The effect of separations on the assessment of Charpy impact tests

Bradley Jacob Davis

University of Wollongong

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THE EFFECT OF SEPARATIONS ON THE ASSESSMENT OF CHARPY IMPACT TESTS
THE EFFECT OF SEPARATIONS ON THE ASSESSMENT OF CHARPY IMPACT TESTS

Authored by
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Supervised by
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Dr. Guillaume Michal, PhD

A thesis submitted in partial satisfaction of the requirements for the degree of Doctor of Philosophy in the field of Materials Engineering within the Faculty of Engineering & Information Sciences at the University of Wollongong

2017
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DECLARATION

I, Bradley Jacob Davis, declare that this thesis, submitted in partial fulfillment of the requirements for the award of Doctor of Philosophy, in the School of Mechanical, Materials, and Mechatronic, and Biomedical Engineering, Faculty of Engineering and Information Sciences, University of Wollongong, Australia, is wholly my own work unless otherwise referenced or acknowledged. The document has not been submitted for qualifications at any other academic institution.

Bradley Jacob Davis
24 July, 2017
Dedicated to

My Grandparents
who taught me to value the simple things in life

My Parents
for allowing me to pursue my dreams even when half a world away

AND

My Beautiful Sisters
for being my greatest inspiration
O ME! O LIFE!

O Me! O life! . . . of the questions of these recurring;
Of the endless trains of the faithless—of cities fill’d
with the foolish;
Of myself forever reproaching myself, (for who
more foolish than I, and who more faith-
less?)
Of eyes that vainly crave the light—of the objects
mean—of the struggle ever renew’d;
Of the poor results of all—of the plodding and
sordid crowds I see around me;
Of the empty and useless years of the rest—with
the rest me intertwined;
The question, O me! so sad, recurring—What good
amid these, O me, O life?

Answer.

That you are here—that life exists, and identity;
That the powerful play goes on, and you will
contribute a verse.

Walt Whitman, *Leaves of Grass* (1892)
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## Acronyms

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<tr>
<td>AGA</td>
<td>American Gas Association</td>
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<tr>
<td>API</td>
<td>American Petroleum Institute</td>
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<tr>
<td>ASTM</td>
<td>American Society for Testing and Materials</td>
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<tr>
<td>AMS</td>
<td>atomic emission spectroscopy</td>
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<td>BP</td>
<td>base plate</td>
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<td>BCC</td>
<td>body-centered cubic</td>
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<td>BN-DWTT</td>
<td>brittle-notch DWTT</td>
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<td>BG</td>
<td>British Gas</td>
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<td>BTCM</td>
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<td>CN-DWTT</td>
<td>Chevron-notch DWTT</td>
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<td>CSM</td>
<td>Centro Sviluppo Materiali</td>
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<td>C(T)</td>
<td>compact tension</td>
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<td>CTOA</td>
<td>crack-tip opening angle</td>
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<td>CVN</td>
<td>Charpy V-notch</td>
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<td>CZM</td>
<td>cohesive zone model</td>
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<td>ductile-to-brittle transition</td>
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<td>DBTT</td>
<td>ductile-to-brittle transition temperature</td>
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<td>D</td>
<td>diagonal</td>
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<td>DIC</td>
<td>digital image correlation</td>
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<td>DWTT</td>
<td>drop weight tear test</td>
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<td>electron backscatter diffraction</td>
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<td>EDM</td>
<td>electrical discharge machining</td>
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<td>fracture appearance transition temperature</td>
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<td>fatigue-cracked-notch DWTT</td>
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<td>finite element analysis</td>
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<td>finish rolling temperature</td>
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<tr>
<td>FPTT</td>
<td>fracture propagation transition temperature</td>
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<td>FSBT</td>
<td>full-scale burst test</td>
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<td>Description</td>
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<tr>
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</tr>
<tr>
<td>GTN</td>
<td>Gurson-Tvergaard-Needleman</td>
</tr>
<tr>
<td>HAZ</td>
<td>heat-affected zone</td>
</tr>
<tr>
<td>HIC</td>
<td>hydrogen induced cracking</td>
</tr>
<tr>
<td>HLP</td>
<td>High-Strength Line Pipe Committee</td>
</tr>
<tr>
<td>HSLA</td>
<td>high-strength, low-alloy</td>
</tr>
<tr>
<td>HV10</td>
<td>Vickers hardness with 10 kgf</td>
</tr>
<tr>
<td>ISIJ</td>
<td>The Iron and Steel Institute of Japan</td>
</tr>
<tr>
<td>ISO</td>
<td>International Organization for Standardization</td>
</tr>
<tr>
<td>L–Z</td>
<td>longitudinal through-thickness</td>
</tr>
<tr>
<td>L–T</td>
<td>longitudinal-transverse</td>
</tr>
<tr>
<td>LEFM</td>
<td>linear elastic fracture mechanics</td>
</tr>
<tr>
<td>L</td>
<td>longitudinal</td>
</tr>
<tr>
<td>LSE</td>
<td>lower-shelf energy</td>
</tr>
<tr>
<td>ODF</td>
<td>orientation distribution function</td>
</tr>
<tr>
<td>OD</td>
<td>outer-diameter</td>
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<tr>
<td>PTW</td>
<td>part-through-wall</td>
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<tr>
<td>PN-DWTT</td>
<td>pressed-notch DWTT</td>
</tr>
<tr>
<td>PRCI</td>
<td>Pipeline Research Council International</td>
</tr>
<tr>
<td>RUS</td>
<td>rising upper-shelf</td>
</tr>
<tr>
<td>RBT</td>
<td>round bar tensile</td>
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<tr>
<td>Z–L</td>
<td>through-thickness longitudinal</td>
</tr>
<tr>
<td>Z–T</td>
<td>through-thickness transverse</td>
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<tr>
<td>SE(B)</td>
<td>single-edge notched bend</td>
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<tr>
<td>SEM</td>
<td>scanning electron microscope</td>
</tr>
<tr>
<td>SI</td>
<td>separation index</td>
</tr>
<tr>
<td>SE(T)</td>
<td>single-edge notched tension</td>
</tr>
<tr>
<td>Z</td>
<td>through-thickness</td>
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<td>SPC-DWTT</td>
<td>static precracked DWTT</td>
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<td>DASW</td>
<td>drawn arc stud welding</td>
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<tr>
<td>T–L</td>
<td>transverse-longitudinal</td>
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<tr>
<td>T–L1</td>
<td>transverse-longitudinal Charpy specimen with 1 incision</td>
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<tr>
<td>T–L2</td>
<td>transverse-longitudinal Charpy specimen with 2 incisions</td>
</tr>
<tr>
<td>Acronym</td>
<td>Description</td>
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<tr>
<td>T-L3</td>
<td>transverse-longitudinal Charpy specimen with 3 incisions</td>
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<td>T-Z</td>
<td>transverse through-thickness</td>
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<td>TW</td>
<td>through-wall</td>
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<td>TMCP</td>
<td>thermo-mechanically controlled processed</td>
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<td>T</td>
<td>transverse</td>
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<td>TSL</td>
<td>traction-separation law</td>
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<td>USE</td>
<td>upper-shelf energy</td>
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<td>WT</td>
<td>wall-thickness</td>
</tr>
<tr>
<td>WJ</td>
<td>West Jefferson</td>
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</table>
NOMENCLATURE

\( A_\ell \) fracture area of the specimen's ligament
\( A_n \) void nucleation rate coefficient
AR aspect ratio
B Charpy specimen thickness
\( \hat{b} \) tangent unit vector
\( B_{\text{eff}} \) effective thickness
\( B_{\text{fail eff}} \) effective thickness across from failed element
\( B_{\max (\sigma)}^{\text{eff}} \) effective thickness across from maximum \( \sigma_{zz} \) element
\( B^* \) Charpy thickness of a full-sized specimen (10 mm)
C backfill constant
\( c_{\text{eq}} \) equivalent defect length for a PTW defect
\( C_v \) absorbed Charpy energy
\( C_{v100} \) absorbed Charpy energy at 100 \%SA
\( C_{vp} \) absorbed Charpy energy at plateau
KCV specific absorbed Charpy energy
\( C_v^* \) absorbed Charpy energy for a full-sized specimen
D damage scalar
\( D_6 \) tensile specimen with 6 mm diameter
\( D_{12} \) tensile specimen with 12 mm diameter
\%DA percent separation area
DA separation area
\( \delta \) cohesive separation
\( \delta_0 \) critical separation value
\( \delta_1 \) defined separation value in traction-separation law
\( \delta_2 \) defined separation value in traction-separation law
\( \delta_F \) maximum cohesive separation defining fracture
\( \hat{\delta} \) cohesive separation vector

xxix
Nomenclature

E  elastic modulus
E  total fracture energy
ε  engineering strain
εₙ  strain where 50% of inclusions are broken
H  backfill constant modification by Rudland and Wilkowski
εₚ  plastic strain
εₚ  effective plastic strain
ε̇  strain rate
ε̇ₚ  plastic strain rate tensor
ε̇ₚ  effective plastic strain rate
ε  true strain
εᵤ  uniform elongation
f  void volume fraction
f₀  initial void volume fraction
fₖ  critical void volume fraction for void coalescence
fₙ  void volume fraction at failure
fₙ  existing void growth rate
fₙ  void volume fraction where damage is nucleated
fₙ  void nucleation rate
f  void volume accumulation rate
f*  effective void volume fraction
fᵤ  void volume fraction at rupture
G  fracture energy
Γ₀  cohesive energy
Hₐₐ₁ₚₐ  actual backfill depth
Hₙₐₐₚₐₐ  backfill depth used in original BTCM expression (30 in)
κ  void growth multiplier
K  cohesive penalty stiffness matrix
K₀  cohesive penalty stiffness scalar
Kᵢₐₐ  critical stress intensity
ℓ  notched specimen ligament length
L₀  initial engineering gauge length
\( \ell_\perp \) element height
\( L_{\text{ND}} \) linear intercept along ND orientation
\( L_{\text{TD}} \) linear intercept along TD orientation
\( L_{\text{max}} \) maximum load
\( M_P \) bulge factor for a PTW defect
\( M_T \) Folias factor for a TW defect
\( \dot{\mu} \) plastic multiplier
\( \hat{n} \) normal unit vector
\( N \) strain hardening exponent
\( \Phi \) yield function
\( P_a \) arrest pressure at the crack tip
\( P_d \) decompressed pressure at the crack tip
\( \phi \) potential
\( \rho_{\text{SP}} \) separation density
\( q_1 \) constitutive equation coefficient
\( q_2 \) constitutive equation coefficient
\( \mathcal{R} \) material resistance
\( \sigma_{\text{true}} \) true stress
\( \sigma \) stress
\( \sigma_{\text{thermal}} \) thermal stress component
\( \sigma_{\text{athermal}} \) athermal stress component
\( \hat{s} \) tangent unit vector
\( \%\text{SA} \) percent shear area
\( \sigma_{\text{eff}} \) effective scalar stress
\( \sigma_{\text{eq}} \) equivalent von Mises stress
\( \sigma_f \) flow stress
\( \sigma_h \) hoop stress
\( \sigma_h \) macroscopic hydrostatic stress
\( \sigma_{\text{arrest}}^h \) arrest hoop stress
\( s_n \) standard deviation on the nucleation strain
\( \sigma_{\text{max}}^p \) maximum principle stress
\( \sigma_P \) effective plastic stress
\( \bar{\sigma} \) Cauchy stress tensor
\(\sigma'\) Cauchy stress tensor deviator
\(R_m\) ultimate tensile strength
\(\sigma_Y\) yield stress
\(R_{P,0.2}\) yield strength at 0.2 % offset
\(R_{P,0.5}\) yield strength at 0.5 % strain
\(\mathcal{T}\) cohesive stress (traction)
\(\tilde{\mathcal{T}}\) traction tensor
\(T\) temperature
\(T_0\) cohesive strength
\(T_b\) tangential traction
\(\mathcal{T}(\delta)\) cohesive stress (traction) function
\(T_n\) normal traction
\(T_s\) tangential traction
\(u\) displacement scalar
\(\Delta u_{AB}\) differential displacement of two points A & B lying opposite on a cohesive interface
\(u_b\) tangential displacement
\(u_n\) normal displacement
\(u_s\) tangential displacement
\(\tilde{u}\) displacement vector
\(\nu\) Poisson’s ratio
\(V_f\) fracture velocity
There are many people that I must acknowledge for helping me reach this point. Unfortunately, I cannot list them all but know that if you’re reading this, there is a great chance you have impacted my life in some way.

First, I would never have been able to complete this work without my supervisors Dr. Cheng Lu and Dr. Guillaume Michal. When I arrived in Australia, it marked the first time in my life I had ever left the United States. It was a momentous step. Both Cheng and Guillaume took me under their wings and always pushed me to expand my horizons. We have shared in many adventures together. From gluing strain gauges to a pipe in minus thirty-degree weather to deep conversations over coffee in the beautiful Australian weather, we have shared many special moments. I am so grateful to have them as my supervisors and look forward to our future adventures.

Immediately after meeting Cheng and Guillaume, the next person I met, who would become one of my best friends, was YouYou Wu. YouYou was the one to show me around the university and teach me the ins-and-outs of everyday life as a Ph.D. student.

Through YouYou, I met Khoa Vo, and the three of us took on the mantle of the “Three Amigos.” Rare was it to find one of us without the other. We had many laughs together over coffee and dinners at our two favorite stops, Fuku and Sakura Sushi. I am so grateful to have met them and am happy to see
their success continue in their careers.

Studying in Australia would not have been financially viable if not for the Energy Pipelines CRC (EPCRC). I am very thankful for the opportunity they provided by funding me for the duration of my studies. In particular, I’d like to give a special thanks to Valerie Linton, Robert Newton, Klaas van Alphen, and Matthew Byers. I’d also like to acknowledge Frank Barbaro, Mark Fothergill, Phil Colvin, Phil Venton for the tremendous insights on the pipeline industry and the even greater stories.

One of the highlights of my study was the opportunity to join Leigh Fletcher on his annual trip to Baosteel of China, where he serves as a consultant. Leigh has been a great mentor to me throughout the years and has always shown great care in building my knowledge of the pipeline field.

One man who I could always count on for a “good morning” greeting was Mr. Bob Wheway. Not only did we have some interesting conversations, he always shared with me the spoils of the countless meat raffles he won.

Behind the scenes, the guys over at the machine shop, the lab technicians, and the IT staff perform a tremendous service, and I do not want to leave them out.

And finally, I’d like to acknowledge my family, who without none of this would ever be possible. Thank you for putting up with my eccentricities through years. Thank you for always helping me when I needed it, and most of all thank you for supporting me as I pursued my dreams in a far away land.
**Publications**


ABSTRACT

In the pipeline industry, the Charpy impact test is the most common method for determining the fracture arrestability (or toughness) of a line pipe steel. Since the 1960s, line pipe steel has increased in both its stress capacity and toughness through advanced steel making and rolling processes. However, due to the lower finishing temperature, pronounced levels of microstructural anisotropy is generated. One consequence of anisotropy is the formation of weakly bonded planes along the rolling plane of the material. This can result in a phenomenon known as separations. Separations are fissures that form along the rolling plane of the material and lie perpendicular to the fracture plane. Separations are observed on all mechanical tests used to characterize a material’s properties (e.g. tensile, Charpy, DWTT, and full-scale burst tests). While separations have been observed on the fracture surfaces of line pipe steels since the 1960s, little research has been done to determine separations’ influence on the Charpy specimen’s ability to capture the effects of separations on full-scale fracture behavior in pipelines.

In this work, an API grade X80 line pipe steel, showing severe separations was used to quantify separation severity and investigate the reliability of Charpy specimens to accurately represent the separation phenomenon. This was accomplished by performing a number of Charpy tests along the longitudinal, transverse, and through-thickness orientations of the pipe. A set of novel Charpy specimens was also created to investigate the influence of separations
on fracture characteristics. The novel Charpy specimens were created by introducing incisions along the rolling direction of the notched specimens. A set of specimens, containing one, two, and three incisions was manufactured and tested to simulate increasing levels of separation severity. Separation severity has been most commonly measured by the separation index metric. This metric considers the ratio of the total length of all separations to the fracture area. The separation index was compared with other metrics found in the literature as well as a newly defined metric introduced in this study, the separation area.

To further investigate the effect of separations on Charpy impact tests, a finite element model was created containing weak interfaces along the separation plane to simulate the formation of separations. The finite element model used the Gurson-Tvergaard-Needleman material behavior to simulate ductile fracture, while the cohesive zone model with a bilinear traction separation law was implemented to capture the brittle behavior of separations. The evolving stress state and geometry changes during full-sized Charpy impact testing was also evaluated for its influence on separation appearance and compared to Charpy specimens of lesser thickness as well as a similar thickness DWTT.

This study revealed that full-sized Charpy specimens, showing high levels of ductility, do not provide an adequate surrogate for assessing separation severity compared to full-scale fracture behavior. This is primarily due to the through-thickness constraint induced by the laterally expanding region as a result of the striker impact. For separation appearance in small-scale, laboratory tests to mirror that of full-scale fracture tests, a steady through-thickness stress state must be achieved for a majority of the fracture process.
Finite element analysis revealed that the specimen’s aspect ratio plays a prominent role on the striker’s ability to induce an artificial through-thickness constraint on the specimen. Lower aspect ratios provide a greater percentage of the fracture process to be influenced by a steady state through-thickness stress state.

By comparing several separation severity metrics to a newly defined separation area metric, which considers the total area of the specimen consisting of separations, the separation severity can be more easily estimated by the naked eye. The study showed that the separation area and commonly used separation index metric are causally related; however, the separation index requires a more detailed, time extensive process, which makes determining the separation area a more useful alternative.
Part I

Literature Review
High-pressure pipelines are currently the safest and most economical method for transporting hydrocarbon products over large distances. Over the last half-century, the demand for oil and gas products has steadily increased, and thus the desire for large diameter pipelines, operating at high pressures has grown substantially. For the most part, the pipeline network goes unnoticed by its consumers, and pipeline engineers make great efforts to keep it this way. As with many large structures (e.g., bridges, skyscrapers, dams), pipelines only make the news or come into the forefront of people’s mind when the pipeline fails, particularly when the failure is catastrophic. Pipelines containing hydrocarbon products store massive amounts of energy, and small errors can lead to large-scale damages to the environment, infrastructure, wildlife, and humans. Therefore, pipeline engineers make ensuring the pipeline’s safety their chief concern.

One element of pipeline safety management is fracture control. In the event of a fracture, pipelines without the proper material characteristics to arrest
the fracture can have fractures propagating along the pipeline for dozens of kilometers in mere seconds. This was especially the case in the early days of pipeline technology, where the most prevalent failure mode was brittle fracture. Luckily for pipeline researchers, brittle failures of ships in WWII led to a surge in fracture mechanics research, so a great foundation already existed to understand and tackle the problem.

Much of the development in fracture control since the early 1960s has been
directed at determining the required fracture resistance of the material with respect to the intended design and service conditions. Fracture resistance is generally quantified by Charpy impact testing, where a section of material is extracted from the pipe, notched, and impacted to measure the fracture (Charpy) toughness. In a span of four decades, pipe mills went from producing steels with Charpy toughnesses of 27 J to steels today that can exceed 300 J [WR08].

Within the realm of gas pipelines, fracture control aims to prevent brittle fracture, optimize the resistance to fracture initiation, and control the extent of ductile fracture propagation. The aspects of fracture control are addressed by specification requirements related to the fracture toughness during pipe production. Brittle fracture is avoided by specifying a high proportion of percent shear area (%SA) in drop weight tear tests (DWTTs). Fracture initiation resistance is optimized by specifying a minimum Charpy energy. The length of a ductile fracture is controlled by requiring a set proportion of heats meet a high enough energy requirement to ensure the arrest of a propagating fracture (termed the arrest toughness). Carefully controlling for these factors ensures that any incidence of fracture will not propagate for a significant distance.

This chapter details the fundamental precepts behind fracture initiation and fracture propagation control. For brittle initiation and propagation, the use of the drop weight tear test to assess the pipe material’s ability to stop an initial brittle fracture is covered. The through-wall and part-through-wall defect criteria are discussed in relation to controlling ductile fracture initiation. Controlling ductile fracture propagation is described by the Battelle Two-
Curve Model and its subsequent modifications along with a brief introduction to proposed alternatives for the Battelle Two-Curve Model. Finally, the Charpy impact test, drop weight tear test, and full-scale burst test are described to provide a background on the pipeline specific testing methods.

1.1 Fracture Initiation Control

Given that pipelines can span distances greater than 1000 km, are often buried, and can be in remote areas that are seldom visited, ensuring that the pipeline can withstand any defects in the pipe wall is critical. Defects can be introduced into a pipeline during manufacturing, transportation, fabrication, installation, deterioration, or external interference by means of drilling and construction. Therefore, defects are inevitable, so pipeline engineers have derived methods to ensure that a defect does not lead to complete failure of the pipeline.

This section outlines the procedures used to prevent the initiation of a brittle or ductile fracture from defects. The pioneers of these methods at the Battelle Memorial Institute used a combination of fracture mechanics and semi-empirical methods from full-scale burst test data. Their work led to the development of the NG-18 equations [Kie+73], which are still widely applied today.

1.1.1 Brittle Fracture Initiation

From the 1930s to the early 1960s, brittle fracture control presented the greatest challenge to high-energy pipelines. A number of pipelines failed between this
period due to brittle fractures that propagated from 15 km to 60 km [WHS13]. These failures were catastrophic, so pipeline researchers made preventing brittle failure their primary concern.

The original methodology employed to improve the arrestability of brittle fractures started with the American Gas Association (AGA) NG-18 program in 1953 [LE10]. The Battelle Memorial Institute in the United States began studying long, running brittle fractures in pipe burst tests in the early 1960s. In 1961, the first burst tests conducted were WJ hydrostatic burst tests named after the place of the test location in West Jefferson, Ohio, U.S.A. The first full-scale fracture propagation test came in 1963 in Athens, Ohio. Duffy and Maxey [DM65] published the first report on full-scale fracture propagation testing in 1965. Six years later, British Gas (BG) presented their results on the arrest of brittle fracture [FJW71]. The final conclusions showed that once brittle fracture initiates, it propagates at much higher speeds than the acoustic velocity of the pipeline’s contents with typical velocities ranging from 450 m.s\(^{-1}\) to 800 m.s\(^{-1}\) [WHS13]. A pipe which has failed by brittle fracture is often distinguishable by the sinusoidal path the fracture takes, shown in Figure 1.2. If the crack-driving force is great enough, multiple brittle fractures will branch from the main fracture path.

In the 1960s, the general method used to evaluate the fracture toughness of line pipe steel was the Charpy impact test [Lei13]. By testing Charpy specimens over a range of temperatures, the DBT region could be determined. This detailed the fracture toughness as a function of temperature, providing the temperature at which the fracture response was brittle, increasingly ductile, and all the way to a fully ductile response. The DBT curve was used to
determine the DBTT relative to a specified %SA. These experiments would indicate whether the minimum pipe temperature was at or above the required 85 %SA, and if so, the fracture was expected to arrest [Eib13].

During the time the Charpy test was being used to assess brittle fracture control, a 2/3-thickness Charpy specimen (6.6 mm) was being used. This was because originally the 2/3-thick Charpy specimen and DWTT had a similar fracture propagation transition temperature [LE10]. As the wall thicknesses of pipes increased beyond 9.5 mm, the 2/3-thickness Charpy specimen no longer accurately predicted the DBTT compared to the full-scale burst test results. To combat this, Eiber [Eib69a] developed the PN-DWTT, which used the pipe’s full wall-thickness to obtain a one-to-one correlation between the laboratory tests and real pipe’s transition temperature. Eiber [Eib13] determined that the DBT curve was better represented by a full-thickness PN-DWTT when compared to full-scale burst test data. The 85 %SA PN-DWTT criterion worked well for line pipe steels in the 1960s.

Even with the issues that have arisen in the last few decades, brittle fracture control still comes down to ensuring that the 85 %SA criterion is met at a temperature below the pipe’s operating temperature. By changing the pipe’s
operating temperature, the steel’s chemistry, or by selecting a steel with a low DBTT, dynamic brittle failure can be avoided.

### 1.1.2 Ductile Fracture Initiation

#### 1.1.2.1 Through-Wall Defects

The Pipeline Research Council International (PRCI) began investigating ductile fracture initiation resistance in the 1960s\[Max74\]. The formulation of through-wall (TW) defects involved an empirical adaptation based on a substantial number of full-scale burst tests along with the strip yield model originally proposed by Dugdale \[Dug60\]. The expression for the TW defect model is expressed as
Fracture Control

\[
\text{KCV} = \frac{8c\sigma_f^2}{1000\pi BE} \ln \left[ \sec \left( \frac{\pi M_T \sigma_h}{2\sigma_f} \right) \right] \quad (1.1)
\]

where \(\sigma_f\) is the flow stress, \(\sigma_h\) is the hoop stress, \(M_T\) is the Folias factor for a TW defect, KCV is the specific absorbed Charpy energy as energy per unit fracture area, E is the elastic modulus, and \(c\) is the half-length of the TW crack. The Folias factor for a TW defect is determined by

\[
M_T = \left[ 1 + 1.255 \left( \frac{c^2}{rt} \right) - 0.0135 \left( \frac{c^4}{r^2t^2} \right) \right]^{\frac{1}{2}} \quad (1.2)
\]

where \(r\) and \(t\) are the radius and wall-thickness of the pipe, respectively.

\(\sigma_f\) was originally defined as \(\sigma_f = \sigma_Y + 69\) MPa. For high-strength materials, the expression \(\sigma_f = (\sigma_Y + Rm)/2\) can be used, because the original expression can provide a flow stress that is greater than the tensile strength of the material.

The TW defect formulation was validated for full-scale burst tests with yield strengths ranging from 220 MPa to >550 MPa, outer-diameters (ODs) from 218 mm to 1219 mm, wall-thickness (WTs) from 4.8 mm to 19 mm, and Charpy full-size equivalent upper-shelf energies (USEs) from 20 J to 200 J. Figure 1.3 shows the schematic of the TW defect introduced by Maxey [Max74] to calibrate Equation (1.1).

Figure 1.4 shows the relationship between the predicted and observed failure stress for 92 tests reported by Kiefner et al. [Kie+73]. The correlation is impressive, especially when considering that Equation (1.1) has very few tunable parameters and primarily relies on experimental data.
Implied in Equation (1.1) is that as toughness increases, the critical defect size increases asymptotically to a toughness-independent value that is a function of $\sigma_f$ and pipe geometry alone. Therefore, for smaller and thinner pipes, the achievable critical defect size through increasing toughness is intrinsically limited.

1.1.2.2 Part-Through-Wall Defects

Part-through-wall (PTW) defects build on the basic equations of the TW defect criterion. Two adjustments were made to account for the change in defect geometry. The first change is the replacement of the Folias factor for a TW defect by a bulge factor for a PTW defect [18], defined as
Fracture Control

Fig. 1.5: Schematic of PTW defect (adapted from Maxey [Max74]).

\[
M_p = \frac{tM_T - d}{M_T (t - d)} \tag{1.3}
\]

where \(d\) is the effective defect depth and \(t\) is the wall-thickness. The half defect length, \(c\), in the TW equation was replaced by the equivalent defect length, \(c_{eq}\), determined by

\[
c_{eq} = \frac{A}{2d} \tag{1.4}
\]

where \(A\) is the area of the PTW defect.

Combining Equations (1.1), (1.3), and (1.4) provides the PTW defect model expressed as
Equation (1.5) was verified using end-capped, full-scale pipe with defects introduced by machining grooves into the pipe wall. 48 full-scale tests were conducted over a range of pipe diameters and wall-thicknesses [LE10]. Figure 1.5 shows a schematic of the PTW defect described by Maxey [Max74] to confirm Equation (1.5). At the time only line pipe steel grades X52 to X65 with Charpy energy values ranging from 20 J to 69 J were available. Figure 1.6 shows the validation of Equation (1.5) by Battelle’s full-scale tests.

\[
\frac{C_v}{A_\ell} = \frac{8\sigma_{\text{eq}}}{\pi BE} \ln \left[ \sec \left( \frac{\pi M_P \sigma_h}{2\sigma_f} \right) \right]
\]  

(1.5)

where \( A_\ell \) is the ligament area of a Charpy impact specimen.
1.2 Fracture Propagation Control

Pipelines are designed with the aim to prevent fracture initiation, and in the strict scientific definition, the pipe is said to fail if a fracture initiates. Thus, a tremendous amount of research and ingenuity has gone towards preventing fracture initiation. Nevertheless, the expanse of the pipeline network makes fracture initiation an inevitable occurrence.

In the early days of fracture control, the primary focus was preventing a running brittle fracture. Brittle fractures are impossible to stop and will run until the fracture meets a tougher material or fitting. Luckily, cases of running brittle fracture have become quite rare and now researchers focus the bulk of their attention on preventing a running ductile fracture. Ductile fractures can, like brittle fractures, propagate for significant distances, so understanding the mechanism behind the driving-force of a ductile fracture is critical.

The following sections discuss the steps taken to minimize the propagation of a brittle and ductile fracture. The steps to prevent a running brittle fracture are also tied to preventing its initiation. However, preventing a running ductile fracture requires a bit more consideration. In terms of ductile fracture control, this section introduces the Battelle Two-Curve Model (BTCM) and its subsequent modifications. Two alternatives to the BTCM are also briefly discussed.
1.2.1 Brittle Fracture Propagation

Considerations used to prevent brittle fracture propagation will also prevent brittle fracture initiation. One important caveat when talking about brittle fracture is that true brittle fracture in a real gas pipeline is rarely achieved. True brittle fracture would have zero percent shear area on the fracture surface, and the fracture velocity would approach the theoretical limit of the longitudinal wave velocity of steel. If such a fracture initiated, fracture arrest would only be possible if the crack front ran into a tougher pipe material or a fitting. Instead, brittle fracture in gas pipelines may have a %SA between 0 % and 85 % on a DWTT specimen. In other words, the fracture is actually a mixture of brittle and ductile failure, which allows for it to be arrested.

In contrast to ductile fracture, brittle fracture can propagate faster than the acoustic velocity of the gas contained with the pipeline. This avoids the necessity to calculate the gas’ decompression behavior, which is required when determining the arrest requirements for a ductile fracture. Another simplicity is that the backfill conditions do not have to be considered because brittle fracture has such little resistance [WHS13]. This provides a rather simple equation for the crack-driving force. Irwin and Corten [IC68] derived the crack-driving force equation by considering the strain-energy release rate, $G$, for a unit fracture area of pipe, written as

$$g = \frac{\pi r \sigma_h}{E} \tag{1.6}$$

where $\sigma_h$ is the hoop stress, $r$ is the pipe’s radius, and $E$ is the elastic modulus.
Crack arrest occurs when the driving force, $G$, is less than the material resistance, $R$.

With the crack-driving force determined, Maxey et al. [MKE83] set out to quantify $R$ for line pipe steels. The first step was to show that a reasonable correlation existed between the %SA values found in full-scale burst tests and either the DWTTs or Charpy tests. Figure 1.7 shows a comparison between the %SA values from full-scale burst tests and DWTTs test at the same temperature. The DWTT curve provided an adequate agreement between the full-scale test data. The Charpy specimens did not provide the same level of correlation with the full-scale burst tests data. The comparison between the DWTT and Charpy data is provided in Figure 1.8. At the time, most pipe mills did not measure...
Fracture Propagation Control

To solve this, Maxey et al. took the %SA from the DWTT at the desired temperature and multiplied it by the Charpy USE. This gave $R$ in terms of energy per fracture area. Maxey et al. reasoned that because shear lips contribute the most to fracture resistance, this was a valid assumption. In fact, for vintage line pipe steel, their assumption held up well because the %SA was proportional to the USE (Figure 1.9).

Wilkowski et al. [WHS13] have done work to extend the original brittle fracture arrest criterion to high-toughness steels and validating the results against full-scale burst tests. For contemporary line pipe steel with high Charpy energies, brittle fracture does not reveal itself in the same manner in PN-DWTTs as it did for older line pipe steel. Ductile fracture begins at the notch and can suddenly change to brittle fracture. This is known as abnormal fracture (or inverse fracture). Because of this, the 85 %SA may be too high a
requirement to arrest a brittle fracture. Wilkowski et al. [WHS13] found results that may in time be able to provide an alternative shear area requirement for brittle fracture control in modern, line pipe steels.

### 1.2.2 Ductile Fracture Propagation

In the instance of initiation of a ductile fracture, preventing a lengthy propagation becomes necessary. To solve this, Maxey [Max74] developed the Battelle Two-Curve Model (BTCM) to determine the arrest toughness for a gas pipeline. The BTCM quantifies the required arrest toughness for a running ductile fracture based on the decompression behavior of the contained gas and the absorbed Charpy energy.
1.2.2.1 The Battelle Two-Curve Model

The crack-driving force for a running ductile fracture was initially reported to be composed of two components [Ald73; FPR76; Hah+73; ISM74; PRK77]. The first component is the hoop stress, and the second is the decompressing gas pushing on the flaps, forming in the fracture’s wake.

The crack-driving force of a propagating (unstable) ductile fracture is not solely dependent on the decompressing gas but also the material’s fracture toughness; the pipe’s radius, thickness, and strength; as well as the medium surrounding the pipe (e.g., clay).

The BTCM was first reported by Maxey [Max74] in 1974. The BTCM considers two uncoupled processes—the gas decompression curve (or crack-driving curve) and the dynamic crack propagation resistance curve (or crack-resistance curve). Both curves are considered as functions of the gas decompression velocity. While the curves are uncoupled, they both relate to the decompressed pressure ahead of the crack tip. Figure 1.10 provides a schematic of the fracture process concerned in the BTCM.
The Battelle Institute was one of the developers of the code GASDECOM to characterize the gas decompression curve. GASDECOM is valid for lean and rich gases [LE10]. The fracture velocity curve is determined by the semi-empirical equation

\[ V_f = C \frac{\sigma_f}{\sqrt{KCV}} \left( \frac{P_d}{P_a} - 1 \right)^{1/6} \] (1.7)

where \( C \) is the backfill constant, \( \sigma_f \) is the flow stress, \( KCV \) is the specific absorbed Charpy energy, \( P_d \) is the decompressed pressure at the crack tip, and \( P_a \) is the arrest pressure at the crack tip. \( C \) has values of 2.75 and 2.34 for no backfill and soil backfill conditions, respectively. The flow stress is determined by \( \sigma_f = \sigma_Y + 69 \text{ MPa} \). \( KCV \) the fracture toughness as the energy per unit area of a full-sized Charpy specimen at the upper-shelf. The arrest pressure is determined by \( P_a = 2t\sigma_{h,\text{arrest}}/D \), where \( \sigma_{h,\text{arrest}} \) is the arrest hoop stress determined by

\[ \sigma_{h,\text{arrest}} = \frac{2\sigma_f}{3.33\pi} \arccos \left[ \exp \left( -\frac{\pi KCV E}{24\sigma_f^2 \sqrt{Dt/2}} \right) \right] \] (1.8)

where \( E \) is the elastic modulus, \( D \) is the pipe’s diameter, and \( t \) is the pipe’s wall thickness. Equation (1.8) was originally developed for fracture initiation control and based on elementary fracture mechanics methods. Equation (1.8) assumes a linear relationship between \( KCV \) and \( G \).

Figure 1.11 provides a schematic of the BTCM process for determining the required toughness to arrest a propagating ductile fracture. The goal of the BTCM is to find a curve using Equation (1.8) which lies tangent to
the gas decompression curve. Figure 1.11 shows three curves describing the hoop stress required to arrest a ductile fracture being driven by the red gas decompression curve. Because the bottom curve lies below the decompression curve, the fracture will propagate continuously because the fracture velocity is faster than the gas velocity. On the other extreme (i.e. the top curve), the fracture will arrest but pipe will have been over-designed, which inevitably means that costs could have been decreased. The middle curve, representing an eventual arrest, is the ideal condition. Here, the fracture will eventually arrest because the fracture velocity speed is slower than the gas decompression velocity at all pressure levels. Equations (1.7) and (1.8) were originally calibrated against experimental data from the late 1960s to early 1970s. At that time only X52 and X65 line pipe steel grades were involved, so
the BTCM as it was originally proposed has been difficult to apply to modern, high-strength, high-toughness line pipe steels. The following sections describe the efforts made by several researchers to remedy this issue.

1.2.2.2 Charpy Energy Methods for Arrest Toughness Determination

Simplification by Maxey et al.

