Chapter 4
Final Reflections

The book has been finished and the authors would like to thank in particular those tolerant and patient readers who have read it to the very end. One can be quite sure that such readers would not mind further brief reflection. In spite of the fact that these final remarks will bring no additional scientific information, the authors believe that selected pieces of knowledge that are assembled in the main text might be worth mentioning again. Because this book preferentially refers to the scientific work of the authors, only the results that arose from their own research will be highlighted. On the other hand, many important things still remain open for further investigations. Some of these tasks related to our research will also be recalled.

4.1 Useful Results

1. The highest achievable tensile strength of a solid of a particular chemical composition, the ideal tensile strength of the perfect crystal, strongly depends on the stress triaxiality. For the majority of metallic crystals, the ideal strength related to the volumetric instability increases almost linearly with increasing transverse biaxial stresses. On the other hand, ceramic crystals with a diamond structure exhibit a sharp maximum either close to the zero biaxial stress (Si, Ge) or in the compressive biaxial region (C).
2. Under uniaxial tension, however, perfect metallic crystals usually fail when reaching the first shear instability. Therefore, their uniaxial ideal strength σ_{iut} does not exceed 10 GPa which is far below their volumetric ideal uniaxial strength and only several times higher than that of the strongest related engineering metallic materials. This is not necessarily true for the case of isotropic (hydrostatic) tension in which the ideal strength σ_{iht} usually reaches values as high as several tens of GPa.

3. The ranking of tensile strengths σ_u of the strongest grades of engineering materials with chemically different matrices nearly follows that of σ_{iht} of related perfect crystals. The same is true for the ductile/brittle response of engineering materials and that of their perfect crystals. Therefore, these mechanical properties are, to a considerable extent, predetermined by those of a perfect lattice. On the other hand, such a correspondence is not apparent for the fracture strain and its dependence on the stress triaxiality. This property is determined instead by crystal defects and secondary phases.
4. Crystals of metals and diamond exhibit a nearly linear decrease in ideal shear strength with increasing superimposed isotropic (hydrostatic) stress σ_h. On the other hand, the dependence $\tau_{is}(\sigma_h)$ of covalent crystals Si, Ge and SiC reveals an opposite trend. These trends also refer to the dependence of ideal shear strength on the superimposed normal stress. This is in agreement with the normal stress influence found for the dislocation nucleation stress (non-Schmid behaviour). However, the ideal shear strength of Cu and Ni crystals becomes lowered by both tensile and compressive normal stresses.
5. By considering the coupling of shear and normal stresses, the values of σ_{iut} can be calculated from those of τ_{is}. This method, proven on a variety of fcc metals, avoids cumbersome examinations of the stability conditions necessary for a standard computation of σ_{iut}.
6. Nanoindentation appears to be the only efficient method for an experimental determination of ideal shear strength values. This method is based on a quantitative interpretation of the pop-in effect on the load–penetration diagram.
7. The geometrically-induced shielding of the crack tip due to crack kinking and branching can substantially increase fracture toughness values of brittle and quasi-brittle engineering materials. This is the reason why, even in the case of severely segregated grain boundaries of very low fracture energy, the fracture toughness of such defect containing steels can reach values higher than $50\,\mathrm{MPa\,m^{1/2}}$. This effect also elucidates the anomalous fracture behaviour of ultra-high-strength low-alloy steels when considering an influence of the size ratio d/r_p (the characteristic microstructural parameter/the plastic zone size). The size ratio is a very important parameter that couples micro (d) and macro (r_p) scales of fracture and fatigue processes. Decreasing size ratio means a decreasing influence of microstructure and thus a decreasing level of geometrical shielding.
8. Enhanced dislocation activity around growing voids gives a steep increase of the strain rate inside the neck during the tensile test of ductile metallic materials. The related model reasonably predicts the value of fracture strain when using the Brown–Emburry or percolation based void-coalescence models.
9. The fracture toughness of steels exhibiting a ductile fracture morphology can be reasonably predicted by means of a simple model based on the frac-

ture strain diagram. This diagram can be easily determined from tensile and compressive tests.

