Mechanical Anisotropy in Steels for Pipelines

M. S. Joo, D.-W. Suh and H. K. D. H. Bhadeshia

Abstract

There has been a huge expansion in the laying of pipelines for the transmission of fossil fuels over large distances and in dire environments. One particular problem associated with the steels used to fabricate the pipes is that of the anisotropy of mechanical properties, especially the toughness. This review represents an attempt to critically assess the steels and the orientation dependence of their mechanical properties, with the aim of establishing a basis for further progress.

1 Introduction

Steels are the material of choice when fabricating pipes for the economic transmission of crude oil and natural gas from remote regions to populated areas where the fuels are exploited in the generation of energy. There are many kinds of steels used in the fabrication of pipes, some of the strongest of which are not fully established. The American Petroleum Institute specifications of the popular alloys which have seen large scale applications are listed in Table 1.

The actual chemical composition can vary within these specification ranges and the details are usually proprietary; for example, it is rare for X65 to contain a carbon concentration which is as high as 0.26 wt%. Table 2 illustrates the intensity of current activity in pipe laying that is currently in progress.

There are many technical challenges for linepipe steels, including a need for homogeneous microstructure and properties, weldability, a low yield to ultimate tensile strength ratio consistent with safe design, and obviously, an adequate level of toughness. Indeed, the quality required is now much more demanding given the often difficult environments where the pipes must be installed, for example in exceptionally deep oceans off the coast of Brazil, on undulating land surfaces, and in freezing climates. It is fortunate therefore, that the steel industry has, with careful microalloying and thermomechanical controlled processing (TMCP), brought the X80 grade to a technological readiness level where the X80 grade is now in its final form and in service – in the aerospace industry, this corresponds to 'flight–proven'. The higher strength associated with X80 when compared with

1

Table 1: API specifications for common pipeline steels. The grade number corresponds to the approximate yield strength in kilogrammes per square inch. Composition is in wt%. YS, TS and YR stand for yield strength, tensile strength and yield ratio respectively.

Grade	C	Mn	P	S	Nb	V
X65	≤ 0.26	≤ 1.40	≤ 0.04	≤ 0.05	≥ 0.005	≥ 0.005
X70	≤ 0.23	≤ 1.60	≤ 0.04	≤ 0.05	–	–
X80	≤ 0.18	≤ 1.80	≤ 0.03	≤ 0.018	–	–

Grade	YS (MPa)	TS (MPa)	YR (%)
X65	≥ 448	≥ 530	≤ 90
X70	≥ 482	≥ 565	≤ 90
X80	≥ 551	620–827	≤ 93

Table 2: High–strength steel pipeline projects over the period 2007–2012 [1]. The X100 grade is at a trial stage.

Region	Length / km	Steel
North America	11,000	X80/X100
Russia	2,000	X80
China	8,000	X80
Europe	500	X80

X65 permits fluids to be transmitted at higher pressures, or depending on design criteria leads to a reduction in self–weight in the case of hanging structures. The toughness is vital, amongst other factors in ensuring the arrest of high–velocity cracks [2–4]. Low–grade petroleum wells contain hydrogen sulphide gas which makes the steels susceptible to hydrogen–induced and stress–corrosion cracking. Further issues arise when changes occur, for example in the yield strength, after pipe–forming, with a resultant increase in the ductile–brittle transition due to the work of hardening [5, 6].

These and other requirements needed to ensure integrity during service are well–known and to a large extent understood and incorporated into practice, both via specifications and detailed analysis. The physical metallurgy aspects have been reviewed, each with different emphasis [2, 7–13]. Our purpose here it to focus on the development of anisotropy in pipeline steels, particularly with respect to the toughness of the steel as measured using Charpy impact tests. Such tests record the energy absorbed on impact by a standard specimen as a function of the test temperature. The function is usually in the form of a sigmoidal curve with an upper shelf corresponding to the maximum absorption of energy by ductile fracture, a lower shelf representing cleavage (brittle) failure, and a transition temperature involving mixtures of the ductile and brittle modes. Anisotropy can arise in both the ductile and brittle modes of fracture.

The orientation dependence of toughness is a well–known phenomenon for hot–rolled steels in general [14, 15]. As will be seen later, there is also considerable work in the context of pipeline steels. There are three particular factors which are emphasised in discussing toughness anisotropy: non–uniform distribution in the size and shape of inclusions, microstructural anisotropy due to chemical segregation with banding or elongated grain structure, and crystallographic texture. In general, the role of inclusions and segregation is most prominent in the X65 or X70 grades steel, whereas the stronger variants are cleaner and transform at larger undercoolings so that they tend to have more homogeneous microstructures in spite of the segregation. Therefore, crystallographic texture may in such materials play the greatest role in determining the anisotropy of properties.

2 Inclusions

Inclusions influence anisotropy because they often are associated with the initial solidification process and any solidification–induced chemical segregation; others may precipitate in the austenite at high temperatures prior to the hot deformation that the cast steel is subjected to. Common inclusions include the manganese sulphides, silicates and alumina and combinations of oxides depending on the particular steelmaking route, and the hot deformation needed to shape the steel has the effect of orienting these phases along the principal plastic–strain directions. Some phases such as manganese sulphides and silicate can deform along with the steel and hence become elongated, whereas others fragment and form stringers along these same directions [16, 17]. The inclusions can as a consequence, lead to a variety of anisotropies, particularly with respect to tensile elongation, bend properties, fracture mode and fracture toughness. The fracture toughness where the Charpy specimen notch is in the rolling direction of the steel, is found to be much lower than when it is normal to that orientation[18–22]. It is the directionality of MnS inclusions that is found responsible for the orientation dependence of toughness on the forging reduction ratio [23]. A typical microstructure of ferritic grains, pearlite bands and shape of flattened MnS is illustrated in Fig. 1

3

[24]. Fig. 2 illustrates impact energy values for two steels with different sulphur contents. The interesting observation is that the degree of anisotropy is less for the higher sulphur steel. This is because the sulphides in the low–sulphur alloy were relatively more elongated [25]. These observations also emphasise the role of inclusions in nucleating and linking voids, since the data illustrated all represent the ductile mode of failure, so that the observed anisotropy is a direct consequence of the shape of the inclusions. It is important to note also that the sulphides may not be uniformly distributed within the steel in which case correlations of properties with the average concentration become less certain [26].
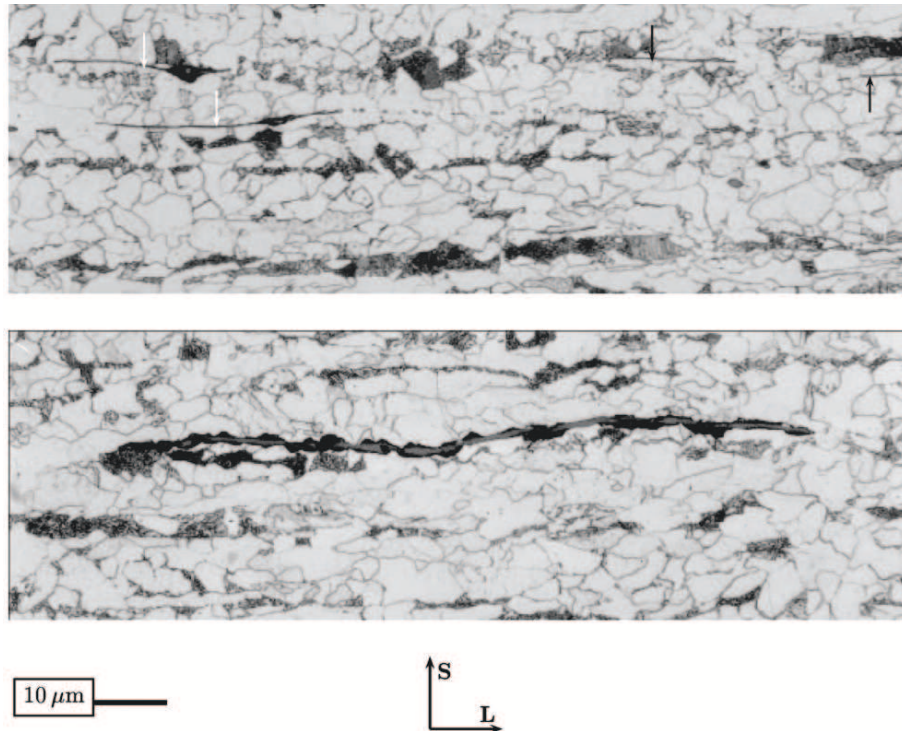