Maxey et al. [MKE75] developed a simplified expression of the BTCM, allowing for the direct estimate of the required toughness to arrest ductile fracture in a pipeline under normal operating pressures for lean gases in the all-gas phase. Using a curve-fitting process, Maxey et al. defined the arrest toughness in terms of a 2/3-thickness Charpy specimen as

\[(C_v)_{2/3} = 7.2 \times 10^{-3} \sigma_h^2 (rt)^{1/3} \quad \text{[US units]} \quad (1.9)\]

where \((C_v)_{2/3}\) is in ft-lb, \(\sigma_h\) is in ksi, and the pipe radius \((r)\) and thickness \((t)\) are in inches. Equation (1.9) and other simplified forms of the BTCM have been implemented in a variety of codes and standards for gas transmission pipeline design [LE10].

The simplified versions became limited once line pipe steels increased in toughness. The BTCM and its simplifications predicted non-conservative results compared to the measured Charpy energy, specifically when \(C_v\) values exceeded 95 J. So, researchers began to propose improvements to the BTCM for high-toughness line pipe steels. These improvements are discussed in the following sections.
**Leis et al. Correlation**

While working on a project for the Alliance Pipeline, Leis et al. [Lei+98] developed a correction for the BTCM based on the energy dissipation principle. This correlation became referred to as the Leis correction method or Leis factor method. Leis et al. assumed that the arrest toughness determined by Charpy impact tests was equivalent to the BTCM method when $C_v < 95$ J. Otherwise, a correction is applied. The expression takes the form

$$
(C_v)_{\text{arrest}} = \begin{cases} 
(C_v)_{\text{BTCM}} & C_v < 95 \text{ J} \\
(C_v)_{\text{BTCM}} + 0.002 (C_v)_{\text{BTCM}}^{2.04} - 21.18 & C_v \geq 95 \text{ J} 
\end{cases}
$$

(1.10)

where $(C_v)_{\text{arrest}}$ is the full-size equivalent Charpy energy, $(C_v)_{\text{BTCM}}$ is the full-size equivalent arrest energy calculated from the BTCM or a simplified version of the BTCM. Equation (1.10) is applicable to line pipe grades at or below X70 [LE10].

**Eiber Correlation**

Eiber [Eib08a; Eib08b] confirmed the Leis correction method when applied to X70 line pipe steels but not when applied to X80 line pipe steels. Eiber; Eiber modified Equation (1.10) by replacing the coefficient 0.002 with 0.003. With this slight modification, Eiber found the predicted arrest toughness requirements to be in good agreement with the experimental data for X80 line pipe steels. Eiber’s expression is
\[ (C_v)_{\text{arrest}} = \begin{cases} (C_v)_{\text{BTCM}} & C_v < 95 \text{ J} \\ (C_v)_{\text{BTCM}} + 0.003(C_v)^{2.04} - 21.18 & C_v \geq 95 \text{ J} \end{cases} \] (1.11)

**CSM Correlation**

Demofonti et al. [DMR07], working for CSM, proposed a correlation simply expressed as a linear relationship between the arrest toughness and the BTCM prediction. Their correlation was applied to high-toughness steel grades X80 and X100. The expression took the form

\[ (C_v)_{\text{arrest}} = k (C_v)_{\text{BTCM}} \] (1.12)

where \( k \) is a constant factor which is 1.43 for X80 and 1.7 or higher for X100 line pipe steel. This is opposed to the previously mentioned Leis correction method which considers a non-linear relationship in this region. The CSM correction method is only valid for the individual grade in consideration.

**C-FER Correlation**

Wolodko and Stephens [WS06] at C-FER modified the BTCM by applying a statistical correction for line pipe grades from X70 to X100. They proposed the following equation

\[ (C_v)_{\text{arrest}} = (1.5 + 0.29n_{\text{std}})(C_v)_{\text{BTCM}} \] (1.13)
where $n_{\text{std}}$ is the multiplier on the standard deviation of the model error, which achieves the desired probability of non-arrest of running fracture. As an example, if $n_{\text{std}} = 1.0, 1.5, \text{and} 2.0$, the factor between the required arrest toughness and the BTCM prediction would be 1.79, 1.935, and 2.08, respectively.

**Backfill Coefficient Modification**

The original BTCM (Equation (1.7)) considered the backfill as one empirically based value. This does not account for different soil conditions, which can affect the fracture behavior. Rudland and Wilkowski [RW06; RW07] conducted a series of burst tests buried under different backfill depths and soil conditions. Based on their data, the BTCM was modified to

$$V_f = C \frac{\sigma_f}{H \sqrt{K_C V/(P_d/P_a - 1)^{1/6}}}$$  \hspace{1cm} (1.14)

with $H$ being the modified backfill coefficient, defined as

$$H = \frac{0.275H_{\text{actual}}}{H_{\text{nominal}}} + 0.725$$  \hspace{1cm} (1.15)

where $H_{\text{actual}}$ is the actual backfill depth and $H_{\text{nominal}}$ is the backfill depth used in original BTCM expression (30 in). $H$ was only calibrated for line pipe steels with $C_V < 100$ J.
**Velocity-Dependent Toughness Modification**

The original BTCM considered the fracture toughness as a constant material resistance, independent of fracture velocity. However, experimental analysis showed that the material resistance is in fact dependent of the fracture velocity, so Duan and Zhou [DZ09b] and Duan et al. [Dua+10] at TransCanada modified the material resistance expression, considering it dependent on the fracture velocity

\[
R = R_0 V_f^{-\alpha}
\]  

(1.16)

where \( R_0 \) is the reference fracture resistance at \( V_f \) and \( \alpha \) is a velocity-dependent index. The effect of \( \alpha \) on the fracture toughness prediction for an X80 line pipe steel in shown in Figure 1.12. When \( \alpha = 0 \), fracture toughness is considered to be constant as with the original BTCM. When \( \alpha = 0.2 \), the prediction was in good agreement with the full-scale burst tests.

The velocity-dependent toughness methods is consistent with the understanding that dynamic fracture is dependent on the loading rate. The method offers a promising improvement to the BTCM; however, quantifying the reference material resistance and velocity-dependent index remains a challenge.

1.2.2.3 **DWTT Energy Methods for Arrest Toughness Determination**

**Early DWTT Methods**

The Battelle Institute developed the DWTT as the first alternative to the Charpy impact test for identification of the DBTT and fracture propagation resistance
measurement for tougher line pipe steels. Because the DWTT specimen is larger than the Charpy specimen and captures the full-thickness of the pipeline, researchers at Battelle believed the DWTT would prove superior when testing high-toughness, high-strength line pipes steel experiencing large plastic deformation.

Wilkowski [Wil79] and Wilkowski et al. [WME77b] developed a linear correlation between a standard PN-DWTT and a Charpy test expressed as

$$\left( \frac{\mathcal{E}}{A_\ell} \right)_{\text{DWTT}} = 3 \left( \frac{\mathcal{E}}{A_\ell} \right)_{\text{CVN}} + 300 \quad \text{[US units]} \quad (1.17)$$

where \( \mathcal{E} \) is the total fracture energy, \( A_\ell \) is the fracture area of the specimen’s ligament. The ratio \( \mathcal{E}/A_\ell \) represents the specific fracture energy (ft-lb/in\(^2\)).
After determining the minimum $C_v$ to arrest a propagating fracture using the BTCM, the minimum DWTT energy can be determined by Equation (1.17). Figure 1.13 shows the relationship between Charpy and DWTT energies for vintage line pipe grades up to X65.

Fearnehough et al. [FDJ76], working with British Gas, showed that the propagation energy in higher toughness line pipe steels was not linearly related to the Charpy energy as claimed in Equation (1.17). Fearnehough et al. conducted a series of tests using DWTT specimens pre-cracked to varying lengths under quasi-static loading conditions. They then impacted the pre-cracked specimens using the traditional DWTT method. This method became known as the interrupted DWTT. Charpy tests from the same parent material were compared to each interrupted DWTT specimen. Figure 1.14 shows the...
results of their investigations. The figure shows that for lower toughness line pipe steels, the relationship between DWTT and Charpy energies is linear but becomes non-linear as the Charpy energy exceeds 70 J.

Leis [Lei02] investigated the linear relationship between the Charpy and DWTT energies of Equation (1.17) for line pipe grades up to X70. Leis found that most burst test data for the X70 steel used in the Alliance Pipeline deviated from the linear relationship. This caused concern for using a linear relationship between the energies of DWTTs and Charpy tests. Experiments have proved that for Charpy energies great than 100 J, the linear relationship breaks down. Wilkowski et al. [Wil+06] showed that the line pipe grade has a significant impact on the correlation between DWTTs and Charpy tests. The slope of the linear function decreases from 2.94 for X60 steels to 1.91 for X100.
At first, researchers believed that the non-linear relationship might have resulted from the large initiation energy measured by the PN-DWTT. So, researchers at the Battelle Institute modified the PN-DWTT with the brittle-notch DWTT (BN-DWTT). As the name implies, the BN-DWTT begins with a brittle fracture and thus reduced the initiation energy. Wilkowski et al. [WME78b] and Wilkowski et al. [WME77b] implemented the BN-DWTT for line pipe steel grades up to X70. Based on this data, the researchers proposed a curve-fitted non-linear function relating the BN-DWTT to the PN-DWTT

\[
\left( \frac{E}{A_l} \right)_{\text{BN-DWTT}} = 175 \left( \frac{E}{A_l} \right)_{\text{PN-DWTT}}^{0.385} - 1500 \quad \text{[US units]} \quad (1.18)
\]

Another alternative form of the DWTT knows as the SPC-DWTT was also developed at the Battelle Institute. The SPC-DWTT specimen is similar to the PN-DWTT but with a static pre-crack introduced via a three-point bend test until the maximum load is just exceeded. SPC-DWTT specimens have been shown to give similar results as the BN-DWTT [Wil+06].

**Wilkowski DWTT Correlations**

Wilkowski et al. [Wil+06] proposed two correlations to accommodate the non-linear relationship between DWTT and Charpy specimens. Wilkowski et al. assumed that the PN-DWTT and BN-DWTT were equivalent since both specimens have a lower initiation energy compared to the total absorbed energy. The PN-DWTT energy from Equation (1.17) replaced the BN-DWTT in Equation (1.18) to obtain the following prediction for KCV
When comparing Equation (1.19) with full-scale burst test data for line pipe steel grades X52, X60, X65, and X70, Wilkowski et al. [WWR00] found that the equation overestimated the arrest DWTT energy compared to the full-scale data. Wilkowski et al. determined a statistical factor of the overestimation and produced a new prediction

\[
\left( \frac{\mathcal{E}}{A_{\ell}} \right)_{KCV(W1977)} = \frac{175}{3} \left[ \left( \frac{\mathcal{E}}{A_{\ell}} \right)_{DWTT} \right]^{0.385} - 600 \quad [\text{US units}] \quad (1.19)
\]

Equations (1.19) and (1.20) became known as the Wilkowski 1977 correction method and Wilkowski 2000 correction method, respectively. Both equations can be used to predict the required DWTT arrest energy when the minimum KCV is found by the BTCM.

**Kawaguchi DWTT Correlation**

Kawaguchi et al. [Kaw+00] improved the Wilkowski correction methods for grade X80 line pipe steel. Kawaguchi et al. attempted the Wilkowski correction methods, finding that it did not match their test data. So, the researchers proposed the following relationship, which correlates the SPC-DWTT and the PN-DWTT energy densities

\[
\left( \frac{\mathcal{E}}{A_{\ell}} \right)_{SPC-DWTT} = 0.9431 \left( \frac{\mathcal{E}}{A_{\ell}} \right)_{PN-DWTT}^{0.9563} \quad [\text{US units}] \quad (1.21)
\]
Following along the same path as Wilkowski et al., Kawaguchi et al. generated a non-linear correlation between the DWTT and Charpy energies for X80 steels. After combining Equations (1.19) and (1.21) their relation became

\[
\left( \frac{E}{A_\ell} \right)_{KCV} = 0.3144 \left( \frac{E}{A_\ell} \right)_{DWTT}^{0.9563} - 100 \quad [\text{US units}] \quad (1.22)
\]

This relation does not deviate far from a linear relationship since the exponent 0.9563 is close to 1.

**Wilkowski C_v Correlation**

Papka [Pap03] at ExxonMobil proposed an alternative correction motivated by Equation (1.10), using the DWTT correlation data of Wilkowski et al. [WME77b]. Using Equations (1.17) and (1.19) and assuming that \((C_v)_{(W1977)} = (C_v)_{BTCM}\), the DWTT term was eliminated, providing the following expression relating the BTCM predicting toughness with the Charpy toughness

\[
(C_v)_{\text{arrest}} = 0.04133 \left[ 0.138 (C_v)_{\text{BTCM}} + 10.29 \right]^{2.597} - 12.4 \quad [\text{US units}] \quad (1.23)
\]

Wolodko and Stephens [WS06] at C-FER converted Equation (1.23) from US units to SI units, giving the expression

\[
(C_v)_{\text{arrest}} = 0.056 \left[ 0.1018 (C_v)_{\text{BTCM}} + 10.29 \right]^{2.597} - 16.8 \quad (1.24)
\]
where the Charpy energy is in Joules. Equations (1.23) and (1.24) attempt to predict the arrest toughness when the BTCM is used for high-strength line pipe steels. Eiber [Eib08a; Eib08b] used Equation (1.24) to calculate the arrest toughness for X70 and X80 line pipe steels, finding that it overestimated the required toughness for arrest.

**Japan HLP Model**

Makino et al. [Mak+01] and Sugie et al. [Sug+82] worked in parallel with the researchers at the Battelle Institute in the late 1970s to correlate the DWTT energy with BTCM. Working within the HLP under the auspices of The Iron and Steel Institute of Japan (ISIJ), these researchers carried out a considerable amount of research, aiming to develop an extended version of the fracture model within the BTCM, where the material resistance ($R$) was defined by SPC-DWTTs.

The HLP model define the fracture velocity as

$$ V_f = 0.670 \frac{\sigma_f}{\sqrt{R}} \left( \frac{P_d}{P_a} - 1 \right)^{0.393} $$

(1.25)

where the flow stress is defined by $\sigma_f = (\sigma_Y + Rm)/2$, the material resistance is defined by $R = \mathcal{E}/A_t$, $P_d$ is the decompressed pressure at the crack tip, and $P_a$ is the arrest pressure at the crack tip. For the HLP model, $\mathcal{E}$ is determined by the estimated total energy of a SPC-DWTT.

The arrest pressure at the crack tip is determined by
\[ P_a = \left[ 0.382 \frac{t_{\sigma f}}{D} \right] \arccos \left[ \exp \left( -\frac{3.81 \times 10^{-7} R}{\sigma t^2 \sqrt{D t}} \right) \right] \]  \tag{1.26}

where \( D \) and \( r \) are the pipe’s diameter and radius, respectively.

Equation (1.25) was calibrated to full-scale burst test data from X70 line pipe steels. When applied to line pipe grades X80 and above, the HLP suffers the same issues as other BTCM models applied to high-toughness steels.

1.2.2.4 Comparison of Charpy Based Arrest Toughness Correlation Methods

Figure 1.15 compares the arrest toughness predictions of the Charpy based BTCM, using Equations (1.10) to (1.13) and (1.24). Full-scale burst test data for X80 line pipe steel has been compiled from [Gra12]. The CSM and C-FER correlations use \( k = 1.43 \) and \( n_{\text{std}} = 1.5 \), respectively, since these values correspond to the predictions for X80 line pipe steel. The actual arrest toughnesses range from approximately 130 J to 400 J. The CSM correlation (Equation (1.12)) provides a reasonable prediction seeing that only two tests propagated above its predicted arrest toughness. The C-FER prediction (Equation (1.13)) is highly conservative, drastically overestimating the required arrest toughness. The Leis correlation (Equation (1.10)) underestimates numerous test results, spanning a majority of the predicted \( C_v \) values. Eiber’s adjustment (Equation (1.11)) to the Leis correlation improves its predictive capability, which is understandable because Leis’ original correlation was suited for line pipe grades up to X70 and Eiber’s correlation extended it to X80 line pipe steels. The Wilkowski correlation shows promising results below predicted
Charpy energies of 180 J; however, beyond this value, the prediction grows increasingly over conservative.

Figure 1.16 shows similar curves as Figure 1.15 but for full-scale burst tests of X100 line pipe steels [DMR07]. The coefficient for the CSM and C-FER correlations have been changed to $k = 1.7$ and $n_{\text{std}} = 2.0$, respectively, in order to reflect the X100 grade. Figure 1.16 highlight the predicament the BTCM faces. As the line pipe grade increases, previous correlation methods become non-conservative. In Figure 1.15, the C-FER represented an overly conservative prediction, yet in Figure 1.16 the correlation shifts to the least
Fig. 1.16: Actual versus predicted Charpy arrest toughness for full-scale burst tests of X100 line pipe steel, along with the various correlation curves.

non-conservative prediction.

Figures 1.15 and 1.16 show that the BTCM provides a non-conservative estimate when predicting the actual arrest toughness required to stop a propagating ductile fracture. Several authors have attempted to remedy this problem by introducing corrections to the original BTCM, seeking to account for the increasing toughness and strength of line pipe steels. Even with these corrections, there continues to be a great deal of uncertainty surrounding the BTCM approach.
1.2.2.5 Comparison of DWTT Based Arrest Toughness Correlation Methods

Figure 1.17 examines the proposed relationships between PN-DWTT and KCV, using Equations (1.17) and (1.19) to (1.21). The data from various pipeline researchers show a trend akin to the one presented in Figure 1.14, but the prediction methods by Battelle, Wilkowski, and Kawaguchi show a divergence from the actual data as the specific Charpy energy increases. For KCV less than roughly 120 J.cm$^{-2}$, the Battelle and Wilkowski (1977) methods
show an ability to predict the PN-DWTT energy from Charpy tests. Beyond this value (which is equivalent to $C_v = 95$ J) the data trend becomes non-linear and bends downward. So, the linear relationship between PN-DWTT and Charpy specimens is only valid when the PN-DWTT energy is less than 420 J.cm$^{-2}$. Thus, it seems that the relationship between PN-DWTT and KCV is troubled analogous to the troubled relationship between the predicted arrest toughness from the BTCM and the actual arrest toughness when the Charpy energy exceeds 95 J.

While their correlation methods were not meant to be applied to line pipe steel grades great than X80, the upward trend of Wilkowski and Kawaguchi’s curves are concerning. Their correlations might be too heavily based on empirical data and neglect the underlying mechanisms behind the PN-DWTTs and Charpy tests. Therefore, their correlations might not provide a beneficial starting point when attempting to predict the fracture behavior moving forward.

After conducting a number of full-scale burst tests, Demofonti et al. [DMR07] concluded that the DWTT propagation energy seems to sufficiently describe the fracture behavior in the full-scale tests. However, Demofonti et al. found that the differences in absorbed energy between the pipes that arrested and propagated were negligible and for practical application could be within the same scatter level of pipe production. With that mindset, the use of Charpy tests or DWTTs could provide the equal predictive capabilities. Demofonti et al. also found the Japan HLP model to be inaccurate when determining arrest toughness for modern line pipe steels.
1.2.2.6 Alternatives to the Charpy and DWTT Based Methods

Several alternatives have been explored to address the Charpy test and DWTT inadequacies in predicting the fracture arrest behavior for high-strength, high-toughness line pipe steels. The issue has been ongoing for many years, going back to late 1970s when Wilkowski et al. [WME78b] addressed the problem.

Alternative approaches face an uphill battle in becoming adopted industry wide. For example, the Charpy test is over 100 years old, but it took 25 years to develop into a draft standard [TRS02], and then possibly another 30 years to become a full ASTM standard [Lan78]. Furthermore, the number of tests that must take place during the production of the plate all the way to forming the pipe is massive. For example, the CSA Z245.1 Standard requires Charpy tests and DWTTs once per heat of steel used. In addition TransCanada’s proprietary specifications require weld and HAZ Charpy tests every production shift. For 300 km of large-diameter pipe, this would equate to approximately 700 DWTTs and 700 Charpy tests for the pipe body, welds, and HAZ. But, there is more—a single DWTT value is the average of two specimens and a Charpy value is the average of three specimens for each specific Charpy test. Therefore, in this particular case, over 3500 specimens would need to be prepared, tested, and analyzed during the normal production run [WR08]. So, any new testing method must provide an immediate, simple, and economic benefit to replace Charpy testing and DWTTs.

An alternative approach known as the crack-tip opening angle (CTOA) is discussed below.
The Crack-Tip Opening Angle Criterion

Kanninen et al. [Kan+79] were the pioneers of the CTOA criterion. They performed the earliest experiments and related FEA simulations to study the CTOA criterion in a thin-walled specimens with simulated, stable crack growth. They found that after a brief starting phase, the CTOA decreased from a high initial value to a lower constant value in a steady-state condition. This meant that cracks propagate at an approximately constant angle over a wide range of crack growth. This excited pipeline researchers who saw the CTOA criterion as a valid candidate to describe a majority of the fracture propagation in a pipeline.

According to the CTOA fracture criterion, the steady-state propagation of a ductile fracture is impossible if the maximum driving force expressed by the CTOA parameter is less than the critical value of the material toughness also determined by the CTOA criterion. The CTOA criterion is similar to the concept of the BTCM but offers an alternative method for determining the required arrest toughness.

The CTOA’s use as a parameter to predict the dynamic, ductile fracture arrest in a pipeline began in the late 1980s. PRCI sponsored a large project, seeking to develop the CTOA criterion as an adequate method to predict the arrest of a running ductile fracture [Kan+92]. This work allowed for a more comprehensive and theoretical approach to describe and predict the ductile fracture process by means of the CTOA parameter. Based on this work, the code PFRAC was developed to determine the crack-driving force in terms of the CTOA parameter as a function of the fracture velocity. Also developed
was a test method to determine the critical CTOA, using two modified DWTT specimens with different initial crack lengths [DVK95].

Roughly 10 years later, Berardo et al. [Ber+00] developed the code PICPRO at CSM by using the CTOA parameter and a cohesive zone model. The developed method proposed that the CTOA could directly relate the crack-driving force to the material resistance, which could then be used to predict the arrest or propagation of a fracture in gas pipelines. Unfortunately, issues arose in measuring the CTOA from a test specimen, understanding the mesh sensitivity of the FEA model, and accounting for the crack-tip singularity [JR97].

More recently, researchers have placed their focus on measuring the critical CTOA value for line pipe steel, using various fracture specimens [Ama+13; And+02; Dar+08a; Dar+08b; Sht+04; XTS13]. The methods used to measure the critical CTOA have resulted in conflicting measurements. Some test methods suggested that the CTOA was sensitive to the ligament length, while others did not [DZ09a], but the CTOA was instead sensitive to the fracture velocity.

Erdelen-Peppler et al. [EHK09] investigated the limits of the existing fracture arrest models. They found that an increase in ligament length led to a decrease in the CTOA. Because of this finding, a criterion had to be developed that could translate the CTOA measured in a laboratory setting to the fracture of a real gas pipeline. Comparison between laboratory and full-scale tests showed large deviations when applied to high-strength line pipe steels. Therefore, they concluded that the fracture propagation issue for modern line pipe steel was not solved with the CTOA approach, and thus empirical corrections
and/or correlations would still be needed.

The CTOA is an engineering fracture parameter within the fracture mechanics methods. Currently, there is no theoretical solution for a CTOA-based crack-driving force for fracture specimens of pipes. To use the CTOA approach, a FEA simulation has to be built with a propagating fracture, where the CTOA driving-force can be calculated. Such a simulation would require a robust FEA model [Zhu15].

1.3 Material Testing

This section will highlight the three industry standard testing methods used to quantify the required fracture toughness to arrest a running ductile fracture. The Charpy, drop weight tear test (DWTT), and full-scale burst test (FSBT) are three material verification test of vastly different size scales.

1.3.1 Charpy Impact Tests

The Charpy impact test has been in use for over a century to characterize a material’s resistance to fracture. The Charpy impact test is an impact-pendulum test method used to fracture notched, rectangular specimens, measuring the energy required to do so. The Charpy impact test used today is associated with the apparatus suggested by Russell [Rus00] in 1898 and Charpy [Cha00] in 1901. The Charpy impact test seems to have taken on Charpy’s namesake in the first half of the 20th century because of his technical
contributions and leadership in developing the procedures that would allow for the test to become the robust, engineering tool it is today [TRS02].

The apparatus found in Charpy’s original paper is shown in Figure 1.18. The process for conducting a Charpy impact test can be summarized by: (1) the specimen is loaded onto a small platform, where the notch is centered between two anvils; (2) the striker is drawn back to a predetermined angle, where the potential energy is known; (3) the striker strikes the Charpy specimen, fracturing it; (4) the final height of the pendulum is measured, which gives the energy lost in the fracturing process; and finally (5) multiple specimens can be tested over a range of temperatures to determine the ductile-to-brittle transition temperature (DBTT).

For the pipeline industry, the Charpy impact test serves as a cost-effective and simple test to determine the arrestability of a line pipe steel. The specimens are relatively easy to prepare and many pipe mills have the capability to
manufacture multiple specimens at once. The Charpy specimen’s dimensions are small enough that they can generally be used along all pipe orientations with the notch placed on any plane. Charpy tests are used as a quality control method to compare different heats of the same steel. As discussed previously, the Charpy test is seminal to the evaluation of a pipe’s ability to arrest a running ductile fracture.

Two prominent standards dictate the testing procedures and geometry of the Charpy system. The ASTM and International Organization for Standardization (ISO) Charpy standards determine the notch, rectangular, striker, and anvil geometry. Little difference exists between the two standard in terms of the
specimen and anvil dimensions; however, the striker dimensions are quite distinct. Figure 1.19 show the standard dimensions according to the ASTM and ISO standards. Seen in the figure are the two striker dimensions. The ASTM striker has an 8 mm radius, while the ISO striker has a 2 mm radius. For Charpy energies below 200 J, the striker type does not show a significant difference. However, above 200 J the ASTM striker consistently provides a larger energy value [Luc08].

Charpy testing is used in the pipeline industry to quantify two important measures—the upper-shelf energy (USE) and the DBT curve. A schematic of these results are shown in Figure 1.20. The lower-shelf region is used to describe fracture that is predominantly brittle where the failure is spontaneous and stress controlled. In the DBTT region, the fracture mode is a mixture between ductile and brittle. Consistently predicting the failure mechanism in
this region is difficult because it is statistically controlled. The upper-shelf region represents a fracture mode that is mostly ductile where the failure mode is a strain controlled, thermally activated process. In the upper-shelf region, the fracture is governed by the development of microvoids, which evolve from inclusions in the material.

Depending the specification and/or the pipe’s wall-thickness, different subsized specimen can be used for Charpy testing. A full-sized Charpy specimen corresponds to a specimen geometry with a thickness of 10 mm and ligament length (length under the notch) of 8 mm. A full-sized Charpy specimen has a fracture surface area, \( A_{\ell} \), of 80 mm\(^2\). As the full-sized specimen is transitioned into sub-sized specimens, only the thickness is modified. These are not subsized specimen in the true sense of the word because the fracture surface area is not linearly related. This, along with a mixture of flat and slant fracture on a broken Charpy sample, has given rise to the need to use a power relationship to convert sub-sized specimens to full-sized specimens and vice-versa [Fer78; Tak+09; Tow86a; Tow86b; Wal01; WKS16].

### 1.3.2 Drop Weight Tear Tests

The pressed-notch DWTT (PN-DWTT) was developed by the Battelle Memorial Institute in the 1960s in conjunction with AGA’s NG-18 Research Program [Cos+09]. Battelle developed the PN-DWTT to overcome limitations with the Pellini drop-weight test, developed at the Naval Research Laboratory in Washington, D.C., U.S.A. The DWTT was developed as an alternative to the Charpy impact test in order to provide a better correlation with full-scale
burst test. As the wall-thickness of pipelines grew, the validity of the Charpy specimen to determine whether a pipe could arrest a brittle fracture came into question [Eib65].

The standard DWTT specimen dimensions are shown in Figure 1.21 along with the striker and anvils. Because of the circumferential geometry of pipelines, procuring a perfectly flat specimen is not always possible. So, specifications allow for curvature of the specimen, termed a “gull-wing” specimen. A “gull-wing” specimen is flattened in the regions to increase the specimen’s stability, but the center portion of the specimen is left curved to avoid any material changes due to plastic hardening.

The DWTT specimen is impacted vertically by dropping a weighted striker from a drop tower, hence the name. The specimen can also be impacted by a pendulum machine similar to the Charpy impact test, only significantly
After the DWTT is broken, the fracture surface is evaluated for its percent shear area (%SA), which is a measure of ductility. From Battelle’s full-scale burst test results, they concluded that the determining factor of the fracture velocity and appearance (cleavage or ductile) was the steel’s temperature relative to its fracture propagation transition temperature (FPTT) [DM65]. FPTT represents the point where the fracture mode transitions from ductile to brittle as the temperature is lowered. The DWTT specimen’s FPTT was shown to correlate well with the same measure on the FSBT (Figure 1.22) [Eib69b].

The FPTT is generally specified as the temperature where the %SA is 85 %. This is a well-established criterion to ensure that brittle fracture will be avoided if the 85 %SA is met at the minimum operating temperature of the pipeline [MGM07]. Figure 1.23 shows a schematic of the DBT for a DWTT, identifying the 85 %SA criterion.

The applicability of the DWTT method has also been demonstrated for larger
diameter, higher grade, and thicker wall-thicknesses than the original Battelle tests [Dem+98]. In the case of controlled-rolled line pipe steels, an alternative to the PN-DWTT was developed, known as the Chevron-notch DWTT (CN-DWTT) [MB91]. Three other forms of DWTTs have also been used—the brittle-notch DWTT (BN-DWTT), static precracked DWTT (SPC-DWTT), and fatigue-cracked-notch DWTT (FCN-DWTT). The BN-DWTT was designed to ensure that the DWTT specimen initiated in a brittle manner by depositing a brittle weld around the notch [FDJ77; WME77a; Yam+75]. The SPC-DWTT is a DWTT specimen that is quasi-statically loaded up to the maximum load to initiate a pre-crack [Jun+77]. The FCN-DWTT was developed to investigate line pipes with heavy wall-thicknesses [Kaw+77].
1.3.3 Full-Scale Burst Tests

Full-scale fracture propagation burst tests, or full-scale burst test (FSBT) for short, are large-scale fracture tests designed to establish the conditions under which a running ductile fracture will arrest. The test is placed under conditions meant to resemble the actual conditions the pipe would experience during its lifespan. FSBTs are generally used for projects where the calculated arrest toughness is high, and thus a FSBT is necessary to validate the required arrest toughness. The approach of FSBTs is included in Annex G of ISO 3183 (2012) [ISO12]. Typically, pipe with a range of toughness are installed in the “burst test section” of the FSBT, with the toughness increasing away from the fracture initiation site. The required toughness is established by noting the location where the fracture arrested and relating it to the charpy energy of the arrest pipe. FSBTs are generally required when the design specifications are outside the existing database of test result. However, because of the enormous expense of FSBT, they are rarely carried out and only justifiable for project with a significant scope [WR08].

Figure 1.24 shows a schematic of a FSBT described by Johnson et al. [Joh+00]. This schematic was used as part of the Alliance Pipeline project commissioned in 2000, which carries natural gas from British Columbia, Canada to Chicago, Illinois, U.S.A. Shown in the diagram are the initiation point, the burst test section, timing wires to measure fracture velocity, as well as the system used to pressurize the pipe.

FSBTs are highly valuable because they provide an opportunity to relate small-scale, laboratory tests such as Charpy and DWTTs to a real-world
pipeline fracture.

1.4 Summary

This chapter focused on the aspects of fracture control as they are applied to pipelines. Fracture control is divided into two areas: fracture initiation control and fracture propagation control. Within both areas, fracture control must account for brittle and ductile fracture modes. Brittle fracture propagation has been solved and now pipeline researchers and engineers focus their efforts on perfecting the understanding of ductile fracture propagation. Over the years, the methods for determining the material characteristics to arrest a running ductile fracture have become less systematic. Thus, a number of
authors have proposed alternative methods for determining the required material characteristics for ductile fracture arrest.

Also included in this chapter was the background and testing methods of the two common fracture toughness characterization methods—the Charpy impact test and the drop weight tear test. Both tests have their advantages and disadvantages, which are briefly discussed. The full-scale burst test is briefly described. In situations where the fracture arrestability of a pipeline is uncertain, this test is used for verification.
HE poised his spear as he spoke, and hurled it from him. It struck the sevenfold shield in its outermost layer—the eighth, which was of bronze—and went through six of the layers but in the seventh hide it stayed.

– Homer

The Iliad

In this work, separations, also known as splits or separations, refer to cracks that form along the rolling plane of line pipe steels. Separations have been observed during mechanical/integrity tests including: tensile, Charpy V-notch (CVN), DWTTs, compact tension (C(T)), single-edge notched tension (SE(T)), hydrostatic, West Jefferson, and full-scale burst tests. Separations are most commonly observed in high-strength, low-alloy (HSLA) line pipe steels that feature a pronounced microstructural anisotropy, resulting from the rolling process. Separations have been attributed to elongated ferrite grain structures [BM77; Kim+08; MMM08; MMM07; Tok+12], a high dislocation density within ferrite grains [MM88], high amounts of sulfur and phosphorus in the steel [Alm70; MM88; Yan+08a], the presence of coarse ferrite grain patches in between the matrix of fine ferrite grains [PSD12], ferrite-pearlite banded microstructures [Alm70; HEE75; Yan+08a], microstructural banding with variations in crystallography between adjacent bands [Joo+12], cube
fiber textures [Bou83; KDD82; Sch+74], and most recently through-thickness texture bands and cube texture clusters on the separation plane [Gho+16]. Separations have been observed in lower grade American Petroleum Institute (API) X60 to higher grade X80 and X100 line pipe steels.

HSLA line pipe is primarily produced from thermo-mechanically controlled processed (TMCP) steels, which are capable of achieving excellent strength and toughness combinations. The drawback of these steels is that they possess a pronounced microstructural and metallurgical anisotropy that complicate the analysis of a material’s mechanical properties [Sha13; Tan81]. This is due in part to the formation of inhomogeneous distribution of inclusions and segregations with banded and elongated grain structures.

There have only been a couple reports of separations causing failure within a pipeline, during full-scale burst tests [Car+13; Pys+14]; therefore, separations do not represent an imminent threat to pipeline failure [KE02]. In the instances where separations have led to pipeline rupture via ductile fracture [Tra94], separations were created by the diffusion of atomic hydrogen at inclusions within the steel during normal operations—the mechanism known as hydrogen induced cracking (HIC) [ASM05; Lam96].

Where separations present the greatest challenge is in the evaluation of fracture toughness by means of CVN, DWTTs, C(T), and SE(T) tests. These tests provide the means to measure fracture resistance, which are affected by specimen geometry. Given that separations change the fracture surface from a continuum to a group of disparate surfaces (shown in Figure 2.1), the measurement of fracture resistance becomes complicated with the onset
of separations. Thus, evaluating ductile fracture in specimens containing separations and their implications on the safety assessment of line pipe steel is critical [RH15].

The separation phenomenon within the realm of line pipe steels has been studied since the 1970s [BM77; GK75; Mar72; Sch+74]. Since that time many researchers have contributed significantly to understanding the genesis and mechanisms behind separations. This has been done through fracture surface observations in tensile and impact specimens [BB78; BM77; Ino+09; Mor75; SPR05; Son+06; Yan+09]. Such studies revealed separations to have an impact on the DBT behavior by lowering the transition temperature and USE of the material as the number of separations increase. Guo et al. [Guo+02], Shin et al. [Shi+09a; Shi+09b], and Hong et al. [Hon+11] added to this knowledge by investigating the separation toughening mechanisms in X70 and X80 line pipe grade steels. They indicated a potential strong interaction between separating weak, transverse planes and a reduction in stress triaxiality within the crack
Separations

Fig. 2.2: Separation types termed by Embury et al. [Emb+67] and Rao et al. [RYR89]

front region due to separation formation. Using a modified form of a CVN test, Pyshmintsev et al. [Pys+14] provided evidence for a reduction in resistance to ductile fracture propagation associated with separation within X80 line-pipe steel. The research of Pyshmintsev et al. show a coupling between separation onset and cleavage or quasi-cleavage fracture identified with grain boundary embrittlement resulting from precipitates formed during hot-rolling.

