the single crystal leads inevitably to the Poisson equation for the ideal random polycrystal. This relationship might be expected to emerge from other methods of averaging also. The discrepancy in the Poisson's ratio, as determined on the polycrystal, of 0·25, makes it possible to gauge the extent to which the Cauchy relations have been satisfied in respect of the single crystal (cf. Table III).

Averages for the *magnetic properties* will be found in (677). Here it is a question of the relative change in length $\left(\frac{\delta l}{l}\right)$ of the crystal parallel to the magnetization vector (magnetostriction), the relative change in the specific resistance $\left(\frac{\delta \rho}{\rho}\right)$ resulting from magnetization, and finally of the energy $\sigma = \int_0^j H_p dJ$ which is necessary for magnetization, in which H_p represents the components of the outer magnetic field parallel to the vector of magnetization. In the event of saturation, the difference in the properties of ferro-magnetic cubic crystals in the direction (S) under investigation and about the cube edge [100]

$$\left(\frac{\delta l}{l}\right)_S - \left(\frac{\delta l}{l}\right)_{[100]},\ \left(\frac{\delta \rho}{\rho}\right)_S - \left(\frac{\sigma \rho}{\rho}\right)_{[100]} \text{ and } \sigma S - \sigma_{[100]}$$

are proportional to the expression $\gamma_1^2 + \gamma_2^2 + \gamma_2^2\gamma_3^2 + \gamma_3^2\gamma_1^2$ [which is also important for the moduli of elasticity and torsion (Fig. 154*a*)], where γ_i is the cos of the angles of the magnetization vector in relation to the cubic axes. The mean value $\bar{f}$ of the three properties of a quasi-isotropic polycrystal is obtained from integration over the entire range of orientation

$$\bar{f} = \tfrac{1}{5}(2f_{[100]} + 3f_{[111]}) \quad . \quad . \quad . \quad . \quad (81/1)$$

in which $f_{[100]}$ and $f_{[111]}$ represent the values for cube edges and the body diagonals. There is close agreement between the observed and calculated results of the change in the electrical resistance of iron.

The same principle of averaging was also applied to the *thermal expansion* (α) and the specific resistance (ρ) of the tetragonal and hexagonal crystals (678). Integration of the expression representing dependence upon orientation (58/1) gives

$$\bar{f} = \tfrac{1}{3}f\| + \tfrac{2}{3}f\bot \quad . \quad . \quad . \quad . \quad . \quad (81/2)$$

in which $f\|$ and $f\bot$ represent the values parallel and perpendicular to the main axis. So far as *specific resistance* is concerned, this result conflicts with the values obtained earlier. It was pointed out in (672) that it was preferable to obtain averages for the constants of conductivity rather than for those of specific resistance, and the expression for the specific resistance of the random oriented polycrystal was stated to be

$$\frac{1}{\rho} = \frac{1}{3\rho_{\parallel}} + \frac{2}{3\rho_{\perp}} \quad . \quad . \quad . \quad . \quad . \qquad (81/3)$$

The conductivity $\left(\frac{1}{\rho}\right)$ has been averaged over the whole range of orientation in (679), from which it follows that for $\rho > \rho_{\perp}$

$$\rho = \sqrt{\rho_{\perp}(\rho - \rho_{\perp})} \arctan \sqrt{\frac{\rho_{\parallel} - \rho_{\perp}}{\rho_{\perp}}} \quad . \qquad (81/4a)$$

and for $\rho_{\parallel} \angle \rho_{\perp}$

$$\rho = 2\frac{\sqrt{\rho_{\perp}(\rho_{\perp} - \rho_{\parallel})}}{\ln\dfrac{\sqrt{\rho_{\perp}} + \sqrt{\rho_{\perp} - \rho_{\parallel}}}{\sqrt{\rho_{\perp}} - \sqrt{\rho_{\perp} - \rho_{\parallel}}}} \quad . \quad . \quad . \qquad (81/4b)$$

A comparison of the methods of calculation will be found in Table XXXVI, which also compares the mean calculated values of some other properties with those determined experimentally.

