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GENERAL METALLURGY

II. ELASTIC PROPERTIES

by

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The electric conduction, including superconductivity, thermal conductivity, magnetic properties, and optical properties, are treated in the FIAT Review "Physics of Solids".

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The investigations in the elastic range of solids are related predominantly to a determination of the modulus of elasticity of metals and alloys, the research on its relations to its constitutional and structural buildup, as well as the influence of chemical and mechanical treatment on the modulus, including a determination of the temperature dependence.

1. Temperature Dependence of the Modulus of Elasticity

The modulus of elasticity, as a function of the temperature, was measured for

31 pure metals from -180°C up to the melting point, ~~respectively to~~ or up to 1000°C (Bibl.1,2). The modulus of elasticity decreases with increasing temperature linearly or at a slightly accelerated pace. At the melting point, the modulus in about two thirds of the metals is 0.4 - 0.5 and, in several other metals, about 0.25 - 0.35 of the value at absolute zero; according to extrapolation, the modulus would have to reach a value of zero at the melting point in some metals. Several temperature curves of the modulus of elasticity coincide as soon as the modulus is related to that at absolute zero and the temperature is related to the melting ~~temperature~~ point (Bibl.3) (see Tables 1 and 2).

Polymorphic transformations are always accompanied by an abrupt change of the modulus. Minor admixtures may lead to a ~~jog~~ bulging in the temperature curve of the elasticity modulus toward lower temperatures. Corresponding observations cover primarily copper, silver, aluminum, and magnesium (Bibl.2). The formation of the ferromagnetic state (Bibl.1,4) leads to an increase in the modulus of elasticity within the temperature interval of the modulus, if the measurement is performed at magnetic saturation; the curve has a break at the Curie temperature.

The observations on the temperature dependence of the elasticity modulus of alloys confirm and extend the above data (Bibl.5,6,7). The occurrence of an ordered atomic arrangement manifests itself in the same manner as that of the ferromagnetic state (Bibl.8,9,10). The increase in energy of the lattice bonds leads in both cases to an increase in cohesion. Plotting of the modulus of elasticity

Table 1

Elastic Properties of Pure Metals

Metal	Compressibility $10^4 cm^2/kg$	Modulus of Compression kg/mm^2	Modulus of Elasticity kg/mm^2	Modulus of Shear kg/mm^2	Poisson Constant
Li	7.2	1390	1170	430	0.36
Na	12.0	832	910	340	0.32
K	24.6	406	360	130	0.35
Rb	52	190	240	-	0.3
Cs	70	140	175	-	0.3
Be	0.855	11700	29280	13500	0.08
Mg	2.952	3390	4515	1770	0.28
Ca	5.697	1750	2000	750	0.31
Sr	8.187	1220	1600	620	0.28
Ba	10.19	980	1290	500	0.28
Al	1.34	7460	7220	2720	0.34
La	3.513	2840	3820	1500	0.28
Ti	0.797	12500	10520	3870	0.36
Zr	1.097	9100	6970	2540	0.37
Hf	0.901	11100	8500	3100	0.37
Th	1.818	5500	7970	3160	0.26
V	0.609	16400	15000	5500	0.35
Nb	0.570	17500	16000	6000	0.35
Ta	0.479	20800	18820	7000	0.35
Cr	0.600	16600	19000	7300	0.31
Mo	0.347	28800	33630	12200	0.31
W	0.293	34000	41520	15140	0.30
U	0.966	10000	13380	4600	0.30
Mn	0.791	12600	20160	7800	0.24
Fe	0.587	17000	21550	8280	0.29
Co	0.539	18500	20380	7630	0.31
Ni	0.529	18900	19700	7500	0.32
Ma	0.34	29000	41500	16500	0.26
Re	0.27	37000	53000	21000	0.26
Ru	0.342	29000	44000	17600	0.25
Os	0.26	38000	57000	22800	0.25
Rh	0.372	27000	38640	15300	0.26
Ir	0.268	37300	53830	21400	0.26
Pd	0.528	19000	12360	4450	0.39
Pt	0.360	27800	17320	6220	0.39
Cu	0.719	13900	12500	4640	0.35
Ag	0.987	10100	8160	2940	0.37
Au	0.577	17300	7900	2820	0.42
Zn	1.66	6000	9400	3790	0.24
Cd	1.99	5000	6350	2460	0.29
Hg	3.4	2900	-	-	-
Ga	~2.0	~5000	1000	430	0.47
In	2.5	4000	1070	380	0.45
Tl	3.48	2900	810	280	0.45
Si	0.41	24400	11500	4050	0.42
Ge	1.41	7100	8000	3000	0.32
Sn	1.87	5300	5500	2060	0.33
Pb	2.37	4200	1600	570	0.44
Sb	1.8	5550	5600	2000	0.33
Bi	2.92	3400	3480	1310	0.33

Table 2

Mean Temperature Coefficient of the Modulus of Elasticity of Pure Metals

Metal	-180° to 0	0 to 100°	100° to 200°	200° to 300°	300° to 400°	400° to 500°	500° to 600°	600° to 700°	700° to 800°	800° to 900°	900° to 1000°
Be	4.1	4.1	4.1	4.2	4.4	4.7	4.9	5.8			
Th	3.4	3.4	3.4	3.4	3.5	3.9					
Ta	2.5	2.5	2.5	2.5	2.7	3.0					
Mo	4.3	4.3	4.4	4.7	4.9	5.2	5.5	6.0	6.6		
W	4.2	4.2	4.2	4.2	4.3	4.4	4.6	5.2	5.8	6.4	7.3
U	6.1	6.1									
Mn	5.5	6.3	7.5	8.0	8.8	9.9	-	10.0	11.5		
Fe	5.4	5.6	6.2	7.0	7.5	8.4	10.0	15.2	28	19.8	11.6
Co	4.0	5.2	ε		γ	6.6	9.1	9.8	9.8	9.8	
Ni	-	-	-	-	-	7.3	7.3	7.3	7.3		
Pd	1.3	1.3	1.3	1.3	1.4	1.8	3.4	8.9	12.8		
Pt	2.3	2.3	2.5	2.8	3.2	3.8	5.0	7.6	10.1	10.1	
Cu	3.8	3.9	4.5	5.2	6.3	7.6	8.6	8.9	9.0	9.0	
Ag	3.4	3.4	3.5	4.0	4.0	5.2	5.5	5.8	5.8	5.9	
Au	1.9	1.9	2.0	2.5	3.0	3.6	4.0	4.2	4.2	4.2	4.2
Mg	1.7	1.8	2.3	3.0	3.5	4.0	4.1				
Ca	0.5	0.5	0.7	1.0	1.5	2.6	1.9				
Ba	0.5	0.5	0.8	1.0	1.6						
Al	3.0	3.1	3.7	4.4	5.2	6.4	6.5				
La	0.9	1.0	1.1	1.3	1.7	3.0					
Zn	3.0	4.0	8.5	12.2	12.3						
Cd	3.3	6.9	18.5	8.5							
In	2.5	3.4	4.0								
Tl	0.6	0.6	0.8								
Sn	5.5	7.7	12.4								
Pb	1.8	1.9	2.2	2.8							
Sb	1.2	1.4	1.8	2.4	3.0	4.7	9.0				
Ti	6.4	6.4	6.4	6.4	6.4	6.4	6.4				
Zr	3.4	3.4	3.4	3.4	3.4	3.4	3.4				
Rh	10.5	10.5	10.5	10.5	10.5	10.5	10.5				
Ir	10.6	10.6	10.6	10.6	10.6	10.6	10.6				

as a function of temperature is a useful method for demonstrating an ordering process, specifically if such a process cannot be defined by roentgenographic examination [for example, Ni_3Fe (Bibl.11), Ni_3Mn (Bibl.12)].