Figure 1: Steel microstructure consisting of ferritic grains, pearlite bands and shape of flattened MnS (arrows) and mixed oxide–MnS (white arrows) both in ferrite in upper image and crushed MnS in pearlite in lower image. 'S' and 'L' stand for the short transverse and longitudinal directions respectively. Reproduced from [24] with permission from Elsevier.

Nonmetallic inclusions influence properties by nucleating cleavage or voids. It is often the case that the inclusions serve to concentrate stress so that adjacent, and more brittle cementite particles can initiate cleavage [27]. This is because inclusions like MnS can lose cohesion with the matrix before the onset of cleavage. The resulting hole provides the greatest stress concentration at the tip of an elongated inclusion, so that loading normal to the plane of the inclusion leads to poor toughness, in contrast to the case where the principal loading is parallel to the long axis when the stress concentration due to decohesion is of minor importance. This of course, leads to anisotropy of toughness [27]. The anisotropy can be reduced by using lower rolling reductions [28] so that the inclusions themselves have smaller aspect ratios (length divided by thickness).

Inclusions are not simply a problem of pipelines or anisotropy, and many methods have been introduced over time to control the shape and size of nonmetallic particles, particularly the manganese
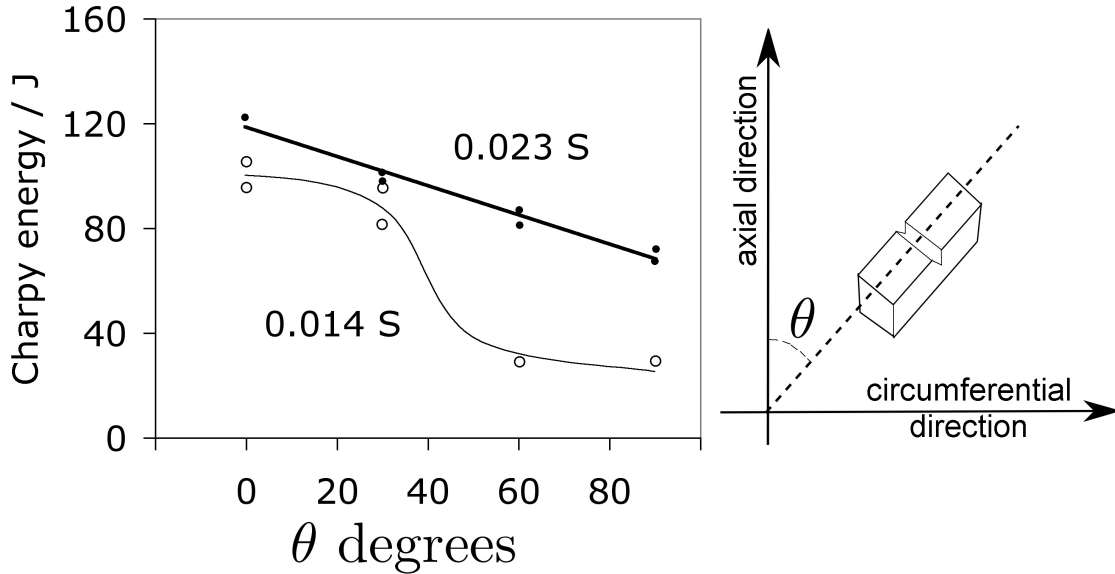
4

Figure 2: Charpy energy as a function of orientation and sulphur content (wt%). The tests were arranged so that all of the samples undergo ductile failure. Data for A106 grade B steel pipe, from [25].

sulphides. The most obvious is to reduce the concentration of sulphides but anisotropic toughness persists even when the concentration is much less than 0.01 wt% [18, 29], presumably, as emphasised in [25], it is the shape of the particles that plays the major role in the orientation dependence of properties. Cross–rolling can mitigate the effects of manganese sulphide even when the sulphur concentration is as large as 0.06 wt% [18], but the method is not practical for most pipelines where the steel plate has to be long in one direction. The addition of cerium, zirconium or titanium helps reduce the plasticity of MnS, thus modifying the inclusion shape into a more spheroidal form [29].

Imai and co–workers did some interesting experiments in which the steel was heat–treated in a variety of ways to modify the deformed manganese sulphide present in the hot–rolled state (Fig. 3) [30, 31]. Although the anisotropy in mechanical properties decreased when the layers were broken up into spherical particles, it was not completely eliminated when the sulphur concentration exceeded about 0.01 wt%. This is because at high concentrations, the spheroidised particles remained aligned as strings ((Fig. 3e). Similar observations have been reported in the context of bearing steels subjected to rolling–contact fatigue, where strings of non–metallic particles aligned normal to the contact surface are much more harmful than those parallel to that surface [32], and there is a strong dependence of life on the length of these strings of inclusions [33].