Separations have also been studied in materials outside the realm of pipelines. Of course, the most common of these is composite materials [Tay03]. Separations have also been studied in ceramic materials, where they have been shown to improve toughness [Cle+90]. Within the metallic realm, Rao and Ritchie [RR89a; RR89b] and Rao et al. [RYR89] discovered separations’ influence on an increase in fracture toughness properties along the L–T orientation for Al–Li alloys at low temperatures. Pilhagen and Sandström [PS13] examined separations’ influence on measured fracture toughness of hot-rolled duplex stainless steels using traditional single-edge notched bend (SE(B)) specimens tested at low temperatures.

Three separation orientations exists during testing of notched specimens—crack divider, crack arrester, and crack separation [Emb+67; RYR89]. Figure 2.3
shows the three separation orientations within a notched Charpy specimen. The crack divider type is associated with specimens oriented along the L–T and T–L orientations. The crack arrester type results from specimens oriented along the longitudinal through-thickness (L–Z) and transverse through-thickness (T–Z) orientations. Lastly, the crack splitting type occurs when specimens are tested along the Z–L and through-thickness transverse (Z–T) orientations. Each of these separation orientations affects the resulting fracture toughness of the specimen in different ways. This work focuses on the crack divider and crack splitting types, but all three are discussed here to provide perspective.

The crack divider type of separation is the most commonly observed during Charpy testing of line pipe steel (Figure 2.1). When evaluating the required Charpy toughness to arrest a running ductile fracture using the BTCM, Charpy specimens oriented along the L–T direction are required. Because of this, understanding the impact separations have on the crack divider type is critical. Crack divider separations reduce the stress triaxiality for each separation
in the newly formed ligament by reducing the through-thickness constraint [BB78]. This takes what can begin as a plane strain stress state and transform it to a plane stress state.

The specimen orientations required to get a crack arrester type separation are rarely tested in the pipeline industry. Therefore, only a brief overview of this separation type will be discussed. Crack arrester separations prevent the advancement of the primary fracture front and can lead to crack turning, shown in Figure 2.3. In line pipe steels, this would provide a misleadingly high fracture toughness value.

The crack splitting type exposes the weakness planes by placing the notch orientation along the separation plane. With respect to pipelines, the crack splitting type comes from specimens oriented along the Z–L and Z–T orientations. These orientations make the fracture susceptible to failure by weak cleavage planes (separations) or segregation bands [Su+16].

Only one numerical model has been performed for line pipe steels, accounting for the effect of separations in a notched specimen [RH15]. Kalyanam et al. [Kal+09] looked at separation cracking in an Al–Li to characterize the crack front stress-strain fields in three-dimensional, small-scale yielding FEA models with and without separations. Inspired by the work of Kalyanam et al. [Kal+09], Ruggieri and Hippert Jr. [RH15] performed a set of numerical investigations on the crack front fields and effects of the crack-tip constraint in conventional fracture specimens. Specifically, Ruggieri and Hippert Jr. evaluated separation’s effect in side-grooved and plane-sided C(T) and clamped SE(T) specimens of X70 line pipe steel.
The following sections summarize the knowledge of separations up to this point with respect to the microstructural origins to the effect separations have on the stress state. Also, discussed is the current understanding of separation’s effect on impact test specimens related to the pipeline industry, as well as the efforts made to quantify separation severity.

2.1 Origins and Consequences of Separations

Ferritic steels show a marked decrease in their fracture toughness values when falling below the DBTT. Steel mills have employed many techniques to reduce the DBTT well below the operating temperature of the structure, thus ensuring the fracture resistance will benefit from the maximum toughness the material can offer. To improve the fracture toughness characteristics, techniques including grain refinement [Kim+08; Mor08; SPR05; TKK01; TO86; Tsu+04; Wan+08], minimizing inclusion and impurities [IG06; Mor08], Ni alloying [Gar86; Mor08], and separation toughening [Car+03; Kum+86; SP96] have been employed.

Separations ability to enhance the impact toughness has been reported for fibrous texture [Ino+09; KIT13; Kim+07; Kim+08; Mor08] and layered structure materials [Car+03; Kum+86; Mor08; RYR89; SP96] produced in mechanical rolling. For example, austenitic stainless steel [ZL96], mico-laminated dual phase steel [Zha+14], and low-carbon steel banded with alternate layers of ferrite and pearlite [SP96; Sha+14], and martensitic steel with a layered structure [Sun+16] show a distinct increase in fracture toughness even at low temperatures. Separation is seen as one of the most effective methods to
Table 2.1: Summary of Proposed Microstructural Causes for Separations [Gho+16]

<table>
<thead>
<tr>
<th>Proposed Cause</th>
<th>Material</th>
<th>Failure Mode</th>
<th>Reference(s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>elongated ferrite grain structure</td>
<td>low-carbon, ferrite-pearlite steel, Nb–Zr alloy</td>
<td>intergranular</td>
<td>[BM77; Kim+08; MMM08; MMM07; Tok+12]</td>
</tr>
<tr>
<td>high dislocation density within ferrite grains</td>
<td>low-carbon, ferrite-pearlite steel</td>
<td>intergranular</td>
<td>[MM88]</td>
</tr>
<tr>
<td>high amount of sulfur and phosphorus</td>
<td>dual phase steel</td>
<td>intergranular</td>
<td>[Alm70; MM88; Yan+08a]</td>
</tr>
<tr>
<td>presence of coarse ferrite grain patches in between the matrix of fine ferrite grains</td>
<td>low-carbon, ferrite-pearlite steel</td>
<td>transgranular</td>
<td>[PSD12]</td>
</tr>
<tr>
<td>ferrite-pearlite banded microstructure</td>
<td>HSLA steel</td>
<td>intergranular</td>
<td>[Alm70; HEE75; Yan+08a]</td>
</tr>
<tr>
<td>microstructural banding with variations in crystallography between adjacent bands</td>
<td>HSLA steel</td>
<td></td>
<td>[Joo+12]</td>
</tr>
<tr>
<td>cube fiber texture</td>
<td>low-carbon, ferrite-pearlite steel</td>
<td>transgranular/intergranular</td>
<td>[Bou83; KDD82; Sch+74]</td>
</tr>
<tr>
<td>through-thickness texture band and cube texture cluster on fissure plane</td>
<td>low-carbon, ferrite-pearlite steel</td>
<td>ductile fracture on main fracture plane, cleavage on fissure plane</td>
<td>[Gho+16]</td>
</tr>
</tbody>
</table>

Improve fracture toughness when separations can be used to mitigate brittle fracture.

The debate of the root cause behind separation formation has been ongoing.
for many years [Alm70; Gho+16; Joo+12; Kim+08; MM07; Sch+74; SPR05; Yan+08a]. Mintz et al. [MMM07] stated that the early stages of separation formation is always ductile, initiating from inclusion/carbide particles seated along the grain boundaries. Studies such as those performed by Almond [Alm70], Mintz and Morrison [MM07], and Song et al. [Son+05] reported that separation formation is primarily ductile with a small amount of cleavage cracking. In contrast, Bourell [Bou83], Joo et al. [Joo+12], Punch et al. [PSD12], Schofield et al. [Sch+74], and Yang et al. [Yan+08a] showed that separation’s mode of fracture is cleavage, resulting from separation along (001) cleavage planes.

Table 2.1 summarizes the causes given by many authors. Seven different causes have so far been proposed: (1) elongated ferrite grain structures, (2) high dislocation density within ferrite grains, (3) high amounts of sulfur and phosphorus, (4) cube fiber textures, (5) presence of coarse ferrite grain patches in between the matrix of fine ferrite grains, (6) ferrite-pearlite banded microstructures, (7) and through-thickness texture bands and cube texture clusters on the fissure plane.

Separations can develop by decohesion at the grain boundaries of deformed ferrite grains, which have an elongated structure in line with the rolling direction [BM77; MMM07] and having a high dislocation density [MM88]. Interactions between dislocations and grain boundaries weaken the grain boundaries, creating a favorable path for separation development [MM88]. High amounts of S and P is steels promote the development of coarse sulfide and phosphide inclusions (e.g. MnS and FeP), which can weaken grain boundaries and promote ductile fracture [Alm70; Yan+08a]. Microstructural
Separations banding has also been reported to lead to separation formation because of the strain incompatibility between adjacent hard and soft phases, promoting separations [Alm70; Has+14; HEE75; Yan+08a].

Bourell [Bou83] suggested that the presence of cube texture (ND)∥⟨001⟩ could be the primary cause for separation formation (ND is the normal direction or the strip/plate thickness direction). Bourell assumed that the weakness of the separation planes emanates from the preferential alignment of ⟨001⟩ planes of the crystals parallel to the rolling plane. Haskel et al. [Has+14] showed that the existence of {011}⟨001⟩ and {110}⟨111⟩ crystallographic orientations contribute to separation development because of the large differences in Taylor factor as well as the elongated grain and banded microstructure. Inoue et al. [Ino+09] attributed separation development to existence of α-fiber texture (RD)∥⟨110⟩ promoted during the warm caliber rolling of ferritic steel (RD is the rolling direction). Ghosh et al. [Gho+16] looked at origin of separations where strong crystallographic texture exists, finding that a high amount of inter-critical deformation led to banding composed of gamma (ND)∥⟨111⟩ and cube (ND)∥⟨001⟩ texture. Ghosh et al. concluded that separation cracks propagated through the main fracture plane in a ductile manner along the interfaces between the previously mentioned bands due to strain incompatibility. Nonetheless, many studies have concluded that crystallographic texture is not responsible for the origin of separations [Alm70; BJC99; HEE75; MM88; PSD12; Shi+09a].

One commonality among all the proposed origins for separations is that they form during the testing of steels which have been extracted from plate with a FRT below the range at which, during cooling, the transformation from
austenite to ferrite begins (generally denoted $A_r$) [Pax80].

The following sections highlight some of the proposed causes of separation formation.

### 2.1.1 Elongated Grain Structures

#### 2.1.1.1 Bramfitt and Marder (1977)

Grain elongation in hot strip steels results from low FRTs, causing different textures to form. Separations have been observed in steels where FRT sat within the two-phase ($\alpha + \gamma$) region [SP96]. When the FRT exceeds the two phase region, the likelihood of separations diminish.

Bramfitt and Marder [BM77] studied the effect of FRT on the separation characteristics of a vacuum-induction melted, high-purity Fe 1\%Mn alloy. Four 225 kg ingots were rolled into 16, 100 mm$^2$ slabs. The slabs were reheated from 1100 $^\circ$C to 1250 $^\circ$C for 2 hours and rolled into 12.7 mm thick plates in 10 passes. The plates were subjected to FRTs from 960 $^\circ$C down to 150 $^\circ$C. Separations related to pearlite banding, carbides, and inclusions were avoided through material processing.

DBT evaluations were carried out for all 16 as-rolled plates. Bramfitt and Marder found that plates with FRTs below 760 $^\circ$C exhibited separations along the rolling plane of Charpy specimens. Figure 2.4 shows the resulting DBT curves for four FRTs: 960, 707, 538, and 316 $^\circ$C. Separations were seen to become more severe as the FRT decreased, which explains the change in fracture behavior. Charpy specimens extracted from the 906 $^\circ$C finish rolled
plate showed a sharp ductile-to-brittle transition around $-4^\circ C$. As the FRT decreased, the DBT zone disappears and the maximum absorbed Charpy energy decreases. Bramfitt and Marder also observed that separation severity increased as the Charpy test temperature decreased. This was the case until the specimen fractured in a predominantly cleavage manner.

Figure 2.5 details example Charpy surfaces for specimens taken from plates at FRTs described in Figure 2.4 with the specimens tested at three different temperatures. Evident in these photographs is the increase in separation severity as the FRT decreased, as well as an increase in separation severity as the test temperature decreased. For Charpy specimens with a FRT around 960 $^\circ C$, the specimen at the USE range did not fully fracture. With a FRT of 707 $^\circ C$, two separations are present for the same test temperature as the 960 $^\circ C$ FRT. At 538 $^\circ C$ FRT the separation count increases to four and at 316 $^\circ C$ FRT the count increase to approximately nine. Bramfitt and Marder also noted that the separation severity was greater for Charpy specimens tested along the T-L
Fig. 2.5: Charpy fracture surfaces for four FRTs tested at three different temperatures (adopted from [BM77]).

orientation compared to the L–T orientation for the same test temperature.

The lowering of the DBTT was explained by the work of Embury et al. [Emb+67]. When separation-type fracture exists, the DBTT is lowered until cleavage fracture controls the fracture process. While there is an improvement in the DBTT with lower FRTs, there is a loss in the USE. Bramfitt and Marder concluded that the decrease in the USE was partly caused by the generation
of delaminations.

Dabkowski et al. [DKB76] introduced the concept that delaminations cause the main fracture area to behave like a cluster of smaller specimens. Given that the width of the plastic zone in front of an advancing shear fracture is proportional to the material thickness, the effective plastic zone width is smaller than a homogeneous specimen. Therefore, the volume of material plastically deformed is reduced, lessening the energy required to fracture the specimen.

Building on the work of Dabkowski et al. and Embury et al., Bramfitt and Marder proposed a relationship between the absorbed Charpy energy, $C_v$, and the number of separations. The relationship took the form

$$C_v = \frac{C_v^*}{n + 1}$$  (2.1)

where $C_v^*$ is the absorbed Charpy energy for a full-sized specimen and $n$ is the number of separations. Effectively, $n + 1$ represents the number of sub-sized Charpy specimens created by the separation process. Bramfitt and Marder were unable to determine $C_v^*$ because their specimen that failed to delaminate exceeded the capacity of their Charpy machine. Thus, they assumed $C_v^* = 500$ J based on the number of splits at or near their machine’s 166 J capacity. The effect of the number of separations on the $C_v$ along with the above equation is shown in Figure 2.6.

It was determined that separations in this case instigated from elongated “pancaked” grain structures and was supported by the works of Mintz et al.
2.1.1.2 Mintz et al. (2008)

Mintz et al. [MMM08] investigated five steels—one X65 line pipe steel and four C–Mn steels. All steels were rare earth treated and produced as 45 kg air melts. The casts were soaked at 1000 °C and rolled into 10 mm plates using one or all of the following rolling schedules:

(i) rolled 40 % in the γ region (i.e. 950 °C), cooled to 760 °C, and subjected to 40 % reduction in the two-phase (α + γ) temperature range. This was followed by air cooling to 600 °C, 500 °C, or 400 °C and subjected to further reduction up to 40 %;

(ii) rolled 40 % from 760 °C to 720 °C in the temperature range where α and γ are present then cooled to 600 °C, 500 °C, or 400 °C, and subjected 40 %
Influence of rolling conditions on the DBT curves of L–T Charpy tests extracted from C–Mn steel, performed by [MMM08]. The parenthetical values provide the number of separations at each test temperature. I < Ductile provides the temperature where the specimen falls below 100 %SA.

reduction;

(iii) and/or rolled 40 % in the γ region, cooled to 600 °C, 500 °C, or 400 °C, and subjected to 40 % reduction.

Increasing the rolling reduction, dramatically decreased the USE, lowered the DBTT, and increased the number of separations (Figure 2.7). Mintz et al. determined that increasing the rolling reduction increased the aspect ratio of the grains as well as the yield stress due to dislocation hardening. Mintz and Morrison [MM88] suggested that dislocation hardening could increase the DBTT by 0.45 °C.MPa⁻¹. Mintz et al. saw an increase of 200 MPa for the 40 % reduction specimen, so the work of Mintz and Morrison [MM88] would suggest an increase of DBTT by 90 °C. Instead, the formation of separations led to a decrease of 30 °C for the DBTT. Reduction ratios of 10 % and 20 %
showed an increase in DBTT. Mintz et al. concluded this was because the separations were not significant enough to offset the increase in DBTT that comes from a higher yield strength. Mintz et al. summarized the study by concluding that while separations initiate via second phase particles, inclusions, or carbides, the subsequent fracture propagation is predominantly along the grain surfaces, which occurs more easily when the grains are large and elongated with a small step height from one grain surface to another. In other words, the major factor controlling separation is grain shape.

### 2.1.2 Non-metallic Inclusions

Several authors have studied the formation of separations via non-metallic inclusions. Specifically, MnS inclusions, which have been elongated because of the rolling process have been of special consideration. The following works looked at the origin of separations with respect to such inclusions.
2.1.2.1 Schofield et al. (1974)

Schofield et al. [Sch+74] looked at the fracture behavior of controlled-rolled line pipe steels. Schofield et al. looked at both DWTT and Charpy test specimens which showed separations. The author proposed that the cause for separations stemmed from non-metallic inclusions or other planar weaknesses located at the mid-thickness, notch tip location where the through-thickness stresses are greatest. The constrained stresses at this position had been shown to allow for non-metallic inclusions to decohere.

The fracture surface of the DWTT specimen showed what Schofield et al. termed “arrowhead” fractures. Figure 2.8 shows the arrowhead fracture appearance on a DWTT specimen. Schofield et al. reasoned that arrowhead fracture originate when a separation occurs at the mid-thickness of the specimen, effectively splitting the specimen into half. This then allowed for further separation in the newly formed half-specimens but the new separations would be less severe than the original. He considered three prerequisites to be necessary for arrowhead fractures: (i) a high DBTT for fracture along the through-thickness direction, (ii) the absence of planar weaknesses to allow for the development of highly constrained stress, (iii) and a relatively high yield stress, which would also allow for the attainment of highly constrained stresses.

2.1.2.2 Mintz et al. (2007)

Mintz et al. [MMM07] looked at inclusion banding in an X52 and X65 line pipe steel. The steel was normalized and controlled rolled, with the controlled
rolled having elongated MnS inclusions. Mintz et al. performed low-blow Charpy impact tests, incrementally increasing the energy until a crack was formed. They then ground the specimen below the notch, examining the surface for the origin of separations. The normalized specimens did not have separations but small crevices were found. The crevices had an average depth between 0.3 mm to 0.4 mm but occasionally a split deeper than 0.8 mm was found in areas of centerline segregation.

The X52 steels had all been normalized, giving similar equiaxed grain sizes, but contained different levels of sulfur. The steels with higher amounts of S showed signs of elongated inclusions, while lower amounts of S provided shorter, more rounded inclusions. The shorter inclusions produced separations occasionally, but the elongated MnS inclusions provided deep separations. Specimens containing elongated MnS inclusions also showed pearlite banding, which can also influence separations (Figure 2.9).
2.1.3 Banded Microstructure

2.1.3.1 Shanmugam and Pathak (1996)

Shanmugam and Pathak [SP96] used varying heat treatments to change the banding concentrations of a hot rolled microalloyed steel. They then tested Charpy specimens to characterize the influence of banding on the fracture properties. The specimens were tested at a temperature range from $-70^\circ C$ to $180^\circ C$.

Optical metallography revealed banding in the L–T and T–L directions of the hot rolled as received plates. Researchers El-Soudani [ElS90], Thaulow et al. [Tha+86], and Thompson and Howell [TH92a] confirmed that banding is primarily due to the microsegregation of manganese, non-metallic inclusions, and hot rolling at low FRTs and cooling rates. Heat treatment determines the banding concentration, which is defined in this context as the number of ferrite bands per mm. The hot rolled as received plate had a banding concentration of 71 and 62 ferrite bands per mm for the L–T and T–L orientations, respectively. When the plate reheaed to $930^\circ C$ for 30 min and furnace cooled, the banding concentration decreased to 28 ferrite bands per mm. The banding concentration was removed by reheating the plate to $930^\circ C$ for 30 min and air cooling. This gave the author a varied range of banding concentrations. It is important to note that the banding concentrations 71, 28, and 0 were for the L–T orientation, while 62 ferrite bands per mm was for the T–L orientation.

Figure 2.10 shows the DBT curves for all the banding concentrations. Evident in this figure is the reduction of USE, which has been observed by previous
authors. A shift in the DBTT and flattening of the transition regions was also confirmed. At lower test temperatures the deleterious effects of separations were minimized. The DBTT for L–T and T–L orientations for the as received plate were estimated to be $-5^\circ C$ and $-2^\circ C$, respectively. For the furnace cooled condition the DBTT rose to $7^\circ C$, and for the air cooled condition, the DBTT was around $11^\circ C$.

The banded microstructure provided the conditions necessary to initiation separations. Both L–T and T–L specimen orientations had separations, but the separations were in greater number and severity for the T–L orientation. This follows along the same lines as the study of Bramfitt and Marder [BM77]. Intermittent separations were observed on the surface of the L–T specimens and were concluded to be the result of a random distribution of ferrite grains across the layers and the spherical shape of non-metallic inclusions. This is opposed to the T–L orientation, where continuous separations were due to
the continuous ordered layer of ferrite and pearlite as well as elongated non-metallic inclusions. The separations disappeared with decreasing temperature for both orientations.

2.2 Effect on Stress State

Traditionally when evaluating a material’s fracture toughness through notched specimens, the fracture area is homogeneous and the through-thickness stress state is continuous. However, when separations occur on the fracture surface, the main fracture plane becomes divided into multiple, disconnected fracture planes and the through-thickness stress state is disrupted. This is what complicates the analysis of fracture toughness in notched specimens having separations.

Below the notch tip in a notched specimen, a high value of stress triaxiality exists. Stress triaxiality is defined as the ratio of hydrostatic stress to the equivalent von Mises stress. Stress triaxiality is well-known to locally reduce the ductility of structural materials. This can be accomplished in two ways. First, stress triaxiality can prevent plastic deformation while the level of stress increases until a failure stress is reached and a cleavage fracture results. The other mechanisms is by encouraging void growth in the material. Preexisting inclusions in the material (e.g., non-metallic inclusions), generate microvoids that enlarge because of plastic straining until the voids coalescence and the material ruptures. Triaxiality stress states are greatly influenced by the material’s initial geometry and any changes that occur during deformation.
A notched specimen is exposed to three, orthogonal stress during deformation. Figure 2.11 shows the three stress states along the x-, y-, and z-direction. These stress states also result in shear stresses $\tau_{xy}$, $\tau_{xz}$, and $\tau_{yz}$. When concerned with the crack divider type separation, the stress along the through-thickness direction, $\sigma_{zz}$, is very important.

The stress state ahead of the fracture front has a great influence on the fracture energy and toughness of the material. Studies have shown that reducing the triaxiality stress, the material toughness can increase [And05; Kno73]. Reducing the triaxiality stress state can be accomplished by minimizing the through-thickness stress. It is for this reason that separations reduce the DBTT of Charpy specimens. With every new separation, the through-thickness stress is further reduced until reaching the plane stress condition, where $\sigma_{zz} = 0$ [MM07]. With a low through-thickness stress the specimen is less likely to fail by cleavage fracture. The downside is that as the specimen is divided into thinner specimens, the Charpy USE reduced [Mor75]. Figure 2.12 shows the
change in $\sigma_{zz}$ as a function of the through-thickness location for a specimen with and without a separation.

Separations in line pipe steels are generally reported to only occur at test temperatures above the DBTT [Bou83]. Separations are in a contest with cleavage fracture on the main fracture plane. If the critical stress for a separation is not met before the critical stress for cleavage fracture, the separation will not form and the specimen will fail in a brittle manner.

Mintz and Morrison [MM07], building on the work of Bourell [Bou83], attempted to create a model to predict when separations will occur with respect to the stress state and test temperature in a Charpy specimen from an X65 line pipe steel. Mintz and Morrison observed that when the temperature was below $-80$ °C, separations did not form as cleavage fracture dominated
Effect on Stress State

The yield stress $\sigma^*_{Y}$ was adjusted for the strain rate of the Charpy test. Arrows show the transition of curve A to curve B once separations are possible [MM07].

Both Bourell and Mintz and Morrison based their model on studies showing that the triaxial stress state adjacent to a Charpy notch loaded in a plane strain condition increases the longitudinal tensile stress at yield by as much as 2.18 times [GH56; TM67; WP66]. In other words, the stress along the $x$-axis in Figure 2.11 can be expressed as $2.18\sigma_Y$. The same studies found that the through-thickness stress along the $z$-axis is less than the $\sigma_x$ value, having a relationship to the yield stress $1.68\sigma_Y$. Therefore, they considered two curves—one describing the longitudinal stress ($\sigma_{xx}$) and one describing the through-thickness stress ($\sigma_{zz}$). They plotted these curves over a range of temperatures shown in Figure 2.13. In this figure, curve A, B, and C correspond to the $\sigma_{xx}$, biaxial stress, and $\sigma_{zz}$, respectively. The curves accounted for the

Fig. 2.13: Schematic showing how separations form over certain temperature ranges. The yield stress $\sigma^*_{Y}$ was adjusted for the strain rate of the Charpy test. Arrows show the transition of curve A to curve B once separations are possible [MM07].
high strain rate of the Charpy test by modifying the yield strength.

Curves $D$ and $F$ represent the fracture stress vs temperature relationship for the main fracture plane and separation fracture plane, respectively. Therefore, the curves consider the fracture stress for brittle initiation along the fracture plane to be independent of temperature [BB78], while the fracture stress for separation formation is dependent on temperature.

The four critical temperatures $T_1$, $T_2$, $T_3$, and $T_4$ are summarized as follows:

(T$_1$) For steels not exhibiting separations, this temperature marks the point where the fracture mode will change from ductile to brittle.

(T$_2$) This signifies the point where the main fracture plane and separation plane stress intersect, so if separations exists, they are expected to occur above this temperature, but below this temperature cleavage on the main fracture plane will dominate.

(T$_3$) At this temperature, the splits will no longer appear in the sample.

(T$_4$) This temperature represents the point where tensile triaxiality is completely removed for brittle fracture.

If separations are present, the temperature at which brittle failure occurs along the main fracture plane will be reduced. Between $T_2$ and $T_1$ the main fracture mode should be ductile, which was confirmed by experiments.
2.3 **Modified Charpy Specimens Evaluating the Effect of Separations**

Several authors have taken steps to control the severity of separations to better understand the impact they have on the fracture properties. Whether through processing the material through various rolling conditions or modifying the test specimen, these authors advanced the understanding of separation’s impact on the perceived fracture toughness of steel. This section is concerned with these studies and seeks to summarize their findings to lay the groundwork for the investigations carried out in this work.

2.3.1 **Embury et al. (1967)**

Embury et al. explored the effect of both crack arrester and crack divider type separations on the fracture behavior of notched impact tests. This summary will focus on the results of the crack divider type separations.

Embury et al. performed notched impact tests of mild steel laminates by bonding soft solder, silver solder, or copper in between sub-sized Charpy specimens. The laminates were produced by brazing or soldering together 8 × 3 inch mild steel plates with varying thicknesses. The laminates ranged from two to six layers. Explosively bonded laminates were used for the crack arrester type, consisting of two 1/8 inch mild steel plates with a 1/32 inch central layer of deoxidized copper.

The authors observed a reduction in the DBTT with increasing number of
Examinations of the fracture surfaces in the DBTT region showed a pair of shear lips on each lamina sub-unit, whereas a single pair of shear lips formed on a non-laminated specimen. Comparing Figures 2.14 and 2.15, the DBTT for the laminated specimens corresponded to the DBTT for their sub-sized specimen counterparts. Embury et al. concluded that the sub-sized specimens behaved similar to the laminated equivalents, only the total energy being the sum of each sub-sized specimen. They felt that the small decrease in USE observed for the laminated specimens reflected the replacement of a small cross-sectional area of mild steel by the weaker silver solder.

Embury et al. provided further evidence that cleavage fracture could be inhibited by laminates because of their ability to suppress triaxial tension. Building of the work of a previous author, Embury et al. proposed that a thick specimen experiencing separations can perform as the sum of its parts. This reveals a belief that the fracture toughness is linearly related to the
Even with the benefits of laminated components, the authors warned that thinner specimens would be weak with respect to tension normal to the laminate plane and such stress should be avoided in service.

### 2.3.2 Ferguson (1978)

Ferguson [Fer78] used a niobium treated, low-carbon plate steel with two rolling schedules to study the effects of separations and Charpy specimen thickness on fracture properties. The steel had two FRTs at 760 °C and 995 °C. The steel finish-rolled at 760 °C suffered from separations while the other FRT did not. Full-, half-, third-, and quarter-sized Charpy specimens were taken from the rolled plate along the L–T orientation. These sub-sized Charpy specimens were tested along with composite Charpy specimens, which
were sub-sized Charpy specimens rivetted together, shown in Figure 2.16. Two configurations of composite Charpy samples were tested—a three-ply specimen made up of three, third-sized specimens; and a four-ply specimen made up of four, quarter-sized specimens.

Focusing on the specimens extracted from the 995°C finish-rolled plate, Ferguson showed a decrease in the USE as the number of composite plies increased (Figure 2.17a). A decrease of \(-37\%\) was observed for the three-ply composite compared to the full-sized specimen. This then decreased by \(-17\%\) for the four-ply composite compared to the three-ply composite. Figure 2.17b shows the trend of $C_v$ as the number of plies increases. Ferguson observed a non-linear relationship between the full-sized Charpy energy and the sub-sized specimens. However, the USE values of the composite specimens were linearly related to their sub-sized counterparts.

Transitioning to the specimens extracted from 760°C finish-rolled plate, Ferguson observed that as the separation severity increased the USE and fracture appearance transition temperature (FATT) decreased. For the full-sized Charpy specimen, Ferguson divided the DBT curves Figure 2.18 into
three regions: (Region I) region where small separations were observed, (Region II) region where severe separations were observed, and (Region III) brittle region where no separations were observed. Within Region III, the full-sized specimen saw an increase from 37 J to 73 J as the number of separations increased from one to three. In Region II, an overlap is observed between the full-size specimen and the four-ply composite specimen, and as the temperature reaches Region I, the $C_v$ diverges. This is explained by the
full-sized specimen no longer exhibiting severe separations.

Ferguson’s work garnered further support for the proposition set forth by Morrison [Mor75]. Separations are seen to accomplish the same goals as subdividing the specimen thickness by lowering the through-thickness constraint and shifting the stress state at the notch root from plane strain to plane stress.

Bluhm [Blu69] defined the relationship between specimen thickness and the force required to extend a crack. He stated that three fracture regions exists—shear fracture, mixed fracture (flat plus shear fracture), and flat fracture. Bluhm’s schematic is shown in Figure 2.19. In this figure, a critical thickness is defined as the point where the specimen no longer fractures in a fully shear manner. Until this point, the fracture toughness is dependent on the volume of the material undergoing deformation. As the thickness increases,
the volume involved in deformation increases along with the toughness measured by $G_c$. Once the critical thickness is passed, strain localization due to the constraint at the crack tip reduces the effect of the deformation zone volume and the surface area becomes the governing factor. $G_c$ decreases as the specimen thickness increases because the plane stress condition gives way to a plane strain condition. Once the specimen is thick enough, shear fracture is removed enough to allow for a dominant flat fracture, approaching the critical toughness level, $G_{ic}$.

Ferguson used Bluhm’s description to argue the effect of separations and specimen thickness on $C_v$. He applied this analysis to the upper-shelf and transition regions. In the upper-shelf region, a reduction in thickness can shift the fracture mode from mixed to pure shear. Therefore, the specific Charpy
energy (i.e. energy per cross-sectional area) can either drop or rise with a decreasing specimen thickness [JPM45]. Once the specimen thickness falls below the critical thickness, further reduction in thickness lowers the specific Charpy energy. The formation of a separation in a full-sized sample reduces the effective thickness and thus must reduce the USE. As the separations become more severe, the effective thickness decreases more and more, lowering the USE even farther. Morrison suggested that the lower limit is reached when the fracture is “woody” in nature. Woody fracture implies a fracture surface infinitely divided into disparate surfaces.

The transition energy is improved by separations because of the thickness reduction, minimizing the plane strain stress state at the notch tip. This lowers the DBTT.

2.4 Rising Upper-Shelf Phenomenon

The rising upper-shelf (RUS) phenomenon refers to cases where the absorbed Charpy energy regularly increases after the specimen has achieved 100 %SA. This phenomenon was reported by several authors dating back to the 1970s [Haw76; HM79; WME78a]. RUS behavior is found in line pipe steels which have been controlled-rolled and contain separations. For conventionally rolled steels, RUS behavior rarely occurs [Zhi02].

Figure 2.20 shows a schematic, comparing the Charpy DBT curve for a controlled-rolled steel to a conventionally rolled steel. The controlled-rolled steel shows the RUS behavior seen by the increasing Charpy energy after the
instance of 100 \text{%}SA. At 100 \text{%}SA, the Charpy energy value is denoted as the absorbed Charpy energy at 100 \text{%}SA \((C_{100}^{v})\). From this point, \(C_v\) rises until it reaches a plateau, denoted as \(C_{P}^{v}\). At \(C_{100}^{v}\) separations are most severe, and as the test temperature increases, the separations lessen in severity until they disappear at \(C_{P}^{v}\).

Because separations cause a reduction in absorbed Charpy energy but exhibit 100 \text{%}SA, the traditional approach to assessing shear area and relating it to the Charpy energy becomes nebulous. As discussed in Chapter 1, Charpy energy is used to determine the required arrest toughness and defect criterion for a ductile fracture. With a minimum \(C_v\) at \(C_{100}^{v}\) and a maximum \(C_v\) at \(C_{P}^{v}\), deciding which value to choose in the model adds further complexity to an already troubled approach. For example, when the Leis correction factor was verified for full-scale burst tests, the material had few separations and the ratio of \(C_{P}^{v}/C_{100}^{v}\) was less than 1.25 [Zhi02]. In fact, Horsley [Hor03] pointed
out that when $C_v^{100}$ is used, highly over-conservative predictions result, yet when using $C_v^P$, non-conservative predictions result.

2.5 Characterizing Separation Severity

Separations are often observable to the naked eye on the surface of Charpy, DWTTs, and full-scale burst tests. Whenever separations form, their major axis is always oriented along the rolling direction of the material. The length, width, depth, and grouping of separations can be quite diverse. In Figure 2.21, three unique separation appearances are shown. The top image shows a DWTT performed by Schofield et al. [Sch+74], where the separations are grouped into what has been called an “arrowhead” appearance. The middle image shows the fracture surface of a WJ test performed by Demofonti et al. [Dem+02]. Here, the separations form a mixture between “arrowhead” and severe, seemingly erratic separations. The bottom image is the surface of a DWTT primarily containing severe separations with minimal “arrowhead” features [Tor+07]. When “arrowheads” are observed, they are always oriented in the direction of fracture propagation.

The following sections summarize the efforts made to characterize and quantify separation severity in Charpy, DWTTs, and full-scale burst tests as well as the relationship between them.
2.5.1 Sugie et al. (1984)

Sugie et al. [Sug+84] carried out five full-scale burst tests of X70 grade line pipe steel. The pipes had an outer-diameter of 1219.2 mm and wall-thickness of 18.3 mm. The burst tests were used to investigate the evaluation method for shear arrestability and to investigate the effect of separations on the arrestability. The pipes were manufactured by two separate methods. Some
pipes were controlled-rolled with two levels of separation severity and others were quenched and tempered not exhibiting separations. The test were carried out at temperatures between $3 \degree C$ to $12 \degree C$.

After the tests were completed, the full-scale surfaces were analyzed and compared to the laboratory testing of Charpy tests, SPC-DWTTs, and PN-DWTTs. The fracture surfaces of the controlled-rolled pipes showed distinct separations. The authors categorized the separations into two feature types. Feature type $A$ showed the fracture surface between separations to be nearly parallel to the main fracture plane, while feature type $B$ showed a $45^\circ$, slant-like surface. Figure 2.22 shows two separation types for a Charpy, DWTT, and full-scale sample. As can be seen in this figure, the Charpy surface is primarily composed of the type $A$ fracture surface, while the DWTT and full-scale surface is a combination of both types $A$ and $B$.

Sugie et al. also measured the separations and defined a criteria which they called the separation index (SI). The SI was defined as the ratio of total
length of all separations to the inspected area, having units mm$^{-1}$. The SI was determined from the fracture surfaces of Charpy tests, SPC-DWTTs, PN-DWTTs, and the full-scale pipe surface. This allowed for comparing the SI of each laboratory test to the actual pipe’s fracture surface. Figure 2.23 shows the results of this investigation. The results have been digitized and a linear fit has been applied to each set. The figure clearly shows that the Charpy’s SI deviates from the full-scale fracture surface, while both types of DWTT shows a similar measure when compared to the full-scale surface.