2. Concentration Dependence of the Modulus of Elasticity

A first survey over the influence of alloying on the modulus of elasticity was obtained on about 70 alloy systems (Bibl.13,14).

In heterogeneous alloy series, the modulus of elasticity varies proportionally to the percentage of the structure components present. In continuous solid-solution series of low affinity, the modulus varies in accordance with the mixture rule. In all other cases, the modulus is located on curves, bent slightly downward or curved slightly or strongly upward. In limited solid-solution series, the modulus usually decreases. This decrease apparently is correlated with a melting-point depression, in which case each metal has approximately a specific value which is independent of the nature of the added alloying element (Bibl.3). The only known exceptions are several alloys of gold as well as alloys with lithium addition. Similarly, the absorption of hydrogen lowers the modulus of elasticity, as demonstrated by Jungnitz (Bibl.15) for the case of palladium-hydrogen alloys.

The modulus of elasticity of intermetallic phases is determined by the crystal structure and type of bond of the latter (Bibl.3,7). At equal chemical composition, the modulus becomes larger in the case of ordered atomic arrangement, i.e., when chemical bonding forces are effective, than in a statistical atomic arrangement; the crystal type with a lower symmetry and a larger elementary cell will have the higher modulus. Structures with large elementary cell and high coordination number usually have a high modulus. Each specific type of bond corresponds to a certain position of the modulus of elasticity in the concentration pattern, so that it is empirically possible to define whether a given crystal type has a high or low modulus. If the metallic bond is predominant, the modulus will largely obey the mixture rule. If a heteropolar or homopolar bonding component is superposed, the modulus will be proportionally increased. At a strongly homopolar bonding, as is the case for example in carbides, the modulus reaches relatively high values.

3. Influence of Mechanical and Thermal Treatment on the Modulus of Elasticity

Cold-working lowers the elasticity modulus. During storage at room temperature or after slight heating, the modulus increases again because of ~~the~~

crystal recovery, but reaches its initial value only after completed recrystallization. In the case of aluminum, the modulus of elasticity increases after an initial decrease at strong cold-working, presumably because of structure formation; in the recovery range, the modulus increases further and then drops again to the initial value at recrystallization. These annealing effects show clearly in the curves of the temperature dependence (Bibl.16,17) of the elasticity modulus of deformed metals.

Low- and high-temperature hardening will lead to minor changes in the modulus of elasticity. Similarly, the solution and segregation processes are reflected in the behavior of the modulus (Bibl.16). The individual steps during an ordered atomic arrangement, produced as a consequence of heat treatment, can also be well defined by measuring the modulus of elasticity (Bibl.8,9,11,17).

4. Influence of the Ferromagnetic State on the Modulus of Elasticity

In the ferromagnetic state, the elastic elongation is accompanied by an elongation due to magnetostriction. The resultant decrease in the elasticity modulus is known as the ΔE-effect. As demonstrated for nickel and iron-nickel alloys, this reduction in the modulus depends largely on the mechanical and thermal pretreatment of the material (Bibl.4,11).

The system iron-nickel contains alloys with technically important values of the temperature coefficient. Fahrenbrach and Meyer (Bibl.18) report on types of steel whose temperature coefficient of the elasticity modulus can be so adapted to the expansion coefficient that the requirements, made on springs for precision instruments, are met. By elimination of the ΔE-effect, due to a sufficiently strong cold-working, practical constancy of the elasticity modulus over a range of about 250°C can be obtained for an alloy with approximately 45% nickel (Bibl.19).

The study of the elastic behavior of nickel contributed to the determination of the quantity σ_1 of the magnetomechanical directional resistance, which plays an important role in the theory of ferromagnetism (Bibl.4).

5. Measurements on Single Crystals

Bender (Bibl.20) determined the main elasticity moduli of polycrystalline lithium and of single crystals of sodium and potassium at low temperatures. He calculated the values for quasi-isotropic polycrystals of the three metals, from the elastic constants. Goens (Bibl.21) investigated single crystals of copper, gold, lead, and aluminum in the temperature interval from -253 to +50°C, by the aid of flexural and torsional vibrations. Both authors compared their results with the quantum-mechanical calculations by Fuchs. The agreement between experiment and calculation was rather satisfactory for the metals sodium, potassium, and copper.

The temperature dependence of the elasticity modulus of single crystals of zinc, with different orientation, was defined by Reinacher and Scheil (Bibl.22).

6. General Considerations

Just like the modulus of shear and the modulus of compression, the modulus of elasticity varies periodically with the atomic number. The same statement applies to the transverse contraction number μ (Bibl.23). Metals in homologous series frequently have the same value of this number in several periods.

Attempts to correlate the elastic constants with the atomic volume and the melting point do not yield satisfactory results. A critical examination of the conventional approximat~~e formula~~ing function $k = T_s/V\chi$ (χ = compressibility, T_s = melting point, V = atomic volume), derived from the characteristic temperature, proves its untenability (Bibl.3).

Nevertheless, in individual cases - as demonstrated in Hume-Rothery phases - the values of the modulus of elasticity of equal structures may become proportional to the quotient of the melting point and atomic volume.

According to Biltz and Klemm, a simple correlation exists between the compressibility χ of the metals and the difference from atomic and ionic volume, the so-called "electron volume" V_E: The values of the metals are approximately located on a straight line when plotting log χ against log V_E. Koppe (Bibl.24), on the basis of simplified assumptions on the kinetic and exchange energy of the electron gas, indicated a way for the theoretical interpretation of this relation. In first approximation, calculation and experiment are in quite good agreement.

The question as to the derivation of the elastic constants of a crystal conglomerate from those of the single crystal, was discussed by Moeller (Bibl.25) on the example of iron. The two limiting assumptions for the calculation - assumption of equal deformation and assumption of equal stress of the crystallites - are excluded as being erroneous. Conclusions are drawn for the roentgenographic stress analysis.