The role of inclusions in promoting the orientation dependence of toughness also is a function of the steel microstructure; a large fraction of pearlite reduces anisotropy by becoming the primary source for void initiation [30].
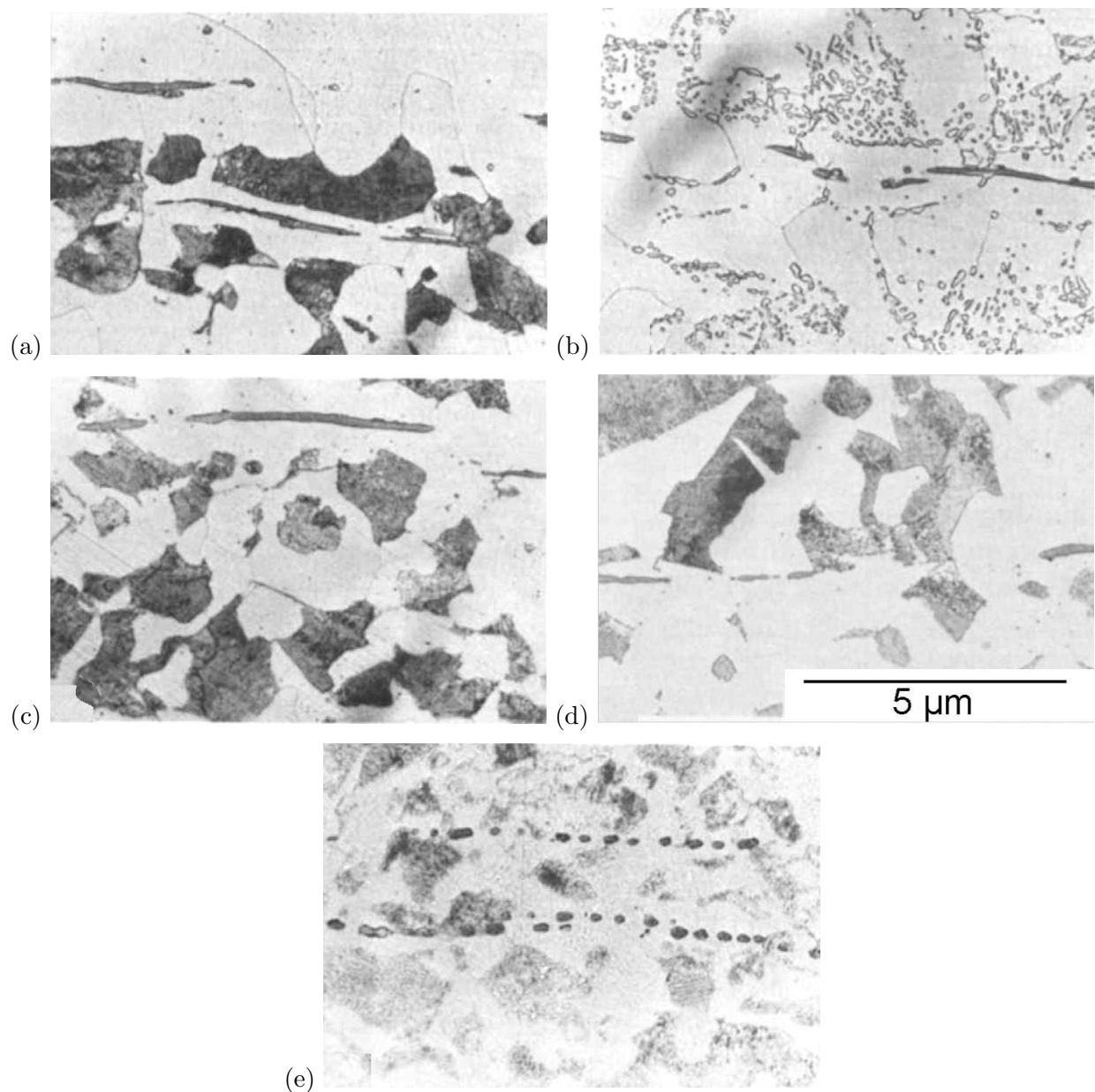
Figure 3: Fe–0.36C–0.22Si–0.73Mn–0.02P–0.025S wt%. (a) As–received hot–rolled condition. (b) Following spheroidisation heat–treatment involving dwell and cycling between 750 and 690°C followed by furnace cooling. (c) Austenitisation at 950°C for 1 h followed by air cooling. (d) Homogenisation at 1300°C for 1 h followed by furnace cooling. (e) Homogenisation at 1300°C for 4 h followed by furnace cooling. Reproduced from [30] with permission from Elsevier.

3 Microstructural Anisotropy

A typical microstructure in a hot–rolled low–alloy steel is illustrated in Fig. 4a [34, 35]. There are planner patches of pearlite parallel to the rolling plane. Microstructural banding is often more pronounced in sections containing the rolling direction than in those containing the transverse direction [36, 37]. Such banding is known to be responsible for the orientation dependence of properties [29, 38], and homogenisation to mitigate the banding reduces anisotropy [39].

Banding occurs primarily because of the segregation of solutes in the last regions of the liquid to solidify during the cooling of steel from the molten state. The low–alloy steels which exhibit banding typically begin solidification as δ–ferrite so that elements such as manganese, silicon, phosphorus and sulphur are partitioned into the interdendritic regions which then solidify with a higher than average concentration of these solutes. Subsequent deformation, for example by hot–rolling, causes these regions to spread out as bands. The segregation of concern is of substitutional solute such as manganese (Fig. 4b). Carbon also segregates during solidification but it diffuses rapidly as the steel cools through the austenite phase field until its chemical potential becomes uniform. The silicon concentration, which is not illustrated in Fig. 4b, was also found to be in phase with the pearlite bands. Although silicon is a ferrite stabiliser, its influence on the transformation in typical steels of interest here is much smaller than that of manganese.

The ferrite–pearlite banding evident in Fig. 4a occurs when the regions which are depleted in austenite–stabilising elements decompose into ferrite, before the transformation can occur in other areas [40]. As a result, carbon is partitioned into the adjacent substitutional–solute rich austenite, which ultimately becomes the pearlite. The microstructural banding therefore correlates with the segregation pattern and the correlation becomes more pronounced when the microstructure is generated by slow cooling. This is because larger cooling rates are associated with greater undercoolings, which permit ferrite to form even in manganese–enriched regions.

A reduction in the carbon concentration also helps make the steel more isotropic because it is the partitioning of carbon into the manganese–rich regions that leads to the formation of pearlite bands [29].

The development of microstructural banding is illustrated in Fig. 5. As pointed out previously, the highly mobile carbon homogenises during cooling through the austenite phase field. However, there are gentle variations which occur in concert with the manganese, as the carbon maintains a uniform chemical potential in the austenite. Manganese lowers the activity of carbon and hence the manganese–rich regions are associated with a somewhat higher carbon concentration [41]. The dependence of the spatial distribution of carbon on that of substitutional solute in austenite was originally thought to be the cause of banding [40]. Bastien [42], however, considered the banding to be due to the substitutional solutes and Kirkaldy *et al.* [41] later showed that this is indeed the dominant effect.

An alternative mechanism is found in steels containing relatively large sulphur concentrations [43]. Manganese sulphides then precipitate in the regions containing a large average concentration of manganese. As a consequence, the manganese is bound in the sulphide which is surrounded by a manganese–depleted zone where ferrite forms. The ferrite partitions carbon into the adjacent zones
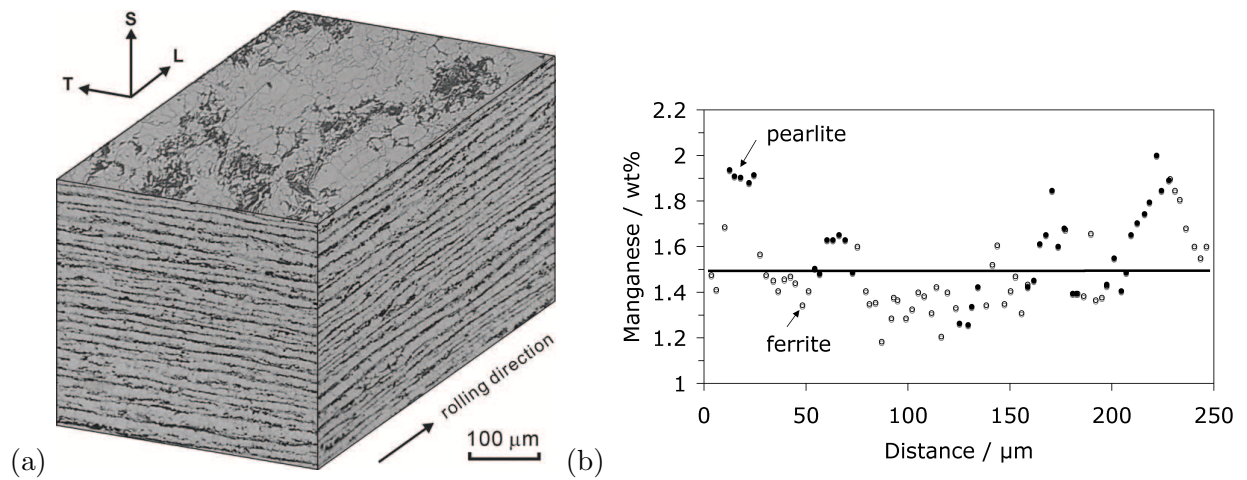
7