Based on the full-scale results, comparing the controlled-rolled (with separations) and the quenched and tempered (without separations) steels, Sugie et al. concluded the arrestability between the two pipes showed no difference. Furthermore, the authors concluded that separations played no role in the arrestability of fracture.
Farber et al. [Far+15] studied an X80 line pipe steel to measure the effect separations had on the fracture characterization. The X80 steel came from four different steel producers and were referred to as steel 1, 2, 3, and 4. They analyzed the fracture surfaces of broken Charpy specimens, noting six characteristics: (1) the initial ductile crack zone ($L_h$), (2) the fibrous zone ($L_f$), (3) the shear lip zone ($\lambda$), (4) the final fracture zone ($L_{ff}$), (5) the area of each separation ($S_{SP}$), and (6) the area of separation relief ($S_{RR}$). The characteristics are shown in Figure 2.24.

Each separation was measured by fitting an ellipse onto the delamination’s major and minor axes. The total separation area was determined by

$$S_{SP} = \frac{\pi l_{SP} b_{SP}}{4} n$$  \hspace{1cm} (2.2)
where \( l_{SP} \) and \( b_{SP} \) were the length and width of each individual separation, respectively, and \( n \) was the number of separations. A similar formula was used to measure the region of relaxation (the regions surrounding the separations) denoted \( S_{RR} \).

The separation density was determined by

\[
\rho_{SP} = \frac{n_{rel}}{S_f} \tag{2.3}
\]

where \( S_f \) was the area taken by the fibrous zone \( L_f \), and \( n_{rel} \) was the relative number of separations determined by dividing the total length of all separations by the length of the shortest separation. The authors used \( \rho_{SP} \) as a more objective measure than \( S_{SP} \).

With these measurements, the specific absorbed Charpy energy could be correlated with the separation severity over a range of test temperatures.
Figures 2.25 and 2.26 show the DBT curve and the separation density for each steel, respectively, for varying test temperatures. The relationship between the fracture toughness and the separation severity is not similar for each steel.

For all steel, the KCV is high with an average value of 350 J.cm\(^{-1}\). Steels 1 and 3 show a decrease in fracture toughness around \(-20^\circ C\), which corresponds with the point where separations first appear. While steel 1 shows the highest separation density of all the steels at \(-20^\circ C\), its fracture toughness value does not decline. The sharp decrease in fracture toughness in steels 1 and 2 began at temperatures below \(-20^\circ C\). This was connected to a sharp rise in separation severity, particularly for steel 1. Steels 1 and 2 showed a similar trend in their DBT curves; however, the separation severity for steel 1 was more than double that of steel 2 at \(-60^\circ C\). Steel 3 showed a very gradual rise in separation density as the temperature was lowered, and steel 4 showed very little separation at all.

From their results, the authors concluded that the onset of separations does not necessarily mean that an immediate decrease in fracture toughness will ensue. They opine that the effect of separations on the fracture toughness is low compared to other metallurgical and mechanical factors.

2.5.3 Igi et al. (2016)

Igi et al. [Igi+16] evaluated Charpy, PN-DWTT, and SPC-DWTT specimens to describe the propagation and arrest properties of X80 line pipe steels
having various Charpy energy values. The authors also evaluated the effect of separations on the absorbed energies and crack velocities of DWTT specimens.

A study done by researchers with Nippon Steel in 1974 looked at the onset of separations, using interrupted PN-DWTTs of an X70 line pipe steel. By applying different energy levels to the DWTT specimens and slicing the specimens afterwards, the onset of separation was able to be observed. Figure 2.27 shows the result of this experiment. From the figure, separations do not begin to appear until striker energy levels around 200 kg.m. As the energy increases, the separation severity increases and at 500 kg.m for a specimen tested at 0 °C, a shear fracture can be seen, joining two separations.

Igi et al. measured the separations on PN-DWTT and SPC-DWTT specimens by drawing lines over the separation lengths. They then summed all the separation lengths and divided that by the total fracture area. This gave the SI, which was the same formulation as Sugie et al. [Sug+84] discussed
previously. Figures 2.28a and 2.28b show the sample surfaces with the separation measurements for the PN-DWTT and SPC-DWTT, respectively. The results of these evaluations are summarized in Table 2.2. The highest SI was seen in the PN-DWTT specimen from Pipe A, but the relative difference between the PN-DWTT and SPC-DWTT specimen from the pipe is minimal. In fact, there does not appear to be any significant different in separation severity between the two DWTT types, suggesting that the separation severity is independent of the DWTT type.

<table>
<thead>
<tr>
<th>Pipe</th>
<th>Process</th>
<th>$C_v$ at $-5^\circ$C [J]</th>
<th>PN-DWTT</th>
<th>SPC-DWTT</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>TMCP</td>
<td>127</td>
<td>0.23</td>
<td>0.21</td>
</tr>
<tr>
<td>B</td>
<td>TMCP</td>
<td>240</td>
<td>0.02</td>
<td>0.03</td>
</tr>
<tr>
<td>C</td>
<td>TMCP</td>
<td>324</td>
<td>0.01</td>
<td>0.01</td>
</tr>
<tr>
<td>D</td>
<td>TMCP</td>
<td>447</td>
<td>0.01</td>
<td>0.02</td>
</tr>
</tbody>
</table>

Using high-speed cameras, Igi et al. were able to measure the average fracture velocity of all DWTT specimens. They then were able to relate the average fracture velocity and total absorbed energy to the separation severity.

Figure 2.29 shows the relationship between the fracture propagation energy and the SI for the PN-DWTTs and SPC-DWTTs. At low SI values, there existed a large scatter in energy. However, at higher SI values, the energy is consistently at the lower end. Figure 2.30 shows a similar relationship but for the fracture velocity. The total fracture propagation was seen to decrease as the fracture velocity and SI increased, yet there did not appear to be a significant impact on the fracture velocity with increasing separation severity.
Added to both Figures 2.29 and 2.30 are the PN-DWTT results of a second FSBT. Figure 2.29 shows the separations severity’s impact on the fracture propagation energy of PN-DWTT specimens. These test show an identical trend to the DWTT specimens of the first FSBT. Figure 2.30 shows the separation severity’s impact on the fracture propagation velocity of the full-scale pipe. It is important to note the difference in fracture velocity between the DWTTs and FSBT. Despite the increased velocities, the measured fracture velocity shows a near linear trend with increasing SI. This is congruent with the results seen in the DWTTs, measuring the same qualities.

2.6 Summary

This chapter explored the current state of knowledge of separations in line pipe steels. The key points covered are

- The origin of separations proposed by several authors.
- The consequences of separations when occurring in material tests.
- The effect of separations on the stress state after initiating.
- A look at the use of modified Charpy test specimens, seeking to better understand the effect of separations.
- A specific look at the rising upper-shelf phenomenon known to occur in controlled-rolled steel exhibiting separations.
- Two separation severity metrics, the separation index (SI) and separation density ($\rho_{SP}$), and how they compare between Charpy, DWTTs, and full-scale burst tests fracture surfaces.
Fig. 2.28: Fracture surfaces of (A) PN-DWTT and (B) SPC-DWTT specimens [Adapted from [Igi+16]].
Summary

Figure 2.29: Relationship between fracture energy and separation severity for PN-DWTTs and SPC-DWTTs [Igi+16].

Figure 2.30: Relationship between fracture velocity and separation severity for PN-DWTTs and SPC-DWTTs [Igi+16].
CHAPTER 3

NUMERICAL MODELING

I should think people would want to know that what they know is truly what the universe is like, or at least as close as they can get to it.

— Isaac Asimov (c. 1920 – 1992)

This chapter will describe the two material models used in the finite element analysis (FEA) of Charpy impact tests with separation. The cohesive zone model (CZM) was used to characterize separations, while the Gurson-Tvergaard-Needleman (GTN) model was used to characterize ductile fracture.

3.1 COHESIVE ZONE MODEL

Before the development of the CZM, the prominent means to describe fracture was through linear elastic fracture mechanics (LEFM). LEFM, however, suffers from several weaknesses. Firstly, LEFM can only describe a body already containing a crack. Secondly, the effect of plasticity at the crack tip must be negligible or confined to a small region around the crack tip. Lastly, as the name implies, the theory is only applicable if the global material behavior is linear elastic [And05].
The foundation of the CZM stems from the work of Dugdale [Dug60] and Barenblatt [Bar62]. Their work with cohesive zones to define crack propagation mirrors that of Griffith’s theory [Gri21], where the surface energy defines the resistance against crack advance. The cohesive zone allowed for the description of material behavior in advance of the crack tip. Both men’s approach was to split the crack surface into two regions—one part stress free and the other loaded by a cohesive stress.

Dugdale assumed the presence of a plastic zone near the crack tip within which a stress equal to the material’s yield strength acts across the crack. He examined the yielding at the end of slits, which were introduced into steel sheets. From this experiment, he was able to deduce a relationship between plastic yielding and an externally applied load, finding that the influence of yielding could be approximated by a long crack extending into the region with a stress equivalent of the yield stress. Dugdale’s theory only holds for plane stress states.

Barenblatt used a similar approach to Dugdale to study brittle materials.
However, instead of a constant stress in the cohesive zone, Barenblatt allowed
the stress to vary with deformation in advance of the crack tip. For Barenblatt’s
formulation, the cohesive zone serves as a binding stress, keeping the disjoint
surfaces together (similar to atomic or molecular attractions). The binding
stress is described by the function $\sigma = T(\delta)$. Figure 3.1 shows a schematic
of a cohesive crack in a body. The crack’s separation is described by $\delta$, the
resulting stress inside the separating membrane is described by $\sigma$, the binding
function is described by $T(\delta)$, and the resulting fracture energy is described
by $\Gamma_0$.

### 3.1.1 Traction-Separation Laws

Within the framework of FEA, the cohesive zone is described by initially
undamaged elements, which when subjected to stresses by surrounding
continuum elements separate until a failure condition is met. The cohesive
element’s damage behavior is generally described by the so-called TSL.

The TSL is a constitutive equation describing the failure behavior of a cohesive
element. Here, traction describes the stress across a cohesive interface as a
function of the crack separation length. As the separation increases, the stress
will reach a maximum before gradually being reduced to zero, signaling
failure or decohesion.

The primary TSL parameters are the cohesive strength, $T_0$, and cohesive
energy, $\Gamma_0$ [SB00; TH92b]. The cohesive strength describes the peak stress
required for separation, while the cohesive energy describes the separation
work per unit area.
Given the vast number of materials the CZM has been used to characterize, numerous TSLs exist. Thus, choosing a suitable TSL is incumbent on the user. The cohesive elements model the initial loading as well as the damage initiation and evolution. Crack propagation is simulated by the chosen parameters of the TSL. The parameters can be based on either the local energy release or separation of the crack surfaces [Che+99].

The bilinear TSL is the most basic formulation to describe the cohesive element behavior. This formulation considers the softening after damage initiation to be linear and is shown in Figure 3.2. The bilinear TSL is defined by two points. The first point, $\delta_0$, is the separation where the maximum traction, $T_0$, occurs; and the second point, is the separation at failure, $\delta_F$, where the traction is zero. Cohesive element damage is initiated when the maximum traction is reached for all cohesive zone models.

Another key parameter is the cohesive penalty stiffness, $K_0$. $K_0$ represents the slope of TSL curve prior to the maximum traction. With a poorly defined
The second portion of the curve, after the maximum traction, describes damage evolution. In the case of linear softening, damage can be defined by the failure separation, \( \delta_F \), or the critical energy release rate, \( \Gamma_0 \). As will be seen later, many TSLs use an exponential-like function to describe damage. Here, softening is described by the failure separation and the desired curvature, which can also be based on energy. The cohesive fracture energy is defined by the integral

\[
\Gamma_0 = \int_0^{\delta_F} T(\delta) \, d\delta
\]  
(3.1)

When used to model brittle fracture, the TSL parameters are easy to determine experimentally (given the ability to determine the fracture behavior). However, for ductile materials like steel, which involve elasticity, plasticity, and damage; the TSL parameters are not easily ascertained. In effect, the parameters are generally assumed [SB03]. The CZM is a phenomenological model and does not purport to model the true fracture process. Therefore, researchers have used several different approaches to implement a realistic cohesive zone method and are described as follows.
3.1.1.1 Needleman (1987)

In 1987, Needleman proposed a CZM which takes finite geometry changes into account and is used to describe void nucleation [Nee87]. Like the GTN model, Needleman’s cohesive zone approach aims to describe the evolution from initial debonding to complete decohesion through the auspices of void growth. Within this framework, decohesion can occur in either a brittle or ductile manner. The fracture manner is governed by the ratio of a characteristic length to the inclusion radius. The characteristic length is introduced by the dimensions of the surrounding material (e.g. dimensions around the crack tip). The cohesive interface’s mechanical response is based on both the critical interfacial strength and the work of separation per unit area.

Interfacial traction is solely dependent on the differential displacement across the interface. Consider two points \( A \) and \( B \), lying on opposite sides of an interface. The interfacial traction is then dependent on the differential displacement between the points. For each point of the interface, Needleman defined the displacements as
\[ u_n = \hat{n} \cdot \Delta \tilde{u}_{AB} \quad u_s = \hat{s} \cdot \Delta \tilde{u}_{AB} \quad u_b = \hat{b} \cdot \Delta \tilde{u}_{AB} \quad (3.2) \]

and thus the tractions as

\[ T_n = \hat{n} \cdot \tilde{T} \quad T_s = \hat{s} \cdot \tilde{T} \quad T_b = \hat{b} \cdot \tilde{T} \quad (3.3) \]

In Equations (3.2) and (3.3), \( \hat{n}, \hat{s}, \) and \( \hat{b} \) represent a right-handed coordinate system so that \( u_n \) corresponds to an increasing interfacial separation and negative \( u_n \) a decreasing separation.

The mechanical response of the interface is described through a constitutive relation relating the tractions to their corresponding displacements. The response is specified in terms of the potential

\[ \phi (u_n, u_s, u_b) = -\int_0^u \left[ T_n \, du_n + T_s \, du_s + T_b \, du_b \right] \quad (3.4) \]

Similar to other TSLs, Needleman’s model exhibits an increasing traction until a maximum is reached at which point the traction incrementally vanishes, signaling decohesion. The specific potential function used is as follows

\[ \phi (u_n, u_s, u_b) = \frac{27}{4} T_0 \Lambda \left\{ \frac{1}{2} \left( \frac{u_n}{\lambda} \right)^2 \left[ 1 - \frac{4}{3} \left( \frac{u_n}{\lambda} \right) + \frac{1}{2} \left( \frac{u_n}{\lambda} \right)^2 \right] + \frac{1}{2} \alpha \left( \frac{u_s}{\lambda} \right)^2 \left[ 1 - 2 \left( \frac{u_n}{\lambda} \right) + \left( \frac{u_n}{\lambda} \right)^2 \right] + \frac{1}{2} \alpha \left( \frac{u_b}{\lambda} \right)^2 \left[ 1 - 2 \left( \frac{u_n}{\lambda} \right) + \left( \frac{u_n}{\lambda} \right)^2 \right] \right\} \quad (3.5) \]
In the case where $u_n \leq \lambda$, where $T_0$ is the maximum traction of the interface experiencing purely normal separation ($u_s = u_b = 0$), $\lambda$ represents the characteristic length and $\alpha$ specifies the ratio of normal to shear stiffness.

Differentiating Equation (3.5) along each displacement vector, $\mathbf{T} = \partial \phi / \partial \mathbf{u}$, yields the tractions

$$T_n = -\frac{27}{4} T_0 \left\{ \left( \frac{u_n}{\lambda} \right) \left[ 1 - 2 \left( \frac{u_n}{\lambda} \right) + \left( \frac{u_n}{\lambda} \right)^2 \right] + \alpha \left( \frac{u_s}{\lambda} \right)^2 \left[ \left( \frac{u_n}{\lambda} \right) - 1 \right] + \alpha \left( \frac{u_b}{\lambda} \right)^2 \left[ \left( \frac{u_n}{\lambda} \right) - 1 \right] \right\}$$

(3.6a)

$$T_s = -\frac{27}{4} T_0 \left\{ \alpha \left( \frac{u_s}{\lambda} \right)^2 \left[ \left( \frac{u_n}{\lambda} \right) - 1 \right] + \alpha \left( \frac{u_b}{\lambda} \right)^2 \left[ \left( \frac{u_n}{\lambda} \right) - 1 \right] \right\}$$

(3.6b)

$$T_b = -\frac{27}{4} T_0 \left\{ \alpha \left( \frac{u_b}{\lambda} \right)^2 \left[ \left( \frac{u_n}{\lambda} \right) - 1 \right] + \alpha \left( \frac{u_b}{\lambda} \right)^2 \left[ \left( \frac{u_n}{\lambda} \right) - 1 \right] \right\}$$

(3.6c)

for $u_n \leq \lambda$ and $T_n = T_s = T_b = 0$ when $u_n > \lambda$.

The potential (Equation (3.5)) is shown in Figure 3.3, where $T_n$ is plotted as a function of $u_n$ with $u_s = u_b = 0$. As with other TSLs, the fracture energy is defined by the area under the curve.

Needleman’s TSL is described by the parameters: $T_0$, $\lambda$, and $\alpha$.

3.1.1.2 TVERGAARD AND HUTCHINSON (1992)

One of the most common TSLs, is the curve proposed by Tvergaard and Hutchinson in 1992 [TH92b]. Their proposed relationship between stress and separation is similar to the bilinear TSL but with an extended region of
maximum traction (see Figure 3.4). As can be seen in this figure, the onset of maximum traction, $T_0$, occurs at the predefined separation nominated $\delta_1$. This is maintained until another predefined separation, $\delta_2$, is sustained at which point softening takes place until the element reaches a critical separation for failure, $\delta_F$.

The dissipated energy for this TSL is defined by the expression

$$\Gamma_0 = \frac{1}{2} T_0 (\delta_F + \delta_1 + \delta_2)$$  \hspace{1cm} (3.7)

The Tvergaard and Hutchinson TSL requires defining the fracture energy and peak traction. $\delta_1$ and $\delta_2$ exist as shape parameters. Also important to define are the parameters used to describe the continuum behavior of the solid. The elastic-plastic strain is characterized by
where $\sigma$ is the stress, $E$ is the elastic modulus, $\sigma_Y$ is the yield stress, and $N$ is the strain hardening exponent.

The TSL proposed by Tvergaard and Hutchinson has found the greatest use in depicting ductile fracture.

3.1.1.3 **Cornec, Scheider, and Schwalbe (2003)**

Cornec et al. proposed a TSL in 2003 aimed at modeling metallic materials [CSS03]. To accomplish this, the model attempted to account for the ductile tearing process, which consists of void initiation, growth, and coalescence. The models implements two separation parameters, $\delta_1$ and $\delta_2$, to modify the
shape of the traction curve. An illustration of the TSL is shown in Figure 3.5 and defined by the function

\[
T(\delta) = T_0 \begin{cases} 
2 \left( \frac{\delta}{\delta_1} \right) - \left( \frac{\delta}{\delta_1} \right)^2 & \delta < \delta_1 \\
1 & \delta_1 < \delta < \delta_2 \\
2 \left( \frac{\delta - \delta_2}{\delta_F - \delta_2} \right)^3 - 3 \left( \frac{\delta - \delta_2}{\delta_F - \delta_2} \right)^2 & \delta_2 < \delta < \delta_F 
\end{cases} 
\]  

(3.9)

The authors suggest the relationship of the additional separation terms to be \( \delta_1 = 0.01\delta_F \) and \( \delta_2 = 0.75\delta_F \). With these terms set, three material parameters had to be determined: (1) the cohesive strength, \( T_0 \); (2) the cohesive energy, \( \Gamma_0 \); and (3) the separation for material failure, \( \delta_F \). Considering that the cohesive energy is a function of the cohesive strength and failure separation

\[
\Gamma_0 = \frac{1}{2} \left( \frac{1}{3} \frac{\delta_1}{\delta_F} + \frac{1}{2} \frac{\delta_2}{\delta_F} \right) 
\]

(3.10)

only two parameters need to be determined to satisfy this model (i.e. \( T_0 \) and \( \delta_F \)).

The exponential curve preceding the maximum traction was chosen to avoid numerical issues between the cohesive elements and the surrounding continuum elements. The rapid softening after \( \delta_2 \) was used to simulate void growth and coalescence.
3.1.1.4 Ren and Ru (2013)

Ren and Ru [RR13] developed a TSL to explore the dynamic relationship between fracture toughness and fracture speed in a DWTT. Recognizing that identifying a specific and reasonably simple CZM for high-speed fracture in line pipe is not trivial, they derived a TSL with the expectation that it would be scalable to high-speed fracture.

Ren and Ru's TSL obeys the constitutive law proposed in Alfano et al. [Alf+04], which states

$$\tilde{T} = (1 - \mathcal{D})K_0\tilde{\delta}$$

(3.11)

where $\tilde{T}$ is the traction tensor, $\tilde{\delta}$ is the cohesive separation vector, $K_0$ is the cohesive penalty stiffness matrix, and $\mathcal{D}$ is the damage scalar. Simplifying the equation to only the normal component yields
\[ T = (1 - D)K_0\delta \quad (3.12) \]

where \( K_0 \) is the cohesive penalty stiffness scalar.

The proposed TSL (shown in Figure 3.6) considers a linear relationship, where no damage occurs until reaching a critical separation value \( \delta_0 \). At this point damage is scaled as \( \delta \) increases until the failure value of \( \delta_F \).

The damage scalar, \( D \), is defined by the piecewise function

\[
D = \begin{cases} 
0 & \delta < \delta_0 \\
1 - \frac{\delta_0}{\delta} \left[ 1 - \left( \frac{\delta - \delta_0}{\delta_F - \delta_0} \right)^\alpha \right] & \delta \geq \delta_0 
\end{cases} \quad (3.13)
\]

where \( \alpha \) controls the shape of the damage evolution curve. Varying values of \( \alpha \) are shown in Figure 3.6. The convex form of \( \alpha \) is generally applied to ductile materials [Sch09; Vol04], and the concave form is generally applied to brittle materials. In other words, \( \alpha < 1 \) is suitable for brittle fracture, while \( \alpha > 1 \) is suitable for ductile fracture.

Using Equations (7.6) and (7.7), the cohesive energy, \( \Gamma_0 \), is shown to be

\[
\Gamma_0 = \int_0^{\delta_F} T \, d\delta = \left( 1 - \frac{1}{\alpha + 1} \right) K_0\delta_0\delta_F + \left( \frac{1}{\alpha + 1} - \frac{1}{2} \right) K_0\delta_0^2 \quad (3.14)
\]

The authors propose the following considerations when implementing their TSL:
(1) The value of $K_0$ should be comparable but not smaller than the Young’s modulus. Calculation accuracy is found to degrade when $K_0$ is too small, while the fracture surface will not perfectly separate when $K_0$ is too large.

(2) $\alpha > 1$ must be given for ductile fracture and $1 < \alpha < 10$ is suggested for line pipe steels. The fracture surface will not perfectly separate if $\alpha$ is too large. FEA results become less sensitive at a large enough value of $\alpha$ (e.g., the results for $\alpha = 6$ is similar to $\alpha = 8$).

(3) $\delta_0$ is chosen by comparing FEA obtained fracture speed with experimental data. Smaller values of $\delta_0$ are associated with higher fracture speeds.

(4) The ratio of $\delta_F/\delta_0$ is chosen based on the comparison of FEA load-displacement curves with experimental data. For a given $\Gamma_0$, a larger value of $\delta_F/\delta_0$ will result in a larger peak displacement.

3.2 Gurson-Tvergaard-Needleman Model

The GTN model is a constitutive model used to describe failure of porous, ductile media. This is done by accounting for the effect of void initiation, growth, and coalescence on the damage behavior in ductile materials. The model is pervasive within the world of numerical modeling of metals.

The role of void evolution on ductile fracture was first identified in 1949 by Tipper [Tip49]. The phenomenon of void nucleation, growth, and coalescence would take nearly a decade before being sufficiently documented by researchers such as Puttick [Put59], Rogers [Rog60], Gurland and Plateau [GP63], and Beachem [Bea63].
At room temperature, voids within a deforming material nucleate by decohesion of second phase particles from the material matrix or by particle fracture. Voids grow as the material plastically deforms, increasing the potential for interaction with other voids. Either through necking down of the material matrix between neighboring voids or by shearing between well separated voids, void coalescence occurs. An idealization of this process is shown in Figure 3.7. Reviews by Garrison and Moody [GM87], Tvergaard [Tve89], Besson [Bes10], Benzerga and Leblond [BL10], and Benzerga et al. [Ben+16] have gone a long way to capture the developments concerning ductile fracture throughout the decades and are prominently used in this section.

Initial studies by McClintock [McC68] and Rice and Tracey [RT69] looked at growth of a single void in an infinite elastic-plastic volume with the void shape being circular cylindrical for the former and spherical for the latter. Needleman [Nee72] performed a numerical study for a material containing a periodic array of circular cylindrical voids, accounting for the interactions between neighboring voids from early growth stages to the near onset of coalescence. Using a representative unit volume incorporating a single void,
necessary boundary conditions were applied to simulate a full material matrix. This set the groundwork for the unit cell analysis, which has become an important tool to analyze various aspects of ductile failure. Unit cells encompassing many voids are able to account for varying void sizes or frequency, as well as localized plastic flow as a result of void clustering or instabilities.

Methods proposed by Gurson [Gur77] and Rousselier [Rou87] incorporate void evolution into the constitutive formulation. Of these, Gurson’s model is the most prominent and exploited. Gurson’s model emerged from micro-mechanical studies, using averaging techniques similar to those applied by Bishop et al. [BHM45]. Several modifications were made throughout the years to Gurson’s model by Chu and Needleman [CN80]; Tvergaard [Tve81; Tve82b]; and Tvergaard and Needleman [TN84]. Hence, the Gurson-Tvergaard-Needleman (GTN) model derives its name. Since then, the GTN model has been applied to a variety of problems and materials to describe porosity effects on material behavior. In many of these applications, the material does not initially contain voids, so characterizing void nucleation as a result of deformation is vital. Porous ductile material models were earlier developed by fitting experiments for powder metallurgy materials [SO76]. The approximate yield surfaces of both methods conform for a given void volume fraction.

Gurson’s model makes a few limiting assumptions. Namely, the voids are considered to be embedded in a standard Mises solid, and the voids remains spherical regardless of the stress state. At low triaxialities, voids tend to elongate, which can have a strong influence on the ductile failure. Several authors have extended the GTN model to accommodate void shape effects
Gurson-Tvergaard-Needleman Model

F/i.sc/g.sc. 3.8: Ductile failure of line pipe steel, showing a dimpled fracture surface with inclusions and voids. [CZ94; GLD93; GLD94; Gol+97]. Benzerga and Besson [BB01] incorporated anisotropic effects via Hill’s criterion [Hil48] into the GTN model. This model initially considered spherical voids but was extended by Keralavarma and Benzerga [KB08; KB10] to account for non-spherical voids.

The final failure stage of porous materials generally occurs by void coalescence. Failure by void coalescence is defined as the stage when neighboring voids neck down to zero thickness, leaving the fibrous fracture surface seen in Figure 3.8. Koplik and Needleman [KN88] have contributed significantly to modeling this mechanism. However, final failure is often associated with shear band instability [NR78; Ric76]. Shear band instability leads to the so-called void-sheet failure, where voids grow to coalescence inside a narrow layer of material [Rog60]. When this occurs, the fracture surface reveals voids being smeared out during coalescence. Materials containing two void or inclusion size scales nucleating, Cox and Low [CL74] as well as Stone et al. [Sto+85]
observed that plastic flow localizes between larger voids and final failure involves void-sheet failure by the smaller voids surrounded by larger voids. Tvergaard [Tve82a] was able to simulate this phenomenon using the GTN model with localization leading to void-sheet failure predicted.

### 3.2.1 Implementation

The GTN flow potential initially proposed by Gurson was unable to represent fracture and coalescence. Additionally, unit cell simulations were unable to accurately predict void growth rates. Tvergaard and Needleman [TN84] modified Gurson’s original yield surface expression to better represent actual experiments, resulting in the flow potential

\[
\Phi(\sigma, f^*) = \left( \frac{\sigma_{eq}}{\sigma_Y} \right)^2 + 2q_1 f^* \cosh \left( \frac{3q_2 \sigma_h}{2\sigma_Y} \right) - 1 - (q_1 f^*)^2 = 0 \tag{3.15}
\]

where \( \sigma \) is the Cauchy stress tensor and \( \sigma' \) its deviator; \( \sigma_{eq} = \sqrt{(3/2)\sigma' : \sigma'} \) is the equivalent von Mises stress; \( \sigma_Y \) is the yield stress; and \( \sigma_h = \text{tr}(\sigma)/3 \) is the macroscopic hydrostatic stress. \( q_1 \) and \( q_2 \) were introduced by Tvergaard [Tve81] to more accurately describe void growth kinetics observed in unit cell computations. Often values of \( q_1 = 1.5 \) or \( q_1 = 1.25 \) and \( q_2 = 1.0 \) are used [KN88]. Faleskog et al. [FGS98] concluded that the \( q \) parameters depend on the plastic hardening exponent and the yield stress to elastic modulus ratio.

The effective void volume fraction, \( f^* \), is a function of the actual porosity, \( f \). The contrived function was introduced to initiation coalescence. When the critical void volume fraction for void coalescence, \( f_c \), is reached, damage
The function is represented by a piecewise representation

\[
f^* = \begin{cases} 
    f & f \leq f_c \\
    f_c + (f - f_c) \left( \frac{f_u - f_c}{f_t - f_c} \right) & f > f_c
\end{cases}
\]  

(3.16)

where \( f_u \) is the void volume fraction at rupture (i.e. loss of ability to carry stress) and is generally considered to be \( 1/q_1 \). \( f_t \) represents the void volume fraction at failure. The value for \( f_t \) at this juncture is entirely heuristic; however, low values of \( f_c \) and \( f_f \) can lead to convergence issues in FEA. Zhang et al. [ZTØ00] provided a method for determining \( f_c \), using unit cell calculations.

The material porosity accumulates via two processes: the growth of existing voids and the nucleation of new voids through plastic deformation. The void volume accumulation rate, \( \dot{f} \), is defined by

\[
\dot{f} = \dot{f}_g + \dot{f}_n
\]  

(3.17)

where \( \dot{f}_g \) is the existing void growth rate, and \( \dot{f}_n \) is the void nucleation rate.

Assuming that the material matrix is plastically incompressible, the existing void growth rate is governed by the mesoscopic plastic dilation

\[
\dot{f}_g = (1 - f) \, \text{tr}(\dot{\varepsilon}^p)
\]  

(3.18)

The plastic strain rate tensor, \( \dot{\varepsilon}^p \), is derived using the normality rule, which has shown to be valid by micromechanical analyses as long as the rule also
applies to the matrix material. Therefore, the plastic strain rate tensor is expressed as

\[ \dot{\varepsilon}^P = \dot{\mu} \frac{\partial \Phi}{\partial \sigma} \]  

(3.19)

where \( \dot{\mu} \) is the plastic multiplier. Assuming that isotropic hardening is described by the effective plastic strain, \( \varepsilon_{\text{eff}}^P \) \[ \text{[SO76]}, \]

\[ \dot{\varepsilon}^P : \sigma = (1 - f) \dot{\varepsilon}_{\text{eff}}^P \sigma_{\text{eff}} \]  

(3.20)

This expression states that the macroscopic plastic work (left-hand side) is equivalent to the microscopic plastic work (right-hand side). The factor \( (1 - f) \) considers that a portion of the macroscopic volume corresponds to pores in which plastic work is nonexistent. Considering that \( \sigma_{\text{eff}} \) is a homogeneous function of degree 1, Euler’s Lemma provides

\[ \dot{\varepsilon}^P : \sigma = \dot{\mu} \sigma_{\text{eff}} \]

so that \( \dot{\mu} = (1 - f) \dot{\varepsilon}_{\text{eff}}^P \) \[ \text{[Bes10]}, \]

The final form of the plastic strain rate tensor is

\[ \dot{\varepsilon}^P = (1 - f) \dot{\varepsilon}_{\text{eff}}^P \frac{\partial \Phi}{\partial \sigma} \]  

(3.21)

The void nucleation rate based on plastic straining, is defined by

\[ \dot{f}_n = A_n \dot{\varepsilon}^P \]  

(3.22)
where $A_n$ represents the void nucleation rate coefficient which is a Gaussian distribution proposed by Chu and Needleman [CN80] and expressed as

$$A_n = \frac{f_n}{s_n \sqrt{2\pi}} \exp \left[ -\frac{1}{2} \left( \frac{\varepsilon_p - \varepsilon_n}{s_n} \right)^2 \right]$$

(3.23)

Herein, three additional parameters are introduced: $f_n$, defining the void volume fraction where damage is nucleated; $\varepsilon_n$, defining the strain where 50\% of inclusions are broken; and $s_n$, defining the standard deviation on the nucleation strain. In the numerical study of Chu and Needleman [CN80], $\varepsilon_n$ was found to be 0.3, and $s_n$ found to be 0.1. By definition, this means that half of the voids are broken at $\varepsilon_{eff}^P = 0.3$, and 98\% of voids are broken at $\varepsilon_{eff}^P = 0.5$. Numerous researchers have used these values without microstructural validation, even though the nucleation rate strongly depends on the material’s characteristics (e.g. chemical composition, thermal treatments). Other forms of $A_n$ have been proposed by Zhang et al. [ZT00], Besson et al. [BDP00], and Prat et al. [Pra+98].

### 3.3 Summary

This chapter looked at the two material damage models to be used in this study to characterize brittle and ductile fracture. The two methods are the cohesive zone model (CZM) and Gurson-Tvergaard-Needleman (GTN) model. Both models have their advantages and disadvantages, which are discussed in their respective sections.
SUMMARY, KNOWLEDGE GAPS, AND OBJECTIVES

To be ignorant of what occurred before you were born is to remain always a child. For what is the worth of human life, unless it is woven into the life of our ancestors by the records of history?

– MARCUS TULLIUS CICERO (106–43 BC)

When investigating the separation phenomenon in the context of fracture control of pipelines, several aspects have been shown to have great significance. This chapter will summarize the findings of the literature review, incorporating the knowledge gaps, and reveal the objectives of the current study.

SUMMARY AND KNOWLEDGE GAPS

To date, fracture control is based on the Battelle Two-Curve Model, which uses notched impact fracture tests to determine the required fracture toughness to arrest a running ductile fracture with a given gas decompression behavior. The primary test for determining the fracture toughness of a line pipe steel is the Charpy impact test. The Charpy impact test has many benefits which
make its use ideal in an industry setting—mainly that the specimen has a small footprint in terms of its geometry, the test is simple, a number of tests can be performed in minutes, and a great deal of data exists for all line pipe steel grades. The drop weight tear test has also been used to determine the fracture toughness, but up until this time, it mainly serves to ensure the material will not fracture in a brittle mode.

With this in mind, understanding how separations affect small-scale, laboratory fracture tests and how this in turn relates to full-scale fracture behavior in pipelines is critical to ensuring that the predictive capabilities of fracture control models are maximized. In an ideal world, small-scale laboratory tests would accurately capture and predict real-world behavior. However, as discussed in the literature review section on ductile fracture control, a great deal of uncertainty surrounds the Charpy impact test’s ability to predict the required arrest toughness of a line pipe steel. Larger and larger correction factors are being applied to correlate what is observed experimentally and what is observed in full-scale burst test trials. While a great deal of work has been done of the origin and consequences of separations, the ability of small-scale testing to mimic the consequences of separations in full-scale fracture testing has yet to be fully explored.

At the microstructural level, the origin of separations has been attributed to many factors: (1) elongated ferrite grains, (2) high dislocation density within ferrite grains, (3) high sulfur and phosphorus contents, (4) coarse ferrite grains in between fine ferrite grains, (5) ferrite-pearlite banded microstructures, (6) microstructural banding, (7) cube fiber textures, and (8) through-thickness texture bands and cube texture clusters on the separation plane. The focus
of this work will not be on the genesis of separations but will look at a case where separations are known to occur and investigate them at the macroscopic level.

Until now, the work done on separations at the macroscopic level have shown separations to affect fracture behavior in several ways. Charpy specimens containing separations have been shown to decrease the ductile-to-brittle transition temperature as the number of separations increase. The number of separations has also shown an effect on the total absorbed energy with the energy decreasing as the separation count increases. With each separation, the specimen effectively transforms into a cluster of sub-sized specimen. Each time the specimen divided, the specimen exhibits a lower through-thickness stress, which minimizes the chances of precipitating another separation. In terms of measuring the percent shear area on the surface of a fractured Charpy specimen, separations result in a phenomenon known as the rising upper-shelf (RUS) phenomenon, where the absorbed Charpy energy is seen to increase with increasing test temperature all the while retaining a 100%SA. During the RUS, the separations are observed to diminish in severity.