The μ-values, calculated from the elasticity constants c_{12} and c_{44} of single crystals agree the better with the values obtained for polycrystalline quasi-isotropic specimens the lower the elastic anisotropy of a given metal. The magnitude of the ratio of both moduli, if not known, can be estimated from the periodic course of the above value (Bibl.23).

The quantities characterizing the structural elasticity, namely moduli of elasticity and shear and transverse contraction factor, are correlated with the lattice structure and are structure-dependent. The quantity characterizing the volume elasticity, namely the compression modulus, conversely varies continuously with the concentration. It is possible that each structural type of equal space lattice and equal type of bonding has the same value of transverse contraction factor (Bibl.3).

7. Determination of the Modulus of Elasticity

In a summary paper, Haupt (Bibl.26) reports on devices and methods for determining the modulus of elasticity by means of vibration processes. The basic principles of the methods are discussed and their applicability is evaluated.

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V. PLASTIC DEFORMATION AND RECRYSTALLIZATION

by

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A. PLASTIC DEFORMATION, SOLIDIFICATION, RECOVERY

1. Summary Reports

During the reporting period, two relatively extensive papers were published on crystal plasticity, by Burgers (Bibl.1) and Kochendoerfer (Bibl.2). Burgers' paper lists and discusses all reports published until the beginning of 1940, while Kochendoerfer makes ~~the~~ an attempt to give a compact theoretical presentation, with emphasis on mostly unpublished own work of the author and his coworkers. He also treats the transition from single crystals to polycrystals and gives a physical interpretation of technical characteristics such as fatigue limit, creep strength, and fatigue strength. Poeschl (Bibl.3) discusses all papers published until the end of 1940.

2. Homogeneous Deformation of Single Crystals

a), Experimental Investigations

Laue photographs of naphthalene crystals (Bibl.2) have shown that Laue asterism occurs only in elongation and upsetting, due to the necessarily resulting distortions of the slip lamellae, but not in pure shear deformation (shear slip according to Bausch) in which the slip planes slide over each other without interference. Consequently, such experiments are specifically useful for investigating the regularities of slip. After a general investigation of the experimental procedure, these experiments permitted an accurate determination of the effect of the rate of slip and thus an experimental construction of the equation of state for homogeneous deformations (Bibl.2).

Poeschl (Bibl.4) constructed a slip-resistance body which gives the required slip-shear stress in any direction; he also prepared a shear-stress body which, in a corresponding manner, characterizes the available shear stress. Slip starts in the direction in which the two bodies first come into contact.

b) Theory of Slip and Solidification

The theoretical investigations are based on earlier concepts by Becker, Orowan, Polanyi, Taylor, Burgers, and Kochendoerfer, and develop these concepts further.

1) True Slip Resistance

The assumption that the true slip resistance of a completely undisturbed (ideal) lattice (known as theoretical slip resistance) is several orders of magnitude larger than the observed slip resistance, known as the critical shear stress, and that therefore an actual crystal lattice must contain mechanically effective voids, is indirectly confirmed by an experimental determination of the true tear strength which, in a similar manner, exceeds the usually observed tear strength by several orders of magnitude. Smekal and coworkers found that, when scratching smooth surfaces of glass, quartz, or corundum with a thin needle, the true tear strength must be overcome since the ~~covered areas~~ areas involved are so small that they cannot possibly contain strength-reducing defects (Bibl.5). Energy analyses have shown that the evolved heat may occasionally lead to temporary melting (Bibl.5,6). Optical-microscopic and electron-microscopic examinations demonstrated that the scratch tracks were free of splinters or fractures (Bibl.7). Stranski (Bibl.8), in investigating isolated rock-salt crystals, first found a definite correlation between the tear strength and the diameter down to about 0.02 mm, where this strength has a value of 180 kg/mm^2 and thus practically approaches the theoretical value of 200 kg/mm^2. During the subsequent growth in supersaturated solution, this high strength disappears again. An energetic study, however, shows that the approach of the observed to the theoretical strength is not merely a surface effect but is due to inhomogeneities in the crystal interior.

2) Influence of the Mosaic Structure

Earlier experiments by Dehlinger and Gisen on the influence of the production process of crystals (from the melt or by recrystallization) demonstrated the effect of the roentgenographically determined mosaic structure (Darwin, Ewald, etc.) on the start of plastic deformation. Kochendoerfer (Bibl.2) was able to interpret this finding by a thermodynamical-statistical calculation and thus prove that the plastically effective voids are located along the mosaic boundaries.

Calculations have shown, in agreement with experimental findings, that an influence of the mosaic structure exists only below the normal critical shear stress, whereas no such influence can be detected above this critical value. Zehender and Kochendoerfer (Bibl.9) were able to demonstrate directly the difference in the mosaic size of cast and recrystallized crystals, from measuring the width of x-ray lines with a refined photographic technique* (Bibl.10) and give~~s~~ their values (about 5 × 10^{-4} cm) for the first time with an error of only 15%. The findings confirm and supplement optical-microscope examinations by Graf (Bibl.11) of growth phenomena on crystal faces.

3) Slip Solidification

The basic question as to the cause of slip solidification was tentatively solved by Kochendoerfer (Bibl.2) on the basis of its temperature and velocity dependence. All known experimental data agree with the assumption that the solidification ~~is~~ ~~represents~~ constitutes an effect of the dislocations, bound to the voids of the lattice. The objection by Masing (Bibl.12) that this effect has a unilateral orientation does not hold, since the effect is not caused by a shear stress opposite in direction to the external shear stress but rather by ~~an increase in~~ a rise of the potential barriers which must be overcome ~~in the~~ during formation and migration of the dislocations and which have an equally solidifying influence in both directions. On reversal of the deformation direction, however, an increase in shear stress first occurs because of the fact that the existing dislocations, may regress at a shear stress lower than required for the new formation of dislocations in the original direction of

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*The experimental setup and the interpretation conditions for a satisfactory individual determination of particle size and lattice distortions from the width of the x-ray lines, required for the mentioned determination of the mosaic size, are discussed in detail by Kochendoerfer and Dehlinger.

deformation. ~~Based on~~ These concepts permitted a qualitatively correct calculation of the experimental findings by Held (Bibl.13) ~~became possible~~, who had investigated alternately stressed crystals during shear deformation and ~~then~~ found that the alternate-solidification curves, in a temperature-dependent manner, have a less steep slope than the unidirectional solidification curves. Kochendoerfer (Bibl.14) was able to design a model which illustrates the solidifying effect of linked dislocations. A calculation of ~~this~~ the effect on this basis seems entirely possible and has actually been started.