Figure 4: (a) Banding in hot–rolled ferrite–pearlite steel, Fe–0.15C–0.16Si–1.07Mn wt%. ’S’, ’T’ and ’L’ stand for the short transverse, transverse and longitudinal directions respectively. Reproduced from [35] with permission from Elsevier. (b) The location of pearlite relative to the manganese concentration [36].
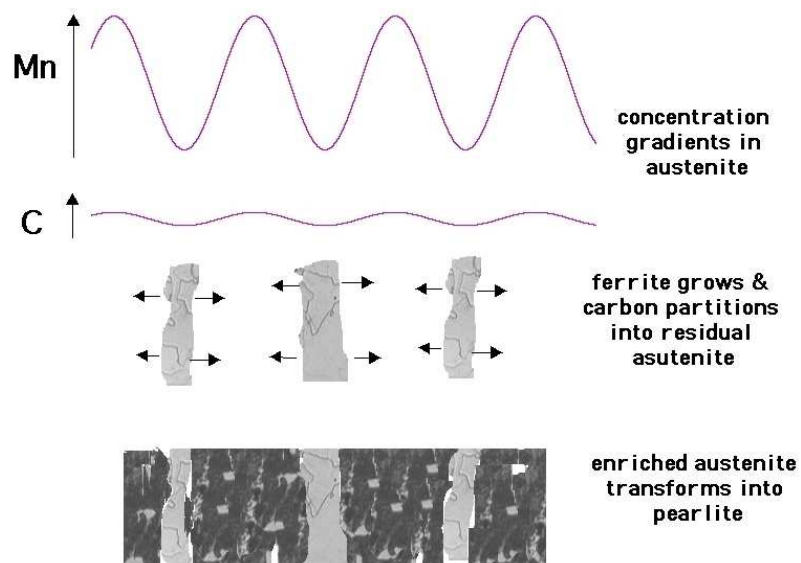

Figure 5: An illustration of the common mechanism of banding. Note that banding has irregularities so it is not entirely an accurate reflection of the chemical segregation pattern [36].

which have a low average concentration of manganese, which transform into pearlite. The position of the ferrite bands is thus shifted into locations where the average Mn concentration is large, but where the Mn is tied up as sulphides (Fig. 6).
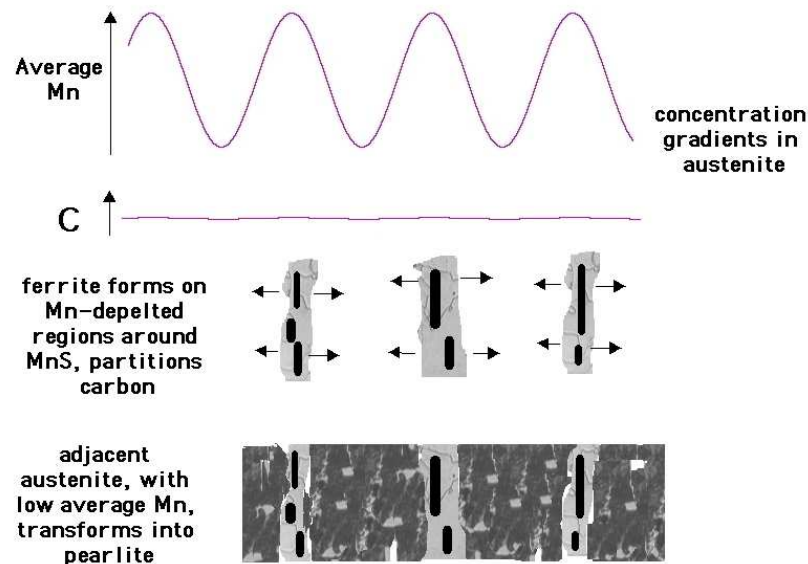

Figure 6: The mechanism of banding in steels containing substantial quantities of manganese sulphides.

Other kinds of microstructural anisotropy also has been shown to correlate with toughness in pipeline steels. Fig. 7 shows the case for the X80 grade pipeline steel where the minimum in Charpy toughness in the ductile–brittle transition temperature range corresponds to the smallest fraction of lower bainite on the fracture plane. Also the study suggested that the non–uniform grain shape is related to the anisotropy of the Charpy toughness [44, 45]. One difficulty with this interpretation is that the fraction should be independent of the plane of section provided that the number of measurements made is statistically meaningful. It is also not clear why the minimum correlates with the smallest lineal intercept since a finer microstructural scale should lead to better toughness. As pointed out by Garcia *et al.*, the real explanation in the context of the X80 steels is likely to be a complex combination of microstructure and crystallographic texture; the latter aspect is discussed in the next section. Kichkina et al. also discussed in ferrite-bainite microstructure that the bainite morphology oriented predominantly in the intermediate direction to the rolling direction, such as 50°, is needed to be considered for anisotropy of toughness in API-X70 steel, since the density of grain boundaries and interphase boundaries in that direction are minimum which is corresponding to the minimal obstacles of crack propagation [46]. Sun and Boyd investigated the effect of processing parameter on anisotropy of cleavage fracture stress in microalloyed linepipe steel. The anisotropy of fracture stress increases with decreasing finish rolling temperature (FRT) above nonrecrystallisation temperature of austenite and the factor for anisotropy was determined from microstructural anisotropy such as elongated grain and banding [47]. Some early study also suggests that an elongated grain structure, for example that generated by warm-rolling with a low

finish-rolling temperature, has been shown to cause variations in the impact properties according to direction relative to the elongated structure [48]. Okuda *et al.* also reported that the ductile–to–brittle transition temperature (DBTT) in longitudinal direction is lower than transverse direction due to elongated grain in the extruded oxide dispersion strengthened (ODS) ferritic steels [49].
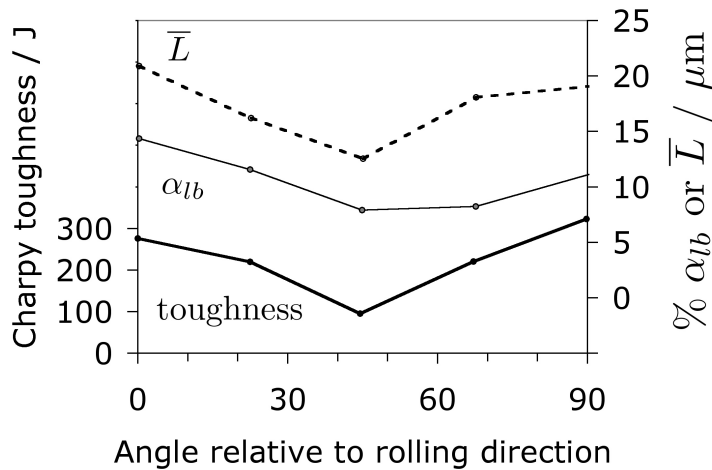

Figure 7: The Charpy toughness at -60°C, percentage of lower bainite (α_{lb}) and mean lineal intercept as a function of the angle in degrees, relative to the rolling direction. Adapted from [44].