Separation severity has been characterized by two metrics. The separation index (SI) measures the total length of all separations and divides it by the total fracture surface area. The SI has been the most often used metric to date. Another measure is that of Farber et al. [Far+15], where the separation density is determined by dividing the total length of all separation by the shortest measured separation then evaluating the ratio of that to the total shear area. The SI has been related between Charpy impact tests, DWTTs, and full-scale burst tests. The results of Sugie et al. [Sug+82] show a consistent
relationship between the measured SI of DWTTs and full-scale bursts tests, while the Charpy specimens showed a higher SI that the full-scale burst tests as the separation severity increased. The cause for the discrepancy between the measured separation severity of Charpy specimens and DWTTs has yet to be explored.

**Objective**

In this study, separations and their effects will be examined in an X80 grade line pipe steel with a wall-thickness of 25 mm and an outer diameter of 1168 mm. The major objectives of this study are summarized as follows:

- The Charpy specimen will be evaluated for its ability to accurately capture the effects of separations when compared to full-scale fracture.
- The window of separation severity will be explored for its ability to predict the fracture toughness requirements.
- The potential cause for differing separation patterns will be explored.
- The effect of separations on thinner Charpy specimen thicknesses will be examined.
- A comparison of separation severity metrics will be performed.

The objectives will be accomplished by means of Charpy impact testing and followed by finite element analysis of Charpy specimens, containing weak interfaces, representing separations. In all, the study seeks to inform the pipeline research community and industry of the effectiveness of Charpy impact tests to correlate materials having separations to their anticipated
affects on full-scale fracture behavior. In other words, do separations effect Charpy specimens in the same manner as they affect full-scale pipe fracture?
Part II

Methods
This chapter outlines the steps taken to characterize the strength and elongation properties of the pipe used in this study.

21 round bar tensile (RBT) tests were carried out at ambient temperature (23°C). Four orientations were tested: (1) transverse, (2) longitudinal, (3) diagonal, and (4) through-thickness, which are shown in Figure 5.1. Two unique dimensions were used. One had a 6 mm diameter (D₆) and 12 mm gauge length, and the other had a 12 mm diameter (D₁₂) and 60 mm gauge length (Figure 5.2). Three specimens were tested for each orientation. All orientations were evaluated for the D₆, while only the T, L, and D orientations were tested for D₁₂. The D₆ T, L, and D orientated specimens were machined to compare the non-conventional Z orientation. The pipe’s wall thickness limited the through-thickness specimen dimensions, so the diameter and gauge length were modified. All testing complied with the ASTM A370-16 [AST16] Standard.
All specimens were tested at an average strain rate of $5 \times 10^{-4} \text{ s}^{-1}$ applied through a constant cross-head displacement rate of $5 \times 10^{-3} \text{ mm.s}^{-1}$ and $2.5 \times 10^{-2} \text{ mm.s}^{-1}$ for the D$_6$ and D$_{12}$ specimens, respectively. Elevated strain rates and varying temperatures were not explored.

Tensile specimens were tested on an Instron®8801 servo-hydraulic machine with a 100 kN axial force capacity. The average strain was recorded using an Epsilon axial extensometer (Model 3542) with a gauge length of 50 mm for the D$_{12}$ specimens. A Dantec Dynamics DIC system was used to monitor average and local strains for all tests. Because the axial extensometer’s gauge length was too large to accommodate the D$_6$ specimen’s parallel section, the DIC system was used to record the average strain for these specimens. The extensometer values were compared to the DIC values to ensure unity.

The following sections describe the procedures for extracting through-thickness specimens and capturing strains using the DIC system. The true stress and true strain values were determined for the D$_{12}$ specimens, using

\[ \text{true stress} \]
To evaluate the through-thickness material properties, a drawn arc stud welder was used to add the material needed to extract tensile and Charpy specimens in the form of welded prolongations. When welded prolongations are required, the ASTM A770/A770M-03(2012)e1 [AST12] Standard proposes four viable welding methods: (a) shielded metal arc, (b) friction, (c) drawn arc stud, and (d) electron-beam welding. The stud welding method was chosen the DIC system. The description for this process is also included.

5.1 **Through-Thickness Specimen Extraction**

Fig. 5.2: Reduced and standard tensile specimen dimensions. [units: mm]
because of its low costs, small HAZ, and repeatability.

Drawn arc stud welding (DASW) is the technique of joining a base metal to a metal stud through an electric arc process. DASW is a popular method used in the aerospace, automotive, construction, electronic, and shipbuilding industries. A DASW system comprises a control unit, welding gun, weld studs, and ceramic ferrule. The control unit provides the welding energy, generally allowing for the adjustment of the current and welding time. The welding gun secures the weld stud and activates the welding process when pressing a trigger. The weld studs are metal rods, formed into varying shapes and sizes. In this study, a cylindrical stud was used. For DASW systems, the weld studs have an ignition tip that precipitates the joining process. The ceramic ferrule serves to concentrate the heat and retains the molten material in the weld area, increasing the bond strength.

Figure 5.3 summarizes the entire process: (1) The weld stud and ceramic ferrule are firmly placed against the base metal. (2) Upon triggering, the welding gun lifts the stud and initiates a controlled electric arc, melting the
ignition tip along with a portion of the base metal. (3) The weld stud is plunged into the work surface and set in place once the weld solidifies.

To improve the weld quality, rectangular BPs were extracted from the pipe wall, using wire-cut electrical discharge machining (EDM). This provided a flat surface for the studs to adhere to as well as improving the ease of alignment for studs on opposing faces. The BPs were dimensioned to capture as much of the pipe’s wall thickness as possible while providing an adequate surface for welding. The BPs were machined to a minimum of 25 mm thickness with a welding surface area of 50 mm × 50 mm (Figure 5.4). Studs were welded to both sides of the BPs.

Portability is a key advantage of a DASW system; however, this presented a problem during the trialing period. Because of the welding gun’s mobility, having adequately aligned studs proved difficult. Extracting tensile and Charpy specimens requires studs that are in alignment and perpendicular to the base plate surface within a narrow tolerance. To remedy this issue, a housing was designed and constructed for the welding gun (Figure 5.5). The housing was composed of medium-density fiberboard and allowed for
Fig. 5.5: Stud welding gun within housing.
vertical translation while limiting horizontal translation. A vice was mounted to the housing for gripping the BP and assisting with alignment. The housing vastly improved the alignment and perpendicularity of the weld studs to the BP surface.

<table>
<thead>
<tr>
<th>Reference</th>
<th>Time (ms)</th>
<th>Current (A)</th>
<th>Reference</th>
<th>Time (ms)</th>
<th>Current (A)</th>
</tr>
</thead>
<tbody>
<tr>
<td>BP-01</td>
<td>400</td>
<td>800</td>
<td>BP-06</td>
<td>500</td>
<td>900</td>
</tr>
<tr>
<td>BP-02</td>
<td>400</td>
<td>900</td>
<td>BP-07</td>
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<td>BP-03</td>
<td>450</td>
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<td>BP-08</td>
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<td>BP-04</td>
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<td>BP-05</td>
<td>450</td>
<td>1000</td>
<td>BP-10</td>
<td>550</td>
<td>1100</td>
</tr>
</tbody>
</table>

After sufficient alignment was achieved, 10 BPs were machined on which trials of varying currents and welding times were explored to determine the best settings for a quality weld and a minimal HAZ. Four welding times were tested—400, 450, 500, and 550 ms. These were coupled with currents ranging from 800 to 1100 A in increments of 100 A. Table 5.1 shows the welding parameters and BP references.

After welding the first stud, the BPs were air cooled for at least 5 min then submerged in ambient temperature water to remove any residual heat before welding the opposing stud. After the final stud was welded, the BPs were air cooled. A small notch was introduced into the BPs during the wire-cutting process to track the L and T orientations.

To evaluate the DASW parameters, the trial BPs were sliced down the middle with a 4 mm thickness. The sliced samples were then ground, polished, and
lightly etched to reveal the HAZ.

To assess the HAZ depth, 41 HV10 indentation values were captured, spanning from one stud to the next (Figure 5.6). Accompanying this were 88 HV10 evaluations on the parent pipe material, arranged in four columns of 22 indentations each. The parent pipe material underwent the same preparation method as the BP samples except etching. The average HV10 value was 221 in the parent pipe material, with values spanning from 215 to 227 within one standard deviation. The maximum and minimum values were 236 and 204, respectively. The high HV10 values were prominently found nearest to the outer and inner surfaces of the pipe. This is a typical observation for line pipe steels due to the rolling procedures.
Figure 5.7 shows the welded region of three etched BPs, ranging from the lowest to the highest weld settings. In this figure, the effect of welding parameters on the weld quality is evident. The salient features are the weld symmetry, visible defects, and HAZ depth. As the welding parameters are increased, the weld symmetry increases and the defects decrease all the while the visible HAZ stays below 4 mm. BP-01 shows a highly unsymmetric weld with a large cavity at the middle and outer portion of the weld region. This trend was similar for BPs 02 to 06, particularly when assessing for cavities. Any BPs containing visible cavities were rejected as viable candidates for the welding parameters. This was critical to minimize the likelihood of failure in the weld region. BP 10 showed the greatest level of weld symmetry with no visible defects. Furthermore, the HV10 values were within a reasonable range of the parent material. Figure 5.8 shows the HV10 for BP-10, comparing it to the parent pipe material data.

With the welding parameters determined to be that of BP10, 23 BPs were stud welded to provide 3 tensile specimens and 20 Charpy specimens. Specifics on through-thickness Charpy extraction will be covered in Section 6.1.

The through-thickness tensile specimens were machined so that the parallel
section and the shoulders lay within the base plate material. The fusion zone between the weld studs and BP was placed in the tensile grip section. This served two purposes: first, so that the parallel section would not be influenced by the stud material and weld zone, and second, to completely remove the possibility of failure in the weld region. Figure 5.2 shows the dimensions for the through-thickness tensile specimens.

5.2 Digital Image Correlation

A digital image correlation system was used to determine the true stress/strain curve for the T and L tensile specimens during the entire test duration.

DIC systems have been in use for decades, but have seen considerable use in the past few years. A DIC system is able to monitor local strains on a wide variety of surfaces and materials. Where traditionally an extensometer, strain gauge, Pi Tape®, or strain grid has been used to account for strains
during material and pipe integrity testing, DIC system are now providing a viable method to gather more precise, localized data. Examples of studies using DIC techniques, during various mechanical tests have be performed by Beardsmore et al. [Bea+13], Chernyatin et al. [CML16], Chu et al. [CRS85], Dehnavi et al. [Deh+14], Dubois et al. [Dub+12], Kim et al. [Kim+13], McNeill et al. [MPS87], Park et al. [Par+17], Sutton et al. [Sut+86; Sut+83], Wenman and Chard-Tuckey [WC10], and Zhang and He [ZH12].

5.2.1 Preliminary Concepts

At its basic level, a DIC system comprises a camera, light(s), and an image processing unit. However, a single camera setup is only able to capture displacements and translations fixed in a two-dimensional plane. Upgrading to a two camera setup enables monitoring of displacements and translations in a three-dimensional sense in much the same way as the human eye. Some DIC systems allow for a multitude of cameras, and the setup used in this study implemented a four camera configuration.

5.2.1.1 Solid Mechanics Deformation

The fundamentals of DIC begin with the basic deformation theory of solid mechanics [Bow09; Mal69]. The following reflects the formulation described in “Applications of digital-image-correlation techniques to experimental mechanics” by Chu et al. [CRS85]. While DIC system manufactures might employ a slightly different formulation,
Consider a body $\mathcal{B}$, which deforms in a Euclidean space $\mathcal{E}$ (Figure 5.9). Now, consider the line segment $PQ$, which deforms to the line segment $P'Q'$, lying on the newly deformed body $\mathcal{B}'$. If $(u, v, w)$ denote the components of displacement of an arbitrary point in the $x$, $y$, and $z$ directions, respectively, the points $P$ and $Q$ are located at $(x, y, z)$ and $(x + dx, y + dy, z + dz)$ prior to deformation, respectively. Deformed points $P$ and $Q$ are represented by $P'$ and $Q'$ defined by

$$
P' = (x', y', z') = \left[ x + u(P), y + v(P), z + w(P) \right]$$  \hspace{1cm} (5.1)

$$\begin{align*}
Q' &= (x' + dx', y' + dy', z' + dz') \\
    &= \left[ x + u(P) + u(Q) - u(P) + dx, \\
    \quad y + v(P) + v(Q) - v(P) + dy, \\
    \quad z + w(P) + w(Q) - w(P) + dz \right] \hspace{1cm} (5.2)
\end{align*}$$
The vector lengths between line segments $PQ$ and $P'Q'$ are defined by

$$|PQ| = (ds)^2 = dx^2 + dy^2 + dz^2$$

$$|P'Q'| = (ds')^2 = (dx')^2 + (dy')^2 + (dz')^2$$

(5.3)

Using the relationships $dx' = u(Q) - u(P) + dx$, $dy' = v(Q) - v(P) + dy$, and $dz' = w(Q) - w(P) + dz$ derived from Equation (5.2), Equation (5.3) can be rewritten as

$$|P'Q'| = [u(Q) - u(P) + dx]^2 + \ldots$$

$$\ldots + [v(Q) - v(P) + dy]^2 + [w(Q) - w(P) + dz]^2$$

(5.4)

A linear Taylor’s expansion of the displacement functions about point $P$ gives

$$u(Q) - u(P) \equiv \frac{\partial u}{\partial x} dx + \frac{\partial u}{\partial y} dy + \frac{\partial u}{\partial z} dz$$

$$v(Q) - v(P) \equiv \frac{\partial v}{\partial x} dx + \frac{\partial v}{\partial y} dy + \frac{\partial v}{\partial z} dz$$

$$w(Q) - w(P) \equiv \frac{\partial w}{\partial x} dx + \frac{\partial w}{\partial y} dy + \frac{\partial w}{\partial z} dz$$

(5.5)

From Equation (5.5), projections of the deformed lengths $dx'$, $dy'$, and $dz'$ are
\(dx' \cong \left(1 + \frac{\partial u}{\partial x}\right) \, dx + \frac{\partial u}{\partial y} \, dy + \frac{\partial u}{\partial z} \, dz\)

\(dy' \cong \frac{\partial v}{\partial x} \, dx + \left(1 + \frac{\partial v}{\partial y}\right) \, dy + \frac{\partial v}{\partial z} \, dz\) \hspace{1cm} (5.6)

\(dz' \cong \frac{\partial w}{\partial x} \, dx + \frac{\partial w}{\partial y} \, dy + \left(1 + \frac{\partial w}{\partial z}\right) \, dz\)

Equation (5.6) are the basis for deriving the finite-strain-tensor equations. If \(P\) and \(Q\) are initially oriented along the \(x\)-axis, \(dz = dy = 0\) and Equation (5.6) generate the final values of \((dx', dy', dz')\). Thus, the strain tensor is equal to the engineering strain shown by

\[
\epsilon_{xx} = \frac{|P'Q'| - |PQ|}{|PQ|} = \frac{\partial u}{\partial x} + \frac{1}{2} \left[ \left( \frac{\partial u}{\partial x} \right)^2 + \left( \frac{\partial v}{\partial x} \right)^2 + \left( \frac{\partial w}{\partial x} \right)^2 \right] \hspace{1cm} (5.7)
\]

For DIC analysis, the observed pattern is a two-dimensional projection of an object onto a plane; therefore, the finite-strain equation used to compute strain is given by

\[
\epsilon_{xx} \cong \frac{\partial u}{\partial x} + \frac{1}{2} \left[ \left( \frac{\partial u}{\partial x} \right)^2 + \left( \frac{\partial v}{\partial x} \right)^2 \right] \hspace{1cm}
\]

\[
\epsilon_{yy} \cong \frac{\partial v}{\partial y} + \frac{1}{2} \left[ \left( \frac{\partial u}{\partial y} \right)^2 + \left( \frac{\partial v}{\partial y} \right)^2 \right] \hspace{1cm} (5.8)
\]

\[
\epsilon_{xy} \cong \frac{1}{2} \left( \frac{\partial u}{\partial y} + \frac{\partial v}{\partial x} \right) + \frac{1}{2} \left[ \frac{\partial u}{\partial x} \frac{\partial u}{\partial y} + \frac{\partial v}{\partial x} \frac{\partial v}{\partial y} \right]
\]

Equation (5.8) is relatively general, but follows along the assumptions inherent in digital image processing. First, the image processing assumes that in-
plane deformations and displacements are not affected by out-of-plane displacements. Second, out-of-plane displacements (i.e. $\partial w / \partial x$) are much less than terms like $\partial u / \partial x$ so that the effect is excluded from Equation (5.8).

5.2.1.2 Image Correlation

Performing a DIC analysis starts with applying a speckle pattern to the object in question. Applying the speckled pattern can somewhat be an art in itself, but the general concept is to apply a stochastic-like pattern composed of two colors in large contrast to one another. The most common colors are black and white. There are several application methods and the choice is largely dependent on the individual test. The most simple method is to use canned aerosol paint, painting the background either black or white with the speckles being the opposing color. The size and distribution of the speckles is largely dependent on the camera’s proximity to the object.

Figure 5.10 shows a schematic of a two-camera DIC configuration. On the object’s surface is a speckle pattern indicative of the pattern generally used for analysis. Consider the measured light-intensity pattern reflected from the object surface. The intensity pattern in its initial state is represented by $f(x, y)$ and after deformation is represented by $f'(x', y')$. Between these states an unique, one-to-one correspondence is assumed. Thus, discretizing the surface into a number of subset images allows for tracking the strains and displacements as the object deforms and/or translates.

An example of a measured light-intensity pattern $f(x, y)$ is shown in Figure 5.11. As the image is scanned over the $x$-$y$ plane, the reflected light is
measured by the camera’s sensor. In regions where a black dot exists, most of the light is absorbed which shows as a dip in $f(x, y)$. Having a sufficiently random speckle pattern is important because repeating patterns can lead to misregistration issues [SOS09]. Correlating for each individual speckle is unfeasible, since the speckles and the surrounding area are continuously deforming. Instead, the domain is discretized, forming a subset of gray value patterns, which are used to track relative displacements.

Consider the initially undeformed subset $\mathcal{B}$ in Figure 5.12. Subset $\mathcal{B}$ lies on a scanning area (light-intensity pattern) discretized by sampling grids. At the center of $\mathcal{B}$ exists point $P$, which after deformation translates to $P'$ on deformed subset $\mathcal{B}'$. Using the theory of deformation discussed in the
previous section, the light-intensity values at points $P$ and $P'$ can be written as

$$f(P) = f(x, y)$$
$$f'(P') = f' [x + u(P), y + v(P)]$$ \hspace{1cm} (5.9)

Similarly, for a point $Q$ at position $(x + dx, y + dy)$ on subset $B$, the light-intensity values for points $Q$ and $Q'$ can be written as

$$f(Q) = f (x + dx, y + dy)$$
$$f'(Q') = f' [x + u(Q) + dx, y + v(Q) + dy]$$ \hspace{1cm} (5.10)

Assuming the local intensity value does not change due to deformation, $f(Q) = f'(Q')$ so that
Referencing Figure 5.12, if subsets $\mathcal{B}$ and $\mathcal{B}'$ are small enough that strain lines remain straight after deformation, Equations (5.5) and (5.6) can be used to describe the position of point $Q'$ as

\[
(x'', y'') = (x' + dx', y' + dy') \left[ x + u(P) + dx', y + v(P) + dy' \right]
\]

\[
= \left[ x + u(P) + \frac{\partial u}{\partial x} dx + \frac{\partial u}{\partial y} dy + dx, \ldots \right.
\]

\[
\left. \ldots, y + v(P) + \frac{\partial v}{\partial x} dx + \frac{\partial v}{\partial y} dy + dy \right]
\]
Using Equation (5.5), Equation (5.11) can be rewritten as

\[ f'(Q') = f \left[ x + u(P) + \frac{\partial u}{\partial x}(P) \, dx + \frac{\partial u}{\partial y}(P) \, dy + \, dx, \ldots \right. \]

\[ \left. \ldots, y + v(P) + \frac{\partial v}{\partial x}(P) \, dx + \frac{\partial v}{\partial y}(P) \, dy + \, dy \right] \]  

(5.13)

Therefore, if the displacement of point \( P \) and its the derivative terms \( \frac{\partial u}{\partial x}(P) \) and \( \frac{\partial u}{\partial y}(P) \) are known, the position of any nearby point \( Q' \) can be determined. Similarly, if values for \( u(P), v(P), \frac{\partial u}{\partial x}(P), \frac{\partial u}{\partial y}(P), \frac{\partial v}{\partial x}(P), \) and \( \frac{\partial v}{\partial y}(P) \) are assumed, estimates for point \( P' \) and all points \( Q' \) can be obtained. This statement forms the foundation for the numerical computation of local deformation.

DIC systems are able to correct for camera lens distortions, lighting conditions, and interpolation/noise biases. Depending on the individual DIC system, the user may be able to apply various image matching and/or interpolation methods.

### 5.2.2 Experimental Setup

The DIC system was used to measure the strain for all RBT specimens. A four camera configuration was used to capture a larger portion of the surface area compared to a two-camera configuration. With a two-camera setup, less than quarter of the surface area could be captured, but upgrading to a four-camera setup allowed for monitoring nearly half of the surface area. The recording rate was set to 5 Hz, which allowed for a detailed look at the strain distribution over the entire test duration.
Figure 5.13 shows the DIC configuration on a D_{12} RBT specimen. The specimens were painted with a black background and white speckles.
This chapter outlines the steps taken to characterize the fracture toughness properties of the pipe in question through Charpy testing. The Charpy specimen’s role in fracture control of pipelines is outlined in Chapter 1.

A total of 147 Charpy tests were carried out over a temperature range from −196 °C to 50 °C. Three orientations were looked at—the transverse-longitudinal (T–L), longitudinal-transverse (L–T), and through-thickness longitudinal (Z–L). As a reminder: the first letter describing the Charpy orientation details the direction of the specimen’s length, while the second letter details the notch orientation. A depiction of the Charpy orientations is shown in Figure 6.1. Along with the previously mentioned Charpy specimens, three novel Charpy specimens were tested to investigate the effect of separations on the absorbed fracture energy. These specimens and their rationale are discussed in a section that follows.

An instrumented Instron® MPX impact tester with a 750 J capacity was used for Charpy testing, shown in Figure 6.4. An ISO dimensioned striker was
used and the testing procedures followed those prescribed by ISO148-1:2016 [ISO16]. The dimensions of a full-size Charpy specimen is shown in Figure 6.2. The Charpy specimens were extracted in columns of two through the pipe’s wall thickness, shown in Figure 6.3. This was done for efficiency and to avoid centerline segregation, which is discussed in greater detail in the next section.

Specimens were tested at temperatures from $-196 \, ^\circ C$ to $50 \, ^\circ C$. Liquid nitrogen (LN$_2$) was used for specimens tested at $-196 \, ^\circ C$. A slush bath of LN$_2$ and isopentane (a.k.a. methylbutane), $n$-pentane, ethanol, and acetone was used for test at $-145$, $-120$, $-105$, and $-85 \, ^\circ C$, respectively. Specimens tested from $-75 \, ^\circ C$ to $23 \, ^\circ C$ were submerged in a refrigerated benchtop bath filled with ethanol. For tests at $50 \, ^\circ C$ the specimens were submerged in slightly heated water.

After the specimens were broken, they were immediately removed from the tester, dried, and stored in a desiccant filled container to preserve the fracture
Through-thickness longitudinal (Z–L) Charpy specimens were extracted using the procedures described in Section 5.1. The primary purpose for testing Z–L Charpy specimens was to examine the fracture toughness properties along the rolling direction of the pipe. The rolling direction corresponds to the cleavage plane, which shows a propensity for separations (see Section 2.1). By testing Charpy specimens along the Z–L orientation, the transition temperature could be compared to the standard L–T and T–L orientation. Furthermore, the fracture resistance of separations could be estimated.

The weld quality of the DASW process was critical for successful Charpy
testing along the Z–L orientation, because of the large stresses and strains exhibited on the Charpy specimen by the striker. Compound this with the low test temperatures special considerations had to be made to mitigate the chance of failure in the weld region.

Unlike the through-thickness RBT tests, the Charpy geometry can not be modified in any manner to reduce the stresses at the weld region. The only available approach was to distance the notch as far as possible from the weld zone. However, the solution was not as simple as placing the notch at the center of the base plate.

A phenomenon known as centerline segregation is commonly observed at the mid-thickness of continuously case steel slabs [ASM08]. Centerline segregation has been shown to have deleterious effects on Charpy impact test
[Bor91; Kya+14; Men+02; STM03; UNY13] and can lead to separations during impact testing. Su et al. [Su+16] performed a study of through-thickness Charpy specimens, investigating segregation’s effect of Charpy testing. They determined that Charpy specimens located at segregated regions exhibited a reduction in Charpy impact toughness and greater variability for Charpy specimens extracted from strips with higher segregation levels.

The focus of this study is on separations not originating from centerline segregation bands; therefore, avoiding their effect was required. This was accomplished by moving the Charpy notch off-center at the quarter distance from the top surface of the base plate (Figure 6.5).

One Z–L Charpy specimen was sliced down the center, ground, polished, and etched to reveal the HAZ in relation to the Charpy notch. Figure 6.6 shows the result of this process. The root of the Charpy notch lies approximately 8 mm from the nearest weld region. In this image, a faint line can be seen below of the
Charpy notch. This line is a centerline segregation region, which confirms that adjusting the notch position was successful in avoiding centerline segregation induced failure.

### 6.2 Incised Specimens

Comparing the effect of separations on similar materials is extraordinarily difficult because any process done to diminish or exaggerate the severity of separations will likewise change the material properties. Therefore, comparing materials with different separation characteristics for similar geometries will inevitably be between materials with differing microstructural properties. This complicates the analysis of separations as a lone variable to be studied independent of the microstructural features.

Following along the lines of Embury et al. [Emb+67], Ferguson [Fer78], and Towers [Tow86b] discussed in Section 2.1, a set of modified Charpy specimens were used to evaluate the role separation plays on fracture characterization. Using wire-cut EDM, slits were introduced along the length of T–L Charpy specimens (Figure 6.7). This method was employed for several
reasons: (1) Making a composite specimen of sub-size Charpy specimens requires extensive tracking to assure that the sub-size specimens are grouped and ordered in relation to their position in the pipe wall. Statistically, this might be inconsequential but was avoided to minimize concern of the pipe’s inhomogeneous through-wall properties obscuring the test results. (2) Having the specimens wire-cut as opposed to riveted, welded, or glued, reduced the effects of lateral contraction displacing the specimens apart. (3) Welding the specimens together would require a greater amount of machining and therefore time.

Figure 6.8 shows the dimensions of the modified Charpy specimens. The total thickness of the Charpy specimens was enlarged to account for the removed material during the EDM process. On average, the wire-cutting process removed approximately 0.3 mm of material, so for every incision made on the specimen, 0.3 mm was added to the specimen’s thickness. The incision depth was determined by looking at the width of critical stresses along the length of a Charpy specimen using FEA.
6.3 Measuring Separations

Using the public domain, Java-based image processing software ImageJ, all fracture surfaces were characterized for their separations. This allowed for acquiring statistics on the quantity, length, and distribution of separations and comparing that with the $C_v$ and %SA for all test temperatures and specimen orientations and geometries. A Python script was written to consolidate the data.

Figures 6.9 and 6.10 show the process for measuring each separation. Ellipses are projected on the broken Charpy surface and their major axis lengths are projected back to an original, unbroken Charpy specimen fracture area. The lengths $h_1$, $h_2$, and $h_3$ describe the separation lengths.
CHAPTER 7

**Numerical Model**

*Define your terms, you will permit me again to say, or we shall never understand one another.*

– Voltaire (1694–1778)

*Miracles*

Finite element analysis (FEA) was used to investigate the effects of separations on the fracture behavior of impacted Charpy specimens. Numerical modeling techniques such as FEA provide a great tool to study specimen behavior when an analytical solution is unavailable or impossible. In the case of Charpy impact tests, where complex stress states exist, the material undergoes local damage, and the rapid nature of the test makes gathering experimental data a challenge, FEA offers a means to explore phenomenon that would otherwise be inaccessible.

In this work, the FEA software LS-DYNA® was used to model Charpy impact tests. LS-DYNA® uses an explicit time integration scheme, allowing for modeling of high-speed, short duration events, where inertial forces play a role. The Charpy simulations were run on a super cluster, consisting of 16 nodes with a total of 128 processors. Depending on the computational requirements of the simulation, between two and eight processors were used, which limited the maximum processing time to 18 hours for each simulation.
A total of 23 simulations were used to analyze the effect of separations and the changing stress state during the fracture process. The 23 simulations was composed of 20 full-sized, 10 mm thick Charpy specimens at four temperatures and a range of critical stress values for separation initiation. Three simulations of sub-sized specimen with thickness of 2.5 mm, 3.3 mm, and 5.0 mm was also explored. One more simulation was added of a 10 mm thick DWTT, using the same parameters defined by the Charpy simulations but with the specimen, anvil, and striker dimensions altered to the DWTT standard dimensions, shown in Section 1.3.2.

While only 23 simulations were used in the analysis, over 100 simulations were performed to tune the GTN and CZM parameters to mimic the experimental observations as nearly as possible. The following sections describe the FEA mesh along with the methods and results of the GTN and CZM element tuning.

### 7.1 Finite Element Mesh

FEA models that incorporate damage show a mesh size dependence when standard FEA techniques are used. Therefore, the mesh size must be considered when using the GTN model [Tan+08]. For the case of Charpy impact test simulations, the mesh length in the direction perpendicular to the fracture plane should be explicitly defined and in many instances considered an adjustable material parameter [SB99].

Figure 7.1 shows the finite element mesh for the Charpy simulations used
in this study as viewed from the $x$–$y$ plane. Shown here are the Charpy specimen, anvil, and striker. Along the fracture plane of the Charpy specimen, a symmetry boundary condition was imposed, restricting displacement along the $x$-axis for nodes lying on the $y$–$z$ plane. The anvil was given a fixed displacement condition along all axes along with a contact condition between the anvil and Charpy surfaces. The striker was fixed along the $x$-axis and given an initial velocity of $5.5 \text{ m.s}^{-1}$. The element density of the striker was scaled to provide 750 J of energy, echoing the capacity of the Charpy impact tester described in Chapter 6. A contact condition was also set between the striker’s leading surface and Charpy specimen’s bottom surface. For all contact conditions, a friction coefficient of 0.1 was used for both static and dynamic conditions. The value was determined by sourcing literature (e.g., Shinohara
et al. [SMB16]).

For the full-sized Charpy specimen simulations, interfaces were introduced along the $x$–$y$ plane, which were used to simulate separations. A total of seven interfaces were evenly dispersed along the through-thickness direction of the specimen. Figure 7.2 shows a schematic of the interfaces. This provided two distinct behavior models. The gross fracture of the Charpy specimen was governed by the GTN model, while the separations (interfaces) were governed by the CZM. Their material parameters will be discussed in the sections that follow.
The element height ($\ell_\perp$) perpendicular to the fracture plane ($x$-axis) was set to 100 $\mu$m. Shown in Table 7.1 the smallest element size used by the various authors was 100 $\mu$m. When using the GTN model to simulate a range of test types (e.g. tensile and Charpy), it is important to keep $\ell_\perp$ equivalent across all specimens once the GTN parameters have been established. Because the FEA performed in this study dealt only with Charpy impact tests, the mesh size was critical only insofar as it provided a high enough resolution of the stress and strain field once separations formed. Therefore, the smallest element size found in literature was applied.

Along the $y$ and $z$ axes, the element length was set as close to 100 $\mu$m as possible. This gave an aspect ratio of nearly 1 for each element along the fracture path. Elements of this size carried out to a depth of 1 mm along the $x$-axis. This was done to provide a consistent element size for the CZM elements, modeling separations. The interface element thickness ($z$-axis) was set to 5 $\mu$m, giving them a sufficiently high stiffness as well as minimizing their footprint. Cohesive elements require a sufficiently high stiffness in order to properly function and not influence the surrounding elements by reducing their stress artificially (see Section 3.1.1).

For the 10 mm thick Charpy simulations, only one symmetry condition was used along the $y$–$z$ plane. For these simulations, the total number of elements composing the specimen was 134,253 hexahedral type elements. For the sub-sized specimen, where separations were not considered, two symmetry planes were used. The $y$–$z$ plane was used as described previously as well as a symmetry plane along the $x$–$y$ plane at the specimen’s mid-thickness. This gave the 2.5, 3.3, and 5.0 mm thick specimens a specimen hexahedral element
count of 27,703, 34,096, and 53,275, respectively.

The same striker and anvil meshes were used in all Charpy simulations. The striker had a total element count of 7068 majority hexahedral elements, while the anvil had 700 hexahedral elements. The striker was considered to have an elastic material behavior with a Young’s modulus of 210 GPa, a Poisson’s ratio of 0.3, and an element density of 0.1391 kg.mm⁻³. The anvil was considered rigid.

7.2 Gurson-Tvergaard-Needleman Elements

The Gurson-Tvergaard-Needleman (GTN) model was used to model ductile fracture behavior in the current work. Section 3.2 details the fundamentals of the GTN model. This section focuses on the specifics of implementation and the steps taken to determine the model’s coefficients.

The GTN model accounts for the change in material behavior as “voids” evolve. The model, however, does not account for each individual void but instead modifies the stress carrying capacity of a continuum element as the element is damaged from void growth. The GTN model is governed by the initial void volume fraction \( f_0 \), void volume fraction where damage is nucleated \( f_n \), critical void volume fraction for void coalescence \( f_c \), void volume fraction at failure \( f_f \), void volume fraction at rupture \( f_u \), standard deviation on the nucleation strain \( s_n \), strain where 50% of inclusions are broken \( \varepsilon_n \), constitutive equation coefficients \( q_1, q_2 \), and element height \( \ell_\perp \). \( \ell_\perp \) is not intrinsic to the GTN model but a parameter to be considered
given the model’s susceptibility to mesh size influences. Additionally, the flow stress behavior of the element must be defined.

Table 7.1: GTN Parameters Used for Line Pipe Steel from Various Authors

<table>
<thead>
<tr>
<th>Author(s)</th>
<th>Steel</th>
<th>(f_0) [10^{-4}]</th>
<th>(f_n) [10^{-3}]</th>
<th>(f_c) [10^{-2}]</th>
<th>(\kappa)</th>
<th>(\varepsilon_n)</th>
<th>(s_n)</th>
<th>(q_1)</th>
<th>(q_2)</th>
<th>(\ell_{\perp}) [\mu m]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dotta and Ruggieri [DR04]</td>
<td>X60</td>
<td>80.0</td>
<td>0.20</td>
<td>1.43</td>
<td>0.83</td>
<td>200</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Oh et al. [Oh+07]</td>
<td>X65</td>
<td>1.25</td>
<td>0.8</td>
<td>1.5</td>
<td>0.0</td>
<td>0.3</td>
<td>0.1</td>
<td>1.5</td>
<td>1.0</td>
<td>150</td>
</tr>
<tr>
<td>Nonn and Brauer [NB14]</td>
<td>X65</td>
<td>1.5</td>
<td>3.0</td>
<td>2.0</td>
<td>4</td>
<td>0.3</td>
<td>0.1</td>
<td>1.5</td>
<td>1.0</td>
<td>200</td>
</tr>
<tr>
<td>Rivalin et al. [Riv+01]</td>
<td>X70</td>
<td>1.5</td>
<td>0.074</td>
<td>3.8</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Chen and Lambert [CL05]</td>
<td>X70</td>
<td>20.0</td>
<td>0.18</td>
<td>0.19</td>
<td>0.3</td>
<td>0.1</td>
<td>1.43</td>
<td>0.95</td>
<td>250</td>
<td></td>
</tr>
<tr>
<td>Qiu et al. [QYZ11]</td>
<td>X80</td>
<td>1.5</td>
<td>0.0</td>
<td>5.36</td>
<td>0.15</td>
<td></td>
<td>1.245</td>
<td>0.88</td>
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<td></td>
</tr>
<tr>
<td>Scheider et al. [Sch+14]</td>
<td>X65</td>
<td>0.5</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>250</td>
</tr>
<tr>
<td></td>
<td>X80</td>
<td>1.5</td>
<td>1.5</td>
<td>2.0</td>
<td>4.0</td>
<td>0.3</td>
<td>0.1</td>
<td>1.5</td>
<td>1.0</td>
<td>250</td>
</tr>
<tr>
<td></td>
<td>X100</td>
<td>5.0</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>200</td>
</tr>
<tr>
<td>Tanguy et al. [Tan+08]</td>
<td>X100</td>
<td>1.35</td>
<td>2.0</td>
<td>4.5</td>
<td></td>
<td></td>
<td>1.6</td>
<td>1.0</td>
<td>200</td>
<td></td>
</tr>
<tr>
<td>Nonn and Kalwa [NK12]</td>
<td>X100</td>
<td>1.5</td>
<td>5.0</td>
<td>2.0</td>
<td>4.0</td>
<td>0.3</td>
<td>0.1</td>
<td>1.5</td>
<td>1.0</td>
<td>200</td>
</tr>
<tr>
<td>Shinohara et al. [SMB16]</td>
<td>X100</td>
<td>2.0</td>
<td>5.0</td>
<td>4.5</td>
<td></td>
<td></td>
<td>3.33</td>
<td>0.8</td>
<td>100</td>
<td></td>
</tr>
</tbody>
</table>

Table 7.1 provides a collection of the GTN parameter values used in literature for varying grades of line pipe steel. The line pipe steel grade in this table ranges from X60 to X100. This table was used extensively when determining the GTN parameters of this work. In some instances, the parameter value was directly pulled from literature (e.g. \(f_c\), \(s_n\), \(\varepsilon_n\)), while in other instances the values served as a guide. The following describes the process for determining each parameter and lastly shows the results of the tuning.