Dehlinger and Kochendoerfer (Bibl.15) demonstrated that the mathematical description of an aperiodic XXXXXX proper motion of atomic series in the crystal lattice, given by Frenkel and Kontorova, represents a special case of migration of a dislocation and chain reactions in rapidly proceeding transformations. In accordance with the experimental findings, the excitation of these processes must take place thermally by individual atoms. It was mentioned that the long-wave vibrations, ~~oscillations~~, observed by Foerster and Scheil (Bibl.16) in transformations and twinnings, must be conceived as an intermediary forms between the mentioned aperiodic proper motions and the normal small thermal ~~oscillations.~~ vibrations.

3. Inhomogeneous Deformation of Single Crystals

a) Quantitative Experimental Investigations

Quantitative experimental investigations of the bending of single crystals were first ~~initially~~ made by Held (Bibl.17) and Loercher (Bibl.18). These investigations showed that the bending curve (bending moment as a function of flexure) has a similar break as the elongation curve at homogeneous deformation, i.e., that a critical bending moment exists after which extensive plastic bending takes place. This moment IX is 20 - 70% greater than the moment which would be expected in the case of purely elastic deformation, if the maximum slip shear stress would be exactly equal to the critical shear stress at homogeneous deformation (denoted in the following as elastic moment). It is impossible by experimental means to determine whether or not the maximum shear stress actually exceeds the homogeneous critical shear stress and whether the deformation in that case, at the critical bending moment, would no longer be elastic.

b) <u>Theoretical Analysis</u>

A theoretical analysis of the bending process, made by Kochendoerfer (Bibl.2), has shown that the second alternative is applicable: The plastic zone, in the case of the critical moment, has closely approached ~~to~~ the neutral fiber. The fact that the bending curve has any jog at all is a consequence of the finding that the plasticity limit ~~above~~ beyond the elastic moment rapidly shifts toward the neutral fiber and that the resultant increase in bending can no longer be measured under conventional experimental conditions.

This again is a consequence of the negligible velocity dependence of the critical shear stress. The special form of this dependence for the various materials shows clearly in the slope of the bending curves in the vicinity of the critical moment. These ratios were discussed in more detail by Loercher (Bibl.18). The latter author used an integral equation for calculating the velocity rate of the critical shear stress from the slope of the bending curve. This yields a mathematically simple correlation between the magnitude of the velocity effect and the formation energy of a dislocation in comparison to the mean thermal energy. The investigations, which promise a further clarification of the concepts on atomic slip processes, are being continued.

c) Equation of State for Inhomogeneous Deformations

The equation of state for inhomogeneous deformations of crystals was constructed by Kochendoerfer (Bibl.2) on the basis of bending tests. The equation, in addition to the slip solidification, contains a solidification component which characterizes the inhomogeneity of the deformation. This component is due to elastic distortions of the slip lamellae and is ~~denoted as~~ known as strain-hardening. The magnitude of this value for single crystals is insignificant in comparison with the slip solidification, but becomes of considerable influence in the case of polycrystals (see Section 4, c, 1)).

4. Deformation of Polycrystals

a) General Remarks

A summary survey, with extensive literature data on the internal processes at unidirectional and alternating stress, has been given by Thum and Petersen (Bibl.19). A justification for a characterization, deviating considerably from the conventional characterization of deformation and hardness curves, was attempted by Spaeth (Bibl.20). Spaeth (Bibl.21) also suggests to relate the elastic limit and the fatigue limit to fixed values of the ratio of permanent set to elastic deformation rather than to fixed values of the permanent set, as had been done previously. In addition, Spaeth (Bibl.22) mentions that studies of relaxation, i.e., the elastic aftereffect ~~under~~ at stress decrease, ~~at~~ and constant elongation, has numerous advantages for materials testing and should not be neglected.

b) Experimental Investigations

1) Testing Device~~s~~

Oehler (Bibl.23) described a tester known as ~~the~~ plasticometer, suitable for microscopic examinations and permitting a study of the structural changes under plastic deformation. As a practical example, the phenomena during deep-drawing were investigated.

2) Slope of the Elongation Curves

Barbier and Loehberg (Bibl.24) as well as Weisse (Bibl.25) investigated the velocity dependence of the strain curves of zinc (including alloys). At high velocities, a sudden load drop at the fatigue limit takes place, which is attributable to twinning. Erdmann-Jesnitzer and Hanemann (Bibl.26) observed that zinc behaves at low temperatures like aluminum and copper at room temperature, i.e., solidification takes place and the strong velocity influence is decreased. Von Rajakovics and Maier (Bibl.27), in experiments with aluminum alloys, obtained an increase in yield limit and tensile strength, approximately proportional to the logarithm of the velocity.

Koerber and Eichinger (Bibl.28) made comparative tensile and compression tests with steels and found that the strain and compression curves, when applying the values of load and deformation related to the respective cross section, agreed within the error of measurement. On alternating direction of loading, the well-known Bauschinger effect occurs.

3) X-Ray Determination of Incipient Flow

Various investigations concerned the problem of the ratio of the roentgenographically determined yield point to the stress-strain limit (Bibl.29,30,31). It was found that the yield point, even in the case of inhomogeneous external dimensional changes, coincides with the stress-strain limit. If the maximum stress at macroscopic initial flow is calculated from the external force, under the assumption that the deformation up to that point had been elastic, the case of inhomogeneous deformations will partly show considerable increases in comparison with the stress-strain limit, similar as found in the bending of single crystals (see Section 3,a). In some instances, this increase is still considered ~~real~~ to exist, and it is believed that a supporting effect of the less stressed areas is involved here. X-ray tests show that this concept is erroneous, a fact mentioned as early as 1936 by Rinagl (see Section 3,b). The experiments showed, in addition, that for cases of deformation beyond the roentgenographic initial flow, the stress at constant deformation does not retain the value of the stress-strain limit but decreases approximately linearly by the same amount as it would have increased on further elastic deformation. These findings are explained by the easier deformability of a surface layer, a concept which seems logical in view of the fact that in one direction, the grains in the surface layer are not hindered by neighboring grains. ~~in one direction~~. Dehlinger, within the scope of his theory on the fatigue limit (see Section 6,b), has developed the same concept.

c) Theory of Deformation of Polycrystals

1) Calculation of Strain Curves

Based on an analysis of the experimental material, Kochendoerfer (Bibl.2), using the obtained equation of state (see Section 3,c), was able to construct the mathematical relation for the slope of the strain curves, with which this slope can be calculated from single-crystal data. A polycrystalline strain curve, accordingly, represents the sum of the mean single-crystal strain curves and the strain-hardening curve which is obtained as a consequence of the elastic lattice distortions occurring during deformation (see Section 3,c).

In highly symmetrical cubic metals the strain-hardening does not increase further after a certain critical stress, since a complete plastic deformability becomes possible due to the activity of several slip planes. Below this critical stress, the strain-hardening increases at a similar rate as the elastic stress, despite the fact that local plastic deformations occur which, however, are connected with lattice distortions. This critical stress determines the yield limit which thus has a definite physical meaning for these metals. Above the critical stress, the polycrystal curve, because of the constant strain-hardening, runs parallel to the mean single-crystal curve. In the case of hexagonal metals, however, with only one slip plane, the strain-hardening increases continuously and no physically definable yield limit exists. The strain-hardening results in ~~the~~ a strong stress increase and ~~the~~ a low elongation at break of such metals unless, as in the case of zinc, the slip plane is constantly returned into favorable positions by twinning.