The great difficulty presented by a comparison of this kind lies in the selection of the observed values. Only in the rarest cases has the texture of the polycrystalline specimen investigated also been determined. The assumed random orientation of the crystals on which the calculations are based is therefore by no means warranted.[1] In the case of specific resistance there is a further element of uncertainty in so far as the determinations on the single crystals and polycrystals were not always carried out on the same type of material, so that the degree of purity may have been different. We believe, however, that the tabulated figures are fairly accurate.

The values obtained by averaging approximate closely to the observed values and so appear to justify the averaging methods

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[1] The random orientation of the crystals can be estimated if the cubic compressibility (K) of the material, or Poisson's ratio (μ), of the specimen is known in addition to the elastic moduli (673). The theory of elasticity yields for the isotropic solid the relationships $K = \frac{9}{E} - \frac{3}{G}$ and $\mu = \frac{E}{2G} - 1$. Satisfaction of these equations is therefore a necessary although inadequate indication of quasi-isotropy.

employed. This should be particularly emphasized in respect of the coefficient of expansion, which in the case of zinc exhibits such exceptional anisotropy $\left(\frac{\alpha\|}{\alpha\perp} = 4{\cdot}51\right)$. The distortions at the grain boundaries, which, as the temperature changes, are necessary for the cohesion of the polycrystal, appear therefore to be confined to such small areas that their effect can be only slight. A decision regarding the average values for specific resistance [the values calculated according to (81/4, *a* and *b*) lie between those given in the table] cannot be reached on the basis of results so far obtained, especially in view of what has just been said regarding the uncertainty of the comparative figures.

In (680) Sachs has furnished an important clue for calculating the *plastic properties* of quasi-isotropic polycrystals from the behaviour of the single crystal. These are arrived at by determining the ratio of the tensile and torsional yield points of the cubic face-centred metals. The shear stress law (cf. Section 40) provides the basis for this calculation. By taking the mean values from graphs it will be found that the maximum shear stress at the start of deformation of the polycrystal is, in the tensile test, greater by 12 per cent., and in the torsional test greater by 29 per cent., than the critical shear strength of the octahedral glide system. From this it follows that the ratio of the torsional yield point to the tensile yield point of the polycrystal (measured by the maximum shear stress) is 1·15. This figure agrees well with the direct experimental determinations made on copper and nickel, which showed the effect of the mean principal stress, and gave for the above ratio an average figure of 1·125 [(681), (682)]. Previously this could be explained only on the assumption that the energy of deformation represented a measure of the probability of flow (683).

An attempt to calculate the yield point of cubic and hexagonal crystals on the basis of an assumed condition of flow of the single crystal, equivalent to a square function of the stresses, will be found in (674). (For objections to this type of flow condition in the single crystal, see Section 40.)

82. *Interpretation of the Properties of Technical Components*

The ideal random arrangement of the crystals, which served as a basis for discussion in the previous section, represents an extreme case which will not normally be present in technical components. As a rule these will be characterized according to their previous history by a more or less pronounced directionality of the grains.

It is on this texture that an exact calculation of the properties of polycrystals must be based, and it is owing to the complicated nature of such actual textures (described in Section 78) that the calculations are difficult to perform. That is why on the whole we still have to be content with a qualitative interpretation of polycrystalline behaviour. However, that this method has added greatly to our knowledge of materials will be apparent from the following examples.