7.2.1 Flow Stress

The flow stress of the material describes its stress response while undergoing deformation or strain. In this study, the experimental results from the tensile
tests were used to define the flow stress of the line pipe steel. In FEA, the flow stress is typically defined as the effective plastic stress ($\sigma_{\text{eff}}^P$) as a function of the effective plastic strain ($\varepsilon_{\text{eff}}^P$). This curve comes into effect once the yield strength of the material is met. The flow stress is defined by the true stress-strain relationship of the material. To determine the true stress-strain relationship, the DIC system and a self-authored MATLAB script was used to determine the stress-strain curve. This process is described in detail in Section 9.2. Here, the results of the study is presented along with the method for fitting the curves.

Figure 7.3 shows the experimental $\sigma_{\text{eff}}^P$ v. $\varepsilon_{\text{eff}}^P$ for tensile tests along the transverse (T) and longitudinal (L) orientations. For each orientation, the Voce equation was fitted to the combined experimental data. A three termed Voce equation was used, taking the form

\begin{equation*}
\sigma_{\text{eff}}^P = A + B \varepsilon_{\text{eff}}^P + C \varepsilon_{\text{eff}}^P \exp(D \varepsilon_{\text{eff}}^P)
\end{equation*}
\[ \sigma_{\text{eff}}^p = \sigma_Y \left( 1 + K_0 \varepsilon_{\text{eff}}^p + K_1 \left[ 1 - \exp\left( -k_1 \varepsilon_{\text{eff}}^p \right) \right] + K_2 \left[ 1 - \exp\left( -k_2 \varepsilon_{\text{eff}}^p \right) \right] \right) \] (7.1)

where \( \sigma_Y \) is the yield stress and \( K_0, K_1, K_2, k_1, \) and \( k_2 \) are coefficients. Equation (7.1) was fitted to the experimental data using the least squares method. \( \sigma_Y \) was defined as the average yield strength at 0.2 % offset (\( R_{p0.2} \)) of the three test for each orientation.

The Voce equation provided a nice fit of the experimental data up to 0.8 mm.mm\(^{-1}\) strain. After this strain value, the material was considered to be perfectly plastic.

### 7.2.2 Strain Rate and Temperature Modification

When performing FEA at high-strain rates and low-temperatures, the stress-strain curve must be modified to account for the changes in measured stress as a result of changing visco-plastic conditions. This is especially important for simulating Charpy impact tests because of the elevated strain-rates resulting from the striker.

For striker velocities in the realm of 5 to 5.5 m.s\(^{-1}\), equivalent strain rates have been reported to be between \( 10^2 \) and \( 10^3 \) s\(^{-1}\) \[RT65; Ser78; Wil66\]. The greatest strain rate is found at the notch root, where values of \( 10^3 \) s\(^{-1}\) have been found using FEA \[Nor79; Ros+99; TN86; TN88\]. An equivalent strain rate of \( 10 \) s\(^{-1}\) averaged over the entire specimen ligament was determined using nuclear-grade pressure vessel steels \[CF02\].
Tensile tests were only conducted at 23°C with a strain rate of $5 \times 10^{-4}$ s$^{-1}$ for the material used in this work. Therefore, empirical or semi-empirical relationships found in literature were explored to account for higher strain rates and changing test temperatures.

For body-centered cubic (BCC) metals, Conrad [Con64] postulated that the flow stress can be considered as the sum of an athermal and thermal stress component. The expression took the form

$$\sigma_f = \sigma_{\text{athermal}} + \sigma_{\text{thermal}}$$  \hspace{1cm} (7.2)

in which $\sigma_{\text{athermal}}$ is related to long-range forces such as long-range stress fields of dislocation pileups, while $\sigma_{\text{thermal}}$ is the stress required for dislocations to overcome short-range obstacles and is dependent on the strain rate and temperature.

Several constitutive equations exist, which modify flow stresses determined at quasi-static conditions to flow stress at elevated strain rates and lower temperatures. Two of the most common for structural steels are the Johnson-Cook [JC83] and Zerilli-Armstrong [ZA87] relations. In recent years, Xu et al. [XBT04] developed a physically-based constitutive equation has been developed which is compatible with the current understanding of the rate controlling mechanism. This constitutive equation was originally developed for ferritic steels but has also been shown to have validity for martensitic steels. The equation takes the form
\[ \sigma_f = \sigma_{\text{athermal}} + \left[ 27.86 - 0.00393T \ln \left( \frac{10^8}{\dot{\varepsilon}} \right) \right]^2 \] (7.3)

where \( \sigma_{\text{athermal}} \) is determined by quasi-static tensile tests at 22°C, \( T \) is the test temperature in Kelvin, and \( \dot{\varepsilon} \) is the strain rate in s\(^{-1}\). The \( \sigma_{\text{thermal}} \) component modifies the flow stress determined by the quasi-static tensile tests for varying strain rates and test temperatures.

Xu and Tyson [XT15] used Equation (7.3) to verify the constitutive equation’s validity for high-strength line pipe steel of grade X70 to X120. Xu and Tyson’s results showed that Equation (7.3) accurately predicted the changing yield strength of line pipe steels over a range of strain rates and test temperatures. Therefore, this study adopted this constitutive equation to account for elevated strain rates and lower test temperatures in the FEA of Charpy impact tests.

The flow stress relationship was provided to LS-DYNA\textsuperscript{®} in the form of a table which contained multiple flow stress curves for varying strain rates. LS-DYNA\textsuperscript{®} used linear interpolation between the curves to obtain the flow stress response of the element at that instance.

### 7.2.3 \( f_0 \)

The initial void volume fraction (\( f_0 \)) is representative of the residual number of voids in the material. In this study, Franklin’s formula was used to determine \( f_0 \). Franklin [Fra69] proposed a formula for determining the initial fraction of non-metallic inclusions in steel, taking the form
\[ f_0 = f_V \frac{\sqrt{d_x d_y}}{d_z} \]  

(7.4)

where \( d_x, d_y, \) and \( d_z \) are the average inclusion dimensions, and \( f_V \) is the void volume fraction. Assuming the inclusions are spherical, \( f_0 \) simplifies to \( f_0 = f_V \).

When considering a steel with inclusions made up of MnS, Franklin’s formula could be used to estimate the initial number of inclusion by the expression

\[ f_V = 0.054 \left( \frac{\%S - 0.001}{\%Mn} \right) \]  

(7.5)

where \( \%S \) and \( \%Mn \) are the weight percent of sulfur and manganese, respectively.

Through the use of atomic emission spectroscopy, the chemical makeup of the X80 line pipe steel was determined. The results are shown in Section 8.1. Pertinent to Equation (7.5) are the \( S \) and \( Mn \) values, which were 0.001 % and 1.69 %, respectively. Using these values Equation (7.5) gave an initial void volume fraction of \( f_0 = 2.2 \times 10^{-5} \).

7.2.4 \( f_n, f_c, f_t, \varepsilon_n, s_n \)

For the GTN parameters \( f_n, f_c, \varepsilon_n, \) and \( s_n \), values from literature were used. Where possible, the value was matched with a similar study concerning an X80 line pipe steel.
For void volume fraction where damage is nucleated and critical void volume fraction for void coalescence, the value used by Scheider et al. [Sch+14] for an X80 line pipe steel was used. This provided that $f_n = 0.0015$ and $f_c = 0.02$.

For the two X80 steel in Table 7.1, a $f_t$ value of 0.15 was used by Qiu et al. [QYZ11] and Scheider et al. [Sch+14]. This value was initially used but overestimated the ductile failure when comparing the simulation to the experimental load-displacement curves. So, the value was incrementally, increased until $f_t = 0.18$ agreed with the experimental observations.

When using $\varepsilon_n$ and $s_n$ to control void nucleation, all authors in Table 7.1 used values of 0.3 and 0.1, respectively. The same values were used in this study.

### 7.2.5 $q_1$ and $q_2$

Falesko et al. [FGS98] provided a means for determining the values of $q_1$ and $q_2$ based on the hardening exponent ($n$) and the ratio of yield strength to the elastic modulus ($\sigma_Y/E$). Using cell model computations, Falesko et al. found that both strength and strain hardening have a relatively strong influence on $q_1$ and $q_2$. Thus, by calibrating the model for different hardening exponents (i.e. the strain hardening response) and ratios of strength to elastic modulus, Falesko et al. provided a set of curves for determining $q_1$ and $q_2$ based on experimental data.

Figure 7.4 shows the resulting calibrated curves of Falesko et al. By using the experimental stress-strain data for the L orientation, the hardening exponent and strength to elastic modulus ratio was determined to be $n = 0.09$ and
\[ \sigma_Y / E = 0.001 \quad \sigma_Y / E = 0.002 \quad \sigma_Y / E = 0.004 \]

By interpolating between the curves in Figure 7.4, \( q_1 \) and \( q_2 \) were found to be 1.4 and 0.96, respectively. These values are relatively similar to the values used in Table 7.1 for studies performed by Chen and Lambert [CL05], Nonn and Brauer [NB14], Nonn and Kalwa [NK12], Oh et al. [Oh+07], and Scheider et al. [Sch+14] of steel grades ranging from X65 to X100.

### 7.2.6 Results

By using semi-emperical relations for the flow stress with strain rate and temperature modifications; common values of \( f_n, f_c, \varepsilon_n, \) and \( s_n \) found in literature; trial and error for \( f_l \); and Faleskog et al.’s relation for determining \( q_1 \) and \( q_2 \), the GTN parameters were determined a fit against the experimental data. A summary of the GTN parameters are provided in Table 7.2.

Figure 7.5 shows a comparison between the FEA and experimental Charpy impact tests for L–T oriented specimens. Four test temperatures are shown
7.2 GTN Parameters Used in This Study

<table>
<thead>
<tr>
<th>$f_0$</th>
<th>$f_c$</th>
<th>$f_n$</th>
<th>$f_f$</th>
<th>$\varepsilon_n$</th>
<th>$s_n$</th>
<th>$q_1$</th>
<th>$q_2$</th>
<th>$\ell_\perp$</th>
</tr>
</thead>
<tbody>
<tr>
<td>$2.2 \times 10^{-5}$</td>
<td>0.02</td>
<td>0.0015</td>
<td>0.18</td>
<td>0.3</td>
<td>0.1</td>
<td>1.4</td>
<td>0.96</td>
<td>100 $\mu$m</td>
</tr>
</tbody>
</table>

at −25, −50, −75, and −100 °C. For all test temperatures, the FEA load-displacement curve is able to capture the maximum load and a majority of the unloading portion of the curve. For $T = -25^\circ C$, the numerical model overestimates the extent of failure. However, as the temperature is lowered, the curve conforms more and more to the experimental data. An important thing to note is that the separation severity grows as the temperature is lowered. It is for this reason that $T = -100^\circ C$ seems to show the numerical model overestimating the ductile capacity. However, for the curve lying beneath the numerical curve at this temperature, the level of separation severity was minimal, which allowed for the simulation to show a better relationship.

Because the simulations is this study took advantage of symmetry along the fracture plane, they were not able “break” in the same manner as the experimental tests. In experimental tests, as the fracture reaches near the striker region, the fracture path often deviates from fracture plane at an angle. The symmetry condition does not allow for this, so the simulations were only considered valid until a little past a striker deflection of 20 mm.

### 7.3 Cohesive Element Behavior

The cohesive elements were used to simulate the separation phenomenon. The cohesive zone model (CZM) is well suited to model separations because
of the CZM’s capacity to capture brittle fracture with little difficulty.

**7.3.1 Maximum Principle Stress Criterion**

In order to obtain the critical stress needed for brittle failure, a \( \sigma_{p1}^{\text{max}} \) criterion was applied to a simulated full-sized Charpy specimen along the fracture plane. The goal was to correlate the \( \sigma_{p1}^{\text{max}} \) with the failure of Charpy specimens at the lower-shelf energy (LSE) in order to obtain a maximum stress value.
Cohesive Element Behavior

which could be used to initiate separation for the cohesive elements.

A full-sized, 10 mm thick Charpy specimen with symmetry conditions along the fracture front and mid-thickness was used to determine \( \sigma_{p1}^{\text{max}} \). The simulation was first ran with \( \sigma_{p1}^{\text{max}} \) ranging from \( 2\sigma_Y \) to \( 5\sigma_Y \). Then, the resulting load-displacement curve was integrated, providing the absorbed Charpy energy \( (C_v) \). The \( C_v \) was compared to values found in the experimental analysis of specimens at the LSE region. Experimentally, the lowest \( C_v \) was 5 J.

After initial trialing, \( \sigma_{p1}^{\text{max}} \) was determined to lie somewhere between \( 4\sigma_Y \) and \( 4.5\sigma_Y \). Therefore, \( \sigma_{p1}^{\text{max}} \) values in this range was explored. Following a few more iterations, a \( \sigma_{p1}^{\text{max}} \) of 2310 MPa (or \( 4.3\sigma_Y \)) was found.

In a study performed by Nonn and Brauer [NB14] of a Charpy simulation at the LSE region of an X65 grade line pipe steel, a \( \sigma_{p1}^{\text{max}} \) of 1980 MPa was determined, which was \( 4.5\sigma_Y \).

At the onset of the study, the through-thickness Charpy impact tests were hoped to reveal the critical stress needed to trigger a separation. However, because of complexities of the Charpy tests (e.g., strain-rate effects and localized adiabatic heating), the results were nebulous. Thus, 2310 MPa was considered as the upper limit for the stress needed to trigger a separation.

### 7.3.2 Traction Separation Law

A simple bilinear traction-separation law (TSL) was used to describe the damage of the cohesive elements. The TSL was modeled after the one
Figure 7.6 shows the TSL when $\alpha = 1$. The proposed TSL considers a linear relationship, where no damage occurs until reaching a critical separation value $\delta_0$. At this point damage is scaled as $\delta$ increases until the failure value of $\delta_F$.

The TSL considers a damage scalar ($D$), which effects the cohesive stress or traction ($T$), described by

$$T = (1 - D)K_0\delta$$  \hspace{1cm} (7.6)

where $K_0$ is the cohesive penalty stiffness scalar.

$D$ is defined by the piecewise function
where $\alpha$ controls the shape of the damage evolution curve.

The cohesive energy ($\Gamma_0$) is shown to be

$$\Gamma_0 = \int_0^{\delta_F} \mathcal{D} \, d\delta = \left(1 - \frac{1}{\alpha + 1}\right) K_0 \delta_0 \delta_F + \left(\frac{1}{\alpha + 1} - \frac{1}{2}\right) K_0 \delta_0^2 \quad (7.8)$$

With $\alpha = 1$, Equation (7.8) simplifies to

$$\Gamma_0 = \frac{1}{2} K_0 \delta_0 \delta_F \quad (7.9)$$

By using the relation $K_0 = \mathcal{T}_0/\delta_0$ and the suggested value of $\delta_F = 4\delta_0$ by Ren and Ru [RR13] in Equation (7.9), $\delta_0$ can be determined by

$$\delta_0 = \frac{2\Gamma_0}{K_0 \delta_F} = \frac{\Gamma_0}{2\mathcal{T}_0} \quad (7.10)$$

$\mathcal{T}_0$ was determined by using the maximum critical principle stress criterion, so only $\Gamma_0$ needs to be determined. In Section 3.1.1, $\Gamma_0$ was shown to be equivalent to the fracture energy $\mathcal{G}$. For plane strain specimens, $\mathcal{G}$ can be written as

$$\mathcal{G} = \frac{(K_{IC})^2 (1 - \nu^2)}{E} \quad (7.11)$$
$K_{IC}$ has been related to the Charpy energy by

$$K_{IC} = \left(12\sqrt{C_v} - 20\right)\left(\frac{25}{B}\right)^{\frac{1}{2}} + 20$$  \hspace{1cm} (7.12)

where $B$ is the Charpy specimen thickness. With $T_0$, $\delta_0$, and $\delta_F$ determined, $K_{IC}$ was adjusted until the CZM provided the same results as the maximum principle stress criterion. The results of the fitting is shown in Figure 7.7.
Part III

Results and Discussion
In this chapter the chemical composition and crystallographic texture of the X80 line pipe steel used in this study is evaluated. The chemical composition was primarily used to determine the initial inclusion fracture for the GTN model. The texture analysis was used to determine the microstructural cause for separations of the steel used.

8.1 Chemical Composition

The chemical composition was determined using atomic emission spectroscopy (AMS) and performed by BlueScope Steel Ltd. in Port Kembla, Australia. Table 8.1 summarizes the results from the AMS results. The steel shows a low sulfur (S) content, which indicates a low inclusion content. The carbon equivalent, $C_{eq}$, was determined by Equation (8.1).
Table 8.1: Chemical Composition (% weight)

<table>
<thead>
<tr>
<th>C</th>
<th>P</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.053</td>
<td>0.007</td>
<td>1.69</td>
<td>0.21</td>
<td>0.0010</td>
<td>0.20</td>
<td>0.22</td>
<td>0.002</td>
<td>0.20</td>
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<tr>
<td>Al</td>
<td>Sn</td>
<td>Nb</td>
<td>Ti</td>
<td>V</td>
<td>Ca</td>
<td>B total</td>
<td>N</td>
<td>C&lt;sub&gt;eq&lt;/sub&gt;</td>
</tr>
<tr>
<td>0.025</td>
<td>&lt;0.002</td>
<td>0.061</td>
<td>0.016</td>
<td>&lt;0.003</td>
<td>0.0005</td>
<td>&lt;0.0003</td>
<td>0.0058</td>
<td>0.441</td>
</tr>
</tbody>
</table>

\[
C_{eq} = C + \left( \frac{Mn + Si}{6} \right) + \left( \frac{Cr + Mo + V}{5} \right) + \left( \frac{Cu + Ni}{15} \right) \quad (8.1)
\]

Table 8.1 was used to determine the initial void volume fraction used in the GTN model.

8.2 Texture

The crystallographic texture of the X80 line pipe steel used in this study were observed by electron backscatter diffraction (EBSD), using the JEOL JSM-7001F at the Electron Microscopy Centre of the University of Wollongong. The JEOL JSM-7001F is a 30 kV analytical thermal field emission gun scanning electron microscope capable of 3 nm spatial resolution. The high resolution EBSD system allows for orientation mapping in scanning and transmission modes. Figure 8.1 shows an image of the JEOL JSM-7001F system. This section discusses the results of the EBSD analyses at both the center and quarter-thickness locations of the pipe wall.

Figure 8.2 shows a schematic of the specimen extracted for the texture analyses. A rectangular base section was extracted from the entire pipe wall-thickness.
The surfaces analyzed were oriented along the ND–TD at two locations—the center and upper-quarter with respect to the ND. The commonly used orientations for texture analysis are the normal direction (ND), transverse direction (TD), and rolling direction (RD). In terms of the pipe orientations, the ND is aligned along the through-thickness axis; the TD is aligned along the transverse axis; and the RD is aligned along the longitudinal axis. The sample was ground and electron polished before EBSD analysis. The EBSD mapping was conducted in a rectangular grid pattern. The software Channel-5 by Oxford Instruments was used for post-processing of the EBSD data.

Figure 8.3 shows the EBSD map of both the center (Figure 8.3a) and quarter (Figure 8.3b) positions. For both figures, the horizontal and vertical axes rep-
resent the TD and ND, respectively. The mapping area was 480 µm × 360 µm with a step size of 1 µm. The map consisted of 1905 grains in Figure 8.3a and 1733 grains in Figure 8.3b when a critical misorientation angle of 15° was used to define the grain boundaries.

In this study, a grain region is defined as being completely enclosed by boundaries that all have a misorientation angle larger than a critical angle—defined here as 15°. Using the software Channel-5, two parameters can be calculated: (1) the equivalent circle diameter and (2) the aspect ratio (AR) defined by the ratio of the major axis to minor axis length when fitting an ellipse. The line intercept method was also used to provide grain size information along the horizontal and vertical axes. Here, 40 lines were drawn over the map and points where lines intercepted a grain boundary were stored. The linear intercept is the length between two adjacent grain boundaries.

Figure 8.4 shows the distribution of the misorientation angle for both the center and quarter locations. The minimum misorientation angle considered was 2°. Two peaks were observed with one located below 5° and the other
between 50° and 55°. Both the center and quarter positions show a similar distribution of misorientation angles and no significant different between the two positions was observed.

Table 8.2 lists the average of the five measured grain characteristics for both the center and quarter positions. Listed are averages of the grain diameter of fitted circle, AR of fitted ellipse, linear intercept along TD orientation (L_{TD}), linear intercept along ND orientation (L_{ND}), and the ratio of L_{TD} to L_{ND}. Both the AR and L_{TD}/L_{ND} ratios are greater than 1, indicating that a large number of grains have pancake shapes. Pancake structures have been
associated with separations formation Mintz et al. [MMM08; MMM07] and Mintz and Morrison [MM07].

Figure 8.5 shows the ODF for center (Figure 8.5a) and quarter (Figure 8.5b) locations with $\varphi_2 = 45^\circ$. Figure 8.6 shows the ideal orientations. Figure 8.6 only covers $\varphi_1$ from 0° to 90°, while Figure 8.5 has $\varphi_1$ from 0° to 180°. Comparing Figures 8.5a and 8.5b to Figure 8.6, shows that the material has (1) a strong RD (often called denoted as $\alpha$-fiber) in which $<110>$ is parallel to the RD; (2) a strong rotated cube orientation ($|001><110>$); and (3) a relatively strong (ND)$\parallel\langle 111 \rangle$ (often denoted as $\gamma$-fiber).

Figure 8.7 shows the density profiles of the texture components along four main fibers—the (ND)$\parallel\langle 001 \rangle$ (Figure 8.7a), (RD)$\parallel\langle 110 \rangle$ (Figure 8.7b), (ND)$\parallel\langle 111 \rangle$ (Figure 8.7c), and TD (Figure 8.7d) fibers. Figures 8.7a to 8.7d indicate that the strongest fibers are along the (RD)$\parallel\langle 110 \rangle$. The (RD)$\parallel\langle 110 \rangle$
fiber primarily consists of the texture components in the $\Phi = 0^\circ$ to $50^\circ$ range or \{001\} $\langle 110 \rangle$ to \{223\} $\langle 110 \rangle$ range. Three peaks appear on the (RD)$\parallel \langle 110 \rangle$ fiber, shown in Figure 8.7b as the \{001\} $\langle 110 \rangle$, \{115\} $\langle 110 \rangle$, and \{112\} $\langle 110 \rangle$ orientations. For the (ND)$\parallel \langle 001 \rangle$ fiber, the main texture components is the rotated cube orientation (\{001\} $\langle 110 \rangle$). The density variation for the (ND)$\parallel \langle 111 \rangle$ in Figure 8.7c is more constant than the other fibers. The TD fiber ($\phi_1 = 90^\circ$) has a main texture of \{554\} $\langle 225 \rangle$ with exception of the rated cube orientation.

Ghosh et al. [Gho+16] conducted EBSD measurements around a separation
crack on a plane perpendicular to the separation propagation plane. The EBSD map reported by Ghosh et al. in Figure 8.8 shows that the separation propagates along the interfaces between the (ND)∥⟨111⟩ and the (ND)∥⟨001⟩ fibers.

Comparing the current study to that of Ghosh et al., Figure 8.9 shows the EBSD map of both the center (Figure 8.9a) and the quarter (Figure 8.9b) position with the fibers (ND)∥⟨001⟩ and (ND)∥⟨111⟩ isolated. The area fraction of the (ND)∥⟨001⟩ and (ND)∥⟨111⟩ is 7.2 % and 17.2 %, respectively for the center location. The area fraction of the (ND)∥⟨001⟩ and (ND)∥⟨111⟩ is 8.1 % and 11.9 %, respectively for the quarter location. Both fibers have band structures parallel to the ND plane, which can facilitate the initiation and propagation
of separations.

The main textures observed for the material used in this study are transformed from the rolling texture and recrystallization texture of the parent austenite phase, which develops during the hot rolling of line pipe steel [RJ90]. The main rolling textures of the austenite phase are the brass (\(\{110\} \langle 112\rangle\)) and the copper (\(\{112\} \langle 111\rangle\)) textures. After the phase transformation, they transform to \(\{332\} \langle 113\rangle\) and \(\{113\} \{4411\}\) \(\langle 110\rangle\), respectively for higher finishing temperatures; while they transform to \(\{554\} \langle 225\rangle\) and \(\{112\} \langle 110\rangle\), respectively at lower finishing temperatures [RJ90].

A very strong \(\{112\} \langle 110\rangle\) texture was observed for the studied material. This indicates that copper is the dominant rolling texture in the parent austenite phase. The strong rotated cube texture observed in this study is believed to be transformed from the recrystallization texture (Cube \(\{100\} \langle 001\rangle\)) of the parent austenite phase. This means that the dynamic recrystallization occurred during hot rolling. The (RD)\(\parallel\) \(\langle 110\rangle\) and (ND)\(\parallel\) \(\langle 111\rangle\) fibers can also be generated during the rolling of the ferrite phase [RJ90].

The observations of Ghosh et al. [Gho+16] showed that separations tend to propagate along the interfaces between (ND)\(\parallel\) \(\langle 111\rangle\) and (ND)\(\parallel\) \(\langle 001\rangle\) fibers. The angle between these two fibers is considered to be a high-angle grain boundary with an angle of 54.7°. Therefore, propagation along the high-angle grain boundary is made easier because propagation along the boundary requires less energy than inside the grain.

Elimination of the (ND)\(\parallel\) \(\langle 001\rangle\) fiber would assist in reducing the occurrence of separations. The (ND)\(\parallel\) \(\langle 001\rangle\) fiber is transformed from the recrystallization
texture of the parent austenite phase, thus eliminating (ND)∥⟨001⟩ fibers may require minimizing the dynamic recrystallization during hot rolling.
Fig. 8.7: Frequency of $\phi_1$ (ND)$\parallel$ ⟨001⟩, $\Phi$ (RD)$\parallel$ ⟨110⟩, $\phi_1$ (ND)$\parallel$ ⟨111⟩, and $\Phi$ TD for the center and quarter positions.
**Fig. 8.8:** (a) Inverse pole figure (IPF) map of the region surrounding a fissure crack on the RD-ND plane and (b) scan performed over the dotted region of (a) [Gho+16].

**Fig. 8.9:** EBSD maps highlighting the (ND)$\parallel$(001) and (ND)$\parallel$(111) for the center and quarter positions.
TENSILE TESTS

All forces occur in pairs, and these two forces are equal in magnitude and opposite in direction.

– Sir Isaac Newton (1643—1727)
Philosophiae Naturalis Principia Mathematica

Tensile testing is a common mechanical test done across all engineering disciplines and is used to quantify material properties such as yield strength, ultimate tensile strength, uniform elongation, etc. Tensile test specimen come in many different sizes depending on the application and testing standard. The tests can be performed over a wide range of temperatures and strain rates.

In this study, tensile tests were performed according to the ASTM A370-16 [AST16] Standard on two different specimen diameters along four orientations. A more detailed description of the tensile testing method can be found in Chapter 5. In the context of pipelines, tensile testing is critical to evaluating the pipe’s material behavior. Tensile tests are also used to define the material behavior response within the FEA framework.

The following sections will discuss the results of the tensile tests by describing the basic mechanical properties, the determination of the true stress-strain...
curves for FEA implementation, the level of anisotropy along the longitudinal and transverse orientations, and the appearance of separations on the fracture surfaces of the transverse and longitudinal orientations.

9.1 Mechanical Properties

The base mechanical properties of any material serve as the first step to understanding its structural capabilities. Knowing the yield strength, ultimate tensile strength, and uniform elongation is the first step to understanding a material’s response to external forces. In the realm of pipelines, the yield strength defines the grade of the line pipe steel, which determines the pipe’s internal pressure capacity. In this study, an X80 grade line pipe material is used. The grade X80 represents a line pipe steel with a minimum yield strength of 80 ksi or 552 MPa. The specified minimum yield strength requirement pertains to the transverse orientation since it corresponds with the hoop stress orientation and where the stresses are greatest during pipe operation.

The mechanical properties in this section derive from tests done on an Instron 8801 servo-hydraulic machine with a 100 kN axial force capacity. The engineering strain was monitored using an extensometer and the DIC system. The tensile specimens with a 12 mm diameter (D_{12}) had an initial gauge length of 50 mm, while the specimens with a 6 mm diameter (D_{6}) had a gauge length of 12 mm due to the reduced parallel section length. Only the DIC system was used to monitor strains for the D_{6} specimens.

Figure 9.1 shows the engineering stress-strain curves for the L, T, D, and Z
orientations. The L, T, and D specimens shown in this figure have a diameter $D_{12}$, while the Z specimen has a diameter $D_6$.

The L, D, and Z orientations have a smooth, rounded transition from yield to ultimate strength. However, the T specimens show a mild yield plateau with one specimen having a drop in load after yield. The T specimens are oriented along the hoop direction of the pipe, which is subject to the straining during pipe formation. This is possibly the reason for the yield plateauing of the T specimens.

Table 9.1 provides the average yield strength at 0.2 % offset ($R_{p0.2}$), yield
Tensile Tests

Table 9.1: Averaged Tensile Properties for Each Orientation

<table>
<thead>
<tr>
<th>Orientation</th>
<th>E [GPa]</th>
<th>Rp_{0.2} [MPa]</th>
<th>Rt_{0.5} [MPa]</th>
<th>Rm [MPa]</th>
<th>(\epsilon_u) [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>T</td>
<td>219</td>
<td>628</td>
<td>627</td>
<td>719</td>
<td>8.8</td>
</tr>
<tr>
<td>L</td>
<td>205</td>
<td>533</td>
<td>541</td>
<td>682</td>
<td>9.0</td>
</tr>
<tr>
<td>D</td>
<td>200</td>
<td>540</td>
<td>544</td>
<td>665</td>
<td>9.5</td>
</tr>
<tr>
<td>Z</td>
<td>219</td>
<td>545</td>
<td>545</td>
<td>677</td>
<td>7.7</td>
</tr>
<tr>
<td>All</td>
<td>211</td>
<td>562</td>
<td>564</td>
<td>686</td>
<td>8.8</td>
</tr>
</tbody>
</table>

strength at 0.5% strain (R_{0.5}), ultimate tensile strength (Rm), and uniform elongation (\(\epsilon_u\)) for all tensile specimen orientations and diameters. The maximum stress values of all specimens are found along the T orientation. The level of strength anisotropy is strongest along the T orientation, as the \(Rp_{0.2}\) is much larger along this orientation. The range between the average \(Rp_{0.2}\) of the L, D, and Z was only 12 MPa. The \(Rp_{0.2}\) difference between the L orientation and the T orientation was 95 MPa, and the difference between the Z orientation and T orientation was 83 MPa.

The D orientated specimens showed the greatest \(\epsilon_u\) with a value of 9.5%, and the Z orientation showed the least with a value of 7.7%. The standard T and L orientations showed a minimal difference in \(\epsilon_u\), only differing by 0.2%. In a later section, the plastic strain anisotropy between these two orientations will be explored.
9.2 Determination of True Stress-Strain Curves

While the engineering stress-strain curve is valid for general mechanical properties, the true stress-strain curve is required for FEA under high plastic straining conditions. The engineering stress-strain curves considers a stress value based on the initial cross-sectional area, whereas the true stress-strain value is updated for a changing cross-section. This section will describe the process for determining the true stress-strain curve that was implemented in the numerical models.

The Dantec Dynamics DIC system used in this study allows for exporting the coordinate data with strain values for each time step. Using this feature, a MATLAB script was created to automatically determine the engineering and true stress-strain curves for the T and L orientations.

This first step was to correct the orientation of the specimen so that the axial strain align vertically during the analysis. Because the DIC system’s cameras are flexible with regard to the object’s orientation, the specimen does not necessarily have to sit perpendicular to the camera lenses. To ensure that the algorithm is measuring strain along the proper axis, a script had to be written to correct specimen misalignment.

This was accomplished by first aligning the coordinate system in the DIC software so that the $y$-axis ran along the parallel length and the $x$-axis ran along the thickness direction. Then, in the MATLAB code, the aligned coordinates were projected to the $x$–$z$ plane, revealing the curvature of the specimen. Figure 9.2 shows a schematic of this process. The coordinate points are projected to the planar surface. A grid pattern is laid over the plane and
each grid containing a coordinate is counted. The object is then rotated using Euler angles, minimizing the number of coordinate occupied grids.

With the object oriented, an initial line gauge \((L_0^{\text{eng}})\) is projected over the surface with a predetermined length (in this case 50 mm). This line segment is monitored over the duration of the test, providing the engineering strain.

In order to determine the true stress-strain curve for the L and T specimens, the line gauge previously mentioned was subdivided into 50 increments. Figure 9.3 shows a schematic of the line gauges used to determine the engineering and true stress-strain curves. As the test progressed, each subdivided line increment \((i)\) was monitored for its true strain value, using the relation

\[
\varepsilon = \ln \left[ \frac{L(t)^i}{L_0^i} \right]
\]

where \(t\) is the time.

As the specimen begins to localize, the true strain of the ligament spanning
the localization region becomes the maximum strain value of all the line segments. Right until the point of failure, this line segment represents the actual true strain of the material. So, with the line segment determined, the data for that segment is retrieved for the entire test duration. This provides the true strain of the material from the initial state up to immediately before rupture.

With the true strain determined, the true stress can be found by assuming a constant material volume, which provides the relation

$$\frac{A_0}{A(t)} = \frac{L(t)}{L_0}$$

(9.2)
where $A$ is the cross-sectional area.

Using this assumption the true stress can be determined by

$$
\sigma_{\text{true}} = \frac{P(t)}{A(t)} = \frac{A_0 L_0}{A(t)} P(t)
$$

(9.3)

For implementation into the Charpy FEA, the effective plastic stress ($\sigma_{\text{eff}}^P$) and strain ($\varepsilon_{\text{eff}}^P$) are required. Figure 9.4 shows the results of the DIC analysis of true stress and strain over the effective plastic region. This simply corresponds to the true stress-strain curve after the yield point.

Figure 9.5 provides the DIC coordinate strain data for a $D_{12}$ transverse tensile specimen used in this study, along with its engineering stress-strain curve. From (A) to (F) the strain field along the vertical direction is shown, starting at the initial state and finishing right before rupture.

Image (B) shows the specimen immediately after reaching the $R_{p0.2}$, where
the strain field over the parallel section is still homogeneous. In image (C), the specimen is at the mid-way point between the yield and ultimate tensile strength. Here, the strain can be seen to be slightly localizing near the lower region of the parallel section. After the ultimate tensile stress is surpassed, image (D) shows a strong localization field in the lower region of the parallel.
section; however, necking at this stage is minimal. At $\epsilon = 18\%$ in image (D) the specimen has a pronounced level of necking, which continues to develop in image (F) until the specimen ruptures.

The true strain line segment for this specimen was contained within the necking region shown in image (F). This allowed for back-tracing the true strain over the entire test duration.