The strain-hardening component of the yield limit, in comparison with the single-crystal component, is considerably greater in cubic face-centered metals than in cubic body-centered metals. In agreement with practical experience (Bibl.32,33), this fact results in a considerable temperature dependence of the yield limit in the case of the latter metals, similar to that of the critical shear stress in single crystals, whereas only a slight temperature dependence occurs for the former metals at low temperatures, followed by a stronger decrease in the ~~range of~~ recrystallization range.

Dehlinger and Kochendoerfer (Bibl.34) demonstrated that the zone of lattice distortions in the vicinity of the grain boundaries can be made proportional to the angle ~~under~~ at which the slip planes of the grains abut. The contribution of the strain-hardening to the yield limit can then be conceived as mean value of the yield limits of the grain-boundary zones, whereas the entire yield limit can be conceived as the sum of the mean values of all yield limits of the grains considered as free single crystals and of the grain-boundary zones, thus covering the reciprocal effect of the grains. This will yield a satisfactory interpretation of the 1:1.15 ratio of the tensile and torsion limit which earlier had been given by Sachs in a numerically correct form but incomplete because he only considered the single-crystal components.

2) <u>General Flow Condition</u>

In continuation of the above considerations, Dchlinger (Bibl.35) proved that the hypothesis, used in technological mechanics, according to which the incipient plastic flow is characterized by a fixed value of morphological-change energy, represents the flow condition ~~which is~~ obtained from the validity of the shear-stress law by Schmid for ~~the~~ individual grain, by forming the arithmetic mean over all grains of an irregularly oriented polycrystal.

3) <u>Energy Ratios</u>

Under the assumption that the work done against the strain-hardening is consumed completely for the formation of additional elastic lattice distortions, a relation derived by Dehlinger (Bibl.36) yields distortions which are about ten times as great as calculated from the broadening of the x-ray lines or, conversely, the stress calculated from the latter is only one tenth of the measured strain-hardening. Dehlinger concludes from this that the stresses which have a broadening effect on the x-ray lines, extend over the entire grains and do not have a noticeably solidifying effect, whereas the solidifying stresses are localized in the neighborhood of the grain boundaries and have no noticeable broadening effect.

Masing (Bibl.37), however, mentioned that, according to experimental data, the work done against strain-hardening is predominantly converted into heat which causes the discrepancy between measured and calculated stress data to disappear. Kochendoerfer (Bibl.38) corrected the equation of state for the deformation of polycrystals, corresponding in its published form to Dehlinger's concepts, insofar as the strain-hardening no longer represents the total but only the partial change in energy of the lattice distortions with deformation; this also yields agreement with experiment, relative to the energy ratios in the sense of Masing's remarks. However, it was found that the solidifying lattice distortions, in accordance with Dehlinger's concept, may nevertheless be considerably greater in small areas near the grain boundaries than the roentgenographically measured distortions.

4) <u>Tensile Strength and Work of Deformation</u>

Using the equation of state for polycrystals, Kochendoerfer (Bibl.39) calculated the tensile strength of metals and their temperature and velocity dependence from single-crystal data. The theoretical results coincide wholly with the experimental findings by Nadai and Manjoine (1941) on aluminum and copper, extending from room temperature up to the melting point and encompassing a velocity range of 1:10^4. In addition to the tensile strength, the work of deformation up to incipient failure was calculated as a function of temperature and velocity, from single-crystal data (Bibl.40). The temperature curves for various velocities, in first approximation, are mutually shifted by an amount proportional to the logarithm of the velocity, i.e., the character of a temperature dependence is quite independent of the velocities. Specifically, the transition from plastic to brittle failure, occurring at decreasing temperature, does not represent a characteristic of a rapidly occurring (abrupt) stress as such, but ~~occurs~~ take place at low temperatures even at loads with normal velocity, a fact previously mentioned by Mailaender (Bibl.41). The further theoretical results agree well with the existing experimental data.

5. Deformed State (Natural Strains, Recovery)

a) Classification and Measurement of Natural Strains

On removal of load from an inhomogeneous deformed single crystal or from an arbitrarily deformed polycrystal, the internal lattice distortions do not return to zero, since the required return could take place only under the effect of external forces, which, except for the sign, coincide with the original ones (Bibl.2). The so-called natural strain or inherent strain, corresponding to the residual distortions, was classified by Dehlinger (Bibl.36) on the basis of the classification by Masing and Sachs in accordance with dynamic viewpoints (solidifying effect) and in accordance with its effect on the x-ray lines (shift, broadening).

The findings obtained in x-ray stress analyses by line shift, covering stresses of the first kind*, showed that even in externally homogeneous deformed polycrystals natural strains of the first kind occur (Bibl.29,31,43) a fact which can be explained by the easier deformability of a surface layer, mentioned ~~already~~ previously in Section 4,b,3. In bent steels, the marginal fibers show inherent strains which have exactly ~~have~~ the ~~magnitude~~ extent by which the stress under load had dropped below the value of the yield limit (cf. Section 4,b,3); this was contrary to the expectation that the value would be twice as high. Dehlinger (Bibl.44)

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*The investigation methods are described by Moeller and others (Bibl.42).

was able to explain this phenomenon quantitatively under the assumption that the outermost layer, after unloading, flows back by the same amount as it had flown in the opposite direction under load.

The rapidly varying stresses of the second and third kind can be calculated from the broadening or intensity variation of the x-ray lines (Bibl.10,45) and from the variation of various magnetic quantities (Bibl.46-50). Measurements of line broadening on freely drawn and rolled copper and silver specimens (Bibl.10), at almost constant particle size of about 5×10^{-6} cm, showed an approximately parabolic increase of the lattice distortions with the degree of rolling, up to about 0.2%; this value corresponds to a mean stress of 45 kg/mm^2.

By means of electron interferences, Kranert and Raether (Bibl.51) were able to demonstrate that the particle size during surface treatment may drop to less than 10 Å in a thin surface layer of the order of magnitude of 50 Å, which cannot be individually detected by x-rays. This state must already be considered as amorphous. Glocker and Richter (Bibl.52) also used electron interference studies for determining that amorphous layers may occur in the polishing of metal surfaces. Smekal (Bibl.6) and Klemm (Bibl.53) assumed that the fracture-free substance shifts, found in scratch tests and connected with local melting processes (cf. Section 2,b,1), occur also during polishing, which would make the formation of an amorphous surface layer plausible.