Work of fracture with microstructural banding has been related with orientation–dependence of impact properties. Joo and co–workers showed the case for the API–X80 grade linepipe steel with the minimum in Charpy toughness in the ductile–brittle transition temperature range corresponds to the absence of 'split' on the fracture plane. It established that Charpy specimens machined at 45° to the rolling direction with notches along the plate normals, performed relatively low impact toughness in the transition temperature regime and that was strongly affected by the occurrence or absence of split. It has been also confirmed experimentally by conducting Charpy tests with notches along the plate parallel and the source of split was discovered as the presence of microstructural banding, with variations in crystallography between adjacent bands [50]. When in exaggerated form, the splitting during mechanical tensile or toughness testing is known sometimes as *delamination*. Such a split or delamination is illustrated in Fig. 8 and has been observed in the high–strength low–alloy steels [51–54], in ferritic stainless steels [55, 56], ausformed alloys [57], marformed steels [58] and low–carbon steels [59, 60]. Delamination is created by weakness parallel to the rolling plane within the steel, either because hot–rolling is terminated at temperatures within the $\alpha + \gamma$ phase field, texture, inclusion alignment on the rolling plane, intergranular failure along prior austenite boundaries, segregation of phosphorus and sulphur and any aspect that leads to anisotropic microstructure [61, 62].

It has been suggested that the partial splitting of Charpy specimens during macroscopically ductile fracture, following an impact test, is related to separation on $\{100\}_\alpha$ cleavage planes. Indeed, delamination is regarded macroscopically as brittle fracture behaviour which is concerned with a decrease of upper shelf energy in Charpy tests [63, 64]. It is suggested to lead to a decrease in toughness in the upper shelf regime, and in a reduction in the ductile-to–brittle transition temperature (DBTT). This does not seem correct; a decrease in DBTT is an increase in toughness, and the evidence has not been presented well to support this notion, and since the split is not parallel to the plane containing the notch, it could be argued that an improvement in toughness may follow due to the additional work of fracture at the split. However, if the separation occurs in the 100%

Figure 8: The feature normal to the length of the Charpy notch is a 'split' which influences the impact energy measured. Reproduced with permission of Elsevier from [63].

ductile range then the impact energy decreases [63], but if it takes place at low temperatures then the enhancement of toughness is observed [65]. So-called delamination toughening has long been known to lead to improved toughness in a wide variety of materials, for example in ceramics [66], carbon [67], and fine-grained steels [68–71]. In particular, delaminations that form parallel to the plate surface can lead to a reduction in the impact transition temperature [72] although excessive fissuring can compromise the upper shelf energy of Charpy tests [73]. Indeed, it is not surprising that delamination is related to the toughness anisotropy.

4 Crystallographic Texture

Texture implies a non–random distribution of crystal orientations and clearly must influence the isotropy of a polycrystalline material. Kotrechko *et al.* investigated the effect of texture on the anisotropy of the cleavage stress of metals by theoretical calculation. It shows that the nonuniform distribution of crystallographic planes is the main reason for the cleavage-stress anisotropy in the textured polycrystals [74]. Mintz *et al.* investigated the influence of texture on the impact properties of controlled rolled steels, and suggested that the role of texture on the directionality of impact properties is small and largely on ductile fracture [75]. Inagaki and coworkers focused on controlled rolled high strength steel with several textured specimens of different chemical compositions. It showed that $\{100\}_\alpha$ cleavage plane is related qualitatively to the anisotropy of toughness. In particular, it pointed out that anisotropy of toughness is induced with the $\{113\}$ texture, which is placed between $\{110\}$ and $\{112\}$ planes, parallel to the rolling plane. Thus, the impact transition temperature peaked in the intermediate directions such as $45°$ to the rolling direction, and correlated this to the texture which favoured cleavage [76]. Bourell and Sherby also discussed the effect of strong $\{001\} < 110 >$ texture component corresponding to lower Charpy energy of diagonal specimens than that of longitudinal specimens at various test temperatures in warmrolled low

11

carbon steel [77]. Joo and co–workers also pointed out that the $\{100\}_\alpha$ cleavage plane is related to the orientation dependence of impact toughness especially in the DBTT regime [50].

The plates used in the manufacture of pipeline steels are thermomechanically processed and hence are textured. Baczynski *et al.* showed that components of the macroscopic texture can be correlated with the observed anisotropy of impact toughness in niobium–microalloyed steels. It is well known that in the ductile fracture region corresponding to relatively high test temperatures, the distribution of $\{110\}_\alpha$ and $\{112\}_\alpha$ slip planes has a greater influence on toughness than that of the $\{100\}_\alpha$ cleavage–planes. The toughness anisotropy associated with ductile fracture correlates with the $\{112\} < \bar{1}10 >_\alpha$ component, and for cleavage fracture with the $\{001\} < 110 >_\alpha$ and $\{110\} < 001 >_\alpha$ components if they exist [78]. Ju *et al.* also reported that the anisotropy of toughness at room temperature might be related to the distribution of $\{110\}_\alpha$ planes as a function of angle relative to the circumferential direction in API-X65 linepipe steel [79].

Mechanical anisotropy depends on many factors other than texture, so it is not surprising that there are studies which reach the conclusion that crystallography has little or no role to play in determining the orientation dependence of properties. Mintz *et al.* found no indication of texture having any influence on the anisotropy of impact toughness in the transition temperature especially when the grain structure itself is directional, so the microstructural anisotropy is severe [48]. Fegredo *et al.* found no significant relationship between toughness and the macroscopic texture in low–carbon steels containing 0.002–0.007 wt% sulphur, produced using different rolling temperatures [80]. Microstructural anisotropy and elongated sulphides seemed to have the strongest influence. Kasada *et al.* investigated the effect of texture and microstructure on the mechanical anisotropy in hot-extruded bar of oxide dispersion strengthened (ODS) ferritic steels with $\{110\}$ textured along the extrusion direction (ED) parallel to the rolling direction (RD), and the toughness anisotropy was induced by the combined effect of elongated grains and small particles along ED, not by texture [81]. Garcia *et al.* reported that both microscopic and macroscopic textures had negligible influence on the anisotropy of properties in X80 steel, the main cause being associated with an uneven distribution of carbide–rich microstructural constituents. Pyshmintsev *et al.* also reported that the modern X80 linepipe tends to show no clear correlation between the $\{100\}_\alpha$ planes and ductile fracture parallel to the rolling plane [82].