10. The ratcheting process in metallic materials starts when reaching a critical plastic strain range during cyclic softening. This threshold value decreases with increasing cyclic ratio. The initial ratcheting rate can be well described by a model based on discrete dislocation theory. This model explains the role of cyclic softening (hardening) in the acceleration (deceleration) of the ratcheting process.
11. Besides geometrically-induced shielding, an increasing size ratio also raises the effect of the roughness-induced crack closure in fatigue. A generalized model of crack tip shielding and closure under mode I loading, based on the concepts of discrete dislocations and size ratio, allows one to separate the intrinsic component from the measured fatigue threshold value. This model correctly predicts a decrease in roughness-induced crack closure with increasing range of applied stress intensity factor as well as the low fatigue threshold in nanostructured materials.
12. Micromechanisms of mode II and mode III crack propagation in metallic materials are generally different. Unlike for the mode II case, the commonly accepted idea of fatigue crack front advance due to an oxygen-assisted creation of new fracture surfaces is rather irrelevant for the mode III case. Therefore, mode II based micromechanisms of the remote mode III crack growth were proposed and experimentally verified for austenitic steel. In notched specimens, the mode I growth by formation of factory roofs (F-Rs) is a typical micromechanism under the remote mode III crack tip loading in the high-cycle fatigue region. These rather complicated micromechanisms are, most probably, the main reason for the low mode III crack growth rates observed in some steels.
13. In the high-cycle fatigue region, the size ratio is rather high and shear mode cracks incline (or branch) to the mode I loading case very easily. This is mainly caused by an increase in both the roughness-induced friction and the microstructurally affected tortuosity of the crack path. Moreover, there is a synergy of modes II and III by a formation of the mode I branch in terms of the crack driving force. The generalized conditions of mode I branching from the shear-mode crack propagation can be defined as follows: (1) the first branch forms at that site of the shear crack front where the value of ΔK_I on a facet of a potential mode I branch becomes maximal; (2) the branching appears at the moment when this maximal value exceeds that of the threshold ΔK_{Ith} for the applied cyclic ratio.
14. A textbook example of the branching behaviour is a formation of the factory roof from mode II+III semielliptical cracks on the surface. The relevancy of a quantitative model of F-R formation based on the synergy effect was proven by stereogrammetrical measurements of the three-dimensional topography of factory roofs. There are two main reasons why the formation of factory roofs is not observed in the low-cycle fatigue region. First, the size of factory-roof patterns decreases with increasing density of crack nu-

clei on the surface which becomes very high in the low-cycle fatigue region. Second, in this region a network of microcracks at microstructural heterogeneities is quickly produced inside a large plastic zone that is coplanar with the notch plane and embraces many characteristic microstructural elements (low size ratio). Coalescence of these microcracks and the main crack front keeps the crack path in their common (shear) plane.

15. Plasma nitriding can substantially raise not only the fatigue limit of mild steels but also that of high-strength steels under various kinds of loading such as push-pull, bending, torsion and combined bending-torsion. Both the improved hardness and the compressive residual stresses in the nitrided layer cause fish-eye cracks to initiate at inclusions inside the specimen and propagate in a near vacuum towards the surface. Both the level and the sign of residual stresses that are present in locations of fish-eye centres can be assessed by utilizing a method based on a determination of bending (or push-pull) stress amplitude corresponding to these centres on the fracture surface.
16. A rather simple method can be employed to reconstitute the applied stress, the cyclic ratio and the number of cycles spent for stable crack propagation during the fatigue process. This method couples quantitative fractography on both microscopic and macroscopic levels and might be very useful for the analysis of failure cases of engineering components and structures.

4.2 Open Tasks

17. Calculations of ideal strength should be corrected by thorough analyses of phonon spectra under various loadings and temperatures. This should finally explain the gaps between the strength of the strongest solids and their upper theoretical limits.
18. Verification of the simple method for recalculating the ideal tensile strength from the ideal shear-strength data of bcc metals and ceramics.
19. The synergy effect found in W-V and Mo-V nanocomposites should be verified for more composite components. A physically justified reason for this phenomenon might possibly be found by means of a precise analysis of stress coupling effects.
20. Nanoindentation experiments should be performed on as many single crystals as possible in order to establish an extended database of experimental values of ideal shear strengths.
21. The dependence of the fracture energy on the grain boundary concentration of phosphorus and other detrimental elements should be established by fracture experiments on well defined segregated bicrystal boundaries. This would finally solve the problem of the severity of individual impurities and set up physically justified impurity limits.

22. Correct measurements of crack growth rates and crack growth thresholds in mode II and mode III should be performed for an extended range of metallic materials. This would solve the practically important question concerning the resistance of materials to crack propagation under both kinds of shear mode loading.
23. The generalized conditions of mode I branching from shear-mode crack propagation should be verified for more metallic materials under various external shear loading modes.
24. Extended experiments on fish-eye crack initiation and growth should be realized for materials with reinforcing surface layers in both standard and gigacycle fatigue regimes. This would decisively complete our knowledge about micromechanisms of interior nucleation and propagation of fatigue cracks.

The authors hope that performing a majority of these tasks might be attempted in a rather near future.