Intensity measurements of the x-ray lines, made by Dehlinger and Schaecke (Bibl.54), gave lattice distortions of about 2% for a 50% degree of deformation. Various reasons speak for the fact that these distortions are actually dislocations which are produced during slip. This assumption ~~could possibly~~ might be confirmed by measurements on single crystals, which are actually scheduled.

b) <u>Recovery</u>

Making use of the above method for determining lattice distortions and performing measurements of mechanical characteristics, the time rate of change of state, due to recovery, was investigated. Mueller (Bibl.48) concluded from the influence on electric and magnetic quantities that the recovery preferably involves the highly disperse stresses of the third kind, an assumption which is also substantiated by the disappearance of the fog in x-ray photographs.

According to Kochendoerfer (Bibl.2), this finding was to be theoretically expected. The further theoretical expectation that, in the zone of recovery, the loss in strength would tend toward one and the same final value at any temperature, although very slowly at lower temperatures, was considered quite probable by Gerlach and Hartnagel (Bibl.46) on the basis of magnetic tests results. Several investigators (Bibl.55,56) confirmed that the recovery represents a timewise continuous process through which the strength and hardness values of polycrystals do not reach the initial values of the undeformed material.

Masing (Bibl.57) made investigations on the possible statements obtainable as to energetic ratios during recovery from the finding that the recovery temperature curves do not overlap for different initial solidification degrees.(Bibl.56,58). Unequivocal conclusions from this finding alone are impossible; a quantitative analysis of earlier results obtained by Kornfeld, however, makes it probable that the activation energy of recovery presumably depends on the present values of solidification but not on the previous values.

6. Endurance Limit at Unidirectional and Alternating Load (Creep Strength and Fatigue Strength, Elastic Limit)

a) Experimental Investigations

The experimental investigations were mainly concerned with the special conditions for certain materials or with questions on the determination of dependable characteristics for the endurance limit in short-time tests. Discussion of these investigations goes beyond the scope of this paper. Of general importance are recent findings that the fatigue strength, within wide limits, is independent (Bibl.59) of frequency and temperature (at sufficiently low temperatures), whereas it depends strongly on the specimen diameter at inhomogeneous stresses, specifically at bending stresses (Bibl.60). Von Philipp and Buchmann (Bibl.61) found that the bending strength first decreases rapidly and then (with increasing diameter) more slowly, approaching asymptotically the tensile-compression fatigue strength.

At elevated temperatures, the fatigue strength, determined by the Woehler curve, with the character of failure at preliminary load, is no longer useful since the unidirectionally acting preliminary load results in impermissibly great elongations (Bibl.62,63). In this case, the permissible fatigue limit assumes the character of a yield point. The maximum permissible preliminary load is equal to the creep strength at which, generally, a superposed alternating amplitude is no longer permissible (Bibl.63) but is still tolerated in special cases up to a certain magnitude (Bibl.64). Since the creep strength, in contrast to the fatigue strength, depends greatly on the temperature, a certain temperature exists (known as the reversal temperature) below which the fatigue strength preferentially has

~~a~~ the character of a failure and above which preferentially the character of a yield point (Bibl.65).

Observations on the variation of x-ray reflections showed that the yield strength represents a sharply defined deformation limit (Bibl.66). Below this limit, only a few reflections are insignificantly broadened; above this limit all reflections become blurred to an increasing degree. In a similar manner, this situation is demonstrated by the fact that, below the fatigue strength, the roentgenographically measured stress ~~reaches~~ rises to the value computed from the external stress at a certain load alternation and then maintains this value; above this ~~yield~~ fatigue strength, however, the progressive plastic deformations cause this stress to remain below the above value after a certain load alternation and decrease continually until actual failure (Bibl.67).

b) Theory of Creep Strength and Fatigue Strength

By evaluating the above-mentioned experimental findings, Dehlinger (Bibl.68) was able to establish a compact theory for creep strength and fatigue strength. As basis, he used the consideration that, in a loaded material, a thermodynamic (i.e. arbitrarily ~~long-existing)~~ protracted) equilibrium can ~~form~~ be established only if the external force is counterbalanced by an internal counterforce of elastic lattice distortions. All solidifying effects by slip, hardening, etc. are without significance here since, during the course of time, they decay because of thermal agitation of the atoms. Consequently, the greatest internal forces compatible with the external conditions, defines the limit at which a primarily occurring plastic deformation will just be permanently stopped. This is considered the true creep limit and is identical with the true creep strength and fatigue strength. Its experimental determination in a continuous-load test is impossible since there is no time to wait for elimination of the mentioned solidification effects and since the sensitivity of the testing instruments has an upper limit. Conversely, in alternating-load tests a dissipation mechanism, whose details are not fully clarified, will cause a decay of these effects because of the alternating stresses, within a relatively short time (Bibl.13). The stress in the horizontal segment of the Woehler curve, consequently, simultaneously represents the practical and true fatigue strength and the true creep limit. The practical creep strength, however, XXXXX represents a limit to be defined from case to case which, depending on the type of material, may be in quite different relations to the true creep strength (Bibl.2). The same statement applies to the practical fatigue strength above the recrystallization point where the true creep strength is zero, since the recrystallization processes eliminate the lattice distortions. The conclusions drawn from this theory agree well with the above-indicated experimental findings.

Poeschl (Bibl.69) developed a theory of fatigue strength, under the assumption that a resistance proportional to the velocity will ~~occur~~ be set up in the material, shifted in phase with respect to the external forces. The fatigue strength is defined by a fixed value of the resultant heat of damping. In this manner, the observed slope of the Woehler curve is predominantly obtained. However, Dehlinger (Bibl.68) and Kochendoerfer (Bibl.2) mentioned that a correct reproduction of the Woehler curve is possible in various manners by using a fixed value of a suitable quantity and, in itself, does not represent ~~a sufficient~~ an adequate criterion for the correctness of the theory.

c) The Question of the Existence of an Elastic Limit

The question as to the existence of an elastic limit is correlated closely with the problem of the true creep limit. Like the latter, the question cannot be basically decided by experimental means (Bibl.70). Theoretically, the concepts discussed in Sections 2,b,2 and 6,b, lead to the result that an elastic limit completely free of hysteresis and aftereffects is nonexistent since, even in the presence of a true creep limit, dislocations are formed, i.e., irreversible processes occur; in addition to this, disordered individual motions of atoms along the grain boundaries occur in polycrystals (Bibl.71). If it is required that each stress is to be coordinated with a uniquely defined deformation, even if the latter takes place only after a certain time, an elastic limit will exist; in the case of single crystals, this is identical with the true creep limit and, in the case of polycrystals, with the creep limit of the grain ~~with~~ having the optimum orientation. The limit is different from zero only for crystals with a well-defined mosaic structure (Bibl.2). It could be assumed at first that the range within which the damping is independent of the amplitude (Bibl.72), represents the elastic range in the above sense. Nevertheless, it is possible that the rapidly alternating load prevents plastic deformations from occurring, such as could take place on long-time unidirectional loadings.