5 Concluding Remarks

Mechanical anisotropy of the steel is essential because of unavoidable chemical segregation and/or thermo–mechanical working. Especially, the anisotropy of toughness is a particular problem, with toughness depending on the orientation relative to the rolling direction. In the context of linepipe steel, large-calibre spiral-welded pipes have not been exploited fully due to their anisotropy of mechanical properties and decrease in strength after pipe forming [5, 83]. The most critical issue is of course the circumferential direction of spiral-welded pipe, which normally experiences the highest load (hoop stress), would have weakest toughness and strength than other orientations [83]. Consequently, this increases the chances of fracture [84, 85].

An attempt has been established here to study the any forms of potential causes for the orientation–dependence of toughness from literatures, in order to understand the phenomenon in linepipe steel.

Three factors connected to the toughness anisotropy were highlighted: non–uniform distribution of inclusions with various sizes and shapes, microstructural anistoropy with banding and/or elongated grains with some work of fracture and non–random crystallographic texture. Therefore the toughness anisotropy is essentially complex due to the possible connections of the culprits. The relationship itself is not clear and significance of the factors is difficult to determine. However, hopefully this review is for the basis of the knowledge and can rouse the thoughts and indeed take one step further to figure out the phenomenon.

6   Acknowledgements

References

[1] A. Liessem, J. Schröder, M. Pant, M. Erdelen-Peppler, M. Liedtke, S. Höhler, and C. Stallybrass. Manufacturing challenges of high strength line pipes. In T. Perez, editor, *New Developments on Metallurgy and Applications of High Strength Steels*, pages 543–555, Materials Park, Ohio, USA, 2008. TMS–AIME.

[2] N. J. Kim. The physical metallurgy of HSLA linepipe steels - a review. *Journal of Metals*, 35:21–27, 1983.

[3] D. P. Fairchild, M. L. Macia, S. D. Papka, C. W. Petersen, J. H. Stevens, S. T. Barbas, N. V. Bangaru, J. Y. Koo, and M. J. Luton. High strength steels — beyond X80. In M. Toyoda and R. Denys, editors, *Proceedings of Pipe Dreamer's Conference, Yokohama*, pages 307–321, U. K., 2002. Scientific Surveys Ltd.

[4] K. T. Corbett, R. R. Bowen, and C. W. Petersen. High-strength steel pipeline economics. *Journal of Offshore and Polar Engineering*, 14:75–80, 2004.

[5] D. H. Seo, W. H. Song, S. S. Ahn, C. M. Kim, and J. Y. Yoo. Microstructure and mechanical properties of api-x80/x100 grade plates and pipes. *POSCO* , 12:20–27, 2007.

[6] K. Kim and J.-H. Bae. Metallurgical and process parameters for commercial production of high toughness API-X80 grade hot rolled strips. In *2008 7th International Pipeline Conference*, volume 3, pages 167–173, New York, USA, 2008. ASME.

[7] A. M. Sage. A review of the physical metallurgy of high strength, low alloy line pipe and pipe fitting steels. In *Steels for Line Pipe and Pipeline Fittings*, pages 39–50, London, U.K., 1981. The Metals Society.

[8] L. E. Collins and R. Kruger. Casing and linepipe steels for sour service applications. *CIM Bulletin*, 91:93–97, 1998.

[9] K. Khulka and S. Aleksandrov. Promising tube steels for gas pipelines. *Metallurgist*, 50:137–143, 2006.

[10] D. B. Lillig, D. S. Hoyt, M. W. Hukle, J. P. Dwyer, A. M. Horn, and K. Manton. Materials and welding engineering for ExxonMobil high strain pipelines. In N. Ames R. Ayer, J. S. Chung and H. G. Wheat, editors, *Proceedings of the 16th International Offshore and Polar Engineering Conference*, volume 4, pages 8–14, California, USA, 2006. International Society of Offshore and Polar Engineers.

[11] N. Bannenberg, A. Streisselberger, and V. Schwinn. New steel plates for the oil and gas industry. *Steel Research International*, 78:185–188, 2007.

[12] C. Yan, W. Li, L. Feng, Z. Xue, S. Bai, and F. Liu. Review of X100 pipeline steel and its field weldability. *Transactions of the China Welding Institution*, 28:105–108, 2007.

[13] C. I. Garcia, K. Cho, M. Hua, and A. J. DeArdo. The alloy design and thermomechanically controlled processing (TMCP) of plate for high pressure, large diameter pipelines. *Materials Science Forum*, 638–642:124–129, 2010.

[14] D. M. Fegredo. Effect of rolling at different temperatures on the fracture toughness anisotropy of a C-Mn structural-steel. *Canadian Metallurgical Quarterly*, 14:243–255, 1975.

[15] A. Ray, S. K. Paul, and S. Jha. Effect of inclusions and microstructural characteristics on the mechanical properties and fracture behavior of a high-strength low-alloy steel. *Journal of Materials Engineering and Performance*, 4:679–688, 1995.

[16] C. E. Sims. The nonmetallic constituents of steel. *Trans. Met. Soc. AIME*, 215:367–393, 1959.

[17] T. J. Baker and J. A. Cameron. Deformation of MnS inclusions in steel. *Journal of the Iron and Steel Institute*, 210:680–690, 1972.

[18] J. M. Hodge, R. H. Frazier, and G. W. Boulger. The effect of sulphur on the notch toughness of heat–treated steels. *Trans. Metall. Soc. AIME*, 215:745–753, 1959.

[19] G. S. Kramer, G. M. Wilkowski, and W. A. Maxey. Flaw tolerance of spiral–welded line pipe. Technical Report L51514, American Gas Association, Washington, D. C., USA, 1987.

[20] V. A. Burnos, T. P. Vaschilo, and L. E. Balandina. Evaluation of the quality of the metal of pipes according to imact toughness with anisotropy taken into account. *Industrial Laboratory (USSR)*, 54:548–550, 1988.

[21] G. M. Wilkowski, J. Ahmad, F. Brust, N. Ghadiali, P. Krishnaswamy, M. Landow, C. W. Marschall, P. Scott, and P. Vieth. Short cracks in piping and piping welds. Technical Report NUREG/CR-4599-Vol.1-No.1; BMI–2173-Vol.1-No.1, Nuclear Regulatory Commission, Ohio, USA, 1991.

[22] R. Mohan, C. Marshall, P. Krishnaswamy, F. Brust, N. Ghadiali, and G. Wilkowski. Fracture behavior of short circumferentially surface-cracked pipe, short cracks in piping and piping welds. Technical Report NUREG/CR-6298, BMI-2183, Nuclear Regulatory Commission, Ohio, USA, 1995.

[23] S. Harada, T. Endo, and M. Kaseda. Effects of forging ratio and specimen orientation on elastic-plastic fracture toughness of thick forged steel plates. In H. Nisitani G. C. Sih and T. Ishihara, editors, *Role of Fracture Mechanics in Modern Technology*, pages 485–496, Amsterdam, Holland, 1987. North Holland.

[24] A. A. Benzerga, J. Besson, and A. Pineau. Anisotropic ductile fracture part i: experiments. *Acta Materialia*, 52:4623–4638, 2004.