### 9.3 Anisotropy

The material shows an anisotropic response along the pipe orientations. Anisotropy is created by the rolling and forming techniques used in pipe production. To describe the level of plastic strain anisotropy, the Lankford coefficient is commonly used. In this study, the Lankford coefficient was calculated for the T and L specimen orientations. The Lankford coefficient was determined by the ratio between the minimum and maximum diameter reduction.

Figure 9.6 shows a schematic, describing the method for determining the Lankford coefficient. Due to the strain anisotropy, both the T and L specimens showed an elliptical shape. Using calipers with an accuracy of 20 $\mu$m, the minor a major axes of the fracture surface was measured. The Lankford coefficient was determined by the ratio $\Delta_{\text{min}}/\Delta_{\text{maj}}$. Table 9.2 summarizes the findings for the T and L specimens. The L orientation showed the greatest level of strain anisotropy nearly having a $1/2$ ratio of $\Delta_{\text{min}}/\Delta_{\text{maj}}$. This is compared to the T orientation which had nearly a $3/4$ ratio. For both orientations, the
\[ \Delta_{\text{maj}} / \Delta_{\text{min}} \]

**Figure 9.6:** Schematic of anisotropy measurements.

**Table 9.2:** Averaged Strain Anisotropy Values

<table>
<thead>
<tr>
<th>Orientation</th>
<th>( \Delta_{\text{maj}} ) [mm]</th>
<th>( \Delta_{\text{min}} ) [mm]</th>
<th>( \Delta_{\text{maj}} ) [%]</th>
<th>( \Delta_{\text{min}} ) [%]</th>
<th>( \Delta_{\text{maj}} / \Delta_{\text{min}} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>T</td>
<td>2.5</td>
<td>3.5</td>
<td>41.3</td>
<td>57.6</td>
<td>0.72</td>
</tr>
<tr>
<td>L</td>
<td>2.1</td>
<td>3.9</td>
<td>35.5</td>
<td>64.2</td>
<td>0.55</td>
</tr>
</tbody>
</table>

\( \Delta_{\text{min}} \) corresponded to the ND (or through-thickness direction) of the material. The T specimen’s \( \Delta_{\text{maj}} \) corresponded to the RD, and the L specimen’s \( \Delta_{\text{maj}} \) corresponded to the TD.
CHAPTER 10

CHARPY TESTS

This is a very complicated case, Maude. You know, a lotta ins, lotta outs, lotta what-have-you’s.

– Jeffrey “The Dude” Lebowski
   *The Big Lebowski*

This chapter presents the results from the Charpy impact tests for specimens along the L–T, T–L, and Z–L orientation as well as the incised Charpy specimens along the T–L orientation. The chapter begins by describing the DBT curves for all specimens. This is followed by discussions on the results of the incised specimens. The Z–L specimens are investigated for their fracture appearance. The separation characteristics are explored along with their effect on the RUS phenomenon. The chapter concludes with a brief look of separations under the SEM.

10.1 TRANSITION TEMPERATURES AND PERCENT SHEAR AREA

Determining the ductile-to-brittle transition (DBT) curve is commonly used in detailed fracture behavior analyses of line pipe steels. Discussed in more
detail in Chapter 1, the DBT is used to determine at which temperature the material transitions from a predominantly ductile fracture to brittle fracture mode. In this section, the DBT curves are analyzed for the Charpy specimens situated along the L–T, T–L, and Z–L orientations as well as the incised Charpy specimens.

10.1.1 L–T, T–L, and Z–L Orientation

A set of Charpy energies taken over a wide range of temperatures are often best-fitted by the sigmoidal (“S” shaped) Boltzmann function. After replacing the base Boltzmann expression with terms related to a Charpy DBT curve, the expression takes the form

$$A_{C_v} = \frac{\text{USE} + \text{LSE}}{2} + \frac{\text{USE} - \text{LSE}}{2} \tanh \left(\frac{T - \text{DBTT}}{\Delta T}\right)$$

where $A_{C_v}$ is the test response, USE is the upper-shelf energy, LSE is the lower-shelf energy, $T$ is the corresponding temperature, DBTT is the ductile-to-brittle transition temperature, and $\Delta T$ is the transition temperature range. The ability to use readily observable characteristics make the Boltzmann function the perfect candidate for describing the ductile-to-brittle fracture process over a range test temperatures.

Using the least-squares fitting method, all Charpy data was fit to Equation (10.1). Figure 10.1 shows the DBT curves for the L–T, T–L, and Z–L orientations. Also shown is the %SA for each specimen. The fitted curves are combined in Figure 10.1d.
The DBTT for orientations L–T and T–L were −116 °C and −99 °C, respectively. This shift in the DBTT is typical for line pipe steels because of the pipe forming process, which creates a pre-strained condition for the T–L orientation. The pre-straining partially exhausts the ductile response for T–L specimens, which is also confirmed by the tensile results shown, where the yield-to-tensile ratio is higher for the transverse orientation compared to the longitudinal orientation. The USE determined by Equation (10.1) was 261 J and 254 J for the L–T and T–L orientations, respectively. The maximum $C_v$ was 279 J and
278 J for the L–T and T–L orientations, respectively. While the DBTT value was slightly different between the L–T and T–L specimens, the average and maximum USEs were virtually indistinguishable.

Figure 10.1c shows the DBT curve for the Z–L orientation. Compared to the L–T and T–L orientations, the Z–L shows a more scatter and the DBTT is much higher. The DBTT for the Z–L orientation exists around −44 °C. This represents a shift of 72 °C and 55 °C compared to the L–T and T–L orientations, respectively. Because this material showed a propensity for separations along the rolling plane and the texture analysis showed weak interfaces parallel to the rolling plane, it is expected that any test with the primary fracture plane parallel to the rolling direction would see an increase in the probability to initiation a brittle fracture at higher test temperatures. Also, because the effect of separations perpendicular to the main fracture plane is removed by the Z–L orientation, the benefit of a lower DBTT can not be realized.

A consistent trend in 100 %SA is found at Z–L Charpy specimens tested above 23 °C. The average USE for the Z–L specimens was 247 J, which is within the realm of the L–T and T–L orientations.

### 10.1.2 T–L, T–L1, T–L2, and T–L3 Orientations

The DBT curves data for the incised Charpy specimens were also analyzed to evaluate the effect an “artificial” separation has on the fracture behavior. The incised specimens T–L1, T–L2, and T–L3 were compared to their full-sized counterpart along the same orientation. The Boltzmann function was used to best fit the data in the same manner described in the previous section.
Figure 10.2 shows the DBT curves for each of the incised Charpy specimens along with a comparison to the non-incised specimen in Figure 10.2d. With each added incision, there is a discernible drop in the USE. This is in addition to a decrease in the DBTT and a flattening of the transition region. T–L Charpy specimens had a DBTT around −99°C. Adding one incision lowered the DBTT to −126°C, which is a change of −27°C. Having two incisions lowered the DBTT to −140°C, and three incisions lowered the DBTT to −144°C. It must be noted that T–L3 had no test where all sub-sized specimen showed 0 %SA;
therefore, the lower-shelf was not achieved in the same manner as all other Charpy tests. This can lead to a conservative estimate of the DBTT, using the Boltzmann equation.

### 10.2 Incised Charpy Specimens

The incised Charpy specimens serve to provide information on the change in Charpy fracture properties in the instance of a severe separation.

#### 10.2.1 Effect on the Upper-Shelf Energy

The USE for Charpy specimens is critical when assessing a pipeline’s ability to prevent fracture initiation and stop a running ductile fracture. Separations
have been shown to reduce the USE of Charpy specimens. Figure 10.3 shows the result of incisions of the USE. The $C_v$ values in this figure correspond to Charpy specimens with 100%SA and without separations. With separations, the energy values can drop even further, so this figure represents the best-case scenario when the Charpy specimen is behaving as disconnected sub-sized specimens for the entire fracture process.

The data in Figure 10.3 was fit the equation

$$C_v = A \left( \frac{B^*}{n + 1} \right)^\beta$$  \hspace{1cm} (10.2)$$

where $A$ and $\beta$ are coefficients, $B^*$ is the Charpy thickness of a full-sized specimen (10 mm), and $n$ is the number of incisions. In essence, the equation relates the number of sub-sized specimens created with the change in absorbed energy. With the introduction of one incision, the $C_v$ dropped by $-36\%$ compared to the non-incised specimen. As incisions were added, the relative change in $C_v$ decayed. This is because as the specimens’ thicknesses are reduced, a larger portion of the specimen is slant fracture. This correlates with the observations made by several authors, looking at the effect of thickness on $C_v$ [Wal01; WKS16].

### 10.2.2 Relationship of Individual Thicknesses

While sub-sized Charpy specimens were not tested, the incised T–L specimens provided a means to analyze the properties of sub-sized specimens. For each incised specimen, the measured Charpy energy was divided by the number
of sub-sized specimens making up the specimen. In other words, the T–L1 comprised of two 5.0 mm thick sub-sized specimens, so the absorbed Charpy energy was divided by two and the thickness considered to be 5.0 mm. Using the same method, the thicknesses of the T–L2 and T–L3 specimens were considered to be 3.33 mm and 2.5 mm, respectively.

In modern, high-toughness steels, a relationship between the $C_v$ and $B$ is non-linear. The BTCM requires the $C_v$ of a $2/3$-thick Charpy specimen. So, for pipes with wall-thicknesses less than 6.6 mm or where only full-sized Charpy specimens are tested, the determined $C_v$ is converted to the $2/3$-size equivalent.

Figure 10.4 compares the power relationship determined from T–L, T–L1, T–L2, and T–L3 specimens with and without separations. These figures
consider only specimen with 100 %SA.

The thicknesses were normalized and fit to the equation

\[ C_v = C_v^* \left( \frac{B}{B^*} \right) ^\beta \]  

(10.3)

where \( C_v \) is the Charpy energy, \( C_v^* \) is the Charpy energy for a full-sized specimen, \( B \) is the specimen thickness, \( B^* \) is the full-sized Charpy thickness, and \( \beta \) is the power coefficient.

As is clear in Figure 10.4, the difference in absorbed energy for specimens with and without separations is minimal for thicknesses less than 5 mm. Using the above relationship, the difference between the calculated \( C_v \) of a 2/3-thick specimen is 12 J, with the power relationship determined by excluding separations giving 135 J and with separations giving 123 J. By neglecting the existence of separations and fitting all values, the difference is roughly split, giving the 2/3-thick specimen an estimated \( C_v \) of 128 J. Compared to the L–T oriented Charpy specimens, the T–L orientation did not exhibit the same measure of separation severity. The separation features will be discussed in greater detail in a later section, but it must be noted that the power relationship is only valid when the separation severity does not reach a point where there is a notch breach or when the separation spans a majority the fracture ligament. This is because the specimen no longer behaves as its full-thickness equivalent and instead acts more akin to a cluster of sub-sized specimens.

Figure 10.4 also shows the normalized Charpy energy/thickness relationship.
developed by Wallin et al. [WKS16]. Wallin’s formulation has not been verified for line pipe steels, but it serves as a good comparison because of the amount of steels the relationship was developed with. Wallin’s relationship is limited to specimen thicknesses between 2.5 mm and 9 mm. The equation was not intended to be applied to the full-sized specimen, the equation has a flaw in that it does not become unity when B is equal to 10 mm. Wallin’s expression takes the form

\[
\frac{C_v}{C_v^*} \left( \frac{10}{B} \right) = 1 - \frac{0.5 \exp(\nabla)}{1 + \exp(\nabla)}
\] (10.4)

where \(C_v\) is the Charpy energy, \(C_v^*\) is the absorbed Charpy energy for a full-sized specimen, B is the specimen’s thickness, and \(\nabla\) is determined by

\[
\nabla = \frac{2(C_v^*/B - 44.7)}{17.3}
\] (10.5)

The dashed band in Figure 10.4 shows the range for one standard deviation. Even though Wallin’s relationship was not designed for line pipe steels, it fits nicely with the steel tested in this study. This is particularly the case at the smaller thicknesses. Wallin’s expression does not directly relate the toughness of one specimen with another. Instead, it relates the toughness a specimen affected by shear lips to a case without shear lips. That is why when \(B = 10\) mm the Equation (10.4) underestimates the total energy. Considering this, when comparing the equation for specimens without separations (Figure 10.4a) to specimen with separations (Figure 10.4b), Wallin’s relationship seem to perform better. Wallin’s expression was not designed to account for
Along the same lines, Figure 10.5 shows the power relationship for varying test temperatures. Using the same method, described in the previous paragraph, specimens tested at similar temperatures were compared for their power relationships. The analysis shows that for temperatures above $-75\,^\circ C$, the power relationship $\beta$ is 1.61, 1.63, and 1.60 for $-75\,^\circ C$, $-50\,^\circ C$, and $-25\,^\circ C$, respectively. The power relationship transitions from approximately 1.1 at $-96\,^\circ C$ to 0 at $-144\,^\circ C$. At $-196\,^\circ C$ the power flips to $-0.7$. The transition from a linear-thickness relationship must take place between $-96\,^\circ C$ and $-75\,^\circ C$. 

Fig. 10.5: Effect of test temperature on the exponential relationship between $B$ and $C_v$. 

separations, which can explain why the fit is better in Figure 10.4a.
Another important metric from the DBT curve is the DBTT. While the Charpy specimens do not give a direct measure of the full pipe’s DBTT, conversion factors have been developed to relate the DBTT of a Charpy specimen to a DWTT specimen, which have a better relationship to the actual pipe’s DBTT. For this reason, exploring the effect severe separations can have on the DBTT of a Charpy specimen is important.

Figure 10.6 shows the change in the DBTT for specimens with incisions. This is described by considering the individual ply’s thicknesses making up the incised specimens. This allows for comparison with Wallin’s expression, relating the drop in DBTT with the drop in specimen thickness. Wallin’s expression takes the form
\[ \Delta \text{DBTT} = 54.1 \ln \left[ 2 \left( \frac{B}{10} \right)^{1/4} - 1 \right] \]  

(10.6)

where \( \Delta \text{DBTT} \) is the change in the DBTT and \( B \) is the specimen thickness. Equation (10.6) was not designed for incised specimens; however, the composite Charpy specimen’s DBTT has been shown to correspond with the sub-sized equivalent [Fer78], so it is reasonable to expect his relationship to hold. Equation (10.6) accurately predicts the change in transition for the quarter-sized Charpy specimen (T–L3), but since this specimen did not experience a full DBT curve (i.e. the lower-shelf was never consistently achieved), this might be a coincidence. Performing the same experiments on a steel with a higher DBTT might show the trend to be more akin to Equation (10.6). With that said, Equation (10.6) under predicted the change in DBTT for the third- and half-sized specimens. This very well may be the result of the separations lowering the DBTT below what Wallin accounted for. Because determining the DBTT when separations are not present for this steels, a comparison cannot be drawn between Equation (10.6) with and without separations as was done in Figure 10.4. With separations, Wallin’s relationship represents an over-conservative estimate of the drop in DBTT.

### 10.2.4 Load v. Deflection Curves

The incised Charpy specimens provided a unique opportunity to observe the effect a severe separation has on the load-deflection curve of an instrumented Charpy test. The load-deflection curve is obtained by a calibrated strain gauge embedded in the Charpy machines striker. This is done at the factory...
and comes with the purchased system. Generally, load-deflection curves are used for research purposes and are rarely obtained at a pipe mill. This is because for industry purposes, only absorbed Charpy energy is required for most specifications. The deflection portion of the load-deflection curve is the displacement of the striker after making initial contact with the Charpy specimen. Integrating the load-deflection curve provides the total absorbed Charpy energy value.

Figure 10.7 compares the Charpy load-deflection curves for T–L, T–L1, T–L2, and T–L3 specimens tested at −25 °C. Several differences can be observed as the number of incision increase. First, the maximum load, $L_{\text{max}}$, is seen to decrease with increasing number of incisions. T–L has a average $L_{\text{max}}$ of 21 kN. For each incision added, $L_{\text{max}}$ decreases to 19, 18, and 17 kN, respectively. Second, the slope after the maximum load is drastically different for the T–L compared to the T–L1, T–L2, and T–L3 specimens. As incisions are added,
the unloading portion of the curve sees a more rapid drop in load.

Figure 10.8 shows the change in maximum load down to the test temperature $-150\,^\circ C$. The test done at $N_2$ are left off because at this temperature the dynamic nature of the load-deflection curve is too great to capture a consistent $L_{\max}$. For each specimen type a linear fit was applied to show the trend in the $L_{\max}$ as the temperature decreases. The increasing maximum load is due to the coupled effect of temperature and strain rate on the Charpy specimen. As the temperature decreases, the yield stress increases, providing a higher load response. Given the amount of scatter in the data, the linear fit accurately describes the trend for all specimens. As is seen in Figure 10.7 the maximum load drops as incisions are added to the full-sized specimen. However, the linear fits show a near parallel trend with respect to each other. This can give some indication to how the maximum load is affected when a separation
forms prior to the maximum load being reached.

10.3 **Tearing Below the Notch in Z–L Charpy Specimens**

The Z–L Charpy tests had an interesting occurrence, where below the notch the specimen showed a small tear before gross fracture (Figure 10.9). Furthermore, on the face opposite to the tear, the fracture surface exhibits a smooth surface distinct from both brittle and ductile appearances below. The tearing occurred for all specimens tested above $-25^{\circ}C$ and for one specimen at $-50^{\circ}C$, corresponding to the highest $C_v$ for that test temperature shown in plot (C) in Figure 10.1.

Figure 10.10 shows representative fracture surfaces of Z–L Charpy specimens over the range of temperatures where tearing occurred. What is evident in Figure 10.10 is that the tearing exists over a wide temperature range, yet its depth is independent of the test temperature. The tearing show signs of being weak planes along the rolling direction, making them conspicuous of separations. What is most interesting is that the tears occur at all temperatures
above $-25 ^\circ C$, but there is a transition in absorbed Charpy energy over this region. This seems to reveal that while the Z–L specimens provide insight into the transition behavior along the through-thickness direction, they do not indicate the fracture toughness of the separations. Seemingly, the separations are much tougher than the Charpy specimen can reveal.

Figure 10.11 shows the fracture surface on the opposite face (described in Figure 10.9) of the tear region, using a SEM. The fracture surface shows elongated dimples typical of ductile fracture in a shear dominated region. Figure 10.12 shows the transition from the tear region to a finely dimpled region and finally to a cleavage fracture region. Figure 10.12 provides the contrast between the fracture surface of the tear region compared to a finely dimpled area.
10.4 SEPARATIONS

In this section, the separation characteristics are discussed. The separation appearance for the L–T, T–L, and the incised specimens are discussed. An novel characteristic termed the separation area is introduced. The complex fracture surfaces generated from severe separations is investigated. Finally, the rising upper-shelf is discussed for each of the specimens.

10.4.1 Summary of Separation Appearance for All Specimens

Because the Charpy specimens were tested over a wide range of temperatures and the incised specimens provided quasi-sub-sized specimens, several important observations were made pertaining to the effect of separations. Figure 10.13 shows a summary of L–T, T–L, T–L1, T–L2, and T–L3. For each
Fig. 10.13: Charpy fracture surfaces for L–T, T–L, T–L1, T–L2, and T–L3 specimens at varying temperatures.

test temperature shared by all the specimens, a sample was chosen that exemplifies the separation behavior at that temperature.

The separations begin to appear in a majority of the specimens at \(-50{^\circ}\text{C}\);
However, the severity is minimal in the T–L2 and T–L3 specimens. At this temperature, both the L–T and T–L show separations in the bottom half of the fracture plane. T–L1 shows a series of separations that lie below the ductile crack front.

When the temperature decreases to $-75^\circ C$, the separations are readily noticeable on most the specimen surfaces. L–T shows a large separation in the bottom portion of the fracture plane, while T–L shows a cleavage “inlet.” The cleavage inlets expose a brittle fracture surface, which renders the %SA to be below 100 %.

Between $-95^\circ C$ and $-145^\circ C$ the separation grow in severity until cleavage fracture dominates the main fracture plane. At $-120^\circ C$ the T–L specimen exhibits a notch breach with a prominent brittle fracture on one half and a majority ductile on the other. A notch breach occurs for the L–T specimen at $-145^\circ C$; however, for this specimen has retained 100 %SA. The incised specimens transition from a majority ductile fracture surface to brittle and in the case of T–L2 and T–L3, ductile fracture below the notch and brittle fracture below the separation length.

At $-196^\circ C$ all but the T–L3 specimens have 0 %SA. For the T–L3 specimens, it was common to have two specimens with 0 %SA and the other to specimens with ductile fractures just below the notch that transitions to brittle before the mid-way point.
10.4.2 Separation Area

A criterion known as the percent separation area (%DA) has been introduced to characterize the separation severity and relate the severity to the absorbed Charpy energy. The %DA criterion was used to characterize the L–T specimens since they showed the clearest separations. These specimens also exhibited a pronounced RUS, which will be discussed in more detail in a later section. The separation area was measured by considering the width of the separations multiplied by the total height of all separations. This gave the separation as a percentage of the total fracture surface. Figure 10.14 shows an example of this process. The advantage of this method as opposed to the separation index (SI) is that it can be estimated in a similar manner to the %SA determination. This would not require taking images of the specimen and measuring each separation.

Figure 10.15 shows the result of measuring the %DA for L–T specimens with 100 %SA. Along with the correlation of %DA with $C_v$, six example surfaces...
are shown. Samples $A$ and $B$ show the notch breach feature; $C$ has a severe separation that spans a majority of the ligament length; and $D$, $E$, and $F$ have separations in the lower portion of the fracture plane. The samples fell into approximately four groupings. Specimens with a %DA below 15% did not show a change in the absorbed energy compared to the average $C_v$ for the L–T specimens without separations, which was 262 J. Specimens having a similar separation severity as $C$ only showed a slight drop, while $A$ and $B$ showed steep drop. The steep drop of $A$ and $B$ is due to the fact that the separation
Examples of notch breach and fracture mode asymmetry. was able to breach the notch, which rendered the specimen more akin to the incised Charpy specimens.

10.4.3 Complex Fracture Surfaces

While testing the full-sized Charpy specimens two unique fracture appearances resulted from separations—the notch breach and fracture mode asymmetry. Figure 10.16 shows an example of this phenomenon occurring in one sample. The figure shows a Charpy specimen that has experienced a separation severe enough to create a notch breach on the left-hand image and a separation that spanned a majority of the specimen ligament on the right-hand image. Because of these severe separations, the specimen was effectively split into two pieces with one side experiencing a fully brittle fracture mode and the other a fully ductile fracture mode. Such a specimen would render a misinterpretation of the %SA because the specimen does not act as one to accumulate the stresses necessary to initiate a cleavage fracture.
These specimens should be rendered invalid when assessing %SA.

### 10.4.4 Charpy Load-Displacement Curves

The effect of separations was not only realized by the absorbed Charpy energy, but also in the load-deflection curves obtained from instrumented Charpy tests. While observing the load-deflection curves, an abrupt drop in load can be seen for L–T specimens having marked separations. This was first noticed during the numerical modeling, which will be discussed in the next chapter.

Figure 10.17 shows four load-deflection curves for four L–T Charpy specimens with separations. Circled on the graph is the point where an abrupt drop in load occurs, signaling the instance a separation popped. As the separation length increases the load-drop point moves further up the curve. This is due to the separation initiating at an earlier stage in the fracture process. In the last graph, where a notch breach occurs, the load can be seen to drop before reaching the maximum load value shown on the previous graphs. For this specimen, the separation occurred early enough to affect the maximum load of the specimen.

When combining the curves as is seen in Figure 10.18, a similar trend is observed compared to the incised T–L specimens in Figure 10.7. However, the drop in load is not as severe for Figure 10.18 compared to Figure 10.7. The chief reason for this is that the separation does not separate the specimen at the onset and the separation does not split the specimen perfectly in two. Instead, the specimen is roughly split into third- and two-thirds-sized specimens.
In this section the separation severity metrics will be examined for the steel used in the current study. The separation index (SI), separation density ($\rho_{SP}$),
and separation area (DA) will be compared.

10.5.1 Separation Index

The SI is the most common metric used to classify separation severity. SI is determined by the ratio of the total length of all separations to the fracture surface area.

Figure 10.20 shows the SI values for all tested specimen orientations, comparing the level of SI to the test temperature in the left column and the $C_v$ in the right column. The data was divided by the specimens average %SA with one set being above 95 %SA and the other set below 95 %SA. For all orientations the majority shear area specimens show a trend of increasing separation severity with lower test temperatures. The increase of SI for the T–L orientation is not as pronounced as the L–T orientation. This is because of the inlet cleavage...
fractures that limited the extent of separation growth. The trend related to $C_v$ is more clear for the L–T orientation above 95%SA but is more erratic for the other orientations.

10.5.2 Separation Density

The separation density derives from the study of Farber et al. [Far+15], where the separation severity was defined as the ratio of the total separation length divided by the smallest separation to the shear fracture area.

In many ways, $\rho_{SP}$ mirrors the results of Figure 10.20; however, for some orientations the trend is erratic (e.g., T–L1). In the L–T orientation, Figure 10.21 shows a $\rho_{SP}$ value equation to approximately 1.5 mm$^{-2}$. This corresponds to the specimen with a notch breach. This point exists as an outlier and is far beyond the scope of the other $\rho_{SP}$ values. This suggests that the $\rho_{SP}$ may potentially struggle with notch breached specimens. The major drawback of the separation density metric is the use of the smallest separation length. Depending on the inspection detail, the smallest separation length can vary greatly, whereas for the SI small separation lengths will make little difference in the final value.

10.5.3 Separation Severity Comparison

A new separation severity metric was introduced previously known as the separation area (DA). DA is defined by the width of the most extreme separation multiplied by the overall height of the separations. Figure 10.19
compares the separation area against the SI and $\rho_{SP}$ for the L–T and T–L Charpy orientations with >95%SA. A linear trend is determined between each of the metrics.

Figure 10.19 shows the relationship between the DA and the SI to be nearly linear. Based on the results of the Charpy surfaces described in this section, it is safe to assume that the greater the amount of separations the more the separation will be spread across the through-thickness direction. Generally,
### Table 10.1: Advantages and Disadvantages of Separation Severity Metrics

<table>
<thead>
<tr>
<th>Metric</th>
<th>Advantage(s)</th>
<th>Disadvantage(s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>separation index (SI)</td>
<td>objective measurement, not heavily influenced by small separations</td>
<td>does not capture shape or distribution of separation clusters, requires high resolution photograph</td>
</tr>
<tr>
<td>separation density (ρ_{SP})</td>
<td>accounts for shear fracture surface area more precisely</td>
<td>does not capture shape or distribution of separation clusters, requires high resolution photograph, requires accounting for multiple fracture types (i.e., initiation, shear, final, brittle), heavily influenced by smallest measured separation</td>
</tr>
<tr>
<td>separation area (DA)</td>
<td>can be obtained without photography, not heavily influenced by small separations</td>
<td>does not capture shape or distribution of separation clusters, not tested for DWTT and full-scale fracture surfaces</td>
</tr>
</tbody>
</table>

The first and largest separation to form is along the centerline, perpendicular to the fracture plane. If this initiates early enough, new separation will spawn in the newly created partitions. So, when using the SI metric, the value will increase as the number of separations along the through-thickness increases. The DA metric captures this by considering the width of the two most extreme separation along the through-thickness direction. Therefore, both the data in Figure 10.19 and the general trend of separation appearance support the conclusion that the SI and the DA are causally related.
Given the difficulty of measuring all separation lengths, the DA metrics provides a much easier method of determining separation severity all-the-while still capturing the more objective measurement of the SI. Table 10.1 provides a summary of the advantages and disadvantages of each separation metric.
\textbf{Fig. 10.20:} SI over varying test temperatures (left column) and $C_v$ (right column) for Charpy specimen orientations L–T, T–L, T–L1, T–L2, and T–L3 (from top to bottom).
Fig. 10.21: $\rho_{SP}$ over varying test temperatures (left column) and $C_v$ (right column) for Charpy specimen orientations L–T, T–L, T–L1, T–L2, and T–L3 (from top to bottom).
This chapter looks at the occurrence of separations during Charpy impact tests as modeled by finite element analysis (FEA). The previous chapter looked at separations through the lens of experimentation, quantifying separation characteristics at the conclusion of the tests. While a great deal of data can be gathered by experimentation and post-analysis, the evolution of separations during the fracture process of a Charpy specimen is unattainable at this point in time experimentally. To solve this issue, separations were modeled using FEA to shed light on the formation of separations and how they affect the stress state and fracture behavior of Charpy specimens.

A number of analyses has been performed to investigate separations and their effects. FEA consist of a meshed geometry composed of elements with a prescribed material behavior. When external forces are applied to the geometry, the body deforms and stresses amount in accordance to the prescribed material behavior. Because separations fracture in a brittle manner, they have been modeled using a bilinear cohesive zone model (CZM). The bilinear CZM lends itself well to modeling brittle fracture behavior. This
study considered the onset of brittle fracture to be governed by a maximum stress criterion. A range of maximum stresses from 1400 MPa to 2310 MPa was implemented. By ranging the critical stress for separation initiation, the numerical model was able to capture observations from the experimental tests.

11.1 Separation Evolution

One of the premier advantages of numerical modeling is the ability to observe phenomenon which would otherwise be unobservable because of the phenomenon’s ephemeral nature. This is certainly the case regarding the formation of separations in Charpy specimen. Separations form within the Charpy specimen and cannot be observed in real-time due to the rapid nature of formation and the test itself. Thus, by using a numerical model that captures the origin and evolution of separations with Charpy impact tests, a greater understanding of separations’ role in influencing fracture behavior can be obtained.

In this section, a general overview of the separation process is given for a full-sized, 10 mm Charpy specimen. The Charpy specimen has cohesive elements distributed throughout its thickness, which simulate separations via a bilinear traction separation law. Figures are provided that show the through-thickness stress state ($\sigma_{zz}$) of a Charpy specimen (1) preceding separation initiation, (2) separation initiation, (3) extension of the first separation that creates newly formed gross fracture centers, (4) the newly built up stress concentrations
preceeding a second set of separations, (5) newly initiated separations, and
(6) the final state of the fracture surface.

Before gross fracture or any separations trigger, the Charpy specimen under-
goes plastic deformation due to the strains exhibits on the specimen by the
striker. Given that the specimen is notched, a stress concentration forms along
the boundary of the notch root. While basic fracture mechanics can provide
insight on the stresses that form at the base of the notch, the real-world
case often lies beyond the scope of fracture mechanics assumptions. This is
certainly the case of high-toughness Charpy specimens which exhibit a great
deal of ductility before fracture initiates. At the perimeter of the notch root the
Charpy specimen laterally contracts, which changes the through thickness stress state of the specimen. Figure 11.1 shows the lateral contraction and the through-thickness ($\sigma_{zz}$) stress concentration at the base of the notch. The largest $\sigma_{zz}$ is found at the center of the specimen, lessening as it approaches the specimen’s sides. It is here where the separation is most likely to initiate if the stress levels are great enough.

As the striker continues to deform the specimen, the region immediately beneath the notch softens through ductile damage via the growth and coalescence of micro-voids. The GTN damage model was used to capture this phenomenon and represents the gross fracture of the Charpy specimen,
shown in Figure 11.2. This leads to an increased stress concentration near the gross fracture’s tip. For this simulation, the elevated through-thickness stress levels are enough to initiate a separation also shown in Figure 11.2. Although only about 0.2 mm in length at this stage, the separation is large enough to disturb the stress field surrounding it.

With the separation initiated, the through-thickness stress are great enough to propagate the separation farther. Figure 11.3 shows the result of the extended separation. With the separation large enough to remove the through-thickness constraint, which before the separation had spanned the entire thickness, the specimen now behaves as a cluster of two, half-sized Charpy specimens. This
is evident by the newly formed fracture centers marked in the figure. With this shift, the ductile damage is now concentrated at the center of the newly formed halves. As the new halves laterally contract, they open the separation plane even greater, which will be shown in subsequent figures. The separation tip is also labeled in Figure 11.3. It can be seen that the total length of the separation is limited by the region below the tip, where the specimen is under compression.

As with Figure 11.1, as the specimen continues to deform, the newly created halves will experience elevated stress levels, concentrated at their center, shown in Figure 11.4. As the gross fracture moves through the halves, its
The front is incrementally constrained by the first separation’s tip. This reduces the halves’ ability to laterally contract and the stress state builds until new separations form. Another important occurrence as separations create new partitions is that the stress triaxiality decreases. Stress triaxiality is as the ratio of hydrostatic stress to equivalent stress, and governs the evolution of voids in the GTN model implemented here. As the stress triaxiality decreases, the material is more capable of undergoing plastic deformation before the material is damaged and thus softened. However, the traditional GTN model has been shown to not perform well at low stress triaxialities [NB14]. This must be considered when evaluating the current models when the separations divide the surface into the smallest partitions.

Figure 11.5 shows the initiation of a new set of separations in the previously created halves. With each newly created partition, the condition for separations becomes less and less likely because the through-thickness stress is proportional to the newly created partition’s thickness. Another important feature of the Charpy specimen as it deforms is the level of lateral expansion that occurs in the region adjacent to the striker. The material’s level of ductility also governs the amount of lateral expansion. Lateral expansion is a condition that does not occur in a full-scale pipe burst test and is the sole result of the striker’s influence on the specimen’s deformation. The relationship of lateral expansion to separation will be discussed in greater detail in the sections that follow.

Figure 11.6 shows the fracture surface once the second set of separations have elongated enough and a stress concentrations have formed in the quarterly divided specimen. All three separations have extended adjacent to
the region where the specimen is under compression stress along the through-thickness direction. Modern, high-toughness line pipe steels are renowned for not completely breaking when at their USE levels. The Charpy specimens instead bend until they lose contact with the anvils. ISO148-1:2016 [ISO16] defines an unbroken specimen as one which the remaining ligament cannot be severed by human hands. The same circumstance occurs in numerical models. Along with the compressive region, this also limits the extent of the separation in full-sized Charpy specimens.

The figures in this section have given a general overview of the fracture process of a Charpy specimen when separations are present. The following sections
will describe in greater detail various measures used to characterize Charpy fracture behavior and how these measures influence separation attributes.

### 11.2 Separations’ Effect on the Charpy Load-Displacement Curve

Typically, in the pipeline industry, only the $C_v$ is recorded to assess a material’s resistance to fracture. However, occasionally the load-displacement curve is recorded, which allows for greater inspection of the fracture process. On a Charpy impact testing machine, the load is obtained by a strain gauge located
within the striker, which is calibrated to translate strain into force and store this data on the computer. The displacement is determined by the distance the striker travels upon impacting the Charpy specimen. By integrated the load-displacement curve, the $C_V$ can be obtained.

To obtain the load-displacement curves for the numerical models, the total force applied to the striker is recorded, giving the load; and the displacement is tracked by recording the translation of the striker. Because the numerical model used an explicit time scheme, recording each time step would result in excessively large data files. So, the data was recorded at a time step of $2.5 \, \mu s$ for the first $5.0 \, ms$, and a time step of $10 \, \mu s$ for the remainder of the simulation. This allowed for capturing the initial dynamic response of the system, which is seen in the following load-displacement curves, as well as having a low enough time step to capture the initiation and propagation of the separations.

Two approaches were taken to investigate the effect of separations on the load-displacement curves of full-sized Charpy specimens. The first approach was to look at Charpy specimens with the same effective strain-stress definition based on the test temperature but with varying maximum stress criterion which initiate separations. The second approach looked at Charpy specimens with the same maximum stress criterion for separation formation but with varying effective strain-stress definitions based on the test temperature. The effective strain-stress definitions ranged in temperatures of $-100, -75, -50,$ and $-25 ^\circ C$. The maximum stress criteria for separation were $1400, 1600, 1800,$ $2000,$ and $2310 \, MPa$. The two approaches allowed for describing the effect of separations as the specimen’s test temperature is lowered as well as the effect
of separations with a range of maximum stress criteria for specimens tested at the same temperature.

11.2.1 Similar Test Temperatures with Varying Critical Stress Criteria for Separations

The first set of load-displacement curves to be examined consists of simulations at similar test temperatures where the critical stress criteria to initiate separations vary. The critical stress criteria to initiate a separation were 1400, 1600, 1800, 2000, and 2310 MPa. 2310 MPa was determined by the matching the maximum traction ($T_0$) of the cohesive elements to a principle stress criterion for the lowest energies of the L–T Charpy specimens. This provided an upper bound for the stress to initiate a separation. $T_0$ was lowered by increments of 200 MPa as long as the resulting fracture surface resembled that which was observed experimentally.

