)	1,520
Mirror glass (4–7 mm)	1,520–1,570
"Sekurit" glass/pre-stressed (5 mm)	1,550
2-Layer safety glass	1,500
Tank glass (20 mm)	1,420
Quartz glass (4 mm)	2,200

Although fracture propagation velocity reaches a material-specific *terminal value*, values for fracture propagation velocity varies during the propagation process. During the propagation process, the fracture propagation velocity is not a constant value, but it depends on loading rate, structure, temperature, etc. This phase—actually an acceleration phase—was referred to as *initial stage* by Smekal [35]. The graphical Wallner–Smekal method can be employed to estimate the velocity of a propagating fracture during that period [17, 19, 35] (see Sect 3). Some experimental results are plotted in Fig. 7. The results were obtained on glass rods with 6 mm in diameter, which were ruptured at a rate of 0.5 MPa/s. The fracture propagation velocities were estimated based on Wallner line analyses. Although the graph was published in 1950, the corresponding tests had already been performed in Halle in 1945. The original very large ("as large as a table") photographs of the fracture planes got lost

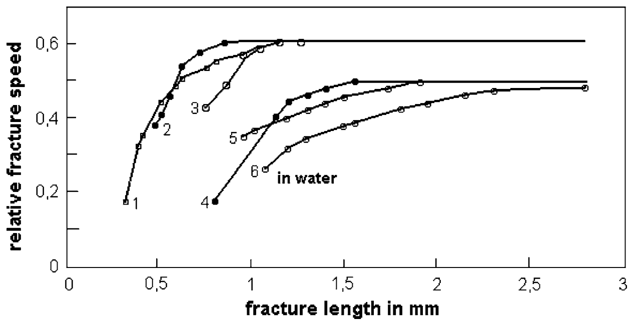

Fig. 7 Fracture propagation velocities in glasses as functions of loading rate and environment [19]. *1* Quartz glass 3.45 kg/mm^2, *2* quartz glass 4.47 kg/mm^3, *3* quartz glass 4.17 kg/mm^2, *4* optical glass 2.94 kg/mm^2, *5* optical glass 3.25 kg/mm^2, *6* optical glass 2.70 kg/mm^2 (in water)

due to "external circumstances", but downsized images (1/10th in size) could be saved. It can be seen from Fig. 7 that the fracture propagation velocity in the early fracture phase was affected by composition and environment. The terminal fracture velocities for the quartz glass (upper curves) were at about 60% of the speed of sound under all conditions, whereby the initial fracture propagation velocities notably depended on the load. The same relationship counts for the optical glass (lower curves). Here, the terminal fracture propagation velocities approached 50% of the speed of sound under all conditions, whereby the values initial fracture propagation velocities depended on load and surrounding medium (water for the very lowest curve). Smekal's approach to separate the fracture propagation process into an *initial stage* and a *final stage* was very important to understand the phenomenology of the fracture process. He also showed that thermal effects, as described in the previous section, act during the initial stage. The terminal fracture propagation speed as measured by Schardin and Struth [30], in contrast, is not affected by any thermal effects. However, Smekal's interpretation of the transition between initial stage and final stage is replaced by other conceptions which attribute the fracture propagation stages to environmental effects, mainly to effects of moisture [16, 36, 37].

Micro-plasticity

In the years between 1941 and 1943, Smekal and co-workers published a series of notes in the journal *Die Naturwissenschaften* about the fundamentals of glass polishing [38–41]. They reported about the appearance of fracture-free grooves on the surface of brittle materials (glass, quartz, corundum) if scratched at very low loads. This phenomenon was referred to as *athermal micro-plasticity*. It occurred only if structural non-homogeneities

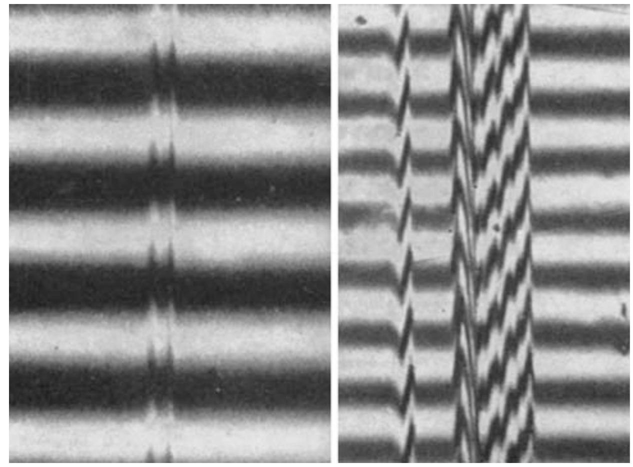

Fig. 8 Micro-plasticity in PMMA (interference microscope images, magnification 920:1, diamond tip 8 g). *Left* quartz glass with network structure and strong principal valency bond; *right* PMMA with chain structure and weak secondary valency bond [43]

did not contribute to the material failure. If they did, the surface showed localized fractures. As the load increased, the probability of fracture formation was found to increase. Two examples are provided in Fig. 8 based on interference microscope images. The fuzziness of the horizontal interference stripes is a sign of surface roughness, which in turn is caused by a high fracture propagation velocity. The sharp interference signals in the center of the images are caused by plastic deformation. This latter effect is more pronounced in the PMMA sample. Today, this phenomenon is referred to as *elastic–plastic transition*. Due to very accurate measurements, and due to the utilization of, at that time, advanced inspection methods, namely interference microscopy, phase-contrast microscopy, and electron microscopy, the very small quantities of displaced material could be quantified. In an analytical paper, the authors clarified the mechanisms of micro-scratching. They mentioned in particular that, if micro-plasticity appears, the molecular strength (chemical bonding strength) of a glass is locally reached and even exceeded [40]. The authors wrote: "*The true (molecular) strength limit can be made visible if the dimensions of the individual surface areas, loaded by compression, are small enough that no fracture initiation occurs, because there are not enough flaws in a sufficient density available*" [41]. It was concluded that the behavior of an ideal lattice was registered. Results of measurements of other authors, however, did not verify this particular conclusion [42]. The controlled mechanical scanning of sub-microscopically small sections of a surface with a small indenter, therefore, would allow for the measurement of cohesion forces. This cohesion force was named *ultra-hardness* by Smekal. This scanning method was first practically applied by Smekal and Klemm in 1951 for the assessment of silica glass (network structure) and