[25] R. Mohan, C. Marschall, P. Krishnaswamy, F. Brust, N. Ghadiali, and G. Wilkowski. Effects of toughness anisotropy and combined tension, torsion and bending loads on fracture behaviour of ferritic nuclear pipe. Technical Report NUREG/CR-6299 BMI-2184, Nuclear Regulatory Commission, Washington, D. C., USA, 1995.

[26] I. P. Medinskaya, Yu I. Rubenchik, T. A. Pisarenko, and E. A. Afanasenko. The influence of property anisotropy on the technological effectiveness of weldable steels. *Chemical and Petroleum Engineering*, 16:658–661, 1981.

[27] T. J. Baker, F. P. L. Kavishe, and J. Wilson. Effect of non-metallic inclusions on cleavage fracture. *Materials Science and Technology*, 2:576–582, 1986.

[28] Y. Tomita. Effect of decreased hot-rolling reduction treatment on fracture toughness of low-alloy structural steels. *Metallurgical and Materials Transactions A*, 21:2555–2563, 1990.

[29] Yu. I. Matrosov and I. E. Polyakov. Increasing the toughness and ductility and decreasing the property anisotropy of low-alloy steels. *Stal*, 2:162–167, 1976.

[30] T. Imai, Y. Nishida, and S. Kogiso. Anisotropy of the Charpy impact value of carbon steels and corrective heat treatment. *Journal of Mechanical Working Technology*, 7:147–161, 1982.

[31] T. Imai, Y. Nishida, and S. Kogiso. Anisotropy of the Charpy impact value of carbon steels. I.–elimination of the anisotropy by heat treatment. *Reports of Government Industrial Research Institute, Nagoya*, 34:123–133, 1985.

[32] J. Courbon, G. Lormand, G. Dudragne, P. Daguler, and A. Vincent. Influence of inclusion pairs, clusters and stringers on the lower bound of the endurance limit of bearing steels. *Tribology International*, 36:921–928, 2003.

[33] J. A. Eckel, P. C. Glawas, J. O. Wolfe, and B. J. Zorc. Clean engineered steels - progress at the end of the twentieth century. In J. K. Mahaney Jr, editor, *Advances in the Production and Use of Steel with Improved Internal Cleanliness*, pages 1–10, Pennsylvania, USA, 1999. ASTM.

[34] L. Tau, S. L. Chan, and C. S. Shin. Effects of anisotropy on the hydrogen induced fatigue crack propagation of a banded ferrite–pearlite steel. *Journal of Marine Science and Technology*, 1:19–22, 1993.

[35] A. A. Korda, Y. Mutoh, Y. Miyashita, T. Sadasue, and S. L. Manan. In situ observation of fatigue crack retardation in banded ferrite–pearlite microstructure due to crack branching. *Scripta Materialia*, 8:1835–1840, June 2006.

[36] S. W. Thompson and P. R. Howell. Factors influencing ferrite/pearlite banding and origin of large pearlite nodules in a hypoeutectoid plate steel. *Materials Science and Technology*, 8:777–784, 1992.

[37] D. Chae, D. A. Koss, A. L. Wilson, and P. R. Howell. Effect of microstructural banding on failure initiation of HY–100 steel. *Metallurgical & Materials Transactions A*, 31:995–1005, 2000.

[38] P. C. Wilson, Y. V. Murty, T. Z. Kattamis, and R. Mehrabian. Effect of homogenization on sulphide morphology and mechanical properties of rolled AISI 4340 steel. *Metals Technology*, 2:241–244, 1975.

[39] S. J. Garwood. The effect of temperature, orientation and constraint on the toughness of A 533 B Class I steel. In *Application of Fracture Mechanics to Materials and Structures*, pages 939–950, Leiden, Holland, 1984. Martinus Nijhoff Publishers.

[40] C. F. Jatczak, D. J. Girardi, and E S Rowland. On banding in steel. *Trans. ASM*, 48:279–305, 1956.

[41] J. S. Kirkaldy, J. von Destinon-Forstmann, and R. J. Brigham. Simulation of banding in steels. *Canadian Metallurgical Quarterly*, 59:59–81, 1962.

[42] P. G. Bastien. The mechanism of formation of banded structures. *Journal of the Iron and Steel Institute*, 187:281–291, 1957.

[43] J. S. Kirkaldy, R. J. Brigham, H. A. Domian, and R. G. Ward. A study of banding in Skelp by electron–probe microanalysis. *Canadian Metallurgical Quarterly*, 2:233–241, 1963.

[44] R. H. Petrov, O. L. Garcia, J. J. L. Mulders, A. C. C. Reis, J. H. Bae, L. A. I. Kestens, and Y. Houbaert. Three dimensional microstructure–microtexture characterization of pipeline steel. *Materials Science Forum*, 550:625–630, 2007.

[45] R. Petrov, O. L. Garcia, N. S. Mourino, L. Kestens, J. H. Bae, and K. B. Kang. Microstructure–texture related toughness anisotropy of API–X80 pipeline steel characterised by 3D–EBSD technique. *Materials Science Forum*, 558–559:1429–1434, 2007.

[46] A. A. Kichkina, M. Yu. Matrosov, L. I. Efron, M. B. Klyukvin, and A. V. Golovanov. Effect of structural anisotropy of ferrite-bainite pipe steel on mechanical properties in tensile and impact bending tests. *Metallurgist*, 54:808–816, 2011.

[47] J. Sun and J. D. Boyd. Effect of thermomechanical processing on anisotropy of cleavage fracture stress in microalloyed linepipe steel. *International Journal of Pressure Vessels and Piping*, 77:369–377, 2000.

[48] B. Mintz, W. B. Morrison, P. I. Welch, and G. J. Davies. The relative contributions of texture and grain shape to the properties of warm-rolled Fe-Mn alloys. In G. Gottstein and K. Lucke, editors, *Texture of Materials*, volume 2, pages 465–474, Berlin, Germany, 1978. Springer–Verlag.

[49] N. Okuda, R. Kasada, and A. Kimura. Statistical evaluation of anisotropic fracture behaviour of ods ferritic steels by using small punch tests. *Journal of Nuclear Materials*, 386–388:974–978, 2009.

[50] M. S. Joo, D.-W. Suh, J.-H. Bae, and H. K. D. H. Bhadeshia. Role of delamination and crystallography on anisotropy of charpy toughness in api-x80 steel. *Materials Science & Engineering A*, xx:xxx–xxx, 2012.

[51] R. Schofield, G. Rowntree, N. V. Sarma, and R. T. Weiner. Arrow–head fractures in controlled–rolled pipeline steels. *Metals Technology*, 1:325–331, 1974.

[52] G. Baldi and G. Buzzichelli. Critial stress for delamination fracture in HSLA steels. *Metal Science*, 12:459–472, 1978.

[53] D. S. Dabkowski, P. J. Konkol, and M. F. Baldy. An investigation of splitting-type fractures in high-strength line-pipe steels. *Metals Engineering Quarterly*, 16:22–32, 1976.

[54] A. J. DeArdo. An investigation of the mechanism of splitting which occurs in tensile specimens of high–strength low–alloy steels. *Metallurgical transactions A*, 8A:473–485, 1977.