Figure 11.7 highlights two of the test temperatures with varying $T_0$. Figure 11.7a shows the load-displacement curves at $-100^\circ$C, while Figure 11.7b shows the load-displacement curves at $-50^\circ$C. For all test temperatures when $T_0 = 2310$ MPa, no separations initiated. Therefore, $T_0 = 2310$ MPa served as the “ideal” specimen where separations do not exist.

When a separation initiates, an immediate drop in load was observed. This was also shown to be the case experimentally. In fact, it was this observation in the numerical model that led to a secondary investigation of the experimental load-displacement curves where a similar drop in load was found. The effect
of the initial separation on the load is readily apparent; however, the following instances of separation do no exhibit the same level of load drop.

For all temperatures, as $T_0$ is reduced, the first instance of separation occurs earlier and earlier in the test. This changes the overall time of the test where the specimen is influenced by a partitioned fracture surface, and therefore, the effect of separations becomes more pronounced.

**Fig. 11.7**: Load-displacement curves at two test temperatures with varying $T_0$. 
The most severe change in the load-displacement curve takes place in the test at \(-100 \, ^\circ\text{C}\) with \(\mathcal{T}_0 = 1400\) MPa (Figure 11.7a). In this case, the separation pops right before the maximum load is reached for the other curves. This results in a notch breach and is shown in Figure 11.6. The notch breach removes the through-thickness constraint and the newly formed partitions undergo more extensive ductile deformation. However, now the specimen behaves as approximately two, half-sized specimens for a majority of the fracture process.

The load-displacement curves in Figure 11.7b at \(-50 \, ^\circ\text{C}\) do not exhibit nearly the difference seen at \(-100 \, ^\circ\text{C}\). Because the separations initiate far enough after the maximum load is reached, a notch breach does not occur. This keeps the separations under a higher lateral constraint, and while the separations can extend for a majority of the fracture surface, the constraint limits the specimens ability to behave as a cluster of “ideally” sub-sized specimens.

Figure 11.8 looks at the change in total cumulative energy as the specimen deforms and the fracture propagates for all specimens plotted in Figure 11.7. Since \(\mathcal{T}_0 = 2310\) MPa did not exhibit separation in either case, this is used as the base specimen. The cumulative energy for the other \(\mathcal{T}_0\) values are subtracted from the \(\mathcal{T}_0 = 2310\) MPa value, giving the incremental change in Charpy energy (\(\Delta C_v\)) for different separation initiation times.

As mentioned previously the greatest change is \(C_v\) is seen by the specimen tested at \(-100 \, ^\circ\text{C}\) with \(\mathcal{T}_0 = 1400\) MPa. For this case \(\Delta C_v\) is nearly \(-25\) J compared to the same specimen without separations. As \(\mathcal{T}_0\) is lowered \(\Delta C_v\) is nearly \(-10\), \(-5\), and \(<1\) J, respectively.
Considering the $\Delta C_v$ between $-100^\circ C$ and $-50^\circ C$ with the same $T_0$ values a shift of approximately 12.5, 5, 4, and 0 J are observed for $T_0$ values of 1400, 1600, 1800, and 2000 MPa, respectively.

### 11.2.2 Similar Critical Stress Criteria for Separations with Varying Test Temperatures

The second set of load-displacement curves to be examined consists of simulations with similar critical stress criteria to initiate separations but for different test temperatures.

Figure 11.9 highlights two $T_0$ values, 1800 MPa and 1400 MPa, at test temperatures $-100$, $-75$, $-50$, and $-25^\circ C$. Because of the difficulty in seeing where
Figure 11.9 shows that increasing the test temperature has the same effect as decreasing $T_0$ in Figure 11.7. By decreasing the test temperature, the point where the separation initiates happens sooner in the load-displacement curve. This causes the specimen to be governed by the separations in a greater way.
Figure 11.10 examines the change in total cumulative energy as the specimen deforms and the fracture propagates for all specimens plotted in Figure 11.9. Since $T_0 = 2310$ MPa did not exhibit separation in either case, this is used as the base specimen. The cumulative energy for the other $T_0$ values are subtracted from the $T_0 = 2310$ MPa value at their similar test temperatures, giving the incremental change in Charpy energy ($\Delta C_v$) for different separation initiation times.

Figure 11.10 shows two distinctive groups based on the $T_0$ value. With $T_0 = 1800$ MPa, $\Delta C_v$ differs by a maximum of 5 J from the case without separations. With $T_0 = 1400$ MPa, a much great disparity in $\Delta C_v$ is observed. This is because the initiation point of separations occurs much sooner in the fracture process, and thus, governs gross fracture for a great portion of the fracture process.
11.2.3 Comparison with Experimental Results

For any numerical model to be of value, it must sufficiently capture the observations from experimental tests. As the observations become more complex, accurately capturing their results grows increasingly difficult. Because the separations form by a brittle fracture process, their initiation is governed by a random, spontaneous process. However, the numerical models considered the formation of separations that were ordered and distributed evenly throughout the specimen’s thickness with the initiation criterion equal for all separations in order to isolate and study the variables at play. The following examples from experiments show that equally distributed separations is seldom the case in reality. Even with used a more controlled approach, the numerical modes were able to capture the general results of separations with different levels of severity and initiation locations. The following paragraphs compare the numerical results to Charpy experiments along the L–T orientation at $-108^\circ$, $-96^\circ$, and $-50^\circ$ C.

Three L–T Charpy specimens tested at $-108^\circ$ C are shown in Figure 11.11. While the specimens were tested at the same temperature, their levels of separation severity and instance of separation initiation on the load-displacement curve are different. This fact provided a great opportunity for comparison between the previously described numerical results.

In Figure 11.11, specimens (A) and (B) have approximately the same level of severity and $C_v$, while their initiation points on the load-displacement curve differ by approximately 1 mm. Comparing this to Figure 11.7a, where similar initiation points occur for the $T_0$ values of 1600 MPa and 1800 MPa, a change
in $C_v$ of 5 J was seen for the numerical model. However, the numerical model’s separations were separated by approximately 2 mm on the load-displacement curve. It is reasonable to assume that if the closer $T_0$ values were used, the difference in $C_v$ would be minimum as predicted by the numerical model. The numerical model predicted that a notch breach (shown by $T_0 = 1400$ MPa at $-100\, ^\circ C$ in Figure 11.8) would lead to the greatest drop in energy, which was nearly $-25\, ^\circ C$. The experimental result of specimen (C) showed a drop of 46 J compared to specimens (A) and (B). This is a much larger drop than the one observed numerically. This can be reasoned in several ways. First, in the experimental result the drop in load occurred approximately 1.6 mm before
the maximum load of specimens (A) and (B); however, in the numerical model, the drop occurred right before the maximum load of the other specimens. While this might seem like a small difference, a notch breach is critical to the absorbed energy difference between specimens and the earlier the specimen experiences a notch breach the more the specimen is governed by partitioned subsized specimens. In the numerical case, a level of global fracture of the full-thickness specimen was experienced before the separation initiated. This would increase the $C_v$ response of the test. Secondly, the asymmetry of the experimental specimen (C) has the potential to decrease the absorbed Charpy energy even further. Lastly, the GTN model used here does not account for low triaxiality stress states, which occur when the specimen becomes increasingly thin. When this occurs, the GTN model can underpredict the growth of voids, thus extending the ductile response. Therefore, the numerical model can overpredict the absorbed energy.

Figure 11.12 shows two L–T Charpy specimens tested at $-96^\circ$C. As with the previous figure the specimens show differing levels of separation severity although being tested at the same temperature. Specimen (D) has the highest level of $C_v$ with a value of 255 J, while specimen (E) has a lower $C_v$ of 235 J.
The difference of 20 J is attributed to an earlier initiation point for specimen (E). This reflects the numerical models’ results for the reasons described in the previous paragraph.

As the temperature increases to \(-50^\circ C\), the separations begin to disappear and are relegated to the lower extremities of the Charpy specimens. Figure 11.13 shows two specimen at this temperature—one with and one without separations. Specimen (F) exhibits two separations at the lower portion of the Charpy specimen nearest to the region impacted by the striker. Specimen (G) does not exhibit any separations. Even though specimen (F) has separations, its $C_v$ is identical to specimen (G). For the specimen in Figure 11.8 tested at \(-50^\circ C\) with $T_0 = 2000$ MPa, the separation occurs late in the fracture process. In this figure, the specimen with a separation only shown a difference of 0.7 J. This is negligible and aligns well with the experimental observation.

The previous paragraphs have compared the experimental results to numerical models with varying critical stress criteria for separation initiation at similar temperatures. Numerical analysis also allowed for comparison of similar $T_0$ values but with varying temperatures. When comparing Figures 11.11, 11.12,
and 11.13, it is evident that the separations decrease in severity and their initiation locations fall to lower regions of the fracture surface. Figure 11.9 shows the results of a similar critical stress values for separation as the test temperature is lowered. This figure compares favorably with the experimental results, since it shows that the separation initiation position lowers as the temperature increases, which is also observed in the experimental figures above.

11.3 Effect of Stress State on Separation Formation and Severity

The occurrence of separations is related to the through-thickness stress state ($\sigma_{zz}$) before and during fracture propagation. Depending on the critical stress of the weak planes where separation is susceptible and the through-thickness stress dependent on the material properties, separations can occur at different points in the fracture process of notched specimens. This section will explore the through-thickness stresses during the fracture process of full-sized and sub-sized Charpy specimens as well as a DWTT specimen with a thickness of 10 mm using FEA. The difference in separation appearance based on the specimen’s geometry will be explained through these models. These results also provide an explanation for the occurrence of the notch breach, which is unique to the full-sized Charpy specimen compared to its sub-sized counterpart parts. Combining these analyses, inferences on the effect of separation during the fracture of full-scale pipe is made.
11.3.1 Full-Sized Charpy Specimen

The full-sized Charpy specimen, which has a thickness (B) of 10 mm is the most common Charpy geometry used to qualify the fracture toughness of line pipe steels. Generally, only when the pipe’s wall-thickness is less than 10 mm are sub-sized Charpy specimens utilized. Therefore, understanding how the stress state of a full-thickness Charpy specimen affects the separation behavior during impact testing is key to gaining insight on the usefulness of Charpy specimens to capture separations’ effects.

As discussed in the previous section and as shown in the experimental results, separations initiate at different locations on Charpy fracture surface for full-sized specimens. Either by the range in critical stress values needed to initiate a separation or by lowering the test temperature, separations can initiate near the notch or occur near the bottom portion of the fracture surface (see Figures 11.11, 11.12, and 11.13). However, wherever the separation initiates, it typically extends from its initiation point to the bottom of the fracture surface. This is particularly the case for the first separation, which generally occurs in the middle of the specimen.

This section focuses on the development of through-thickness stress for a full-sized Charpy specimen as it deforms and the fracture propagates. The stresses will be analyzed for varying test temperatures and from the vantage point of the striker’s displacement and the fracture front.

Figure 11.14 shows the load-displacement curve for a Charpy specimen (Figure 11.14a) and the maximum $\sigma_{zz}$ in the central cohesive elements (Figure 11.14b) with the point where the critical stresses of 1400, 1600, 1800, and
2000 MPa are met for each curve. These curves are from a numerical model with a test temperature of −100 °C where separations did not occur.

Figure 11.14b reveals that as the specimen is impacted by the striker and the specimen fractures, the through-thickness stress steadily increases. With this being the case, a wide range of $\sigma_{zz}$ stress values are experienced at the center of the specimen. After the gross fracture begins a range of through-thickness stresses between approximately 1400 MPa to 2300 MPa is experienced by the specimen. This provides a wide range of through-thickness stresses that can pop a separation susceptible to these values of stress. After the maximum $\sigma_{zz}$ value the stress decreases due to the fracture front running into a highly compressed region, where the maximum stress value diverts from the specimen's center. This is the point where the specimen fracture may deviate from the initial fracture plane and tunnel into the specimen. This can be seen at the bottom of many Charpy specimens (see the lower portion
The previous figure looked at the through-thickness stresses for a case where the specimen’s stress-strain conforms to a test temperature at $-100^\circ$C. Figure 11.15 looks at the maximum $\sigma_{zz}$ for specimens resembling test temperatures at $-100^\circ$, $-75^\circ$, $-50^\circ$, and $-25^\circ$C to evaluate the effect of differing stress-strain relationships on the through-thickness stress during fracture. Evident in this figure is that as the temperature increases and thus the yield stress decreases, the through-thickness stress decreases over the entire duration of the Charpy test. This explains why increasing the temperature tends to delay the initiation point for separation, which was observed in the previous section as well as experimentally.

While the largest through-thickness stress lies at the mid-thickness of the specimen during deformation, some experimental fracture surfaces showed separation initiating away from the center. Figure 11.16 shows Charpy fracture
surfaces at temperatures $-120$, $-108$, $-96$, and $-75^\circ$C. The specimens at $-120^\circ$C both have their initial separations off the mid-thickness line, and one specimen at $-108^\circ$C is off center. As the temperature increases, the separations consistently form at the specimen’s centerline. As the specimen deforms, through-thickness stresses are distributed throughout the entire thickness. This $\sigma_{zz}$ stress gradient is shown in Figure 11.1. With the information provided by Figure 11.15, showing how lower temperatures effect the maximum through-thickness stress state, it can be noted that lowering the temperature will not only increase the stress in the center but also increase the stresses at its periphery. Because separations are brittle and been shown to have a range of critical stress criteria, a separation with a low enough critical stress can still pop outside the center region. As the test temperature increases, the deformation required to provide an adequate stress level forces the separations to occur in the center of the specimen at a greater frequency.

Seeing that the through-thickness stress of a full-sized Charpy specimen increases as the specimen fractures, it is important to understand the origin
of this increase. Because the material properties in a Charpy specimen do not change as the fracture propagates, the primary culprit of the increasing $\sigma_{zz}$ stress state is the lateral expansion, which is typically observed in the region impacted by the striker. To investigate this reasoning, a Python script was written to monitor the fracture front of the Charpy specimen, while measuring the lateral contraction and state of stress at the fracture front.

Figure 11.17 provides a schematic of this process along with the measurement taken. The monitored elements were located along the mid-thickness of the Charpy specimen, since this is where the leading edge of the fracture front and largest stresses are located. The fracture front is defined by the most recently deleted element, which has reached a critical void volume fraction and lost
its stress carrying capability. Along the through-thickness direction, defined here as the effective thickness ($B_{\text{eff}}$), the lateral contraction can be monitored with respect to the fracture front. As the fracture moves through the Charpy’s fracture ligament ($\ell$), the location of the maximum stress element with respect to the fracture front is monitored and defined as $d\ell$. The lateral contraction across from the fracture front is defined by $B_{\text{eff}}^{\text{fail}}$, and the lateral contraction across from the maximum stress element is defined by $B_{\text{eff}}^{\text{max}(\sigma)}$.

Figure 11.18 shows the lateral contraction of a full-sized Charpy specimen measured from the fracture front and the element with the maximum $\sigma_{zz}$ as the fracture front progresses through the ligament. Both curves shown that less than 25% of the specimen’s ligament is the specimen allowed to laterally contract. Instead, at approximately 25% for the fracture front and 20% at the maximum $\sigma_{zz}$ element, the lateral expansion from the striker creates a condition where the specimen can no longer contract. This is the
reason for the ever-increasing through-thickness stress state in the full-sized Charpy specimen, shown in Figure 11.15. The lateral expansion at the seat of the specimen expands such that the specimen is not allowed to contract sufficiently, thus producing an artificial constraint of the specimen’s through-thickness deformation. The ideal scenario would be that the specimen is not influenced by the laterally expanding region via the striker’s contact with the specimen.

Figure 11.19 shows the increasing through-thickness stresses as the fracture front moves through the specimen’s ligament. This figure reiterates the idea of the maximum $\sigma_{zz}$ increasing as the fracture moves through the specimen. The effect of the increasing through-thickness stress state means several things with regard to separation formation. First, separations that form in the lower half of a full-sized specimen are likely to initiate because of the increased through-thickness stresses induced by the lateral expansion from the striker’s impact. Second, the total length of separations for full-sized
specimen is influenced by the lateral expansion sufficiently to produce a more exaggerated separation length. Lastly, when comparing the formation of separations between specimens with differing geometries, it is important to consider the effect of the lateral expansion via impact from the striker.

### 11.3.2 Alternative Specimen Geometries

In the previous section, the changing through-thickness stress state for a full-thickness Charpy specimen was shown to effect separation formation. This section will look at other specimen geometries in order to assess their susceptibility to this same phenomenon. Sub-sized Charpy specimens with thicknesses of 2.5 mm, 3.3 mm, and 5.0 mm were explored as well as a 10 mm thick DWTT, which were compared with the full-sized Charpy results.

#### 11.3.2.1 Sub-Sized Charpy Specimens

Sub-sized Charpy specimen are typically used when the pipe’s wall thickness can not accommodate a full-sized, 10 mm Charpy specimen. In this study, sub-sized Charpy specimens made up the incised Charpy specimens. An important feature of separations was observed for the sub-sized Charpy samples compared to the full-sized sample. The full-sized Charpy samples typically had a single separation ranging across a majority of the fracture ligament, whereas the sub-sized samples had a cluster of separations that occurred periodically over the fracture ligament.

Figure 11.20 shows examples of fracture surfaces of a full-sized (B = 10 mm)
Charpy specimen along with sub-sized (B = 5, 3.3, and 2.5 mm) Charpy specimens. On the each surface, examples of repeated separations are indicated by arrows. What is apparent in these fracture surfaces is that as the specimen’s thickness decreases, the level of repeated separations increases. The 10 mm specimen shows a large separation around the center of the specimen, which extends for a majority of the specimen ligament. However, with the large separation effectively reducing the specimen into a cluster of sub-sized specimens, repeated separations can be seen in the newly formed sub-sized specimen on the left-hand side of the fracture surface. This appearance mirrors the surface of the 5 mm specimen to the right. The specimens with thicknesses of 3.3 mm and 2.5 mm have less severe separations, but their separations are repeated along the fracture ligament.

In the previous section, the cause of separations to span the entire ligament in full-sized specimens was related to the increasing through-thickness stresses induced by lateral expansion at the contact region of the striker. As Figure 11.20 shows, separations are not as long and often repeated in the sub-sized Charpy
specimens. The same method of following the fracture front and monitoring the lateral contraction and maximum through-thickness stresses was used to show that the thinner Charpy specimens do not experience an increase in $\sigma_{zz}$ at the same level of the full-sized specimen.

Figure 11.21 shows the value of the maximum through-thickness stress as the fracture front moves through the specimen’s ligament for Charpy specimen thicknesses of 10, 5, 3.3, and 2.5 mm. As the specimen’s thickness decreases, the through-thickness stress fall as well. This is not out of the ordinary and well-known to occur as the thickness is reduced and the specimen converges to a state of plane stress, where the through-thickness stress approach zero. Here, the salient feature is the shape of each curve as a function of the fracture’s position in the specimen. As discussed previously, the full-sized, 10 mm Charpy specimen sees an increase in through-thickness stress at the fracture
propagates. However, as the specimen’s thickness is reduced, the slope of the curve approaches zero for a longer span of the specimen ligament.

The specimen with $B = 5$ mm shows an immediate drop in $\sigma_{zz}$ after fracture propagation begins. After this, the through-thickness stress increases similarly as the 10 mm specimen. This suggests that while the lateral expansion’s influence is not as pronounced as the full-sized specimen, lateral expansion still plays a major role in the through-thickness stresses.

When looking at the specimen with $B = 3.3$ mm the specimen sees a slight drop in $\sigma_{zz}$ after fracture initiation but by the point the fracture front reaches 20% of the ligament, a steady through-thickness stress state is achieved. This steady-state lasts for a little over 30% of the fracture ligament.

For the 2.5 mm thick specimen the through-thickness stress stays constant immediately after fracture initiation. Over 65% of the fracture front experiences a through-thickness stress bounded between 900 MPa and 1100 MPa. This suggests that the laterally expanding region at the bottom of the specimen provides little influence on the through-thickness stress state. For fracture evaluation this represents the ideal case in that a majority of the fracture propagation through the specimen is akin to the conditions in a full-scale pipe burst test, where there are no forces to promote lateral expansion downstream from the fracture front.

Figure 11.22 reiterates this conclusion by analyzing the effective thickness of the Charpy specimen with respect to the location of maximum $\sigma_{zz}$ stress. Shown previously, the 10 mm specimen begins to see an increase in $B_{\text{eff}}$ at around 20% of the fracture front. The same is seen for the 5 mm thick
specimen but the increased ability to laterally contract reduces the specimen $\sigma_{zz}$ enough to distinguish between the two. As with Figure 11.21, the 3.3 mm and 2.5 mm thick specimens show an expanded region of relatively stable contraction.

Keeping these conclusions in mind, the prolonged stable region of fracture propagation seen in thinner specimens helps to explain why the repeated separation pattern shown in Figure 11.20 is possible. The thicker specimens are more heavily influenced by the laterally expanding region, which promotes separations that extend farther across the fracture ligament. The thinner specimens’ stable through-thickness stresses allow for a more natural occurrence of separations with a repeated pattern, which is more commonly observed on the fracture surfaces of full-scale burst tests.
11.3.2.2 Drop Weight Tear Test

To further investigate the conclusions made in the previous section regarding the separation appearance of sub-sized Charpy specimens and the separation behavior with relationship to the through-thickness stress state, a drop weight tear test (DWTT) specimen was modeled using the same parameters as the numerical Charpy models. The DWTT was created with a thickness of 10 mm in order to compare with the full-sized Charpy specimen. A dimensional description of the DWTT is provided by Figure 1.21. Compared to a fracture ligament of 8 mm in a Charpy specimen, the DWTT has a fracture ligament of 70 mm.

In this study, experimental DWTTs were not performed, but to compare the fracture surface of a DWTT specimen to a Charpy specimen, Figure 11.23 is used. Figure 11.23 is adopted from the work of Fujishiro and Hara [FH11],
which looked at separations occurring on DWTTs of grade X80 line pipe steel (the same grade used in this study). Figure 11.23 reveals the DWTTs show similar separation features to Charpy specimens. As the test temperature is lowered, the initial separation moves closer and closer to the notch root. Also shown is that the separation occur in a repeated pattern similar to the patterns seen on the sub-sized Charpy specimen in Figure 11.20.

If the conclusions made about separation appearance holds true, through-thickness stress and lateral contraction results should be obtained for the DWTT specimen and the 2.5 mm thick Charpy specimen. As done earlier, the fracture front of the DWTT specimen was monitored along with the effective thickness referenced across from the maximum through-thickness stress, \( \sigma_{zz} \).

Figure 11.24 shows the value of maximum \( \sigma_{zz} \) as the fracture front progresses.
through both a 10 mm Charpy and DWTT specimen. An immediate difference is apparent between the evolving through-thickness stress states for each specimen. Already discussed is the rising through-thickness stress value for the 10 mm Charpy specimen. However, with a similar thickness in a DWTT dimensioned specimen, the through-thickness stress is far more stable in comparison. Between approximately 5% and 70% of the fracture ligament, the through-thickness stress only varies between 1500 MPa and 1600 MPa. This range in $\sigma_{zz}$ is smaller even than the $\sigma_{zz}$ range for the 2.5 mm specimen.

Overlaid on Figure 11.23 is the approximate position of the fracture front with respect to the specimen’s ligament length. This is done to compare the separation behavior over the fracture surface to Figure 11.24. Assuming that a repeated separation pattern is consistent with a steady, through-thickness stress state, the pattern should be observed between 5% and 70% of the fracture surface. The DWTTs done at 0°C and 20°C show precisely this result. At 0°C repeated separations are observed from around 5% to 80%. At 20°C, a repeated separation pattern is seen from 10% to 70% of the ligament length. The specimen tested at −20°C appears to have captured both the separation appearance under steady $\sigma_{zz}$ conditions as well as the effect of lateral expansion on separation severity. Repeated separations are observed to occur between just below the notch root and 50% of the ligament length. Between 40% and 85% a separation of much greater length has occurred. In this region of the fracture ligament and at a test temperature that would increase the through-thickness stress compared to the other specimens, the laterally expanding region can exaggerate the separation severity. This aligns with the same reported observation for the full-sized Charpy specimen.
The numerical models of the full-sized and sub-sized Charpy specimens along with the DWTT simulation has revealed an important factor when assessing the ability of a notched specimen to accurately reflect the separation appearance observed during full-scale burst tests. The relationship between the specimen’s fracture ligament length and the specimen’s thickness, or the fracture area AR, is an important factor for separations to occur in a manner conforming to their generation in full-scale burst tests. This is due to the influence of lateral expansion at the base of the notched sample, which results from impact via the striker.

Figure 11.25 plots the range of aspect ratios for varying Charpy and DWTT specimen thicknesses. Two regions are highlighted on the figure, where the AR of the Charpy specimens with B equal to 2.5 mm and 3.3 mm are located. These were the two Charpy specimens least effected by the influence on the lateral expansion caused by the striker. For Charpy thicknesses of 2.5 mm,
3.3 mm, 5 mm, and 10 mm, the fracture surface ARs are 0.3125, 0.4125, 0.625, and 1.25, respectively. However, even a DWTT specimen with a thickness of 30 mm only as an AR of 0.4286. This is only slightly larger than the 3.3 mm Charpy specimen’s AR.

Given the conclusions made to this point, the ability of the Charpy specimen to capture the effect of separations like that observed in a full-scale pipe fracture is highly unlikely. The influence of lateral expansion on the appearance of separations in full-sized Charpy specimens exaggerates the separations’ severity measure as a percentage of the total fracture surface. By reexamining the comparison between the measured separation severity between Charpy and DWTTs by [Sug+84], an indication for the discrepancy in separation severity can be explained by the conclusions stated previously.

Figure 11.26 shows the correlation between separation severity between Charpy and the full-scale pipe as well as the correlation between two types of DWTTs and full-scale pipe. If considering a separation index (SI) of 0.1 mm$^{-1}$
for the Charpy specimen in Figure 11.26a, this would equate to a cumulative separation length of 8 mm. The specimen’s fracture ligament is 8 mm long, so considering the general appearance of separations on a full-sized Charpy specimen, it is reasonable assume that the centerline separation would span a majority of the ligament length. As the SI increases, its deviation from the full-scale surface diverges. At \( SI = 0.2 \text{ mm}^{-1} \), the total separation length increases to 16 mm, which suggests that there were multiple separations spanning a majority of the fracture ligament. This conveys that if Charpy specimens are able to capture the separation severity seen in full-scale burst tests, the window of validity is quite small.

With that in mind, the DWTT specimen serves as a much better surrogate for assessing the separation severity of full-scale burst test behavior. The numerical model shows that the DWTT specimen experiences a consistent stress state from approximately 10% to 70% of the fracture surface. The schematic in Figure 11.27 summarizes the proposed valid range for assessing separation severity. This proposal is based on the region of the DWTT specimen least influenced by the laterally expanding thickness at the striker contact region as well as the fracture initiation region below the notch root. Ideally, by
identifying the region of steady-state fracture propagation and evaluating the fracture surface before reaching the striker’s influence, an accurate description of separation formation and appearance can be obtained.

Figure 11.28 highlights the region of the load-displacement diagram where the steady through-thickness stress state exists. Conveniently, this is also the region generally regarded as the propagation energy region. By using this region to determine the fracture toughness, a more accurate representation of the effect of separations can be obtained.
Conclusions and Further Work

A written word is the choicest of relics. It is something at once more intimate with us and more universal than any other work of art. It is the work of art nearest to life itself. It may be translated into every language, and not only be read but actually breathed from all human lips; – not be represented on canvas or in marble only, but be carved out of the breath of life itself. The symbol of an ancient man’s thought becomes a modern man’s speech.

– Henry David Thoreau (1817 – 1862)
Walden

In this work, the separation phenomenon was examined, using an X80 grade line pipe steel with wall thickness of 26 mm and an outer diameter of 1168 mm. The separations were evaluated from four perspectives: texture analysis, tensile tests, along with experimental and numerical Charpy impact tests. The bulk of the analysis focused on the Charpy impact testing, where a set of novel specimens was used to evaluate the effect of separations on fracture behavior, and finite element analysis was used to simulate the inception and growth of separations during the fracture process of a Charpy sample.

This chapter summarizes the finding of the texture, tensile, experimental Charpy, and numerical Charpy results. Conclusions pertaining to the objective described in Chapter 4 are specifically highlighted. Also included are recommendations for future work.
Conclusions

The texture analysis revealed that the material used in this study showed clusters of $(\text{ND})\parallel\langle111\rangle$ and $(\text{ND})\parallel\langle001\rangle$ fibers, which were shown by Ghosh et al. [Gho+16] to create separation initiation sites at interfaces between these fibers. In order to reduce the amount of $(\text{ND})\parallel\langle001\rangle$ fibers, the critical strain promoting the onset of dynamic recrystallization during hot rolling will need to be maximized.

Using a DIC system, the true stress-strain curves were determined for implementation in the Charpy finite element model. The approach considered the true strain within the necking region of the tensile specimen and determined the true stress by implementing the constant volume assumption. This eased the process of determining the GTN parameters.

The experimental Charpy impact tests focused on three main aspects. First, through-thickness (Z–L) Charpy specimens whose fracture plane was aligned along the separation plane were compared to standard Charpy specimens along the L–T and T–L orientations. Second, a set of novel Charpy specimens was manufactured with the goal of capturing the effects of separations by introducing one, two, and three incisions into T–L oriented Charpy specimens. Finally, separation severity measurements were taken on all specimens in order to compare the method of determining the separation index to the separation density as well introducing a new measurement, the separation area.

The Z–L oriented Charpy specimens showed an increased DBTT compared to the traditional L–T and T–L orientations. Where the DBTT of the L–T
specimen was −116 °C and −99 °C for the T–L specimen, the Z–L specimen had a DBTT of −44 °C. Furthermore, a distinctive tearing occurred for all Z–L specimens tested above −50 °C. This was consistent in appearance with separations; however, the tearing only spanned from the notch root to the mid-ligament length of the Charpy specimen. This suggests that the fracture energy of the separations is in fact lower than the test reported since the fracture mode of the lower half of the specimen was sometimes ductile.

The incised Charpy specimens were manufactured to explore the changing fracture behavior as the number of incisions increased. While the incisions represented an extreme case of separations, where the specimen is effectively divided into sub-sized specimens once load is applied by the striker, the incised specimens served to provide information on the decreasing slope after fracture initiation as seen on the Charpy load-displacement curve. When comparing the incised specimens to L–T Charpy specimens showing a wide range of separation severity, the L–T specimens were seen to behave in a similar manner as the incised specimens. This was particularly the case when separations initiated early in the fracture process. The decreasing slope is believed to be related to the change from a majority flat fracture to slant fracture. When separations effectively divide a full-sized specimen into quasi-sub-sized specimens, the fracture surface transforms from a majority flat fracture to a majority slant fracture due to the quasi-sub-sized specimens approaching a plane stress state.

Until now, two separation severity metrics have been proposed. The most often used metric is the separation index (SI), which considers the total length of separations divided by the fracture area. Another metric introduced by
Farber et al. [Far+15] is the separation density ($\rho_{SP}$), which considers the ratio of the total length of separations divided by the smallest separation to the shear fracture area. These metrics were determined for the tests in the current study and compared to a newly defined metric, the separation area (DA). The SI and $\rho_{SP}$ suffer from similar issues in that they require a time exhaustive process to measure all separations lengths. However, when using the DA metric, which is simply the percent area of the fracture surface determined by the width of the two separations closest to the specimen’s edges multiplied by the span of separations along the ligament length, the separation severity could be quickly estimated with the naked eye. Moreover, relationship between the DA and SI was nearly linear, suggesting that there is causal relationship between the two. Determining $\rho_{SP}$ presents the greatest challenge because the metric can change too easily depending on the level of detail applied when finding the smallest separation.

Another important phenomenon is the occurrence of a notch breach in full-sized Charpy specimens. This relates to the case where the separation is severe enough that it severs the notch. When this occurs, the absorbed Charpy energy drastically decreases. The notch breach is easily identified by the appearance of two distinct ductile initiation zones. When this occurs the specimen cannot be said to behave as a full-sized specimen and should not be regarded as a valid Charpy test.

FEA was used to simulate the occurrence of separation during Charpy impact tests. The numerical model provided tremendous insight into separations effects on fracture behavior as well as the specimen geometry’s effects on separation formation and severity.
The Charpy simulations involved full-sized specimens with seven cohesive zone planes distributed evenly along the through-thickness of the specimen. The cohesive interfaces were given varying critical stress criteria for failure and tested for plastic stress-strain conditions at four temperatures. The simulations revealed that separations can initiate at different locations based on either a range of critical stresses at one test temperature or a range of temperatures with one critical stress value. This is a result of an increasing through-thickness stress state brought on by an increasing constraint condition as a result of lateral expansion due to the striker. In effect, as the fracture moves along the ligament, the specimen’s ability to laterally contract near the fracture tip is minimized by an expanding region at the base of the Charpy specimen where the striker contacts. This creates a higher through-thickness stress state which can promote separation initiation as seen in experimental Charpy tests once the test temperature nears the absorbed Charpy energy at plateau ($C_P^*$) of the rising upper-shelf phenomenon.

Another set of simulations was carried out which looked at sub-sized Charpy specimen of thickness 2.5 mm, 3.3 mm, and 5.0 mm along with a DWTT simulation with a thickness of 10 mm. These simulations did not have cohesive elements, but were instead used to compare the evolving through-thickness stress state with that of the full-sized Charpy specimen. The simulations revealed that as the Charpy specimen thickness was reduced, the constraint effect caused by the striker diminished. This allowed for a greater portion of the Charpy specimen to be governed by a steady through-thickness stress state. Adding to this, the 10 mm DWTT simulation showed a steady through-thickness stress state for nearly 70% of the fracture process. Furthermore, the
steady stress state corresponded with a majority of the propagation energy region of the DWTT’s load-displacement curve. It was determined that the specimen’s aspect ratio (AR) played a major role in the ability of the striker to influence the through-thickness stresses. The 2.5 mm Charpy specimen showed the most consistent through-thickness stress during the fracture process. The 2.5 mm thick specimen had an AR = 0.3125, which compares well with the DWTT aspect ratios being only 0.3333 for a specimen thickness of 25 mm. When comparing the DWTT simulation to data fracture surfaces found in literature, the consistent separation appearance corresponded with the range predicted by the simulation.

In summary, the full-sized Charpy specimen does not provide an adequate surrogate when trying to relation separation severity between small-scale, laboratory testing to what is expected in a full-scale pipe fracture. Because the striker introduced an artificial constraint of the specimen that would not exist in a full-scale fracture, the Charpy specimen exaggerates the separation severity. This explains the results of Igi et al. [Igi+16], where the researchers showed the Charpy specimen to have a larger SI when compared to DWTTs and full-scale burst tests.

**Future Work**

- The author is currently involved in a project where two full-scale burst tests of X65 grade line pipe steel will be carried out. As part of this project, a number of Charpy and DWTT will be performed. The author will take this opportunity to use the same techniques described in this
work regarding separations and apply it to the full-scale tests. This will provide a complete set of data regarding separations from small-scale laboratory testing to the full-scale behavior.

▷ In order to improve the numerical model, several avenues can be explored:

- The current model does not consider changing triaxial stress states. As separation form, they reduce the stress triaxiality as a result of an effectively thinner specimen. The GTN model used in this study did not account for low triaxial stress values, which can lead to an overestimation of fracture toughness. Incorporating the model’s response to lower triaxial stress values would provide a better comparison in the steep drop in load upon initiating a separation which was observed experimentally. Works done by Morgeneyer and Besson [MB11], Simha et al. [SXT15], and Xue [Xue07] can provide a great starting point for incorporating this phenomenon.

- Separations were not explored for their thermal or strain rate dependent characteristics.

- Calibrating separations to the Beremin model may provide a way to simulate separations in DWTTs and full-scale tests.

▷ Traditional fracture mechanics specimens such as single-edge notched tension (SE(T)) and compact tension (C(T)) specimens may be used to better characterize the fracture toughness of separations where the influence of a striker can be mitigated.

▷ The issue discussed in the literature review regarding the inability of Charpy impact test specimens to predict the fracture arrestability of a
pipeline at the full-scale for a range of line pipe steel grades has not been addressed by this thesis. Further work must be done to correct for this error and determine what role separation may or may not play.
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