PMMA (chain structure) samples [43] (see Fig. 8). Moreover, amount and statistical distribution of microflaws (non-homogeneities) in a solid can be estimated based on the methodology developed by Smekal and his co-workers. This general view comes very close to the statistical approach published by Weibull a few years later.

Precise measurements on the walls formed along the crack-free grooves and theoretical calculations [44] have shown that they were the result of local melting and successive re-solidification [40]. This process allowed for the formation of a very thin pressurized layer on the material surface [45].

Fragmentation and comminution

In the 1930s and 1940s, Smekal was the chairman of the Working Group "Bruchphysik" (fracture physics) in the Reichsforschungsrat. He was also personally involved in the "Fachausschuss Verfahrenstechnik" (Department for Process Engineering) of the "Verein Deutscher Ingenieure" (Association of German Engineers) for many years. He saw his major contribution in providing a sound physical background to the practices of fragmentation and comminution.

The scratching tests reported in the previous section showed that the walls formed during the scratching at low loads exhibited a better solubility in acid [46]. This phenomenon could be attributed to *mechanical activation* of the structure due to the rather high amount of energy stored in the deformed material sections. This particular aspect of materials response is today an independent and important research field in process engineering [47].

In a series of papers in the late 1930s, Smekal developed a theoretical framework for the analysis and interpretation of comminution processes [48–50]. He reconsidered the results of his theoretical and experimental work on fracture physics (see Sects. 1, 2) and utilized the findings for relevant comminution problems. In analogy to his work on fracture physics, he distinguished between the *physical comminution* (see Eq. 1) and *technical comminution*, whereby the latter process is dependent on specimen geometry and loading situation [50]. Physical comminution is expressed by Eq. 1, which characterizes the minimum load required for a comminution process, whereby technical comminution requires an "overload". The relationship between these two loads (works) can represent the efficiency of a comminution process [49]. Smekal claimed that machines for coarse fragmentation of solids must deliver energy for no-load operation and for elastic deformation only, whereas fine comminution (grinding) machines must additionally deliver energy for grinding stock movement and for external and internal friction processes. The latter machines must, therefore, exhibit lower efficiency values. From a comparison of literature data available at that time, Smekal concluded that the Rittinger law of comminution is not a physical law, but rather an expression of the wastage of work in grinding machines. He called this regularity "*technical working law for comminution*" [48]. Smekal recognized that the specific work of comminution must depend on the specimen size, because fracture always originates from the most effective individual flaw [49, 50]. Larger specimens contain more flaws, and the probability that an effective flaw is activated is much higher. Today, this relationship statement is common sense. For rather low specimen sizes, the specific work of comminution must approach a certain saturation value, which is characterized by a statistical average flaw size. For very small specimens with dimensions lower than the average distance between two flaws, relative work of comminution can be expected to increase again [49]. Approximations for the efficiency of comminution processes delivered values of about 1% for physical comminution and about 0.1% for technical comminution [48, 49].

Theoretical considerations convinced Smekal that compressive comminution must be the most efficient method for the size reduction of at least glass [49], and he performed a number of experimental investigations into the response of glass to compressive comminution [51, 52]. Test on crystalline and glassy quartz particles in a roller frame revealed equal grain size and grain shape parameters of the crushed samples. This result has shown that anisotropy did not have an influence on the grindability of the materials [52]. These findings verified the fundamentals of the exponential grain size distribution law of Rosin, Rammler, and Sperling, that was developed just a few years ago [53]. They showed that the comminution of particles is based on probabilistically determined fracture processes. Finally, Smekal introduced a model for the formation of fines during the compressive comminution of brittle materials [51], which could very well describe results of experimental investigations. A view on the drawing printed in Smekal's paper (which is redrawn in Beke's book [54]) reveals signs of self similarity of the fragments. The model suggested that the Kick law of comminution must fail for non-homogeneous (compact-disperse) materials, which was in fact found for magnetite and cement mortar. For homogeneous materials, basalt and granite, Kick's law was found to apply.

Concluding remark

Certain, today classical, phenomena of materials science are associated with the name of Adolf Smekal; this includes flaw statistics, lancet fractures, mechanical activation, micro-plasticity, molecular fracture, Raman spectroscopy, and Wallner lines. Smekal has mainly influenced the work of German materials scientists and engineers, such as Hubert Schardin, Hans Rumpf, and Frank Kerkhof. Unfortunately, Smekal's publications appeared almost completely in German language, and his pretentious writing style made his papers not easy to read. This may be one reason that the work of Adolf Smekal is not sufficiently appreciated. Another reason may be that the experimental studies performed in Halle in order to support Smekal's fracture theory were not published under his name, but under the names of his students only (although they all gave personal credit to Smekal in all publications). Only very recently, English speaking materials scientists started to re-consider Smekal's pioneering work at least in some of his research areas. Quinn [55] for example, in his very readable essay on the history of the fractography of brittle materials, credited the work of Smekal and his co-workers. The 50th anniversary of Adolf Smekal's death seems to be an appropriate opportunity to think of the work of this outstanding scientist.

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