Before we examine the individual results, however, let us glance at a group of polycrystalline properties which can be plausibly interpreted only on the basis of the general behaviour of single crystals [cf., for instance, (684)]. The properties in question are the elastic after-effect, hysteresis and the Bauschinger effect. The occurrence of these phenomena in the case of single crystals of metals has been described in Section 55, while a very tentative explanation of them will be found in Section 76. The reason for their occurrence in the polycrystal is thought to be completely different from that in the single crystal. The different resistance to flow exhibited by the individual grains furnishes a clue to an understanding of the phenomenon. If a polycrystalline specimen is subjected to increasing stress, the yield point will not be exceeded simultaneously in all grains. Whereas those crystals which are very favourably oriented for deformation are soon plastically deformed, the grains which are less favourably placed remain within the range of purely elastic deformation. Thus the individual crystallites do not uniformly resist the externally applied force. Whereas the soft grains, by deforming plastically, are not heavily stressed and are subjected to a load which only slightly exceeds the yield point, the strength of the material resides principally in the grains which have not yet been plastically deformed. In these grains considerable elastic stresses predominate. Even when the stress has been increased to such an extent that the strongest grains become plastically deformed, differences still persist in the stress content of the various crystals. If the load is removed from the solid, the average value of the stress reverts to zero, but not so the stress in the individual grains. In order that the crystals which have already experienced purely elastic deformation can become completely unstressed they must first induce reverse deformation in their plastically deformed neighbours. This is possible only to a limited extent. Consequently, after the removal of the load, there will remain in the grains which resisted the previous load a stress appropriate to this load, while in the plastically deformed crystals there will be a stress of opposite sign. It is in this way that Heyn's concealed elastic stresses (685) can be explained structurally (686).

Since the crystals which have already been deformed plastically continue to be the source of further deformation, it follows from the stress distribution in the unloaded specimen that there will be a reduction in the resistance to a stress applied in the reverse direction to that of the original stress (Bauschinger *effect*) and an increase in the resistance to deformation in the same direction (*work hardening*).

In view of the plastic inhomogeneity of the crystal, the *elastic*

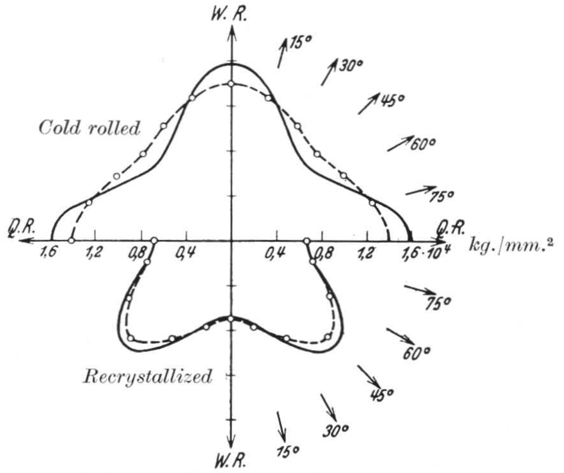


FIG. 218.—The Modulus of Elasticity in Relation to Direction in Cold-rolled and Recrystallized Cu Sheet (690).

WR = rolling direction. *QR* = transverse direction.

——— : Calculated. ○- - -○ : Measured.

after-effect, *i.e.*, the gradual emergence of a condition of equilibrium after application or removal of the load, can now be explained satisfactorily as a *plastic* process. This phenomenon, which normally is absent in the single crystal, consists in an equalization, over a period of time, of the various stresses present in the grains—a process which, as already stated, is possible only with the aid of plastic deformation (687). The quantitative development of this theory of after-effect, and its combination with the earlier phenomenological theories, will be found in (688).

In accordance with this conception the dependence of the shape, which a specimen must possess in order to achieve a certain condition of stress, upon the path which has brought about this condition—a phenomenon known as *hysteresis*—can also be attributed to plastic deformation in those individual crystals which are favourably placed. In this connection the appearance of glide bands in individual crystals at stresses far below the limit of proportionality is highly important (689).

In order to illustrate the connection between the properties of the polycrystal, the texture and the behaviour of the single crystal, diagrams are reproduced in Figs. 218 and 219, which show the dependence upon direction of the *modulus* of elasticity of copper and iron sheets. The curve which has been calculated on the basis

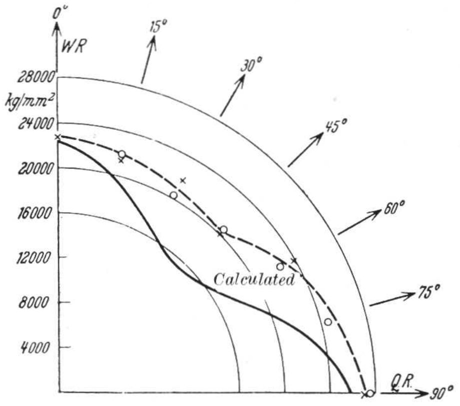


FIG. 219.—The Modulus of Elasticity in Relation to Direction in Cold-rolled Fe Sheet (691).