[55] B. Mintz. Influence of grain boundaries on fissure formation during impact testing of ferritic stainless steels. *Metals Technology*, 7:127–128, 1980.

[56] H. C. Chao. Mechanism of anisotropic lamellar fractures. *Metallurgical transactions A*, 9:509–514, 1978.

[57] A. J. McEvily and R. H. Bush. An investigation of the notch–impact strength of an ausformed steel. *Trans. ASM*, 55:654–666, 1962.

[58] R. K. Ray and S. Basu. Mechanism of splitting of a few Fe–Mn–Ni–base alloys subjected to thermomechanical treatment. *Steel Research*, 56:341–345, 1985.

[59] B. L. Bramfitt and A. R. Marder. A study of the delamination behaviour of a very low–carbon steel. *Metallurgical transactions A*, 8:1263–1273, 1977.

[60] S. Y. Shin, S. M. Hong, J. H. Bae, K. S. Kim, and S. H. Lee. Separation phenomenon occurring during the Charpy impact test of API X80 pipeline steels. *Metallurgical & Materials Transactions A*, 40:2333–2349, 2009.

[61] D. L. Bourell. Cleavage delamination in impact tested warmrolled steel. *Metallurgical Transactions A*, 14:2487–2496, 1983.

[62] I. Tamura, H. Sekine, T. Tanaka, and C. Ouchi. *Thermomechanical processing of high–strength low–alloy steels*. Butterworth & Co., London, U.K., 1988.

[63] J. I. Verdeja, J. Asensio, and J. A. Pero-Sanz. Texture, formability, lamellar tearing and HIC susceptibility of ferritic and low-carbon HSLA steels. *Materials Characterization*, 50:81–86, 2003.

[64] P. Brozzo and G. Buzzichelli. Effect of plastic anisotropy on the occurrence of separations on fracture surfaces of hotrolled steel specimens. *Scripta Metallurgica*, 10:235–240, 1976.

[65] R. Song, D. Pong, and D. Raabe. Mechanical properties of an ultrafine grained C–Mn steel processed by warm deformation and annealing. *Acta Materialia*, 53:4881–4892, 2005.

[66] W.J. Clegg, K. Kendall, N. Alford, J.D. Birchall, and T.W. Button. A simple way to make tough ceramics. *Nature*, 347:455–457, 1990.

[67] M. Sakai, R. C. Bradt, and Fischbach D. B. Fracture toughness anisotropy of a pyrolytic carbon. *Journal of Materials Science*, 21:1491–1501, 1986.

17

[68] Y. Kimura, T. Inoue, E. Yin, and K. Tsuzaki. Inverse temperature dependence of toughness in an ultrafine grain-structure steel. *Science*, 320:1057–1060, 2008.

[69] Y. Kimura, T. Inoue, E. Yin, and K. Tsuzaki. Delamination toughening of ultrafine grain structure steels processed through tempforming at elevated temperatures. *ISIJ International*, 50:152–161, 2010.

[70] T. Inoue, F. Yin, Y. Kimura, K. Tsuzaki, and S. Ochiai. Delamination effect on impact properties of ultrafine-grained low-carbon steel processed by warm caliber rolling. *Metallurgical & Materials Transactions A*, 41:341–355, 2010.

[71] B. Mintz, E. Maina, and W. B. Morrison. Origin of fissures on fracture surfaces of impact samples of HSLA steels with ferrite/pearlite microstructures. *Materials Science and Technology*, 23:347–354, 2007.

[72] E. Maina, D. N. Crowther, J. R. Banerjee, and B. Mintz. Influence of directionality on strength and impact behaviour of high strength steels. *Materials Science and Technology*, http://dx.doi.org/10.1179/1743284711Y.0000000061, 2012.

[73] B. Mintz, E. Maina, and W. B. Morrison. Influence of dislocation hardening, precipitation hardening, grain elongation and sulphides on fissure formation in HSLA steels having a ferrite/pearlite microstructure. *Materials Science and Technology*, 24:177–188, 2008.

[74] S. Kotrechko, N. Stetsenko, and S. Shevchenko. Effect of texture smearing on the anisotropy of cleavage-stress of metals and alloys. *Theoretical and Applied Fracture Mechanics*, 42:89–98, 2004.

[75] B. Mintz, W. B. Morrison, P. P. Morris, and G. J. Davies. The influence of texture on the tensile and impact properties of controlled steels. In G. J. Davies, editor, *Texture and Properties of Materials*, pages 224–234, London, U.K., 1976. The Metals Society.

[76] H. Inagaki, K. Kurihara, and I. Kozasu. Influence of crystallographic texture on the strength and toughness of the controlled rolled high tensile strength steel. *The Iron and Steel Institute of Japan*, 7:991–1011, 1975.

[77] D. L. Bourell and O. D. Sherby. Texture induced cleavage delamination of warm–rolled low carbon steel. *Metallurgical transactions A*, 14:2563–2566, 1983.

[78] G. J. Baczynski, J. J. Jonas, and L. E. Collins. The influence of rolling practice on nothc toughness and texture development in high–strength linepipe. *Metallurgical & Materials Transactions A*, 30:3045–3054, 1999.

[79] J. B. Ju, J. S. Lee, and J. I. Jang. Fracture toughness anisotropy in a API steel line-pipe. *Materials Letters*, 61:5178–5180, 2007.

[80] D. M. Fegredo, B. Faucher, and M. T. Shehata. Influence of inclusion content, texture and microstructure on the toughness anisotropy of low carbon steels. In *Strength of Metals and Alloys*, volume 2, pages 1127–1132, Oxford, U. K., 1985. Pergamon Press.

[81] R. Kasada, S. G. Lee, J. Isselin, J. H. Lee, T. Omura, A. Kimura, T. Okuda, M. Inoue, S. Ukai, S. Honuki, T. Fujisawa, and F. Abe. Anisotropy in tensile and ductilebrittle transition behaviour of ods ferritic steels. *Journal of Nuclear Materials*, doi:10.1016/j.jnucmat.2010.12.069, 2011.

[82] I. Pyshmintsev, A. Gervasyev, R. H. Petrov, V. C. Olalla, and L. A. I. Kestens. Crystallographic texture as a factor enabling ductile fracture arrest in high strength pipeline steel. *Materials Science Forum*, 702–703:770–773, 2012.

[83] D. W. Kim, R. S. Qin, and H. K. D. H. Bhadeshia. Transformation texture of allotriomorphic ferrite in steel. *Materials Science and Technology*, 25:892–895, 2009.

[84] K.-B. Kang, J. H. Bae, and W.-Y. Choo. Effect of thermomechanical processing parameters on mechanical properties and microstructure of api-x80 grade hot rolled strips. In *4th Int. Conf. on Pipeline Technology*, volume 4, pages 1689–1699, Ostend, Belguim, May 2004.

[85] D. Stalheim, K. Barnes, and D. McCutcheon. Alloy designs for high strength oil and gas transmission linepipe steels. In *Proc. Int. Symp. on Microalloyed Steels for the Oil and Gas Industry*, pages 73–108, Araxa, MG, Brazil, January, 2006 2007. The Minerals, Metals & Materials Society (TMS).