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B. RECRYSTALLIZATION

1. Recrystallization in Worked Bodies

a) Summary Report, Denotations

In 1941, a detailed report on this particular field was published by Burgers (Bibl.1). The paper contains an extensive list of all articles published until 1940. The entire experimental material of general significance is clearly arranged, and the theoretical concepts are presented in a concise form.

The denotations in the literature for a given process differ widely. We are using here the denotations introduced by Burgers. Primary recrystallization: microscopically or roentgenographically detectable occurrence of new crystallites in the deformed configuration (nuclei formation) and growth of the nuclei up to mutual contact (nuclei growth). Secondary recrystallization: distinct growth of a few crystallites in the primarily recrystallized configuration. Grain enlargement: approximately uniform coarsening of the primarily recrystallized configuration. Selective recrystallization: recrystallization in unworked bodies.

b) <u>Recrystallization Diagrams</u>

During the reporting period, the following recrystallization diagrams were constructed: Mg (Bibl.2); Mg-Mn (2%) (Bibl.3,4); Mg-Mn(~~2%~~)-Ce (0.5%) (Bibl.3); Mg-Al (Bibl.5); Zn(99.99%) and Zn-Cu(0.1 - 3%) (Bibl.6); Fe-Ni-Cr; Fe-Ni-Cr-Ti; Fe-Ni-Cr-Co-W steels (Bibl.7).

c) <u>Recrystallization Point and Recrystallization Time</u>

Bungardt and Osswald (Bibl.8) investigated technical Al-Cu-Mg alloys and the same type alloys melted from pure substances and found a distinct lowering of the temperatures of incipient and completed recrystallization (recrystallization interval*)~~XXXXX~~ produced by ~~soft~~ dead-annealing before rolling. In the case of ~~weak~~ minor deformations, the temperature of completed recrystallization shifts with decreasing

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*These concepts, naturally, are useful only when based on a certain recrystallization time. This time differs from case to case, for which reason the results of different authors cannot be directly compared (Bibl.1). It should be specifically considered that the rate of recrystallization strongly increases with the degree of rolling, the temperature, and the degree of purity (Bibl.10,11).

degree of deformation toward increasing temperatures and presumably reaches the melting point in the limiting case of vanishing deformation. With increasing degree of rolling, both temperatures decrease, while the interval width remains constant or decreases slightly. Conversely, VosGuehler (Bibl.12) observed a linear increase of the interval with the degree of rolling for the case of an Al-Mg alloy, since the temperature at the beginning of recrystallization decreases more strongly than at the end. With increasing annealing time, the interval width first increases but then remains constant. A large interval width, such as exists in ~~existing in~~ cold-rolling, may be absent in hot-rolling, as determined by Jan and Hofmann (Bibl.3) on a Mg-Mn alloy. According to Masing and Wallbaum (Bibl.13), the ~~fear~~ possibility ~~as to~~ of harmful recrystallization in the usual zinc alloys ~~are unfounded~~ does not exist, since the recrystallization either starts ~~only~~ at high temperatures only or else ~~or~~ results in a fine grain and then becomes harmless. Coarse-grain formation was observed in only one case.

Several investigations were made as to the influence of impurities and additives on the temperature level and width of the recrystallization interval. Manganese increases the temperature level of aluminum and of alloys of the Al-Mg and Al-Cu-Mg type (Bibl.14,15,16) whereas Mg, Si, and Fe lower the level, in which case a complete compensation of the increasing effect of manganese may take place (Bibl.14). An addition of copper to aluminum has a nonuniform effect (Bibl.8), whereas an addition of copper to zinc will shift the incipient recrystallization toward higher temperatures (Bibl.6). Cerium will increase the temperature level of the interval of Mg-~~Mo~~ Mn at decreasing width (Bibl.3). Additions of vanadium and chromium to Al-Zn-Mg alloys, at low degrees of rolling, will result in a strong increase of the temperature level of the interval, whereas their effect at higher degrees of rolling is negligible (Bibl.17).

In technical Al-Cu-Mg alloys and in the same type alloys melted from pure substances, cold-tempering without hot-tempering has a minor influence on the recrystallization interval (Bibl.8). According to observations by Mueller (Bibl.18), however, the recrystallization in the case of segregatable Fe-Ni-Cu alloys takes place only at temperatures at which the heterogeneous components have almost gone into solution. Near ~~In the vicinity of~~ 40% Fe, 40% Ni, and 20% Cu, an anomal~~ies~~ous recrystallization takes place close to ~~already near~~ 600°C (i.e.,300°C below the normal recrystallization point), under formation of a characteristic cubic structure. Between 700° and 900°C, no recrystallization takes place, while normal recrystallization occurs above 900°C. These phenomena are investigated by examining the variations of the electrical, magnetic, and mechanical properties. For their interpretation it is assumed that the heterogeneous segregations, at a temperature of 600°C, are surrounded on all sides by the recrystallization nuclei, whereas they coagulate between 700° and 900°C and thus prevent recrystallization; only at higher temperatures, when these segregations have again gone into solution, will a new recrystallization become possible.

d) <u>Influence of Recrystallization on the Mechanical, Electrical, and Magnetic Properties</u>

The generally unsteady character of recrystallization (see Section e,1), in comparison with the steady character of recovery, manifests itself also in a usually ~~generally~~ strong fluctuation of the mentioned properties at ~~with~~ incipient recrystallization, tending ~~in a direction~~ toward the initial values which existed before working; after completed recrystallization, these values may ~~can XXXX be~~ actually be attained. ~~reached.~~ Generally, a grain enlargement is accompanied by an minor change in these values. Consequently, for the beginning of recrystallization the instantaneous values are of less importance than their changes with time and temperature. Various investigations were concerned with these problems (Bibl.10,12,19,30). It has been proved that, after recrystallization at high temperatures, which proceeds at great velocity, lattice disturbances still exist. For example, Mueller (Bibl.11) found that nickel, after recrystallization at 900°C, showed a strong increase in density and ~~in electric~~ resistivity which ~~were~~ was eliminated by subsequent cold-rolling. Presumably, gases trapped in the material play a certain role which, at the high temperatures involved, may lead to void formation in the structure (Bibl.21). From measurements of electric and magnetic quantities it can be concluded that recrystallization at low temperatures will only partly remove the lattice distortions produced by working of the material, prefer~~ably~~entially the rapidly varying types.(Bibl.11).

According to Dreyer and Seemann (Bibl.16), the mechanical properties of specimens of Al-Cu-Mg alloys, which were only molded, differ from the properties ~~of~~ specimens which were both molded and afterdrawn. It was found that this molding effect is the lower (and, in the critical case, disappears entirely) the more complete the recrystallization had been at the solution annealing temperature from which quenching took place.