WR = rolling direction. *QR* = transverse direction.

TABLE XXXVII

Dependence of the Strength and Elongation of Recrystallized Sheets upon Direction

Metal.	Parallel to the rolling		Strength (kg./mm.^{2}). Angle to the rolling direction			Elongation (%). Angle to the rolling direction.		
	Direction.	Plane.	0°.	45°.	90°.	0°.	45°.	90°.
Aluminium (641) .	[100]	(001)	(5·8 =) 1	1·13	1·03	(6·1 =) 1	1·97	1·21
Copper (692) . .			(18·5 =) 1	1·08	0·99	(20 =) 1	3·40	1·15
Calculated . .			1	1·06	1·00	1	2·30	1·00
Silver (693) . .	[112]	(311)	(21 =) 1	0·92	0·90	(44 =) 1	1·00	0·96
α-Brass (72% Cu) (693)			(39·5 =) 1	0·95	0·94	(50 =) 1	1·09	1·12
Calculated . .			1	0·92	0·89	1	1·10	1·12

of the principal positions of the texture agrees closely with the experimentally determined curve for recrystallized copper sheet, while for the cold-rolled sheets also the approximation to the observed results is good. No doubt the agreement would be still more marked if closer regard could be paid to the actual textures [cf., in particular, the detailed discussion in (689*a*)].

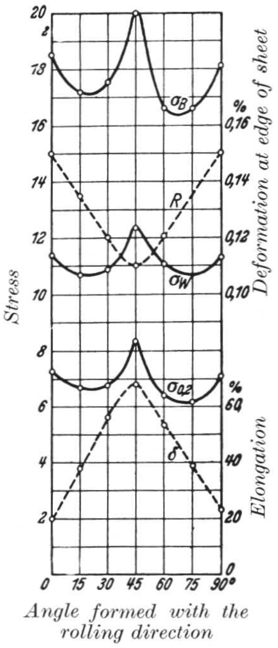


FIG. 220.—Anisotropy of the Static and Dynamic Mechanical Properties of Annealed Cu Sheet (692).

σ_B = U.T.S.;
$\sigma_{0 \cdot 2}$ = yield point;
δ = elongation;
σ_W = bending fatigue strength (10^6 stress reversals);
R = edge deformation.

Table XXXVII contains the *mechanical properties* of recrystallized sheets of cubic face-centred metals. The observed dependence upon direction is contrasted with a calculated figure derived from the properties of correspondingly oriented aluminium crystals (cf. Fig. 99). Consequently, although in this case it is a question of transferring to other metals results already obtained with aluminium, and although only the main position of the recrystallized texture has been considered, the calculated values correspond in general quite closely to the behaviour of the sheets. In this table the low maximum for strength, the high maximum for the elongation in the 45° direction of sheets with the cube texture, and in the case of sheets with the [112]—(311) texture, the slight decrease in strength and increase in elongation with increasing angle to the rolling direction, are shown very clearly. The graph shown in Fig. 220 contains in addition the yield point and fatigue strength of recrystallized copper sheet. The dependence of these properties upon orientation corresponds substantially to that of the static tensile strength. It seems probable, therefore, that fatigue strength, like the tensile strength, is related to the orientation of the crystals. It is thus essential that the texture should be taken into account not only in the case of the static properties but also when assessing the fatigue strength.

The *deep-drawing properties* of sheets are also largely influenced by the texture; in fact, the cupping test is merely an intricate tensile test in which the fibres are stretched in all possible directions (694). In the deep drawing of hollow components the anisotropy of the sheet will reveal itself in " earing " (" Zipfelbildung ") and in a reduced capacity for deep drawing [aluminium (632), copper (695), (693), iron (696)].

The zonal type of structure which is characteristic of cold-drawn wires (cf. Fig. 209) suggests that in different layers of the same wire

Table XXXVIII

Directionality of Certain Properties of Zinc Sheets. [*Taken from* (637)]

Property.	Sheet thickness, mm.	Parallel to the :		
		Rolling direction.	Transverse direction.	Normal direction.
Coefficient of expansion × 10^6 between 30° and 50° C.	0·65 2·27	21·0 30·5	14·1 18·7	 36·7
Modulus of elasticity in kg./mm.^{2} [taken from (701)]	0·28 2·27	9180 8200	10110 10100	
Poisson's ratio [taken from (701)]	2·27	0·299	0·226	0·320
Yield point (0·2%) in kg./mm.^{2}	0·65 2·27	9·7 14·8	13·1 18·8	
Ultimate tensile strength in kg./mm.^{2}	0·28 0·65 2·27	24·4 20·4 28·5	31·6 27·2 35·9	
Elongation, %	0·28 0·65 2·27	2 10 12	$\sim$2·5 7 3	

there will be differences in those properties which are dependent on orientation. Such differences were, in fact, observed in the tensile strength. Tensile tests on etched copper wires revealed that the tensile strength of the core zone is about 10 per cent. higher than the average value for the original wire. In spite of the greater distortion caused by deformation in the surface zone, the work hardening of the more heavily deformed core sections predominates (629). Compression tests on drawn aluminium bars also revealed this difference between the core and surface zones (697).

Attention has been drawn in (696) to the connection between the anisotropy of the *magnetization* of iron and nickel sheets on the one hand, and the texture and crystalline properties on the other. The significance of texture in the *corrosion resistance* of sheets is clearly shown in the case of copper with random and regular recrystallized texture (698).

In regard to the *specific resistance* and the *coefficient of expansion* we will examine in the first place the changes which accompany the plastic working of hexagonal metals and tin. In every case the changes could be related to reorientations of the crystals, whether by gliding or by the development of deformation twins [(679), (699), (700)]. The coefficient of expansion of zinc and the hexagonal zinc alloys in the rolling, transverse and normal directions, is examined with particular care in (645). The maximum value for thermal anisotropy (the maximum differential expansion between the rolling and transverse directions) was found to occur at rolling percentages of about 20–40 per cent.; beyond 80 per cent. the anisotropy disappears. This behaviour is explained by the opposing effects of gliding and twinning (twinning increases, basal glide diminishes anisotropy in the rolling plane).

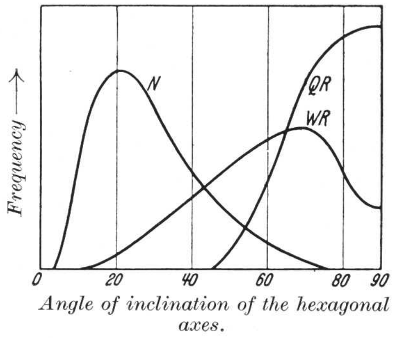


FIG. 221.—Distribution of the Angles of the Hexagonal Axes for the Rolling, Transverse and Normal Directions in Zinc Sheets.

WR = Rolling direction.
QR = Transverse direction.
N = Normal direction.

Table XXXVIII summarizes the directionality of properties of zinc sheets of various thickness. In view of the substantial range of scatter (Fig. 206*a*) the results cannot be explained in terms of the rolling structure if only one main position of the hexagonal axis is used. Fig. 221 gives for the rolling, transverse and normal directions an approximate picture of the distribution of the angles at which the hexagonal axes are located in the sheet under investigation. These distributions must be taken into account when deriving the properties of the sheet from the behaviour of the single crystal and the texture. On a qualitative basis this operation leads to very satisfactory agreement in the case of the coefficient of expansion and modulus of elasticity. The absence of anisotropy in specific resistance in the plane of the sheet can be readily understood (accuracy of measurement 1 per cent.). For the mechanical properties the comparison is less accurate; the reason is that hexagonal crystals glide on the basal plane only; thus they are quite unable to cope with the deformation demanded by the crystal aggregate (cf. Section 79). Nevertheless, if the behaviour of the single crystal is taken into account (Figs. 87 and 96) and if allowance is made for the diversity of orientation in the sheet, it is possible to account roughly for the observed anisotropy of the yield point, ultimate tensile strength, elongation and bending capacity (702).