Recrystallization also has an influence on the plastic flow insofar as, during ~~the~~ recrystallization, the crystallographically determinate slip processes are accompanied by crystallographically indeterminate individual agitation of atoms (so-called amorphous or position-change plasticity). This furnishes an explanation for the observation that strongly deformed specimens, in the temperature range of recrystallization at equal load, show a greater flow than undeformed specimens (Bibl.22) and that fine-grain specimens show a greater flow at elevated temperatures than coarse-grain specimens (Bibl.23).

e) Fundamental Processes in Recrystallization

1) Nuclei Formation and Nuclei Growth

Bungardt and Osswald (Bibl.24) made x-ray investigations to demonstrate that the recrystallization, at medium and high degrees of deformation, is a process taking place at unsteady nuclei formation but that, at low degrees of deformation, a steady sequence of intermediate states of lower stability occurs before build-up of undisturbed lattice elements; this had earlier been proved by Carpenter and Elam on aluminum. Masing (Bibl.25) mentioned that the observed changes of the x-ray reflections do not necessarily imply such a steady nuclei formation but that such a process is highly probable. In experiments with tin, Masing and Long (Bibl.26) demonstrated that the primary process of recrystallization consists in the formation of stable nuclei which are formed in the zones of greatest lattice irregularities and grow until mutual contact, at the expense of their still disturbed surrounding. Dehlinger (Bibl.27) developed a theoretical pattern of recrystallization processes, on an atomistic basis. He starts from the premise that the distortions of entire lattice sectors, which are always present in deformed polycrystals, cannot be canceled by individual motions of the atoms as is the case for dislocations produced on slip during recovery, but can be canceled only by simultaneous motions of several atoms which, in part, proceed as chain processes. For thermodynamic-static reasons, such atomic motion takes place at points of maximum lattice distortions. This causes a relaxation of the less distorted points, directly adjacent to the above points, resulting in the formation of viable nuclei whose position thus determines the orientation multiplicity of the recrystallization structure. Consequently, in contrast to a frequently held opinion, the first atomic motions, in analogy to the nuclei formation in gaseous phases, do not ~~lead~~ directly produce ~~to the formation~~ nuclei but merely have an initiating ~~character~~ effect. The conclusions from this theory agree well with the general experimental data, specifically with respect to the correlation between the orientation multiplicity of the deformed and of the recrystallized structure.

2) Secondary Recrystallization

Masing and Long (Bibl.26) made investigations on zinc and aluminum to demonstrate that the secondary recrystallization, like the primary type, starts by nuclei formation and that, consequently, they are identical in nature. According to experiments by Mueller (Bibl.11) with nickel, the secondary recrystallization is initiated by lattice distortions which had not been eliminated during the preceding primary recrystallization at lower temperatures. By a gradual increase in temperature after primary recrystallization, however, a slowly ~~proceeding~~ progressing grain enlargement may take place, which results in a distortion-free lattice so that no secondary recrystallization can take place. According to Masing and Staunau (Bibl.6), the beginning of secondary recrystallization in the case of zinc, ~~shifts~~ at higher degrees of rolling shifts toward lower temperatures. On addition of copper, it is impossible to make a clear differentiation between primary and secondary recrystallization. The linear rate of growth of secondary crystals, in first approximation, is independent of the time but strongly dependent on the rate of heating within the ~~range of~~ secondary recrystallization range.

Masing and Long (Bibl.26) and Schmidt (Bibl.28) ~~both~~ came to the result that the so-called forced secondary recrystallization merely represents a normal recrystallization after weak deformation and that, therefore, this concept should not be used. An accurate interpretation of the conditions is still lacking in the case observed by Mueller (Bibl.29) on a Fe-Ni alloy where large secondary crystals were detected at the point of fracture when the specimens, after rolling, were fractured on one side and then recrystallized.

Schoeneck and Verleger (Bibl.30) gave a summary report on the production of recrystallization single crystals.

3) Grain Size and Grain Enlargement

According to Masing and Staunau (Bibl.6), the grain size in zinc and Zn-Cu alloys ~~decreases~~ is reduced during primary recrystallization ~~with~~ at decreasing size of the initial grain, and is strongly decreased by copper additions in the zone of solid-solution formation. The amount of segregated components, as also found by Bollenrath and Bungardt (Bibl.31) on Al-Mg alloys, has no influence on the grain size. However, the rate of heating (Bibl.5,31) has a certain influence, and may lead to the anomalies at high and low values.

The tendency toward coarse-grain formation due to grain enlargement, present after minor deformation (critical zone) in Al-Mg and Al-Cu-Mg alloys, can be eliminated in various manners: by a low degree of rolling before the final annealing, which yields a coarse initial grain (Bibl.32); by the addition of manganese (Bibl.14,15,31); or, according to Dreyer and Hansen (Bibl.33), by a suitable pre-annealing independent of the final degree of rolling (see also Bibl.5). Temperature and duration of this pre-tempering must be so selected that the previously deformed structure is just able to recrystallize. The optimum temperature range is shifted with increasing manganese content toward higher temperatures and, with increasing final degree of rolling, toward slightly lower temperatures. In this manner, a structure is obtained in which no coarse-grain formation occurs even after repeated critical deformation and recrystallization. Similarly, in other types of alloy suitable additives will reduce the tendency toward coarse-grain formation; for example, an addition of cerium to Mg-Mn alloys (Bibl.3), vanadium and chromium to Al-Zn-Mg alloys (Bibl.17). According to Bungardt and Osswald (Bibl.8), technical Al-Cu-Mg alloys have a much lesser tendency toward coarse-grain formation than alloys melted from pure substances. Dehlinger (Bibl.27) mentioned that the distinct mosaic structure in recrystallized materials may be responsible for the grain enlargement because of its surface energy (see also Section e,2).

2. Recrystallization in Unworked Bodies (Cumulative ~~Selective~~ Recrystallization)

The basic question as to whether recrystallization, in a material which is everywhere uniform except ~~for~~ at the grain-boundary zones, is able to occur in such a manner that individual grains will grow at the expense of others, cannot be answered unequivocally. After the finding by Roehrig in 1935 that ~~selective~~ cumulative recrystallization will occur in aluminum only if the specimens ~~have~~ had previously been cooled to room temperature but not if they had been annealed to slightly below the melting point, it was considered probable (Bibl.1) that the above question should be answered in the negative since disturbances of various types may occur also on slow cooling. Benedicks (Bibl.34) assumed that, during tempering at slightly below melting point, a system with grain boundaries in capillary equilibrium will form. Bulian and Fahrenhorst (Bibl.35), conversely, observed selective recrystallization on cast magnesium specimens of 99.9% purity which were kept 15° below the melting point for a period of 20 days. On the basis of these findings, the authors believe it probable that the above question can be answered in the affirmative. In electrolytically deposited metals, the presence of stresses cannot be doubted (Bibl.21), and the nature of recrystallization is the same as that of a normal recrystallization after cold-working.

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