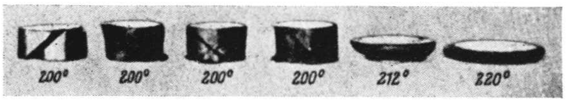
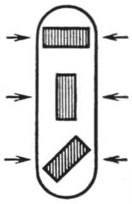


FIG. 222.—Sudden Change in the Plastic Properties of a Magnesium Alloy (*AZM*) at about 210° C. (703).

We will mention in conclusion the example of an Elektron alloy (AZM—6·3 per cent. Al, 1·0 per cent. Zn, remainder magnesium). It will be seen from Fig. 222 that a sudden change takes place in the plasticity of the technical alloy at a working temperature of about

TABLE XXXIX

Tensile and Comprehensive Yield Point of a Deformed Elektron Alloy (*AZM*). [*Taken from* (703)]

Component.	Yield point (0·2%), kg./mm.^{2}.	Compressive yield point (—0·2%), kg./mm.^{2}.
Extruded rod.	23·0	13·0
Forged Billet. 	12·0	18·0
	18·0	12·0
	8·0	8·0

210° C. This behaviour results from the activity of new glide elements at higher temperatures in the hexagonal solid solutions of magnesium (cf. Table VI), as a result of which the material deforms more readily.

A very characteristic effect of the mechanisms by which hexagonal crystals deform (gliding *and* twinning) can also be observed in the behaviour of the tensile and compressive yield points of worked material. Table XXXIX contains particulars of bar material and forged discs. The appreciable discrepancies can be readily explained in terms of the deformation textures (hexagonal axis normal to the longitudinal direction of the bar, or in the forging direction). Basal glide can occur only in specimens taken at an angle of 45° to the forging direction. In other cases deformation must normally start by twinning. This leads, however, to a very definite change of shape (Section 39). If this change is in the direction of the external stress, the flow resistance will be low, but if it is in the opposite direction to the imposed deformation, then the resistance will be high. Where basal glide is possible, deformation takes place in both directions at the same low stress. By producing a specially fine grain in the extruded rods, and by frequently changing the direction of forging and so avoiding formation of too pronounced a texture, it is possible to eliminate plastic anisotropy—usually a technically undesirable property (703).

TABLE XL

Crystalline Lattice Structure of the Elements

□ Cubic face centred.
⊡ Cubic body centred.
⊠ Cubic Diamond.
⬡ Hexagonal close packed.
△ Rhombohedral.

1 H hex.																	2 He □
3 Li ⊡	4 Be ⬡	5 B	6 C ⊠ D hex G											7 N kub. hex.	8 O rhomb. hex.	9 F	10 Ne □
11 Na ⊡	12 Mg ⬡	13 Al □	14 Si ⊠											15 P kub. △	16 S rhomb.	17 Cl	18 Ar □
19 K ⊡	20 Ca □ α ⊡ β	21 Sc	22 Ti ⬡	23 V ⊡	24 Cr ⊡ α ⬡ β kub. γ	25 Mn kub. α kub. β tetr. γ	26 Fe ⊡ αβ □ γ	27 Co ⬡ α □ β	28 Ni □ (⬡)	29 Cu □	30 Zn ⬡	31 Ga rhomb.	32 Ge ⊠	33 As △	34 Se hex. monoklin	35 Br	36 Kr □
37 Rb ⊡	38 Sr □	39 Y ⬡	40 Zr ⬡ α ⊡ β	41 Nb ⊡	42 Mo ⊡	43 Ma	44 Ru ⬡	45 Rh kub. □	46 Pd □	47 Ag □	48 Cd ⬡	49 In tetr.	50 Sn ⊠ α tetr. β	51 Sb △	52 Te hex.	53 J rhomb	54 X □
55 Cs ⊡	56 Ba ⊡	57–71 Selt. Erden	72 Hf ⬡	73 Ta ⊡	74 W ⊡ α kub. β	75 Re ⬡	76 Os ⬡	77 Ir □	78 Pt □	79 Au □	80 Hg △	81 Tl ⬡ α □ β	82 Pb □	83 Bi △	84 Po	85	86 Em
87	88 Ra	89 Ac	90 Th □	91 Pa	92 U ⊡												

TABLE XLI

Type and Dimensions of the Lattices of the Important Metals

Metal.	Lattice type.	Parameter,* Å.	Metal.	Lattice type.	Parameter,* Å.	
					a.	c.
Al	Cubic face-centred A1 †	4·040	Be	Hexagonal close packed A3	2·281	3·577
γ-Fe ‡		3·56	Mg		3·202	5·199
β-Co		3·54	α-Co		2·51	4·07
Ni		3·517	Zn		2·659	4·935
Pd		3·881	Cd		2·974	5·606
Pt		3·915				
Cu		3·608	α-Sn	Cubic face-centred (diamond) A4	6·46	—
Ag		4·078				
Au		4·070				
Pb		4·939	β-Sn	Tetragonal body-centred A5	5·819	3·175
Ta	Cubic body-centred A2	3·29	Bi	Rhombohedral body-centred A7	4·736	ω = 57° 16′ u = 0·474
Mo		3·140	Sb		4·50	ω = 57° 5′ u = 0·466
W		3·158				
α-Fe		2·861	Te	Hexagonal A8	4·445	5·912 u = 0·269

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* Measurements taken from M. C. Neuburger, *Z. Kristallogr.*, Vol. 86 (1933), p. 395.

† Designation of the lattice types according to P. P. Ewald and C. Hermann, *Strukturbericht*, Leipzig, 1931.

‡ Extrapolated at room temperature.

TABLE XLII

Type and Dimensions of the Lattices of some Ionic Crystals *

Salt.	Lattice type.	Parameter, Å.	Salt.	Lattice type.	Parameter, Å.	
					a.	c.
			β-ZnS	Wurtzite B4 (hexagonal)	3·84	6·28
			CdS		4·14	6·72
NaF	Rock salt B1 (cubic)	4·62	BeO		2·69	4·37
NaCl		5·628	ZnO		3·24	5·18
NaBr		5·96				
NaI		6·46	NiAs	Nickel Arsenide B8 (hexagonal)	3·61	$5{\cdot}0_3$
KF		5·33	NiS		3·42	$5{\cdot}3_0$
KCl		6·28				
KBr		6·59	CaF_2	Fluorspar C1 (cubic)	$5{\cdot}45_1$	—
KI		7·05				
RbCl		6·54				
NH_4I		7·24	TiO_2	Rutile, C4 (tetragonal)	4·58	2·95
MgO		4·20				
PbS		5·9				
			CdI_2	Cadmium iodide C6 (hexagonal)	$4{\cdot}2_4$	$6{\cdot}8_4$
			$Mg(OH)_2$		$3{\cdot}1_2$	$4{\cdot}7_3$
			α-Al_2O_3	Corundum D51 (rhombohedral)	5·12	α = 55° 17′
CsCl	Cæsium chloride B2 (cubic)	4·11				
NH_4Cl		3·86				
NH_4Br		4·05	$CaCO_3$	Calcite G1 (rhombohedral)	6·361	α = 46° 7′
			$MgCO_3$		5·61	$48^\circ\ 1'_0$
α-ZnS	Zinc blende B3 (cubic)	5·42	$ZnCO_3$		5·62	$48^\circ\ 2'_0$
CuCl		5·41	$MnCO_3$		5·84	$47^\circ\ 2'_0$
CuI		6·05	$FeCO_3$		5·82	$47^\circ\ 4'_5$
			$NaNO_3$		6·32	$47^\circ\ 1'_5$

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* According to P. P. Ewald and C. Hermann, *Strukturbericht*, Leipzig, 1931. Selection from Table XXIV.

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